FIBRE - CEMENT HYBRID COMPOSITES

PH. D.

XU GUODONG

October 1994
A thesis submitted for the
degree of Doctor of Philosophy

by

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Department of Civil Engineering
University of Surrey

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Dedicated to my late father
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SUMMARY

The theoretical stress-strain behaviour of individual fibre reinforced cement composites is reviewed. Based on the multiple cracking concept of the existing theory, analytical expressions are developed to describe the tensile stress-strain behaviour of a fibre-cement hybrid composite consisting of three components, i.e. two reinforcing fibres with different moduli, strengths and strains to failure and a common cement binder. The model predicts that the tensile stress-strain curve of the hybrid composites consists of five stages, instead of three stages of the existing models for individual fibre cements, and relates the tensile behaviour of each stage to the component properties of the components and the test system parameters.

A description is given of the physical and mechanical properties of four types of reinforcing fibres used in the study. These were fibrillated polypropylene film, alkali-resistant glass, polyvinyl alcohol fibres and carbon fibres. A small number of direct tensile tests on continuous glass, carbon and polyvinyl alcohol were performed.

The tensile stress-strain behaviour of four types of fibre-cement hybrid composites was studied with particular emphasis on that of the glass-polypropylene hybrids for which the flexural load-deflection behaviour was also examined. It is shown that the fibre-cement hybrid composites yield superior engineering properties over their parent composites and the improvements are sensitive to volume fractions of each of the two fibres. The measured tensile stress-strain curves of the hybrids were compared with the theoretical predictions and satisfactory agreement in general is obtained.

Implications from the present work for the design of fibre-cement hybrid composites are assessed.
I would like to express my sincere gratitude to my supervisor, Professor D.J. Hannant, for his valuable advice and encouragement and tremendous help throughout the duration of the research. I am fortunate to have Professor Hannant as my supervisor.

I am indebted to my collaborative supervisor, Dr. S. Magnani. In addition to many useful technical discussions, his helpfulness in many aspects during my stay as a collaborative research student in Fibronit S.R.L. in Italy is sincerely appreciated.

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APPENDIX I
There are two undesirable properties of unreinforced brittle materials that greatly limit their applications in building components. Since brittle materials usually fail by sudden unstable propagation of pre-existing flaws, no warning is given before failure occurs. Moreover, the size of pre-existing flaws is difficult to control and usually varies significantly from one component to another. As a result, the strength of components made with the same brittle material using similar processes can be very different. Brittle materials are thus also materials of low reliability.

The introduction of fibres into a brittle cementitious matrix can greatly improve its performance. In addition to the enhancement of composite strength by incorporating fibres, which can be well over that of the cement matrix, of far greater importance is the enhancement of the tensile strain capacity of the composite. The ability of the fibres to inhibit unstable crack growth and transform a rapid brittle type of failure into a slow stable fracture, with post-cracking ductility and significantly improved energy absorption capacity prior to failure is well established. The service reliability is therefore improved.

In modern times, the first widely used reinforced composite was asbestos cement, which was developed in about 1900 with the invention of the Hatschek process (1). Asbestos cement is still produced and consumed in some parts of the world more than other fibre reinforced cements due to its unique properties largely resulting from the excellent compatibility between the fibre and the cement matrix.

Asbestos cement can also be a dangerous material (2-3). It was the health risks caused by the production and use of asbestos cement components that
stimulated the world-wide research efforts looking for replacements for the asbestos fibres, and developing asbestos-free fibre reinforced cements. Some asbestos-free products are already produced in continuous industrial processes on a commercial basis. The challenges, difficulties and the philosophy behind the development efforts of the industry to find replacements are reviewed in Refs 4-6. It seems that developments of these asbestos-free products are at present limited by the price level acceptable by the market and by the necessity of having a sound understanding or knowledge of those materials including their long-term behaviour.

One asbestos-free composite using layers of continuous fibrillated polypropylene film, opened up to form networks, to make thin reinforced cement sheets was developed at the University of Surrey during the late 1970's. In the fibrillated opened netlike form, the polypropylene fibre can provide an effective mechanical bond between the fibre and the cement matrix by a mechanism of so-called "misfit" sufficient to transfer stress between them and produce closely spaced multiple cracking and thus great energy absorption capacity. Another feature of the material is its high ultimate strength which results if the critical fibre volume in the composite is well exceeded (7-10). The concept of using fibrillated polypropylene net as a reinforcement in thin cement sheet had led to the patent specification in 1981 under the tradename NETCEM (11). In 1984, Montedison S.P.A., through its subsidiary Moplefan S.P.A. in Milan, announced the pilot production of the fibrillated polypropylene net given the name RETIFLEX (12). Fibronit S.R.L. subsequently started their trial production of the corrugated NETCEM sheets based on RETIFLEX in their own specially developed machine in Avenza (13). The sheets were put on Italian market in 1987 under the tradename RETICEM.

Although the advantage of fibrillated polypropylene film reinforced cement, in terms of high strength and high reliability, is now well established, it is recognized that this high strength is achieved at the expense of large extensibilities (i.e. 7% strain) of the composite because of the high strain
to failure and low elastic modulus of the fibres (around 4-8 GPa). In practice, such a high strain is rarely reached and therefore it is very unusual for the high strength to be able to be sufficiently utilized. Other unfavourable characteristics of the composite which result from the low elastic modulus polymer fibres (compared with glass fibres) are: a) a lower first cracking strength, b) a lower load bearing capacity following the first crack, and c) less effective control of crack opening. In certain applications, to suppress or control crack opening is of technological importance. For example, cracks in a RETICEM roofing sheet may permit access of water through the material. This decreases the service quality of the composite. It was towards the solution of the problems mentioned above that this study is directed.

The tensile behaviour of fibre reinforced cement composites can differ considerably between the two cases where the elastic modulus of the reinforcement is higher than that of the matrix and where it is lower. Glass and carbon fibre reinforced cements can be sited as representative examples of the former, while the polypropylene fibre reinforced cement may be given as a typical example of the latter. Compared to polypropylene cement, glass fibre reinforced cement has the advantages in sustaining higher applied load at a much smaller strain (i.e. at 1% strain), following the matrix cracking, which is related to the high modulus and high tensile strength of glass accompanied by a considerably lower strain to failure. The limited strain capacity of glass fibres, however, results in a lower composite toughness. Doubts remain about the long term strength and durability of glass fibres in alkali cement environments, especially when the composite is used outdoors (14).

By using the concept of "hybridization" with two or more types of reinforcements incorporated in a common resin matrix, it seems possible to have greater control of specific properties and achieve a more favourable balance between the advantages and disadvantages inherent in their parents composite materials (15-17). This has been well evidenced in the carbon-
glass-epoxy composites. The incorporation of two or more fibres within a single matrix is known as hybridization and the resulting material generally referred to as a "hybrid" or "hybrid composite". In the carbon-glass-epoxy composite, where the failure strain of the epoxy matrix is much greater than its reinforcements, the low elongation (LE) carbon fibre will fail first and the failure strain, and hence the strength of the carbon fibre appears to be greater in a hybrid than in an all carbon fibre composite structure. Catastrophic failure of the hybrid will not occur on the failure of the LE fibres since the load will be transferred onto the high elongation (HE) glass fibres and the HE fibre will continue to carry load until its failure strain is reached. As a result, toughness of the LE fibre based composite is improved.

In recent years, there has been a lively interest in hybrid composites based on a brittle cement matrix. A number of investigators (18-23) have used steel, glass and carbon to achieve strength combined with chopped polypropylene or chopped polyethylene to improve toughness. The fibre types, volume fractions used and the properties investigated are collected in Table 1.1. In these hybrid composites, the high modulus and low elongation fibres tend to increase strength with only modest improvements in toughness, while the low modulus fibres lead to a considerably higher toughness with hardly any improvement in strength.

Another category of hybrid composites which have been actively developed in recent years are non-asbestos roofing and cladding sheets and sewerage and drainage pipes in conjunction with the continuous Hatschek and Magnani processes. Here it is difficult to use one type of non-asbestos fibre which would provide simultaneously a reinforcing effect and filtering and cement particle retention properties and thus a blend of two or three types of fibres are used. In such hybrid composites, fibres of high modulus polyvinylalcohol (PVA) and polyacrylonitrile (PAN) or alkali-resistant (AR) glass with a fibre length of 4-6 mm are usually used as reinforcing and toughening elements. The pulps of polyethylene or polyacrylonitrile are
used mainly for processing purpose while the cellulose (normally soft wood) acts as both a processing aid and a weaker reinforcing fibre.

In the present work, the main fibre-cement hybrid composites to be studied are those containing fibrillated polypropylene networks and continuous or chopped AR-glass fibres. The concept of using a combination of polypropylene networks and AR-glass fibres in a cement matrix has led to the commercial production of the composite by Fibronit S.R.L. since 1989 under the tradename RETIVER (24). Other three types of hybrid systems, namely, AR-glass and continuous PVA yarns, PVA yarns and polypropylene networks, and polypropylene networks and continuous carbon rovings, are also studied in an attempt to establish the optimum choice of fibres in cement based hybrid composites.

The hybrid composites studied in this thesis are different from the above described fibre-cement hybrids (Table 1.1) in three aspects:

1) Physical properties of fibres: In this study, the toughening fibre of polypropylene networks or PVA yarns were continuous rather than chopped monofilaments; and the stiffening fibres (glass or carbon) were either chopped or continuous.

2) Functions of fibres: Since both the two types of fibres in a hybrid were reinforcing fibres with different elastic moduli, tensile strengths and strains to failure, they were expected to contribute their major reinforcing effects at different composite strain levels and thus the load-strain behaviour would not be the same as those in Table 1.1. The ultimate strength of the hybrids would be determined by the HE polypropylene and PVA fibres rather by the LE glass and carbon fibres.

3) Load-sharing: Since both types of fibres were reinforcing fibres, the additional load after the matrix cracking would be shared by the two fibres rather than only by one fibre as in individual fibre cements or by
Table 1.1 Hybrid fibre cements and concretes investigated in Refs 18-23

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<th>Fibre volume fractions</th>
<th>Properties measured</th>
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**Walton et al. (18)**
- 2.0-5.0% AR-glass strand (32mm)
- 0.5-2.3% PP filament (20-51mm) (by wt.%)
- 4.0% AR-glass strand (32mm)
- 0.2-0.9% nylon fibre (12-51mm) (by wt.%)  
- 0.7% carbon mat
- 1.0% PP filament (50mm) (by wt.%)  
- 1.0% asbestos fibre
- 1.0% nylon fibre (12mm) (by wt.%)  

**Kobayashi et al. (19)**
- 1.0% steel fibre (30mm)
- 1.0-3.0% PE filament (40mm) (by wt.%)
- 1.0% steel fibre (30mm)
- 1.0-2.0% PP filament (20-60mm) (by vol.%)
- 2.41% steel fibre (16mm)
- 0.04% PP filament (57mm) (by wt.%)
- 1.0% steel fibre (25mm)
- 1.0-3.0% PE filament (40mm) (by vol.%)
- Continuous steel bar
- 0.1-0.5% PP filament (19mm) (by vol.%)

the stiffer fibre as in those hybrids in Table 1.1. The ACK theories cannot therefore be directly applied to the fibre-cement hybrid composites studied in this thesis.

This work is an initial investigation of the characteristics of hybrid reinforcement intending to provide an understanding of the load-strain behaviour of the new type of fibre-cement hybrids. In addition, to look for the optimum choices of fibres and their relative proportions in the hybrid composite.

Tests were carried out basically in uniaxial tension using small scale coupons, although limited experimental results of flat and corrugated specimens under flexural loading are also presented. It is believed that a direct tensile test is more fundamental in characterizing the properties of fibre reinforced cement, in comparison with the flexural tests.

A theoretical model based on the conventional ACK model was proposed to predict the tensile stress-strain behaviour of the hybrid composites. The applicability of this model was tested with the experimental data obtained in this study.

It is hoped that the work presented in this thesis will contribute to an understanding of these types of hybrids and lead to further improvements of the commercial products.
Fibre reinforcement is a well established method of improving the mechanical properties of a variety of matrices. In the case of hardened cement pastes, their brittle characteristics in terms of low tensile strength and poor extensibility can be improved by incorporating fibres. It is necessary to understand the way in which the fibres work, and the interaction between the fibres and cement matrix when loads are applied, which will allow sensible design of the composite. It is because of these considerations that so much effort has been applied in the past 20 years to develop theories to describe the properties of the composite materials and their relationship to the properties of the constituents, the fibre volume fractions and the way the fibres are arranged.

Romualdi and Baston (1) were the first to use a fracture mechanics approach to analyse the restraining effect on crack propagation of closely spaced fibre reinforcement of concrete. They predicted that the cracking stress should be inversely proportional to the square root of the fibre spacing as long as the fibres remain bonded to the cement. Although their theories have been disputed by some other workers (2-3), their work produced a great stimulus in the development of fibre reinforced cement both in the academic and in the practical fields.

The theory of fibre reinforced cementitious composites was not well established until 1971 when Aveston, Cooper, Kelly and co-workers initiated a series of papers giving detailed analytical treatment of the tensile behaviour of brittle matrix composites which fail by multiple cracking (4-6), and set the understanding of the mechanism of fibre reinforcement on a
sound footing. The analytical expressions proposed by Aveston et al could be related to the properties of the constituent fibres and matrix, and could thus predict the tensile behaviour of such composites. Henceforth it will be referred to as the ACK theory.

It is clear that the ACK theory was developed for the single type of fibre reinforcement only, and could not be directly used for those composites containing more than one type of fibre. Nevertheless, the ACK theory and its subsequent modifications are reviewed in some detail, since it is fundamental to further theoretical developments for their application in the hybrid composites discussed in this thesis.

2.2 Theoretical assumptions and idealised tensile stress strain curve

The ACK theory is founded on the following assumptions:

(a) The composite consists of continuous aligned fibres;
(b) The matrix has a well defined single-valued breaking strength;
(c) The fibres are linear elastic;
(d) The stress transfer between fibres and matrix is purely frictional and linear with distance from the crack.

There are some obvious limitations about these assumptions.

The "continuous aligned" assumption appears to be justified for the cement reinforced with continuous long fibres such as glass or PVA rovings, but is inappropriate to the majority of fibre cements in which fibres are short and randomly aligned. For the fibrillated polypropylene networks, they are unlikely to be as ideal as this assumption due to their complex network pattern. The cement matrix is known to be highly variable in terms of strength and elastic modulus, depending on the size and distribution of voids in it. The assumption of linear elastic fibres for steel and glass may
be adequate but polymer fibres will deform inelastically. In addition, the stress transfer between fibres and matrix is almost certainly generated by more complex interaction rather than the purely frictional slip pattern.

The ACK theory yields an idealised tensile stress strain curve as shown in Figure 2.1. According to the ACK model, the overall mechanical behaviour of the composite reinforced with one type of fibre can be described in terms of the three stress-strain stages:

1. Elastic range, up to the point of first crack; the matrix and the fibres are both supposedly in their linear elastic range.
2. Multiple cracking range, in which the composite stress remains constant and the strain has exceeded the ultimate strain of the matrix.
3. Post-multiple cracking stage, in which all the load is carried by the fibres and a relative straight-up curve results.

In the following sections it is intended to examine all the ranges of the curve based on the ACK theory, and some of the modifications of it in relation to practical composites.

2.3 Composite behaviour below matrix cracking stress

2.3.1 Law of Mixtures approach

Before the matrix cracks, the composite stiffness may be assessed by the Law of Mixtures which relates the mechanical properties of the individual phases and their volume fractions. The Law of Mixtures is based on the following assumptions: there is no slip between the matrix and fibres; the fibres, the matrix and the composite are all strained equally, although in practice none of the assumptions is likely to be true. Thus, the elastic modulus, $E_c$, of the uncracked composite with aligned continuous fibres can be predicted by:
Figure 2.1 Idealized tensile stress-strain curve for a brittle matrix composite predicted by ACK model
where, $E$ - elastic modulus
$V$ - volume fraction
subscripts c, m and f refer to composite, matrix and fibre, respectively.

Substituting values typical for glass fibre reinforced cements into equation 2.1 ($V_f$ of 4%, $E_f$ of 70 GPa) readily demonstrates that the improvement in the elastic modulus over that of the matrix ($E_m = 20-30$ GPa) cannot exceed 5 to 10%, which from a practical point of view is insignificant. In FRC composites, the major role played by the fibres before cracking is to enhance the tensile strain capacity of the matrix and, after cracking, to toughen and strengthen the brittle matrix composite.

2.3.2 Enhancement of matrix failure strain

In the early 1960s Romualdi and Batson (1) suggested that the cracking strain of concrete could be increased by the addition of fine wires. Their numerical calculations led to the conclusion that the tensile cracking strain of the matrix would be inversely proportional to the square root of the fibre spacing, at a given fibre volume fraction. The theory was based on a fracture mechanics approach assuming elastic continuity across the fibre-cement interface and was apparently verified by the results of flexural tests.

The Romualdi theory has a number of limitations. For instance, some experimental results based on direct tensile tests showed considerable deviation from the theoretical predictions and the large scale improvement of the matrix cracking strain expected by the Romualdi theory did not occur (2-3). It would appear that Romualdi et al did not properly account for in their theory the effects of fibre geometry and fibre-matrix bond. The fully
elastic bonding assumption is simply not realistic for practical fibre cement, especially for short fibre cement.

An alternative approach to calculate the matrix cracking strain was proposed by Aveston et al (4-6), using a fracture mechanics approach in the form of energy balance. Their theory led to the following expression for the cracking strain in continuous aligned fibre composites:

\[
e_{mu} = \left[ \frac{12r\gamma_m E_m V_f}{E_f E_m^2 \gamma_m} \right]^{\frac{1}{3}} \quad (2.2)
\]

where \(r\) is the frictional shear strength and \(\gamma_m\) is the surface energy to fracture of matrix.

Equation 2.2 predicts that the matrix failure strain will be increased when the fibre diameter is decreased for a constant fibre volume fraction. Subsequent published results appear to confirm the ACK theory for steel and carbon fibre cements (5).

An anomaly in the ACK theory is that it predicts zero matrix failure strain at zero fibre volume. Another evident shortfall is that the ACK energy balance only considers the energy change of the system before and after cracking; no consideration is given to the mechanics of crack growth. Thus, as pointed by Hannant et al (7) that the ACK equation always gives a lower limit to the first cracking strain. This strain must be exceeded for cracking to occur.

Recently, strain-release models (7-8) have been developed to account for the enhanced strain of brittle matrices containing flaws and reinforced with fibres. These approaches, assuming frictional shear bond as the ACK model, are capable of predicting the stabilizing effects of fibres on matrix crack growth by calculating the rate of strain energy release (by matrix crack propagation) and the rate of energy absorption (by sliding friction) with
increasing length of a matrix crack. Unstable growth of the matrix crack will occur when the rate of release of strain energy with increasing crack length is greater than the rate of energy absorption. These models predict a continuous increasing curve relating matrix cracking strain to fibre volume fraction and the unreinforced matrix is therefore the end point. The difference in the approaches of Hannant et al (7) and the others (8) lies in the assumptions made about the size of the relaxation zone around a flaw in the unreinforced matrix.

The models (4, 7-8) reviewed above for predicting the matrix failure strain were derived for composites with aligned continuous fibres. For composites with discontinuous short fibres, an efficiency factor, η, is usually applied to the strain model (5, 7). Recently, Tjiptobroto et al (9), using similar energy components with that of Aveston et al (4), has derived a model which predicts the elastic strain at first cracking of composites with high volume fraction of randomly distributed short fibres. One of the major differences between the Tjiptobroto model and the ACK model appears in that the former assumes no debonding at the onset of matrix cracking and thus the energy term associated with fibre-matrix interfacial friction is not included in his model, as shown in Figure 2.2. However, in fact, Tjiptobroto used the ACK assumption of constant frictional stress and linear stress transfer during the derivation of the strain model and this appears self-contradictory with his no-debonding assumption. Thus, an error has resulted in his model against his basic assumption, since the strain distribution of a bridging fibre around the crack with elastic or nonlinear stress transfer shall differ. Tjiptobroto resolves this contradiction by suggesting to apply a ratio A/B to the derived model, where A and B are the solutions to the integral for the increased fibre strain energy, $\Delta U_f$, for linear and nonlinear strain distribution, respectively. Tjiptobroto's model agrees well with his experimental results from flexural tests using fibre reinforced densified small particles (DSP) containing high volume fraction of fine and short steel fibres but appears to predict lower matrix failure strain for the ordinary fibre reinforced cements (9). Unfortunately,
Figure 2.2 Energy terms considered in the strain models of ACK (for continuous fibre composites) and Tjiptobroto (for discontinuous fibre composites) after Tjiptobroto (9)
Tjiptobroto did not examine his theory using direct tensile tests.

Once the matrix failure strain $\epsilon_{mu}$ is known, the composite stress, $\sigma_c$, at this strain may be given by:

$$\sigma_c = (E_m V_m + E_f V_f) \epsilon_{mu}$$

(2.3)

Since it is assumed that there is no slip between the matrix and the fibres, the strain values of the fibres, the matrix and the composite are all the same before the matrix cracks.

Equation 2.3 is not entirely compatible with equation 2.2. If $\epsilon_{mu}$ is enhanced according to equation 2.2 elastic continuity between fibre and matrix may not be maintained. Thus some debonding may occur along the matrix-fibre interface and some of the microcracks in the matrix imposed from imperfect fabrication and drying shrinkage may already start to propagate.

2.4 Multiple cracking in brittle matrix composite

2.4.1 Conditions for multiple cracking

In a composite, if the elongation to failure of the fibre, $\epsilon_{fu}$ is less than that of the matrix, $\epsilon_{mu}$, as in the case of fibre reinforced plastics (FRP), for multiple fracture of the fibre to occur, the load per unit area of composite at the failure strain of the fibres must be less than the strength of the matrix, $\sigma_{mu}$, times its volume fraction, i.e.

$$E_c \epsilon_{fu} \leq \sigma_{mu} V_m$$

(2.4)

which, for a unidirectional composite, means (10)

$$\sigma_{fu} V_f \leq \epsilon_{mu} E_m V_m - \epsilon_{fu} E_m V_m$$

(2.5)
where $\sigma_{fu}$ is the ultimate strength of fibres. When equation 2.5 is satisfied (i.e. when the fibre content, $V_f$, is sufficiently low), the fibres will successively fracture into a number of segments until the matrix attains its failure strain. In this case, the FRP composite will be weaker than the unreinforced matrix and so of not much practical value.

Our more interesting case is that the elongation to failure of the fibres is much greater than that of the matrix, so that as cracking occurs in the brittle matrix, the load is transferred to the fibres; if failure is to be prevented at this stage, the load bearing capacity of the fibres, $\sigma_{fu}V_f$, should be greater than the load on the composite at first crack:

$$\sigma_{fu}V_f \geq E_o e_{mu} \qquad (2.6)$$

that is,

$$\sigma_{fu}V_f \geq e_{mu} E_m V_m + e_{mu} E_f V_f \qquad (2.7)$$

When equation 2.7 is satisfied (i.e. when the fibre content, $V_f$, is sufficiently high), the first crack to occur in the composite will not lead to catastrophic failure, but redistribution of the load between the matrix and the fibres. Consequently, the matrix is fractured into a series of blocks, separated by cracks. The multiple cracking process is extremely important for the fibre reinforced cement composite, not only because it contributes to the composite a quasi-ductile behaviour but also because it controls the crack spacing and hence the crack width at this stage which has a considerable influence on the serviceability of the material.

Equation 2.6 may be used to evaluate the so-called critical fibre volume, $V_f(crit)$, which is defined as the minimum fibre volume needed to allow the multiple cracking of the matrix to happen (11),
When $V_f > V_{f(crit)}$, the fibres in the composite will contribute to both toughness and strength, the mode of fracture is characterised by multiple cracking of the matrix. In contrast, at $V_f < V_{f(crit)}$ the first crack of the matrix will lead to a catastrophic failure of the composite; the mode of failure will be by the propagation of a single crack, since there is an insufficient volume of fibres to support the load that was carried by the matrix as it cracked.

Critical fibre volumes calculated from equation 2.8 are 0.4% and 0.8% for glass and polypropylene networks reinforced cements (11). However, equation 2.8 was derived for continuous and aligned reinforcement. In practical composites, fibres, in particular the glass fibres, are frequently used in short, randomly-oriented forms. In the latter case, length and orientation factors have to be taken into account and thus the critical fibre volumes will be considerably greater. The fibre length and orientation efficiency will be considered in a subsequent section.

2.4.2 Multiple cracking in brittle matrix composite

If fibre strength and content are all sufficient for the left-hand terms in equation 2.6 or 2.7 to exceed the matrix cracking stress, then, when the matrix cracks, the load carried by the matrix per unit area of composite, $\sigma_{mu}V_m$, will be thrown onto the fibres bridging the crack and will then be transferred back into the matrix by the action of a constant frictional shear stress, $\tau$. If the matrix has a single-valued failure stress, then further matrix failure will occur, outside the transfer length $x'$ either side of the first crack, until the matrix is eventually broken down into a series of blocks of lengths between $x'$ and $2x'$ (4). By a simple force balance of the load, $\sigma_{mu}V_m$, needed to break unit area of matrix and the load
carried by \( N \) fibres across the same area after cracking, the transfer length or crack spacing, \( x' \), can be derived:

\[
\sigma_{mu} V_m = \frac{P_f}{V_f} N \cdot \tau \cdot x'
\]

and since \( N = \frac{V_f}{A_f} \)

\[
x' = \frac{V_m \cdot \sigma_{mu} A_f}{V_f \cdot \tau P_f} \quad (2.9)
\]

where \( P_f \) is fibre perimeter and \( A_f \) is fibre cross-section area. For circular fibres of radius \( r \), equation 2.9 can be written as:

\[
x' = \frac{V_m \cdot \sigma_{mu} r}{V_f \cdot 2 \tau} \quad (2.9a)
\]

For fibrillated polypropylene films the cross-section of the fibres approximates more closely to a thin rectangle (11), for which:

\[
\frac{A_f}{P_f} = \frac{bt}{2(b + t)}
\]

where \( b \) and \( t \) are the fibre breadth and thickness respectively. In this case:

\[
x' = \frac{V_m \cdot \sigma_{mu} bt}{V_f \cdot 2 \tau (b + t)} \quad (2.9b)
\]

The average crack spacing has been quoted by Aveston et al (4) as \((1.364 \pm 0.002)x'\), from work by Gale based on the analogous problem of minimum average spacing between cars of length \( x' \) parked at random in a given space. More recently Kimber and Keer (12) have shown that for composites having large lengths the theoretical value for the average gap between cracks is \( 1.337x' \). The use of this value in preference to Gale's has little effect upon the theoretical stress strain curve for the average crack spacing and
the value of $1.364x'$ is used in this thesis.

The strain distributions in fibre and matrix associated with one crack are shown in Figure 2.3a. In the ACK theory, the frictional shear stress is assumed to be uniformly along the fibre length, so that the additional load is then transferred linearly from the fibre to the matrix as can be seen in the illustration.

The additional stress on the bridging fibres due to cracking of the matrix varies between $\sigma_{mu}V_m/V_f$ at the crack and zero at distance $x'$ from the crack, so that the maximum additional strain, $\Delta \varepsilon_f$, in the fibres at the crack due to this additional stress will be

$$\Delta \varepsilon_f = \frac{\sigma_{mu}V_m}{E_fV_f} = \frac{E_aV_a}{E_fV_f} \varepsilon_{nu} = \alpha \varepsilon_{mu}$$

The average additional strain in the fibres which is equal to the extension per unit length of composite, $\Delta \varepsilon_c$, at constant stress $E_c\varepsilon_{mu}$ is therefore

$$\Delta \bar{\varepsilon}_f = \Delta \varepsilon_c = \frac{1}{2} \alpha \varepsilon_{mu} \quad (2.10)$$

where $\alpha = E_aV_a/E_fV_f$

Thus, the strains in the composite, $\varepsilon_{mc}$, at the end of multiple cracking is the average strain of the fibres at that time, i.e. (5)

$$\varepsilon_{mu}(1 + \frac{1}{2} \alpha) < \varepsilon_{mc} < \varepsilon_{mu}(1 + \frac{3}{4} \alpha) \quad (2.11)$$

The upper and lower limits correspond to the minimum crack spacing $x'$ and maximum spacing $2x'$, respectively.

Similarly, for cracks developing at the average crack spacing of $1.364x'$
(a) strain distributions associated with one crack

(b) $2x'$ crack spacing

(c) $x'$ crack spacing

(d) $1.364x'$ crack spacing

Figure 2.3 Strain distributions in fibre and matrix (after Keer, (13))
(Figure 2.3d) then the strain at the completion of multiple cracking is (5):

\[ e_{m\sigma} = e_{mu} (1 + 0.659\alpha) \]  \hspace{1cm} (2.12)

The crack width in the multiple cracking range, \( w \), can be calculated by multiplying the crack spacing by the strain at the end of multiple cracking by taking into account the matrix strain relaxation. For a crack spacing of \( 2x' \), the crack width is given by (11):

\[ w = 2x' \left( \frac{e_{mu}}{2} + \frac{\alpha e_{mu}}{2} \right) \]

or

\[ w = e_{mu} (1 + \alpha) x' \]  \hspace{1cm} (2.13)

2.5 Post multiple cracking behaviour in brittle matrix composite

When the distance between the cracks is too small to allow sufficient stress to be transferred from the fibre to the matrix to break it further the multiple cracking process is complete. Then according to the ACK theory, no more cracking can take place beyond the multiple cracking stage (although in practice this is not completely true) and the fibres alone take any further increase in stress and are stretched through the matrix. When the local strain in the fibres at a crack reaches the fibre failure strain the composite breaks.

Assuming stretching of continuous and aligned fibres, the elastic modulus of composite in the post multiple cracking zone is \( E_fV_f \), i.e. no contribution from the cracked matrix. The composite ultimate strength, \( \sigma_{cu} \), is solely the contribution of fibres:
The ultimate composite strain, \( \varepsilon_{cu} \), is not the failure strain of the fibres, but is reduced because of the load sharing along part of the fibre length. There is in effect a shifting of the stress strain curve of the fibre alone (of Figure 2.1) as a result of this load sharing, but the slope is unchanged. The composite strain to failure is the average strain of the fibres at failure and is given by (5):

\[
\left( \varepsilon_{fu} - \frac{\alpha \varepsilon_{mu}}{2} \right) < \varepsilon_{cu} < \left( \varepsilon_{fu} - \frac{\alpha \varepsilon_{mu}}{4} \right) \tag{2.15}
\]

The upper and lower limits correspond to the minimum spacing \( x' \) and maximum spacing \( 2x' \), respectively.

Similarly, the ultimate composite strain for the average crack spacing of 1.364\( x' \) is given by:

\[
\varepsilon_{cu} = \varepsilon_{fu} - 0.341 \alpha \varepsilon_{mu} \tag{2.16}
\]

It should be mentioned that the equations 2.14-2.16 are based on the assumption that failure by fibre fracture will precede failure by fibre pull-out. This is less likely to occur in short fibre reinforced composites.

2.6 Efficiency of fibre reinforcement

In many practical composites, fibres are not continuous and aligned in the direction of applied stress but are used in short, randomly-oriented forms. The contribution of such short, randomly-distributed fibres to the mechanical properties of the composite is smaller than that of long fibres oriented parallel to the load, i.e. the efficiency of the short and randomly-oriented fibres is less.
The efficiency of fibres is usually expressed in terms of an efficiency factor, which is a value between 0 to 1, being the ratio between the reinforcing effect of the short, randomly-oriented fibres, and that expected from aligned continuous fibres. The factors $\eta_L$ and $\eta_o$, for length and orientation efficiency, respectively, can be determined either empirically or on the basis of analytical calculations. The various efficiency factors proposed in the literature are briefly reviewed below.

2.6.1 Length efficiency factor

2.6.1.1 Uncracked composite

Kelly (14) has derived the following length efficiency factors:

\[
\begin{align*}
\text{for } L > L_c & \quad \eta_L = 1 - \frac{L_c}{2L} \\
\text{for } L < L_c & \quad \eta_L = \frac{L}{2L_c}
\end{align*}
\]  

(2.17)  

(2.18)

where $L_c$ is the critical fibre length defined as the minimum length of the fibre required to achieve the full strength capacity of the fibre. For a frictional shear stress transfer, $L_c$ is given by:

\[
L_c = \frac{\sigma_{fr}x}{\tau}
\]  

(2.19)

The efficiency factors of equations 2.17 and 2.18 are the ones generally used in the theory of polymer or metal matrix-based composite materials which are not cracked. Allen (15) and Aveston al et (5) have however used similar equations for the cement composite containing discontinuous and two-dimensionally aligned glass fibres.

Laws (16) has indicated that, for frictional stress transfer, the average stress in the fibre, $\bar{\sigma}_f$, and thus its efficiency, is a function of the composite strain, $\varepsilon_c(x)$.
where $L_{x}/2$ is the length required for a stress of $E_{f} \epsilon_{c}(x)$ to be built up in the fibre. At matrix failure

$$\eta_{L} = 1 - \frac{L_{c}}{2L} \epsilon_{mu}$$  \hspace{1cm} (2.21)

The matrix failure strain is generally of the order of a few hundred microstrain for cement composites. Thus the length efficiency factor is unlikely to be less than 0.98 and nearly unity according to equation 2.21. Equation 2.21 also indicates that the efficiency factor of $1-L_{c}/2L$ shown in equation 2.17 would be valid for uncracked fibre reinforced cement composites only if the ultimate strains of the fibre and the cement matrix were the same, which is obviously not the case. Thus, Laws approach (16) discussed above would appear to be the more correct method.

2.6.1.2 Cracked composite

Laws (16) has derived length factors for ultimate composite strength for two cases:

(i) static frictional bond ($\tau_{s}$) only;

(ii) combined static frictional bond ($\tau_{s}$) and dynamic frictional bond ($\tau_{d}$) which initiates after the static frictional bond fails.

For the former case, Laws derived the following length efficiency factors which are essentially the same as those of Krenchel (17), related to composite strength:

$$\text{for } L > 2L_{c} \hspace{1cm} \eta_{L} = 1 - \frac{L_{c}}{L}$$  \hspace{1cm} (2.22)

$$\text{for } L < 2L_{c} \hspace{1cm} \eta_{L} = \frac{L}{4L_{c}}$$  \hspace{1cm} (2.23)
In the derivation of equations 2.22 and 2.23, it was assumed that the portion of the fibres which slip and pull-out do not contribute to strength. These relations would change if one were to assume that some frictional resistance \( (\tau_d) \) between fibres and matrix would initiate:

\[
\begin{align*}
\text{for } L > 2L'_{c} & \quad \eta_L = 1 - \frac{L_c}{2L} (2 - \frac{\tau_d}{\tau_s}) \quad (2.24) \\
\text{for } L < 2L'_{c} & \quad \eta_D = \frac{L}{2L_c (2 - \frac{\tau_d}{\tau_s})} \quad (2.25)
\end{align*}
\]

where \( L'_{c} = (1/2)L_c (2 - \frac{\tau_d}{\tau_s}) \). The ratio of \( \tau_d/\tau_s \) is assumed in the range of 0 to 1. However, Laws estimated \( \tau_s \) to be 2/3 of \( \tau_d \), on the measurements of de Vekey and Majumdar (18). Equations 2.24 and 2.25 indicate the importance in increasing the dynamic frictional bond strength and the fibre length for the efficiency of fibres, as graphically demonstrated in Figure 2.4 (19).

2.6.2 Orientation factor

2.6.2.1 Uncracked composite

Krenchel (17) calculated the orientation factor, \( \eta_o \), for the condition where the composite is constrained so that it does not deform in the lateral direction. The value he derived for a composite containing two-dimensionally (2-D) random array of fibres is 3/8. Instead, Cox (20) analyzed the case for the unconstrained condition and derived the factor to be 1/3. For a three-dimensionally (3-D) random arrangement of fibres the corresponding factors are 1/5 and 1/6 respectively. These values indicate that the differences in the results obtained using the two assumptions are small.

2.6.2.2 Cracked composite

Aveston and Kelly (21) analysed the behaviour of a random, long fibre distribution across a crack, the fibres align themselves across the crack
Figure 2.4 Effect of the ratios $L/L_c$ and $\tau_d/\tau_s$ on stress-strain curves of aligned short fibre reinforced cements. $E_r = 70$GPa, $\sigma_{fb} = 1200$MPa, $V_f = 5\%$. (after Laws (19))
with the direction perpendicular to the crack face. The crack spacing was shown to be \( \pi/2 \) or 2 times the aligned case for a random 2-D or 3-D fibre arrangement, respectively; and the number of fibres crossing the crack are \( V_f/\pi r^2 \), \( 2/\pi(V_f/\pi r^2) \) and \( 1/2(V_f/\pi r^2) \) for aligned, random 2-D and 3-D fibre distributions, respectively. They suggested that, for the frictional stress transfer case, the strength of a random fibre brittle composite would be related to that of the aligned fibre composite by the same factors as the number of fibres crossing a plane, i.e. \( 2/\pi \) for 2-D fibre arrangement, and \( 1/2 \) for the 3-D case. These orientation factors are considerably higher than values of the order of \( 3/8 \), \( 1/3 \) or \( 1/5 \), \( 1/6 \) respectively normally quoted in the literature and based on a Krenchel/Cox type of elastic analysis (17, 20). This difference is likely caused by the consideration made by Aveston et al of the local fibre bending around the crack, although later Aveston et al (5) proposed that a value of 50% of the aligned strength was more appropriate, with the post-cracking stiffness a similar proportion of the aligned case, i.e. \( E_f V_f/2 \).

Laws (16) and Majumdar and Laws (22) concluded that the combined efficiency factors, \( \eta \), due to both length and orientation, cannot be simply calculated as the product of \( \eta_0 \eta_L \) since the fibre length required to transfer stress from the matrix, and therefore the contribution to \( \eta \), depends on orientation. However, the error in using the simple factor \( \eta_0 \eta_L \) is not significant in practice for fibre reinforced cement because the fibre volume fraction is small, i.e. only about 5%. It is therefore convenient to express the total efficiency factor as \( \eta_0 \eta_L \).

2.6.3 Empirical measurement on the efficiency of short, random 2-D aligned glass strand

In practice, the most widely used glass fibre reinforced cement composites contain fibres in the form of chopped strands with about 200 filaments in each strand of a length 4-50 mm. The configuration of the fibres in a composite can vary, depending on the size and shape of the bundle and its
treatment during manufacture and fabrication, as well as on the aging evolution. The fibres may expect to disperse completely in fibre composites made in industrial processes (i.e. in the Hatschek sheet process or in the Magnani pipe process). With the commonly used fabrication techniques such as spray suction or the type used in Retiver process, the bundles themselves are approximately randomly oriented in the plane of sheets with only the outer perimeter of the bundle in contact with the cement matrix at the early ages of the composites, although with aging some of the voidage between fibres within the bundle may be eventually filled by continuing cement hydration (23-26).

Figure 2.5 shows the most common configuration of integral strands in a young composite (27). Obviously, to be taken into account are not only the orientation and length effects but also the strand effects concerning the bonding area of fibres and the voidage within the strand, as long as the fibre efficiency in a composite is concerned. Strictly, the strand effect should also be considered in those composites reinforced with aligned continuous glass rovings. In the case, the overall efficiency of the aligned continuous glass rovings would be less than 1, at least at the early age of the composite. No detailed consideration has been given on this aspect as far as the author knows.

Oakley and Proctor (28) assessed the strand efficiency factors using an empirical approach. The samples they used were cut from spray suction dewatered sheets with 4-5% by volume glass strand content. The glass was in the form of 38 mm long chopped strands of Cem-FIL.

Combining equation 2.2 in the form of matrix cracking stress with equation 2.9a, allowing $E_c = E_m$ and $\bar{x} = 1.364 x'$, the following equation results (28):
Figure 2.5  Schematic description of the internal structure of a glass fibre cement composite, presenting an idealized unit cell (after Nair (27))
where $K_2$ is the crack suppression efficiency factor concerning the orientation and strand geometry effects, and $\bar{x}$ is the average crack spacing. Equation 2.26 was used with the empirically measured values of $\sigma_{mc}$ and $\bar{x}$ to estimate $K_2$ (28). Obtained results shown that the efficiency factor, $K_2$, strongly depends on the fibre orientation. This implies an agreement with Laws (16) who predicts a unity length efficiency factor for uncracked composite (equation 2.21). The values of $K_2$ they obtained ranged from 0.3 to 1.0 for the composite with sand-cement ratio up to 1.5 tested in the longitudinal direction. There was a marked loss of fibre effectiveness at the high sand levels. In their calculation a value of 4 J/m² for $\gamma_m$ and of 70 GPa for glass modulus were used.

Once multiple cracking is complete all the load is carried by the fibres. The post-multiple cracking stiffness, $E_z$, and the ultimate strength of the composite, $\sigma_{cu}$, are given by (28):

$$E_z = K_3 E_f V_f$$  \hspace{1cm} (2.27)

and

$$\sigma_{cu} = K_4 \sigma_{fu} V_f$$ \hspace{1cm} (2.28)

where $K_3$ and $K_4$ are the empirical post-stiffness efficiency factor and the strength efficiency factor, respectively, introduced by Oakley and Proctor (28) to describe the overall effectiveness of the random short fibres in strand form. The post-multiple cracking stiffness and strength efficiency factors were found to be similar and have the values in the range of 0.2-0.3, determined using the experimental values of the post-multiple cracking stiffness, composite strength and fibre stiffness. These values may be compared with the experimental estimate of 0.26 for strength efficiency by
Allen (29) in a glass reinforced cement and 0.21 for strength efficiency by Aveston et al (5) in a carbon fibre reinforced material.

2.6.4 Effects of fibre efficiency on predicted stress-strain curves based on ACK model

Subsequently, Proctor (30) extended the empirical efficiency factor approach to other aspects of the composite stress strain curves predicted by the ACK theory for aligned continuous fibre composites. Taking an efficiency factor $K_o$ for orientation, and $K_s$ for strand effects, the strain at the end of multiple cracking, $\varepsilon_{mc}$, according to Proctor (30) is given by:

$$\varepsilon_{mu}(1 + \frac{\alpha}{2K_oK_s}) \leq \varepsilon_{mc} \leq \varepsilon_{mu}(1 + \frac{3\alpha}{4K_oK_s})$$

for crack spacing $2x'$ and $x'$, respectively. Correspondingly, The composite failure strain, $\varepsilon_{cu}$, and strength, $\sigma_{cu}$, are given by:

$$K_s\varepsilon_{fu} - \left( \frac{\alpha\varepsilon_{mu}}{2K_oK_s} \right) \leq \varepsilon_{cu} \leq K_s\varepsilon_{fu} - \left( \frac{\alpha\varepsilon_{mu}}{4K_oK_s} \right)$$

$$\sigma_{cu} = K_oK_s\sigma_{fu}V_f$$

In equation 2.30, the $\varepsilon_{fu}$ term is only modified by the $K_s$ factor, and Proctor argues that fibres lying at an angle to the load are strained less than those aligned with the load and the latter will be the first to fail, thus the $K_o$ factor does not need to be applied to the $\varepsilon_{fu}$ term. The predicted stress strain curves using equations 2.29 to 2.31 are shown in Figure 2.6. It is notable that the modified equations by Proctor lead to a stiffer post-multiple cracking curve (dashed lines in Figure 2.6) when shorter fibres are used, which appears less reasonable. The reason for this may be due to that Proctor used the same length factor ($K_s$) at the end of multiple cracking and at fibre failure and this efficiency factor would be smaller at the fibre
Figure 2.6 Typical stress-strain curves for glass fibre cement composites calculated from equations 2.29 to 2.31. $E_r = 70\text{GPa}$, $E_m = 20\text{GPa}$, $\sigma_{fu} = 1200\text{MPa}$, $V_f = 4\%$. (after Proctor (30))
failure.

Laws (19) used an analytical approach to account for the effect of the fibre efficiency on the stress strain curve predicted by the ACK model. Assuming the stress is transferred to the matrix linearly from the crack and the crack spacing is $2x'$, the average strain in the fibres at the end of multiple cracking derived by Laws is:

$$
\epsilon_{mc} = \frac{\epsilon_{mu}}{2} (1 + \frac{K_{mu}}{K_{mc}}) + \frac{\alpha \epsilon_{mu}}{2 \eta_o K_{mc}} \tag{2.32}
$$

where $\eta$ is the orientation factor, and $K_{mu}$ and $K_{mc}$ are the corresponding length efficiency factors (to strains). Approximated by Laws, $K_{mc} \approx K_{mu} = 1$, equation 2.32 becomes:

$$
\epsilon_{mc} = \epsilon_{mu} (1 + \frac{\alpha}{2\eta_o}) \tag{2.33}
$$

The composite failure strain, $\epsilon_{cu}$, and strength, $\sigma_{cu}$, are written as:

$$
\epsilon_{cu} = \frac{\epsilon_{cu}}{2} (1 + \frac{\eta_L}{K_{cu}}) - \frac{\alpha \epsilon_{mu}}{2 \eta_o K_{cu}} \tag{2.34}
$$

$$
\sigma_{cu} = \eta_o \eta_L \sigma_{fu} V_f \tag{2.35}
$$

where $K_{cu}$ is the corresponding length factor and equal to $\eta_L = (1 - c l_c / L)$ at the ultimate strain, thus

$$
\epsilon_{cu} = \epsilon_{fu} - \frac{\alpha \epsilon_{mu}}{2 \eta_o \eta_L} \tag{2.36}
$$

The predicted stress strain curves for composite with random 2-D arrangement of fibres using equations 2.33 to 2.36 are shown in Figure 2.7.
In comparison, Laws' equations with that of Proctor, both treatments indicate that for shorter fibre lengths, the multiple cracking strains increase but the composite strength and strain to failure decrease. However, Laws' equations yield lower multiple cracking strain and higher ultimate strain than those of Proctor's, as well as non-linear post-multiple cracking curves. The main reason for the differences is that the length efficiency factor (K) used by Laws is strain dependent, which appears to be theoretically more correct, since the efficiency of reinforcement of short fibres depends on the stress.

2.7 Fibre-cement bond and stress transfer

2.7.1 Elastic stress transfer, combined elastic and frictional stress transfer

ACK theory (4-6) and the later modifications by Proctor and Laws (30, 19) outlined earlier are all based on the assumption so-called pure frictional shear stress transfer at the fibre-cement interface. Aveston and Kelly (21) extended the ACK model by considering an alternative of pure elastic stress transfer mechanism. Under this assumption, the matrix remains bonded to the fibre after it has cracked, except for the deformation at the crack surface. In this case, the additional stress on the fibre at the crack is assumed to decrease exponentially from the crack face (Figure 2.8). In order for a new crack to form, however, subject to the condition of a small increase in the additional load. The bonded case, therefore, leads to a smoothly rising stress strain curve in the multiple cracking region without the additional assumption of a unique matrix strength as being made in the frictional bond case.

Aveston and Kelly (21) have demonstrated that the shape of the stress strain curve is not very sensitive to the assumption regarding the nature of bond (Figure 2.9). Assuming the interfacial elastic bond strength to be equal to the matrix tensile strength, Aveston and Kelly indicate that the elastic
Figure 2.7 Predicted stress-strain curves for composites with 2-D arrangement of fibres. Curve A is for continuous fibres, Curve B, C and D refer to decreasing fibre lengths. $E_f = 70\text{GPa}$, $E_m = 20\text{GPa}$, $\sigma_{fu} = 1200\text{MPa}$ and $V_f = 4\%$. (after Laws (19))
Figure 2.8 Elastic stress transfer: strain distribution in fibre and in matrix (after Keer (13))

Figure 2.9 Idealized stress-strain curves assuming elastic or frictional stress transfer (after Aveston and Kelly (21))
bonding can be maintained during the entire multiple cracking process only if the fibre volume fractions exceed 30%, 38% and 50% in cement-based composites reinforced with carbon, steel and glass fibres, respectively. Obviously, these values are outside the range of practically interesting fibre reinforced cements for construction purposes. Thus, in practice, pure elastic bonding is highly unlikely, and the multiple cracking model should be based on combined elastic and frictional shear mechanism. However, additional analysis (21) has indicated that the crack spacing in typical FRC composites, calculated using the partial bonding assumption, will not differ by more than 15% from the value determined by the simplified model, which assumes that frictional shear is the only stress transfer mechanism. Thus, it appears that the ACK theory assuming friction shear stress transfer adequately describes the behaviour of fibre reinforced cement, as long as the elastic shear bond strength is not much greater than the matrix strength (21). In a later analysis, a similar conclusion was reached by Laws (31).

2.7.2 Fibre-matrix bond strength

There are two alternative approaches for the determination of fibre-cement interfacial bond strength. One is the pull-out tests, in which either a single fibre or an array of fibres (32-37) or, either a single strand (or tape, in case of polypropylene) consisting of a number of filaments or an array of strands or tapes (16, 28, 38-41) is pull out of a matrix. Another method is based on equation 2.9, relating the interfacial bond strength to the empirically measured crack spacing (28, 38, 42-46).

The information obtained from the pull-out tests, however, doesn't always represent the condition in the actual composites (43). The main criticism originates from the fact that fibre stress in a pull-out test is dependent on the embedded fibre length. For example, the average bond strength determined from the pull-out test of a fibre of a specific length cannot be used to predict the load that can be supported in a similar system with a fibre of different length, just by multiplying by the fibre length ratio.
Nevertheless, the pull-out test is a useful method for the comparison of the bond properties among types of fibre cement composites.

While in a pull-out test with a single filament it is possible to obtain a reliable estimate of the fibre perimeter in contact with the cement matrix, situation becomes more complicated when a multifilament strand or a polypropylene film which has an irregular cross-section is involved, since the fibre surface area or the fibre surface in contact with the matrix is not well defined. Take a glass strand in cement as example, the external filaments in the strand are bonded to a greater extent than the inner ones (Figure 2.5). Thus there is a difficulty in interpreting the results of pull-out tests of strands in terms of bond stress, even if the bond of a single filament surrounded by the matrix is known. To avoid the determination of perimeters or surface areas of the fibres, it was suggested (36, 40) that the value of the shear flow, expressed in the shear force per unit length of the fibre embedment, should be used.

Oakley and Proctor (28) estimated the perimeter of glass strand in real contact with the matrix by a microscopy technique and reported a bond strength value of 1.1 MPa. The authors (28) also estimated the bond strength on the basis of the crack spacing (equation 2.9) in 2-D random glass strands reinforced cement sheets and the values they deduced were of the order of 1 MPa or less, which is in agreement with the values reported later by Laws et al (41) who used the shear flow approach in their pull-out tests of the glass strands. On the other hand, de Vekey and Majumdar (18) reported bond strength values of ~ 10 MPa using glass rods of ~ 1 mm diameter but it appears that such value is less likely to be achieved in glass fibre reinforced cements.

It is extremely difficult to precisely determine the bond strength between cement and fibrillated polypropylene film. This is because of the complex nature of the hairy irregular surface shape and the multiple splits in the film. Also, it is generally agreed that the bonding mechanism between the
fibrillated film and the cement matrix is a mechanical one rather than an adhesive one. Attempts have been made to use a technique of Krypton gas absorption (38, 43-44, 46) or a method of image analysis (45) for measuring the specific surface area of the fibrillated polypropylene film, and the average bond strength were calculated from the final average crack spacing $x$ in the composite by using the rewritten equation 2.9:

$$
\tau = 1.364 \frac{V_F}{V_e} \frac{\sigma_{mu}}{X} \frac{A_F}{P_F} \quad (2.37)
$$

Based on this approach, Hannant (43-44) reported a value of the bond strength about 0.4 MPa of a composite containing about 6% by volume of the fibrillated polypropylene networks. Similar values to Hannant's of the bond strength were also reported by Hughes (38, 46). Ohno (45) however obtained much higher values of the bond strength of ~1.5 MPa (~6% fibre volume). The differences in the bond strength are considered due to the different methods used by the authors for the determination of the specific surface area of the fibrillated polypropylene films. It would appear that the calculated bond strength values by Hannant and Hughes represents the lower bound of the actual bond strength, since the gas absorption method measures the whole specific surface area of the film including the microvoids which have been shown to not be filled by any cement matrix - at least for those whose width are less than 2 $\mu$m (45).

In practice, it is doubtful whether the bond strength between the polypropylene fibres and the cement matrix is the main function of their contact area (equation 2.37). Because of their high Poisson's ratio, low elastic modulus and hydrophobic nature, polypropylene fibres tend to contract and separate from the matrix, thus eliminating the normal stresses which are required to generate frictional shear resistance. Consequently, multiple cracking might not occur (47-49). Hannant's ten year results (44) show that the bond strength between the fibrillated polypropylene networks and the cement matrix is constant, regardless the storage time and storage
conditions. Thus in a fibrillated polypropylene fibre composite, the contact area may be not the controlling parameter (38, 43) and effective bonding may be induced by mechanisms other than simple interfacial shear such as mechanical anchoring or misfit (38, 43 and 50), and interfacial slip within the film elements (38, 43). The latter point was further confirmed by the photographic evidence by Ohno (45). The contribution of the variable profile and incomplete fibrillation of the film are considered highly important for generating such effects (38, 43, 45 and 50).

2.8 Mechanism of fibre reinforcement in flexure

Fibre reinforced cement composites are, in practice, frequently subjected to flexural loading and are usually tested in bending. The bending capacity is commonly expressed in terms of the flexural strength, $\sigma_b$, calculated from the elastic bending theory assuming that the material at failure is behaving elastically with equal elastic moduli in tension and compression and with neutral axis at mid-depth. For an ideally elastic material, the flexural strength is equal to the ultimate tensile strength, $\sigma_u$. For fibre reinforced cement composites with pseudo-ductile behaviour under loading, the calculated flexural strength, $\sigma_b$, can be much greater than the tensile strength, $\sigma_u$, due to the neutral axis moving upward to the compressive surface (11). Thus, the obtained flexural strength is a "notional" flexural strength.

Edgington (51) and Allen (52) have shown that the neutral axis in steel or glass fibre reinforced cement composites may be as much as 4/5 of the beam height $D$ from its tensile surface, at the ultimate stage of flexural loading. Using a more conservative estimate being 3D/4, and the stress blocks in flexure (53, 11) in Figure 2.10, Hannant (53, 11) derived the notional flexural strength, $\sigma_b$, can be 2.44 times of the tensile strength, $\sigma_u$. This may suggest that the critical fibre volume for bending can be about half of the critical fibre volume for tension, and fibres which do not lead to an increase in tensile strength can lead to an increase in the load
carrying capacity of the composite in bending. Conversely, a composite may remain tough and ductile in flexure but could become brittle in tension (54). These implications of the theory have to be taken into consideration when a composite is designed. It first has to be determined whether the proposed use for the material will result in the applied stresses being more closely approximated by uniaxial tension or by flexure and then the appropriate fibre volume should be incorporated by accounting for not only the appropriate stress system but also the ageing effects.

The greater efficiency of the fibres in flexure is graphically demonstrated in Figure 2.11 by Laws (55) based on the tensile stress-strain curves predicted by the ACK model.

Aveston et al (5) have calculated the ratio between notional flexural strength, $\sigma_b$, and direct tensile strength, $\sigma_{cu}$, from the theoretical tensile stress-strain curve of Figure 2.1. Failure in flexure occurred when the strain, $\varepsilon_{cu}$, at the tensile face of the beam reached the ultimate strain derived from equation 2.15 of the ACK model. Aveston et al (5) were able to show that the notional flexural strength may be up to 2-3 times the ultimate tensile strength. The stress distribution is shown in Figure 2.12, and the analysis results, presented as a set of curves which can be used to determine the ratio of notional flexural strength to ultimate tensile strength if matrix and fibre properties, volume fraction, etc are known, is shown in Figure 2.13. There are difficulties in using the curves in Figure 2.13 to determine the $\sigma_b/\sigma_{cu}$ ratio, and hence to calculate the tensile strength from the results of a flexural test, if the data of the properties of fibre and matrix etc is not available. Furthermore, the analysis of Aveston et al (5) does not take into consideration a stress capacity beyond tensile failure which could have a significant effect on bending behaviour, in the case of fibre pull-out.

Laws discussed the relationship between the notional flexural strength and fibre volume fraction (55), which, as graphically shown in Figure 2.14, is
Figure 2.10  Stress distribution in flexure of an elastic-plastic material at the instant of failure (after Hannant (11))

Figure 2.11  Tensile stress-strain curves for continuous and aligned fibre reinforced cement predicted by the ACK theory (full lines) and the flexural response calculated from them.  
$E_f = 76\text{GPa}, \ E_m = 25\text{GPa}, \ \epsilon_{mu} = 0.04\%, \ \epsilon_{fu} = 2\%, \ V_f = 0.7\%$  
(curves A, A'), 1%(curves B, B'), 2%(curves C, C').  
(after Laws (55))
Figure 2.12 Stress distribution at failure in flexure of a composite with tensile behaviour modelled by the ACK theory.
(a) specimen geometry, (b) stress distribution
(after Aveston et al (5))

Figure 2.13 Ratio of bending strength to ultimate tensile strength versus $\alpha = \frac{E_m V_m}{E_f V_f}$ for composite with different $\epsilon_u/\epsilon_m$ ratios
(after Aveston et al (5))
Figure 2.14 Predicted flexural strength versus fibre volume curves for aligned and continuous fibre composites. Full lines relate to fibre volume fraction above the critical volume in tension, dashed lines refer to extrapolated response below this critical fibre volume. (after Laws (55))
not linear. However, over a practical fibre content range this non-linearity is small. If the best fit straight line is used over a limited fibre volume range there will be a positive intercept on the stress axis and the $\sigma_b/V_f$ relationship may appear to follow an apparent mixture rule. By using the flexural strength data for glass reinforced cement for the straight line of best fit, Laws (55) concluded that, even though the $\sigma_b/V_f$ relationship follows a simple mixture rule, this apparent mixture rule for the flexural strength is not a true mixture rule, since it results from the changing shape of the tensile stress-strain curve as the fibre volume fraction increases, and the positive intercept that results from fitting the best straight line to bending data is not the matrix strength in flexure. However, it is possible that this positive intercept could be a nearly approximation of the notional flexural strength and further experimental evidence is needed for this.
CHAPTER 3

THEORETICAL PROPOSALS FOR PREDICTING THE TENSILE BEHAVIOUR OF FIBRE-CEMENT HYBRID COMPOSITES

3.1 Introduction

Little theoretical background is available in the literature for the tensile behaviour of hybrid composites based on a cement matrix. Although the understanding of the interaction between fibres and matrices in single fibre reinforced composites is now reasonably satisfactory the interaction between the two fibres in a common matrix is not understood. One reason for the unavailability of theory for the tensile behaviour of the hybrid composites may be simply attributed to the fact that little attention has been paid to uniaxial tensile properties. Nearly all the authors (1-6) reported empirically their results on the properties in impact and flexure on which theoretical development has been similarly lacking. Another reason for the lack of theoretical development in the hybrids may be attributed to the inability to accurately describe the way and amount of load sharing by the two fibres.

If the potential applications of the hybrids are to be realized, the interaction between the two fibres and the failure mechanism under different load histories must be understood and characterized. In this chapter, an attempt is made to model the tensile behaviour of cement based hybrid composites in which one low-elongation (LE) fibre (i.e. glass, carbon) and one high-elongation (HE) fibre (i.e. polypropylene, PVA) are jointly used in the form of aligned continuous fibres. Of primary concern is the load sharing and the interaction between the two fibres which determines the amount of stresses or strains each fibre may carry in the hybrid at each stage of a loading process.
3.2 Theoretical prediction of tensile behaviour of fibre-cement hybrid composites

A fibre-cement hybrid composite consists of more than one type of reinforcing fibres in the same cement matrix. The conventional ACK theories (7-9) usually used to predict the tensile behaviour of individual fibre cements cannot be directly applied to the fibre-cement hybrid composite. A theoretical model based on the existing multiple cracking theories (7-9) is therefore developed to predict the tensile behaviour of such composites containing more than one type of fibre in a common matrix. The theory is developed mainly to predict the tensile behaviour of the glass-polypropylene hybrid composite studied in this thesis.

3.2.1 Single and multiple fracture of matrix

Basic assumptions in the derivation of the theory are similar to that in the ACK theory, that is, stress transfer between the fibres and the matrix is linear, and the frictional shear stress developed at the fibre-cement interface is uniformly distributed and is maintained until the complete failure of the fibre component. However, before the matrix cracks elastic bond is assumed to exist and hence the strain in each component in the hybrid is the same and external load is shared in proportion to the stiffness and volume fraction of the three components. Thus, the elastic modulus, $E_c$, and the first cracking stress, $\sigma_c$, of the hybrid can be calculated using the composite materials approach or the rule of mixtures:

$$E_c = E_m V_m + E_{f-L} V_{f-L} + E_{f-H} V_{f-H} \quad (3.1)$$

and

$$\sigma_c = \sigma_{mu} V_m + \sigma'_{f-L} V_{f-L} + \sigma'_{f-H} V_{f-H} \quad (3.2)$$

where $\sigma_{mu}$ is the tensile strength of the matrix, $\sigma'_{f-L}$ and $\sigma'_{f-H}$ are the stresses of the LE fibre and the HE fibre, respectively, at a strain of $\epsilon_{mu}$.
and the subscripts m, f-L and f-H refer to matrix, LE fibre and HE fibre, respectively.

Equation 3.2 can be written as

\[ \sigma_c = E_m \varepsilon_{mu} + E_{f-L} \varepsilon_{mu} + E_{f-H} \varepsilon_{mu} \]  \hspace{1cm} (3.3)

where \( \varepsilon_{mu} \) is the failure strain of the matrix. Then

\[ \sigma_c = E_c \varepsilon_{mu} \]  \hspace{1cm} (3.4)

After the initial cracking in the brittle matrix, the additional load, \( \sigma_{mu} \), is transferred to the fibres bridging the crack. If failure is to be prevented at this stage, the total load bearing capacity of the two fibres at the first crack should be greater than the composite cracking stress, that is,

\[ \sigma_{f-L}^{''} V_{f-L} + \sigma_{f-H}^{''} V_{f-H} \geq \sigma_{cu} \]  \hspace{1cm} (3.5)

where \( \sigma_{f-L}^{''} \) and \( \sigma_{f-H}^{''} \) are the stresses in the LE and HE fibres at the first crack face respectively, after the additional load (\( \sigma_{mu} \)) is thrown onto the two fibres. When equation 3.5 is satisfied, the first crack in the hybrid will not lead to catastrophic failure, but will result in redistribution of the load between the matrix and the two fibres. As the force balance must be satisfied on every cross-section of the composite, the sum of forces shared by the two fibres at the first crack face must be equal to the force at a cross section of the composite where there is no crack. Thus

\[ E_{f-L} V_{f-L} (\varepsilon_{mu} + \Delta \varepsilon_{f-L}) + E_{f-H} V_{f-H} (\varepsilon_{mu} + \Delta \varepsilon_{f-H}) = E_c \varepsilon_{mu} \] \hspace{1cm} (3.6)

where \( \Delta \varepsilon_{f-L} \) and \( \Delta \varepsilon_{f-H} \) are the additional strains in the LE fibre and HE fibre at the crack face, respectively.
As the additional load is thrown onto the fibres, the two fibres bridging the crack will tend to deform or slip differently because of the difference in bond strength between the two fibres. However, since the two fibres are aligned and continuous and are working in a common matrix, the average deformation occurring in the two fibres over the whole length of the specimen will be the same. The strains in the fibres at the crack face may differ but for the purpose of this analysis, which is a first approximation to the real solution, the problem has been simplified by assuming that fibre strains are equal. That is, $\Delta \varepsilon_{f-L} = \Delta \varepsilon_{f-H} = \Delta \varepsilon_f$. Combining equation 3.6 with equation 3.1, we obtain

$$\Delta \varepsilon_f = \frac{E_m V_m}{E_{f-L} V_{f-L} + E_{f-H} V_{f-H}} \cdot \varepsilon_{mu} \quad (3.7)$$

Let

$$\alpha' = \frac{E_m V_m}{E_{f-L} V_{f-L} + E_{f-H} V_{f-H}} \quad (3.8)$$

hence

$$\Delta \varepsilon_f = \alpha' \varepsilon_{mu} \quad (3.9)$$

The additional load shared by the two fibres at the crack must be transferred back to the matrix. According to the equation of ACK theory (7-8), the average crack spacing, $C$, in an individual fibre cement is given by

$$C = 1.364 \frac{V_m \sigma_{mu}}{V_f} \cdot \frac{1}{\tau} \frac{A_f}{P_f} \quad (3.10)$$

or

$$C = 1.364 \varepsilon_{mu} E_f \alpha' \cdot \frac{1}{\tau} \frac{A_f}{P_f} \quad (3.11)$$

where
According to equation 3.11, at a given fibre volume fraction in a given matrix, the average crack spacing \( C \) in various individual fibre reinforced cements should depend mainly on the average frictional bond stress \((\tau)\) and the ratio of perimeter to cross-section area or the so-called "specific surface area" \((P_f/A_f)\) of the fibre, but is independent of the fibre modulus. The fibre modulus may have some effect on \( \epsilon_{nu} \) but the extent is unlikely to be significant at the limited fibre volume fractions practically incorporated. Thus, theoretically speaking, the modulus of the fibres would have little effect on the crack spacing of the composite. The same applies to stress transfer length.

In hybrid composites with aligned continuous fibres, even if the two fibres have different stress transfer lengths which would result if the two fibres have different \( \tau \) or \( P_f/A_f \), one important point is that the additional load, \( \sigma_{nu}V_n \), shared by the two fibres at the crack must be wholly transferred back to the matrix before a second crack can form at a minimum distance \( x \) from the crack. Thus, as we have assumed that the two fibres have equal additional strain \( \Delta \epsilon_f \) at the crack face, the average additional strains in the LE fibre and in the HE fibre within length \( x \) would be the same because they are continuous throughout the specimen:

\[
\Delta \epsilon_{f-L} = \Delta \epsilon_{f-H} = \Delta \epsilon_f
\]  

(3.12)

Based on the assumptions of linear stress transfer and that the two fibres have equal additional strain \( \Delta \epsilon_f \) at the crack face, the strain distribution of each of the components in the hybrid around a crack for a maximum crack spacing 2\( x \) may be described as in Figure 3.1. In the figure \( \Delta \epsilon_f \) represents either \( \Delta \epsilon_{f-L} \) or \( \Delta \epsilon_{f-H} \) and varies between \( \alpha' \epsilon_{nu} \) (equation 3.9) at the crack face and zero at distance \( x \) from the crack. Therefore, the average additional
Figure 3.1 Strain distribution in fibres and in matrix
- 2x stress transfer length
strain $\Delta \varepsilon_f$ in the fibres, which equals the extension per unit length of the composite $\Delta \varepsilon_c$ at constant stress $E_c \varepsilon_{mu}$, is given by

$$\Delta \varepsilon_f = \Delta \varepsilon_c = \frac{1}{2} \alpha' \varepsilon_{mu} \quad (3.13)$$

where $\alpha'$ is determined by equation 3.8.

According to equation 3.13, the additional strain $\Delta \varepsilon_c$ in the hybrid specimen as a result of multiple cracking of the matrix depends mainly on moduli and volume fractions of the two fibres. Thus, for an equal volume glass-polypropylene hybrid composite for example, its $\Delta \varepsilon_c$ and hence the strain at termination of multiple cracking, $\varepsilon_{c-m}$, would mainly be controlled by the LE glass fibre due to the much greater elastic modulus of glass. Reducing $\Delta \varepsilon_c$ by incorporating LE fibres is one of the major benefits obtained in a hybrid composite. Also, according to equation 3.13, the existence of the high-elongation fibre would also reduce the strain at completion of multiple cracking in the hybrid composite.

Similarly, it can be derived for the minimum crack spacing of $x$ that

$$\Delta \varepsilon_f = \Delta \varepsilon_c = \frac{3}{4} \alpha' \varepsilon_{mu} \quad (3.14)$$

Correspondingly, strains at the end of matrix multiple cracking, $\varepsilon_{c-m}$, become

$$\varepsilon_{c-m} = \varepsilon_{mu} \cdot (1 + \frac{1}{2} \alpha') \quad (3.15)$$

$$\varepsilon_{c-m} = \varepsilon_{mu} \cdot (1 + \frac{3}{4} \alpha') \quad (3.16)$$

for crack spacings $2x$ and $x$ respectively. Where $\alpha'$ is defined by equation 3.8.
In practice, crack spacing \( C \) (equation 3.10) is determined by counting the crack numbers within a certain length of a test specimen. A single value of \( r \), the average frictional bond strength, is therefore obtained using equation 3.10 with the measured \( C \) and the measured specific surface area \( P_f/A_f \). As discussed in Chapter 2, the measurements of \( P_f/A_f \), the specific surface area per unit volume of the fibre, or \( P_r \), the perimeter of fibres in contact with matrix, are extremely difficult tasks, since the surface of fibres in contact with the cement matrix are not well defined. For example, when glass fibres are incorporated in a cement matrix in the form of bundles, the outer filaments in the bundle would be bonded to a greater extent than the inner ones and the degree of bonding probably varies from bundle to bundle depending on the treatment during incorporation. Because of the complex nature of bonding, the assessed perimeter \( (P_r) \) values (10-12) of the glass strands by means of an optical microscope may be treated to be approximate but by no means accurate. The accurate determinations of \( P_r \) and \( A_r \), or \( P_r/A_r \) of the fibrillated polypropylene film are even more difficult. This is because of the complexity of the hairy irregular surface shape and the multiple splits in the film and, additionally, of the discontinuous nature of the interfacial bonding (13). Attempts have been made to use a technique of Krypton absorption (13) or a method of image analysis (14) to measure the specific surface area but the resulted \( P_r/A_r \) values vary considerably from one method to the other. Even with the same method of the image analysis, the difference in the \( P_r/A_r \) values calculated from the independently measured \( P_r \), \( A_r \) compared with that directly measured could be as large as 400\% (14). Thus, before an appropriate method for measuring \( P_r/A_r \) or \( P_r \) is developed the frictional bond strength \( r \) obtained from equation 3.10 is, to a great extent a notional value.

Nevertheless, based on equation 3.10, Oakley and Proctor (12) have reported \( r \) values for glass fibres to be in the range of 0.8 to 1.6 MPa (in agreement with Laws et al (15)) using measured crack spacings of 1.0 to 1.9 mm and an assessed \( P_r/A_r \) value 2.83/(25160 x 10^{-6}) = 112.5 mm/mm². The composites Oakley and Proctor used contained 4-5\% volume fractions of 38 mm long
chopped glass strands. On the other side, using the same Surrey matrix at about 6% $V_f$ of polypropylene, Hannant (16) and Hughes (13) have reported $\tau$ values of 0.4 to 0.6 MPa corresponding to $P_f/A_f$ values ranging from 86 to 482 mm/mm$^2$ and crack spacings about 1 mm (16). For polypropylene, Ohno (14) has measured $\tau$ values of 0.6 to 1.5 MPa with corresponding $P_f/A_f$ values of 59 to 67 mm/mm$^2$ and crack spacings about 2 mm. Thus, it would appear from the above three groups of reported values that the differences in average frictional bond strengths, $\tau$, and in the crack spacings $C$ and hence the stress transfer lengths, between the all polypropylene cement and the all glass cement are not very significant for similar volume fractions of the two fibres. However, one significant difference between the two composites is that $\tau$ in the glass composite increases with aging (15) while in the polypropylene composite it does not (16) since the contact area between the polypropylene and the matrix is thought not to be the controlling factor (13, 16). Thus, a theoretical model capable of predicting the stress transfer mechanism in a young composite is not necessarily equally applicable to the aged composite.

3.2.2 On the failure of LE fibres

Assuming that the cement matrix has a single-valued failure strain. Then after the multiple cracking ceases at a stress $\sigma_f$, the two fibres alone take any increase in load and are pulled through the matrix as they extend resulting in another rising-stress curve. The slope stiffness or elastic modulus of the cracked hybrid composite, $E_1$, is determined by

$$E_1 = E_{f-L}V_{f-L} + E_{f-H}V_{f-H} \quad (3.17)$$

As the LE fibre and the HE fibre have very different strains to failure, the LE fibre in the hybrid composite must fail prior to the HE fibre when its failure strain is reached. Up to the first failure event of the LE fibres there should be a contribution to the overall composite stress, $\sigma_1$, from the HE fibres. In the present study, the stress $\sigma_1$ at which the LE fibre fails
is referred to as the "first failure stress" and the corresponding strain, \( \epsilon_1 \), as the "first failure strain". This point is shown by "c" on Figure 3.2. The first failure stress is thus given by

\[
\sigma_1 = \sigma_{f_u-L} V_{f-L} + \sigma_{f-H} V_{f-H}
\]  

(3.18)

where \( \sigma_{f-H} \) is the stress developed in the HE fibre at the first failure strain \( \epsilon_1 \)

\[
\sigma_{f-H} = \epsilon_1 E_{f-H}
\]  

(3.19)

and may be better approximated by

\[
\sigma_{f-H} \approx \epsilon_{f_u-L} E_{f-H}
\]  

(3.20)

because \( \epsilon_{f_u-L} \) is the maximum strain at a crack.

Equation 3.19 or 3.20 will generally give lower limits of the stress \( \sigma_{f-H} \) in the HE fibre, since the fibre strain, and hence stress developed at a crack opening could be much greater than the average stress calculated by the equations, especially, when the volume fraction of the HE fibre in the composite is low.

Assuming that the composite stress supported by the HE fibre at \( \epsilon_1 \), \( \sigma_{f-H} V_{f-H} \), is the same as that carried in an all HE fibre reinforced composite for a given \( V_{f-H} \) at \( \epsilon_1 \), any increase in stress over \( \sigma_{f-H} V_{f-H} \) at \( \epsilon_1 \) in the hybrid may be considered solely to be a contribution from the LE fibre incorporated in the HE fibre based composite. Thus, by measuring the HE fibre stress from the all HE fibre composite using equation 3.21 at the failure strain of the LE fibre and using this value in equation 3.18 with the assumed LE fibre strength, the hybrid stress \( \sigma_1 \) may be predicted.
\[ \sigma_{f-H} = \frac{\sigma_{f-H}}{V_{f-H}} \]  

(3.21)

where \( \sigma_{f-H} \) is the stress of all HE fibre cement at the failure strain of the LE fibre. Alternatively \( \sigma_{f-H} \) could be calculated from \( \sigma_f \) and equation 3.18 by a similar procedure.

Obviously, the use of equations 3.18 to 3.20 and 3.21 are limited by the assumptions which have to be made regarding the failure strain of the LE fibre in the hybrid.

Assuming that the fibres of a single type in the hybrid fail at a definite strain then when the local strain in the fibres at a crack reaches the failure strain of the LE fibre, \( \varepsilon_{fu-L} \), the LE fibre in the hybrid will fail. The first failure strain, \( \varepsilon_1 \), of the hybrid should be the ultimate failure strain of the LE fibres less an amount due to matrix restraint, i.e.

\[ \varepsilon_1 = \varepsilon_{fu-L} - \frac{1}{2} \alpha' e_{mu} \]  

(3.22)

for a crack spacing of 2x. For the minimum crack spacing of x

\[ \varepsilon_1 = \varepsilon_{fu-L} - \frac{1}{4} \alpha' e_{mu} \]  

(3.23)

Thus according to the above theory, up to the first failure event of the LE fibre, the stress-strain curve shape of the hybrid is predicted as that in Figure 3.2 consisting of three stages: the elastic stage (point o to a), at which all the three components in the hybrid are in their linear, elastic range, the matrix multiple fracture range (point a to b), in which the composite strain has exceeded the ultimate strain of the cement matrix and the matrix breaks into a series of parallel blocks, and the post matrix multiple-fracture range (point b to c), during which the two fibres alone
Figure 3.2 Predicted stress-strain curve shape of fibre-cement hybrid composite: up to the failure of LE fibres
are carrying load until failure of the LE fibre.

3.2.3 Critical volume fraction of HE fibres, $V_{\text{crit-H}}$ and multiple fracture of LE fibres

After the LE fibre fails at one crack, the additional load $\sigma_{fu-L}V_{f_L}$ would be thrown onto the HE fibres bridging the crack. In order to prevent catastrophic failure occurring in the hybrid at this stage, the load per unit area of composite at the first failure strain $\epsilon_1$ (at point c on Figure 3.2) of the hybrid must be less than the strength of the HE fibre times its volume fractions, i.e.

$$\sigma_1 \leq \sigma_{fu-H} V_{f-H}$$

That is, at the strain at point c on Figure 3.2

$$\sigma_{fu-L} V_{f-L} + \sigma_{f-H} V_{f-H} \leq \sigma_{fu-H} V_{f-H}$$

or

$$\sigma_{f-H} + \frac{\sigma_{fu-L} V_{f-L}}{V_{f-H}} \leq \sigma_{fu-H}$$

where $\sigma_{fu-L} V_{f-L}/V_{f-H}$ is the maximum additional stress on the HE fibre.

Thus, the minimum volume fraction of the HE fibres needed to sustain the additional stress without suffering instant failure, or the critical fibre volume fraction of the HE fibres, $V_{\text{crit-H}}$, is defined by

$$V_{\text{crit-H}} = \frac{\sigma_1}{\sigma_{fu-H}} = \frac{\sigma_{fu-L} V_{f-L}}{\sigma_{fu-H} - \sigma_{f-H}}$$

For example, for a continuous glass-polypropylene cementitious hybrid composite with a glass volume $V_{f,L} = 0.5\%$, according to equation 3.27, a polypropylene volume of 2.6\% is the minimum needed to avoid catastrophic
failure when the glass fails. In the calculation, \( \sigma_{fu,L} = 1200 \text{ MPa} \), \( \sigma_{fu,H} = 400 \text{ MPa} \) and \( \sigma_r = 170 \text{ MPa} \) are assumed.

Thus, if there is not enough HE fibre present in the hybrid or, in other words, if the volume fraction of the HE fibre incorporated is lower than its critical fibre volume \( V_{f_{crit-H}} \) defined by equation 3.27, catastrophic failure of the hybrid will be instant following the failure of the LE fibre. In this case, the complete stress-strain curve of the hybrid will be the same as that in Figure 3.2.

If the volume fraction of the HE fibre exceeds the critical fibre volume at point c on Figure 3.2 there are two possibilities available to predict the shape of the stress-strain curve at strains exceeding \( \epsilon_1 \).

**Simultaneous fracture of all glass filaments**

If all the glass filaments fracture simultaneously at all cracks, the sudden transfer of load to the HE fibre will result in a rapid extension of the crack at a rate faster than the test machine cross-head speed. The load cell will therefore unload from c to d on Figure 3.3a before further loading of the HE fibre along path d-e.

**Progressive fracture of glass filaments**

When glass fibres are incorporated in a composite in the form of large bundles or rovings consisting of a number of strands per roving, a bond gradient and thus a stress gradient may exist between the outer fibres and the core fibres in relation to the cement matrix. The outer fibres will be better bonded with the matrix and more highly stressed when the external tensile load is applied, compared with the core fibres. Therefore, at \( \sigma_1 \), the outer fibres of the roving will fracture first at a single crack and the additional load due to glass fibre breakage will be thrown onto the unbroken core fibres of the roving.
This progressive fracture of the glass at only one crack will slow down the stress transfer to the HE fibre reducing its extension rate to nearer that of the cross-head. Thus length c-c' in Figure 3.3b will be less than length c-d in Figure 3.3a. If the stress transfer rate to the HE fibre is sufficiently reduced to equal to the rate of cross-head movement then there will be no sudden load drop at c on the tensile stress-strain curve on Figure 3.3b and Figure 3.3c results. Additional fracture of the LE fibre (glass) may also occur at cracks where it was previously intact. Then c'-d curve on Figure 3.3b, or c-d curve on Figure 3.3c, will be horizontal because the stress, the strain and the modulus balance in the HE fibre at d must be satisfied before further loading of the HE fibre along curve d-e.

**Prediction of stress-strain curve from c to e on Figure 3.3**

Because of the interaction of the test machine and load cell stiffness and stress transfer rates in the hybrid composite, it is not possible to predict the position of c' on Figure 3.3b because this point is not related to fundamental material parameters. Difficulties exist regarding the quantitative prediction of length c'-d on the stress-strain curve on Figure 3.3b (or c-d on Figure 3.3c) because of the extensively cracked nature of the hybrid and because the exact stress transfer mechanism between the components at this loading stage is more complicated and unknown. However, it should be possible to predict slope d-e and point e.

After point c' on Figure 3.3b, the stress at a given average strain in the composite will be generally greater for a hybrid composite than for a HE single fibre composite for the following reasons:

At the first crack where the LE fibres fail, the local strain in the HE component at the crack must be equal to the local strain in the HE single fibre composite at the same stress in order to satisfy compatibility between stress, strain and elastic modulus of the fibre. However, we do not measure local strain at the crack but only the average strain over many cracks. In
Figure 3.3 Predicted stress-strain curve shapes of fibre-cement hybrid composites, provided that there is:
   a) simultaneous fracture of all glass filaments,
   b) progressive failure of glass filaments, and
   c) stress transfer rate to HE fibre on the failure of glass fibre equals to the rate of cross-head movement.
other cracks, before \( \sigma \) is again exceeded, some LE fibres will still be intact or will have residual frictional pull-out stresses which reduce crack width and hence will reduce the average strain for a given composite stress. Thus, the average strain until all the glass fibres have failed at every crack, will be smaller for a given hybrid stress than for the single HE fibre case. This type of behaviour is shown in the experimental results in Chapter 5.

3.2.4 Ultimate tensile strength and strain

When the fracture of the LE fibres is complete, the HE fibres will alone take any increase in stress until they break at their failure strength. If the failure of the LE fibres causes no immediate breaking of the adjacent HE fibres in the hybrid and the LE fibre no longer contributes to the load carrying capacity of the hybrid then the ultimate tensile strength of the composite, \( \sigma_{cu} \), should be that of the HE fibre multiplied by its volume fraction:

\[
\sigma_{cu} = \sigma_{fu-H} V_{f-H} \tag{3.28}
\]

The ultimate tensile strain of the hybrid, \( \varepsilon_{cu} \), is predicted from the conventional single fibre theory alone. That is:

\[
\varepsilon_{cu} = \varepsilon_{fu-H} - \frac{1}{2} \frac{E_n(1-V_{f-H})}{E_{f-H}V_{f-H}} \varepsilon_{mu} \tag{3.29}
\]

and

\[
\varepsilon_{cu} = \varepsilon_{fu-H} - \frac{1}{4} \frac{E_n(1-V_{f-H})}{E_{f-H}V_{f-H}} \varepsilon_{mu} \tag{3.30}
\]

for crack spacing 2x and x respectively.
3.3 Summary

Based on the assumptions of constant frictional bond and equal average additional strain in the two fibres at the first matrix crack and up to the first failure event of the LE fibre, the stress-strain curve of the fibre-cement hybrid composite is predicted as that in Figures 3.4 a), b) and c) but up to the point "c" only, consisting of three stages:

1) Elastic stage (curve o-a), at which all the three components, the LE fibre, the HE fibre and the cement matrix, in the hybrid are in their linear, elastic range.

2) Matrix multiple fracture range (curve a-b), in which the composite strain has exceeded the ultimate strain of the cement matrix and the matrix breaks into a series of parallel blocks.

3) Post matrix multiple-fracture range (curve b-c), during which the two fibres alone are carrying load until failure of the LE fibre.

If there is not enough HE fibre in the hybrid, catastrophic failure of the hybrid will occur instantly following the first failure event of the LE fibre. In this case, the end point of the complete stress-strain curve of the hybrid will be at point c on Figure 3.4 solely consisting of three stress-strain stages. Further toughening by the HE fibre will not occur and the ultimate tensile strength, $\sigma_{cu}$, of the hybrid will equal to its first failure stress, $\sigma_1$.

If there is sufficient HE fibre present in the hybrid, the stress-strain curves after the failure of the LE fibres will depend on the way in which the LE component fractures and the relative rates of extension of the composite and the test machine cross-head. Three types of stress-strain curves corresponding to the assumptions of simultaneous fracture (Figure
3.4a) or progressive LE component fracture (Figure 3.4b) or equal rate of stress transfer and cross-head movement (Figure 3.4c) are predicted.

In Figure 3.4, the quantitative prediction of curve c'-d (or strain $\varepsilon_{c_d}$) cannot be determined because at this loading stage the hybrid composites have become extensively cracked and the exact stress transfer mechanism between the components is complicated and unknown. It is not possible to predict the position of c' on Figure 3.4 because this point is not related to fundamental material parameters.
Figure 3.4  Predicted stress-strain curve of a fibre-cement hybrid composite
Figure 3.4 Predicted stress-strain curve of a fibre-cement hybrid composite
Definitions of the symbols shown in Figure 3.4 are listed below.

\( E_c \)  
Elastic modulus of uncracked composite.

\( \epsilon_{\text{mu}} \)  
Matrix cracking strain.

\( \sigma_c \)  
Composite cracking stress.

\( \epsilon_{c-m} \)  
Strain at termination of multiple fracture of cement matrix.

\( E_1 \)  
Elastic modulus of cracked hybrid composite at the range of post multiple fracture of matrix.

\( \epsilon_1 \)  
Strain of hybrid composite at which the LE fibre fails, referred to the "first failure strain" in the present study.

\( \sigma_1 \)  
Stress of hybrid composite at which the LE fibre fails, referred to the "first failure stress" in the present study.

\( \epsilon_{c-L} \)  
Strain at which fracture of the LE fibre terminates and the stress and the strain balance in the HE fibre.

\( E_2 \)  
Elastic modulus of cracked composite at the range of post multiple fracture of LE fibre.

\( \epsilon_{\text{cu}} \)  
Ultimate tensile strain of composite.

\( \sigma_{\text{cu}} \)  
Ultimate tensile strength of composite.

\( \epsilon_{\text{fu-H}} \)  
Ultimate tensile strain of HE fibre only.
In this chapter, the constituent materials used in the study and the determination of their properties are described. Details concerning experimental programmes for the various composites are located in the appropriate sections of the subsequent chapters.

4.1 Constituent materials of matrices

4.1.1 The Surrey matrices

The materials forming the matrices used at Surrey (i.e. the normal Surrey matrix and the modified Surrey matrix) were cement, silica sand, pulverised fuel ash (fly ash), water and a superplasticiser.

The cement used was ordinary Portland cement manufactured to BS12 obtained from Blue Circle. The silica sand has a particle size range between 100-300 μm supplied by Buckland Sand & Silica Ltd. The fly ash tended to be compacted within the bag and was placed over a No. 25 sieve to break up the lumps. The superplasticiser used was a sulfonated melamine formaldehyde resin with the trade name Melment L10.

The normal Surrey matrix, with mix proportions listed in Table 4.1, has been used at Surrey university for many years for manufacturing the fibrillated polypropylene network reinforced composites. This matrix has been shown to have excellent workability which gives easy penetration of the fresh matrix into the fibrillated polypropylene networks during fabrication. In the present study, this matrix was modified (modified Surrey matrix) to have the mix proportions shown in Table 4.1. The modifications were made mainly based on two considerations: 1) Most of the composites used in the present study
were to contain AR-glass fibres which have been shown to deteriorate in high alkali environments with aging (1-2). This deterioration may be slowed down by reducing the overall alkalinity of the matrix. 2) Product quality and production cost are also two primary considerations in industry. Increasing the content of fly ash and silica sand not only reduces the overall alkalinity of the matrix and the raw material cost but also reduces drying shrinkage (notably by silica sand, Refs. 3-4) which is highly desirable in thin cladding and roofing materials. Other important improvements in composite properties by adding fly ash are: better retention in strength and toughness of glass-cement composites, greater increase in long-term matrix strength and reduced permeability (5-9). The greater amount of fly ash and silica sand may insert a strength penalty at early stages which has been shown to be less obvious pronounced if the replacements are kept with certain ranges (3-4, 8-9).

Table 4.1 Mix proportions of modified Surrey matrix by weight compared with normal Surrey matrix

<table>
<thead>
<tr>
<th>Material</th>
<th>Normal Surrey matrix</th>
<th>Modified Surrey matrix</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ordinary Portland cement</td>
<td>1.00</td>
<td>1.00</td>
</tr>
<tr>
<td>Silica sand</td>
<td>0.19</td>
<td>0.35</td>
</tr>
<tr>
<td>Pulverised Fuel ash (fly ash)</td>
<td>0.25</td>
<td>0.30</td>
</tr>
<tr>
<td>Water</td>
<td>0.34</td>
<td>0.38</td>
</tr>
<tr>
<td>Superplasticiser</td>
<td>0.02</td>
<td>0.02</td>
</tr>
</tbody>
</table>

It was the modified Surrey matrix which was used in the present study. Compared with the normal Surrey matrix, the modified matrix has higher (sand+fly ash)-to-cement ratio and higher replacement of cement by fly ash. In the modified matrix, the proportion of fly ash in the total cementitious materials (cement+fly ash) content was 0.23, and the (sand+fly ash)-to-
cement ratio was 0.65. By using superplasticiser, the water-to-solid(cement+fly ash+sand) ratio was kept at 0.23 for all the composites made in the Surrey laboratory.

4.1.2 The Broni matrix

Although the modified Surrey matrix had similar sand content to that of the commercial Retiver product, the Retiver matrix didn't contain fly ash. Also, the two matrices had raw materials from different sources. It was suggested that a matrix similar to Retiver matrix should be used in the fibre-cement hybrid composites made in the Broni laboratory of Fibronit srl using the same raw materials as in Retiver so that a closer idea of the effect of the use of combined types of fibres on the product might be obtained. We named this matrix after Broni and the mix proportions by weight are given in Table 4.2. All the composites manufactured in the Broni laboratory used the Broni matrix.

<table>
<thead>
<tr>
<th>Material</th>
<th>Mix proportion by weight</th>
</tr>
</thead>
<tbody>
<tr>
<td>Portland cement</td>
<td>1.00</td>
</tr>
<tr>
<td>Silica sand</td>
<td>0.38</td>
</tr>
<tr>
<td>Water</td>
<td>0.36</td>
</tr>
</tbody>
</table>

The cement used in Broni matrix was 425 Portland cement, manufactured by Italcementi. Its physical properties, provided by the producer, are shown in Appendix I. The silica sand, supplied by Sive 4, had a particle size range between 100-300 μm, a similar grade as that in the Surrey matrix. The Broni matrix had a sand-to-cement ratio of 0.38 and a water-to-solid(cement+sand) ratio of 0.26.
Water-reducing admixtures were not used in the Broni matrix because problems in using water-reducing admixtures may occur in vacuum dewatering process like Hatschek and Magnani machines (10), since the accumulative built-up of the plasticiser in the recycled water may eventually over dose the cement slurry and make continuing production impossible.

4.1.3 Preparation of samples of unreinforced cement matrix

One plain matrix sheet composed of modified Surrey matrix was prepared by casting the fresh mortar into a wood-frame mould with a thickness about 7 mm and a piece of polythene film acting as its base. The fresh mortar was trowelled flat with a little pressure and left in laboratory air for 24 hour covered by polythene film to avoid rapid drying. Then the sheet was stored in water at 20 °C and at the age of 28 days it was cut into strips about 300x25 mm for tensile tests and 150x50 mm for flexural tests.

One plain matrix sheet composed of Broni matrix was prepared by casting the fresh mortar into a multi-framed steel mould as shown in Figure 4.1. With this mould, the sample coupons with above mentioned dimensions can be moulded without cutting. The samples covered with the polythene were demoulded after 24 hours and immersed in 20 °C water until testing at 28 days.

4.1.4 Determination of matrix properties

The coupons of the unreinforced matrices were tested under tension and flexure (4-point loading) at Surrey in an Instron 1122 machine at 28 days. The test procedures were the same as that for composites described in Section 5.2.1 (tensile test) and Section 7.2.1 (flexural test). Here only the test results are provided. Both sets of tensile and flexural values of the plain specimens were quite scattered, so that a range of the measured values are given, as shown in Table 4.3. Each value in the table was the average of at least four or five test results.
Figure 4.1 Moulds for preparing plain matrix coupons
The flexural elastic modulus of the plain specimens was determined using the following formula (11) assuming the conventional beam theory to apply:

\[ E = \frac{W a^2}{6 D I} (3L - 4a) \]  

(4.1)

Where

- \( W \) - total load (N)
- \( a \) - distance of loading line from support, \( a=L/3 \) mm
- \( D \) - deflection under either load (mm)
- \( I \) - second moment of area
- \( L \) - length of span, \( L=135 \) mm

When comparing the modified Surrey matrix with the Broni matrix in Table 4.3, the slightly higher strength shown by the Broni matrix may be the result of smaller content of the additives (sand+fly ash). The higher strain to break of the Broni matrix might be due to the higher sand content used.

4.2 Fibres and properties

4.2.1 Fibrillated polypropylene networks

Since the material of fibrillated polypropylene networks has been studied and reported by several researchers (12-15), only a general review of the polypropylene networks used in this study is presented.

4.2.1.1 General description of polypropylene networks

Polypropylene film is made of high molecular isotactic polypropylene produced by the basic procedure of heating, extrusion, quenching and drawing. The drawing of such film increases the crystalline orientation and improves the mechanical properties of the subsequently fibrillated fibres. The fibrillation of the film is operated by passing the film over rollers which are fitted with a designed pattern of pins and blades. Using
Table 4.3 Matrix properties

<table>
<thead>
<tr>
<th>Property</th>
<th>Modified Surrey matrix</th>
<th>Broni Matrix</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Tension</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Tensile strength  (MPa)</td>
<td>2.6-4</td>
<td>3.3-5</td>
</tr>
<tr>
<td>Tensile modulus   (GPa)</td>
<td>22.30</td>
<td>17.29</td>
</tr>
<tr>
<td>Failure strain    x10^-6</td>
<td>70-140</td>
<td>120-200</td>
</tr>
<tr>
<td><strong>Flexure</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Flexural strength (MPa)</td>
<td>7-11</td>
<td>9-11</td>
</tr>
<tr>
<td>Flexural modulus  (GPa)</td>
<td>27.32</td>
<td>22.31</td>
</tr>
<tr>
<td>Maximum deflection (mm)</td>
<td>0.25-0.30</td>
<td>0.30-0.40</td>
</tr>
</tbody>
</table>

A specially designed machine, the fibrillated networks are then opened and stabilised. The products of fibrillated polypropylene networks are industrially supplied in the form of rolls or bobbins on which a few layers of networks have been combined as one pack.

4.2.1.2 Polypropylene networks used in the study

In this study, an industrial product of fibrillated polypropylene networks with the trade name Retiflex was used. The polypropylene was made up of 12 layers of opened networks for each pack, 8 being in the longitudinal direction and 4 at right angles to this (Figure 4.2). To prevent deterioration from ultraviolet rays, a carbon powder was added to the polymer and thus the networks were black.

4.2.1.3 Properties of polypropylene film, unfibrillated and fibrillated

Table 4.4 shows the main properties of the polypropylene films. In the
Figure 4.2  Fibrillated polypropylene networks (Retiflex)
Figure 4.3 Schematic geometry of fibrillated polypropylene networks (15)
second column of the table are listed the properties of unfibrillated film measured by the producer (16) and by Ohno (15), respectively. The properties of the unfibrillated film were measured by direct tension using a film strip of 5 mm in width. It is almost impossible to measure the properties of the fibrillated film alone by direct tensile test due to its highly irregular geometry and network form (Figure 4.3). Thus, the mechanical properties of fibrillated polypropylene networks have been assessed by the tensile test on the fibrillated network reinforced cement composites (12-15). In the third column of Table 4.4, properties of fibrillated polypropylene networks were collected from Ohno's work (15) who used same type of polypropylene networks in his study as the current author. In the table, the modulus of the fibrillated film was referred as "secondary modulus", as it was determined at 5% strain in the tensile stress-strain curves of the composites (15). Instead, the initial modulus values for the unfibrillated polypropylene film were determined within 0.5% strain on the tensile stress-strain curves of the film. Up to 0.5% strain a linear relationship of stress and strain could be maintained but beyond this strain the film modulus decreases rapidly. For instance, modulus of the unfibrillated film was about 12 GPa at 0.5% strain and at 5% strain the modulus decreased to about 7 GPa, compared with 2.5-3.5 GPa of the fibrillated film at the same strain (Table 4.4). It would appear from the previous work (12-15) that fibrillating polypropylene film doesn't affect much its tensile strength and strain to failure but significantly reduces elastic modulus of the unfibrillated film. This reduction in elastic modulus for the fibrillated film may be attributed to the reduction of fibre alignment due to its netlike form and to the earlier breakage of some of the fine fibrils or hairs held in the matrix near the edge of the main fibres during the tensile process (15).

It can be seen in Figure 4.3 that the fibrillated polypropylene network has a very complicated geometry, in addition to its highly uneven surface where the asperities could be in the range of 1-20 microns (13, 15). The uneven surface and the complicated geometry have considerable influence on the way of bonding between the networks and the cement matrix and this is the
The primary reason why the film is fibrillated. The bond is not only the result of interfacial adhesion, a considerable contribution may be due to mechanical anchoring and interlocking. Hughes called such kind of bonding as "misfit" (13).

Table 4.4 Main properties of the polypropylene films, unfibrillated and fibrillated

<table>
<thead>
<tr>
<th>Property</th>
<th>Unfibrillated</th>
<th>Fibrillated</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Moplefan(16)</td>
<td>Ohno(15)</td>
</tr>
<tr>
<td>Tensile strength (MPa)</td>
<td>500 - 550</td>
<td>400</td>
</tr>
<tr>
<td>Initial modulus within 0.5% strain (GPa)</td>
<td>12 - 14</td>
<td>12 - 13</td>
</tr>
<tr>
<td>Secondary modulus at 5% strain (GPa)</td>
<td></td>
<td>7</td>
</tr>
<tr>
<td>Strain to break (%)</td>
<td>8.1</td>
<td>6</td>
</tr>
<tr>
<td>Fibre thickness (µm)</td>
<td>60 - 80</td>
<td></td>
</tr>
<tr>
<td>Specific gravity (g/cm³)</td>
<td>0.93</td>
<td></td>
</tr>
<tr>
<td>Wettability (dyne/cm)</td>
<td>40</td>
<td></td>
</tr>
</tbody>
</table>

* range of polypropylene volume fractions used: 2.6-14%

4.2.2 Alkali-resistant glass fibres

4.2.2.1 Manufacture of glass fibres

Glass fibres are manufactured in a process in which molten glass is drawn in the form of filaments, through the bottom of a heated platinum tank or bushing. Usually, 200 filaments with 10-20 µm in diameter are drawn
simultaneously and they solidify while cooling outside the heated tank. The filaments then are wound on a rotation drum into a "strand". Prior to winding, the filaments are coated with a size which binds the filaments into bundles for incorporation into composites as well as decreasing the deterioration rate of glass fibres in the cement environment if a chemical inhibitor such as pyrogallol is incorporated in the size formulations (17-18). A number of strands may be collected together to form a "roving". The glass fibres can be marketed as either continuous rovings or chopped strands, as shown in Figure 4.4.

4.2.2.2 Alkali-resistant (AR) glass fibres

Originally, E-glass fibres, the common type used in the fibre reinforced plastics industry, were directly adopted to reinforce cements. It was however found that the exposure of the E-glass fibres to an alkaline environment led to a rapid deterioration of the glass fibres which involved strength and weight losses and reduction in the filament diameter (19). This process can be attributed to the break-down of silicon-oxygen-silicon bonds in the glass network, by the hydroxyl ions which are highly concentrated in the alkali pore solution:

\[ \text{Si-O-Si + OH}^- \rightarrow \text{Si-OH + SiO} \quad \text{(in solution)} \]

Early work by Biryukovich et al (20) attempted to overcome this problem by using the E-glass fibre in combination with a matrix of low alkali cement or cement with polymer. A different approach adopted by Majumdar at the Building Research Establishment (BRE) was to incorporate higher than 16% ZrO₂ in the glass composition, based on observations that fibres made from glasses in the Na₂O-SiO₂-ZrO₂ system are more chemically stable in alkali solutions (19, 21). Early commercial developments on AR-glass fibres were undertaken by Owens-Corning Glass Works, Pilkington Brothers Ltd and Nippon Electric Glass Co. Ltd (NEG). In the late of 1970's, China Building Materials Academy succeeded in developing AR-glass fibres with the combined
Figure 4.4 Forms of glass fibre reinforcement
   a) Continuous roving   b) Chopped strand
use (> 20%) of TiO₂ and ZrO₂. There are several AR-glass fibre manufacturers in China at present.

The AR-glass fibres used in this study were supplied by Pilkington Co. Ltd. (Cem-FIL 2, chopped 24 mm strand) and by Nippon Electric Glass Co. Ltd. (continuous roving), respectively. Typical physical properties from the technical data sheets (23-25) are listed in Table 4.5.

It is notable from Table 4.5 that the tensile strength of bundled strand and single filament can be significantly different. One explanation for this difference may be that, the extractions of single filaments from a strand for the tensile test inevitably has selected the stronger ones, since the weaker filaments are prone to damage and fracture in the process. On the other hand, when a strand is tested, the 200 filaments in the strand can not be bundled under completely the same tension and relative slip might occur. The weaker filaments in the strand during the tensile process will break first and the load originally carried by the weaker fibres is transferred to the surviving fibres. This sudden increase in load will promote more fibres breakage in the bundle than in the case where the unbroken fibres are under static loading and result in decrease of load to failure. It may be that the value of single filament should be used in case of chopped glass strands used in Hatschek process and in Magnani process because the fibres are expected to disperse individually in the cement slurry and function independently. In the cases where the glass fibres used as chopped strands and can not be dispersed individually in cement (say, in the process of spray-up, hand lay-up, Retiver process, etc.), the value of the strand strength instead of filament strength should be taken into account. In the present study, continuous glass fibres were incorporated in cement in the form of "roving" which consisted of 32 strands, the strand strength may not be reached in the roving due to the bundle-size effect.
<table>
<thead>
<tr>
<th>Property</th>
<th>Cem-FIL 2</th>
<th>Nippon glass</th>
</tr>
</thead>
<tbody>
<tr>
<td>Filament diameter (µm)</td>
<td>13-20</td>
<td>13.5 ± 2</td>
</tr>
<tr>
<td>Number of filaments per strand</td>
<td>200</td>
<td>200</td>
</tr>
<tr>
<td>Number of strands per roving</td>
<td></td>
<td>32</td>
</tr>
<tr>
<td>Specific gravity (g/cm³)</td>
<td>2.68</td>
<td>2.70</td>
</tr>
<tr>
<td>Tensile strength of single filament (MPa)</td>
<td>3700</td>
<td>2500</td>
</tr>
<tr>
<td>Tensile strength of glass strand (MPa)</td>
<td>1700</td>
<td>1400</td>
</tr>
<tr>
<td>Elastic modulus (GPa)</td>
<td>72-76</td>
<td>74</td>
</tr>
<tr>
<td>Strain at failure (%)</td>
<td>up to 2.4%</td>
<td>up to 2%</td>
</tr>
<tr>
<td>Thermal expansion coefficient /°C</td>
<td>7.9x10⁻⁶</td>
<td>9x10⁻⁶</td>
</tr>
</tbody>
</table>

4.2.2.3 Determination of glass fibre properties

A simple direct tensile test on Nippon glass strand was carried out using special clamps (Figure 4.5) equipped in the Instron 1122 machine to assess its mechanical properties. Theoretically, the cross-section of a strand can be determined by multiplying the cross-section of a filament with the total number of filaments in the strand which can be known from the manufacturer's data sheet. However, it was found by weighing the strands in a roving that the weight per unit length of individual strands was not always constant but varies a lot. This may imply that the number of filaments in a strand was not constant which might be the result of splitting of a strand or breakage of some of the filaments. It was then decided that the cross-section area of
Figure 4.5  Set-up for direct tensile tests of long fibres
a strand should be determined by dividing the actual strand weight per meter length by the specific gravity of the fibres. This method at least eliminated the calculation error caused by the difference in strand weight but ignored the coating size weight on the fibres.

A meter length of strand was cut equally into three portions after weighing in a balance to 0.0001 g. Each portion of the strand had a length 300 mm. A 25 mm length of each strand end was treated with an epoxy resin covered by masking tape and left in laboratory air overnight. This treatment was aimed at alleviating relative slip of filaments and preventing the ends from rupturing during the tensile test. The sample was then placed and centralised in the clamps of the Instron and tested at a constant speed of 2 mm/min. The load output from the Instron load cell and the cross-head movement were recorded on the chart. The strain was taken to be the cross-head displacement divided by the "effective" sample length of 250 mm, assuming that there was no slip of the clamped yarn ends. This treatment would give greater than actual strain values, if slip and extension of the end parts do occur in practice. On the other hand, if the tensile strain is calculated based on the whole sample length (300 mm), smaller than true strain values would result because the end parts would extend considerably less than the unrestrained central part of the sample. The true tensile strain might lie between the values of the two cases. The strand strength was determined from the breaking load divided by the cross-section area of the strand, and the elastic modulus of the strand was determined from the linearity of the stress-strain curve using Hook's Law.

With a similar procedure described above, strands of 650 mm length (i.e. 600 mm "effective" length) were also tested in order to check the length effect. It was thought that the error introduced for strain caused by relative slip, if any, between the clamped parts of the strand and the clamps, and between the inner part and the outerpart of the strand might become negligible if a longer strand length was used for the testing.
The test results are included in Table 4.6. Each result was the average of 20 individual test results for the 250 mm Nippon glass strand and of 6 individual test results for the 600 mm Nippon strand. Representative tensile stress-strain curves are shown in Figures 4.6 a) and b) for the samples of 250 mm and 600 mm test lengths, respectively.

The direct tensile test was unable to be carried out on the Cem-FIL 2 glass, as it was supplied in the form of chopped strand. However, Kakemi (26) has measured the properties of continuous Cem-FIL 2 glass strand using a similar method described above and his reference is taken as shown in the same table.

It can be seen from Figure 4.6 that the tensile stress-strain curve of the glass strand was essentially linear and this was as expected. The elastic modulus measured (Table 4.6) was slightly smaller than that given by the producer (Table 4.5) but generally in agreement. The tensile strength

<table>
<thead>
<tr>
<th>Glass type</th>
<th>Nippon glass strand</th>
<th>Cem-FIL 2 strand</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>250 mm</td>
<td>600 mm</td>
</tr>
<tr>
<td>Tensile strength (MPa)</td>
<td>935 (109)</td>
<td>883 (81.4)</td>
</tr>
<tr>
<td>Elastic modulus (GPa)</td>
<td>69 (2.2)</td>
<td>70 (2.9)</td>
</tr>
<tr>
<td>Strain to failure (%)</td>
<td>1.47 (0.16)</td>
<td>1.34 (0.12)</td>
</tr>
</tbody>
</table>

( ) standard deviation
Figure 4.6 Tensile stress-strain curves of glass strand
a) 250 mm test length,   b) 600 mm test length
values were scattered. Both the strength and the strain to failure measured in the study were much lower than that given in the data sheets (Table 4.5). One of the reasons might be due to different methods and devices used for testing.

It would appear from the limited testing results that the failure strain and strength of the 600 samples were smaller than that of the 250 mm samples. The sample length effect on the measured properties needed to be further clarified.

4.2.3 Polyvinyl alcohol fibres

The polyvinyl alcohol (PVA) resin was discovered in 1924, and the water-soluble PVA fibre and the water-insoluble PVA fibre were firstly manufactured in 1931 and in 1939 (27), respectively. For a long period of time since PVA fibres were invented, they have been mainly used in the fields of PVA fibre reinforced rubber, PVA fibre reinforced plastics and PVA fibre reinforced asphalt. It is only in the past ten years that the specially developed high strength, high modulus PVA fibres have been used in non-asbestos fibre cement products.

4.2.3.1 Polymer structure and properties

It is known that polyvinyl alcohol is synthesized from acetylene or ethylene under certain conditions and of the following molecular structure:

\[
\text{OH} \quad (-\text{CH}_2\text{-CH-})_n
\]

As the intermolecular attraction is strengthened because of the hydrogen bond formation between two hydroxyl groups of neighbouring chains, crystallisation can occur easily during manufacture and crystallites are oriented very highly in the chain direction. The polymer has a melting point
about 240 °C which is much higher than polypropylene. Chemical resistance and the stability to sunlight of the polymer is good. In contrast to many other synthetic polymers, PVA is hydrophilic in nature due to the presence of -OH groups. These are good fundamental reasons for the polymer being chosen as a raw material to manufacture fibres for cement reinforcement.

4.2.3.2 Industrial production of PVA fibres

So far only a few companies in the world including Kuraray Co. Ltd and Unitika Kasei Ltd in Japan, Beijing Vinylon Plant in China and 28 Vinylon Union Corporation in North Korea can produce high modulus (say, over 30 GPa) and high strength PVA fibres specially for the cement products industry, using the wet-spinning process, although there are several companies in China producing so-called "Modified PVA fibres" for cements with modulus 20 GPa or lower (28). The manufacturing technology may differ from one producer to the other but the main manufacturing procedure is principally similar consisting of four operations, namely, dissolution, spinning, drawing and finishing (27, 29-30).

The purified polymer is dissolved in hot water and the solution is extruded through spinnerets in a spinning machine into an alkaline coagulating bath to form continuous filaments, which are then dried, heat drawn and heat treated to obtain required mechanical properties and dimensional stability. A suitable finishing agent is then applied to the fibre surface to enhance its bonding with cement and improve its dispersability in cement slurry in the case of short fibres. Figure 4.7 shows PVA fibres in the form of continuous rovings.

There are two critical contributing factors for achieving high strength and high modulus of PVA fibres during manufacture, namely, boric acid and alkali spinning. Detailed information can be found from Refs. 27 and 29.
Figure 4.7  Continuous polyvinyl alcohol (PVA) fibre roving
4.2.3.3 Cement products containing PVA fibres

High strength PVA fibres have been developed mainly for asbestos replacement in the Hatschek process manufacturing flat or corrugated sheets. Another recent application of the PVA fibres is in the Magnani or Mazza process to produce asbestos-free non-pressure sewage pipe. In these industrial processes, no single type of fibre so far can provide simultaneously a reinforcing effect and the filtering and solid retention characteristics of asbestos fibres. Thus, a blend of at least two types of fibres was used. In such non-asbestos cement systems PVA fibres, with a fibre content between 1.5%-2.5% by weight and a cut length of 4 or 6 mm, function as reinforcing fibres. PVA fibres have little support for cement particles and require a carrier to give uniform distribution in a cement slurry. By the combined use of cellulose pulp or other synthetic pulp - which are carriers of PVA fibres, and processing aids such as mineral fillers and polyacrylamide flocculant, the problem of poor retention of cement particles by PVA fibres can be alleviated and a homogeneous fresh fibre slurry can be obtained. The flexural strength and especially the stiffness of the resulted fibre cements are lower than asbestos cements, but the energy absorbing ability or toughness could be two or three times that of the asbestos cement (31). It has been reported that, for a fibre cement sheet produced in a Hatschek process containing 2% PVA fibres, 3% cellulose pulp and 1.5% slag wool, a flexural strength over 30 MPa has been resulted (29, 32). For a non-asbestos sewer pipe containing 2% PVA fibres, 3-4% cellulose and 1-2% polyethylene pulp, a flexural crushing strength over 40 MPa has been obtained (33).

4.2.3.4 Long-term stability of PVA fibres

According to some investigators (29, 34-36), the PVA fibres can be adequately durable in high alkali cement environments. After immersion in water at 60 °C for 57 days of the PVA fibre-cement specimens, the ratio value of bending moment after aging and before aging was greater than 1 (36), indicating the excellent strength retention of the PVA fibres. This
aging behaviour of PVA is superior to glass fibres. Zhou et al (35) reported that PVA fibres are thermally stable and insensitive to biological attack. This was reflected in the weathering performance of PVA-cement composites. After 6 years natural weathering, a 11% and a 23% increase in flexural strength of the PVA-cement composites have been reported by Akers et al (34) and by Tan et al (36), respectively. At the same time however, the toughness was reduced. Hikasa (29) has reported a reduction in toughness of nearly 50% after one year exposure of the composite in natural weathering. This reduction, combined with the small increase in strength may be due to improved fibre-matrix bond, this proposed mechanism must be validated by more study.

4.2.3.5 Mechanical properties of PVA yarns

In the present study, continuous PVA rovings rather than short PVA fibres were used. The PVA roving was formed by five individual yarns with 1000 filaments per yarn. Table 4.7 shows the typical properties of the PVA yarns and the data were taken from the data sheets (37-38) of the producer (Kuraray Co. Ltd). The T-5501 type is the normal type "high strength and high modulus" PVA fibre and much used in the form of short fibres in Hatschek process. The other two types of the PVA yarns, T-5516 and T-7901, are the very recent development by Kuraray.

The properties in terms of strength and modulus of PVA shown in Table 4.7 would be expected to be higher for single filaments. Even in the form of yarns, their tensile strength and modulus are much higher than many other types of synthetic organic fibres such as polyolefin fibres, nylon fibres, polyester fibres and polyacrylonitrile fibres. It is interesting, in addition, to note that the tensile strengths of the yarns can be of the same order of magnitude as that of the glass fibre strand, and their elastic modulus similar to or greater than the cement matrix. The elongation at break ranges from 5.4% to 9% for the currently marketed high modulus PVA fibres, which are still too high to allow the maximum strength of the fibre to be used in pressure pipes as major reinforcement.
Table 4.7 Properties of PVA yarns (37-38)

<table>
<thead>
<tr>
<th>Yarn type</th>
<th>T-5501</th>
<th>T-5516</th>
<th>T-7901</th>
</tr>
</thead>
<tbody>
<tr>
<td>Diameter (μm)</td>
<td>14</td>
<td>14</td>
<td>14</td>
</tr>
<tr>
<td>Total denier*</td>
<td>1800</td>
<td>1800</td>
<td>1800</td>
</tr>
<tr>
<td>Number of filaments per yarn</td>
<td>1000</td>
<td>1000</td>
<td>1000</td>
</tr>
<tr>
<td>Number of yarns per roving</td>
<td>5</td>
<td>5</td>
<td>5</td>
</tr>
<tr>
<td>Tensile strength (MPa)</td>
<td>1300</td>
<td>1490</td>
<td>1900</td>
</tr>
<tr>
<td>Elastic modulus (GPa)</td>
<td>25</td>
<td>32</td>
<td>38</td>
</tr>
<tr>
<td>Elongation to break (%)</td>
<td>6.7</td>
<td>6.3</td>
<td>5.4</td>
</tr>
<tr>
<td>Specific gravity (g/cm³)</td>
<td>1.3</td>
<td>1.3</td>
<td>1.3</td>
</tr>
</tbody>
</table>

* weight in gram per 9000 meter yarn

4.2.3.6 Direct tensile test on PVA yarns

Direct tensile tests on the PVA yarns for determining their mechanical properties were carried out using the same equipment and test arrangement as that for testing glass strand described in Section 4.2.2.3. The PVA sample length used was 300 mm with an "effective" test length of 250 mm used for calculating the tensile strains (see Section 4.2.2.3).

Figures 4.8 a), b) and c) show typical tensile load-strain curves of the three types of PVA yarns. As can be seen in these curves, within a tensile strain of approximately 0.5% the stress-strain curves were linear and we define the modulus within the 0.5% strain as the "initial modulus". Then there appeared a "yield stage" within which the strain increase became much faster. The curves eventually became stiffer again following the yield stage and we define the elastic modulus measured after the yield stage as the
"secondary modulus". Thus, a schematic description of the tensile stress-strain curve of PVA yarn may be expressed as in Figure 4.9. As can be seen in the figure, the curve slope of PVA appeared to increase with increasing tensile strains after the yield stage and was not completely linear as the idealized stress-strain curve shown (Figure 4.9). This tensile behaviour was different from that of polypropylene to which curve stiffness decreases with increasing tensile strains. Table 4.8 contains the measured results of the PVA yarns. In comparing the measured values with that in Table 4.7 from the data sheet, it is obvious that the tensile strength and strain to failure of the PVA yarns in Table 4.8 are lower than the corresponding data provided by the producer. One of the reasons for this may be due to the different test device and method used. The initial elastic moduli of the yarns measured are in good agreement with the moduli given by the producer. For the T-7901 type yarn, its initial modulus and secondary modulus were found quite similar, to be 39 GPa and 37 GPa, respectively.

Table 4.8 Average tensile test results of PVA yarns

<table>
<thead>
<tr>
<th>PVA type</th>
<th>Initial modulus (GPa)</th>
<th>Corre. strain (%)</th>
<th>Secondary modulus (GPa)</th>
<th>Failure strain (%)</th>
<th>Tensile strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>T-5501*</td>
<td>29</td>
<td>0.52</td>
<td>20</td>
<td>5.47</td>
<td>936</td>
</tr>
<tr>
<td></td>
<td>(2.0)</td>
<td>(0.05)</td>
<td>(0.7)</td>
<td>(0.12)</td>
<td>(40)</td>
</tr>
<tr>
<td>T-5516+</td>
<td>34</td>
<td>0.48</td>
<td>24</td>
<td>5.00</td>
<td>1092</td>
</tr>
<tr>
<td></td>
<td>(0.9)</td>
<td>(0.01)</td>
<td>(0.7)</td>
<td>(0.12)</td>
<td>(48)</td>
</tr>
<tr>
<td>T-7901*</td>
<td>39</td>
<td>0.48</td>
<td>37</td>
<td>4.06</td>
<td>1328</td>
</tr>
<tr>
<td></td>
<td>(2.0)</td>
<td>(0.02)</td>
<td>(2.4)</td>
<td>(0.17)</td>
<td>(71)</td>
</tr>
</tbody>
</table>

* average of 10 test results
+ average of 5 test results
( ) standard deviation
Figure 4.8 Tensile stress-strain curves of PVA yarns (1000 filaments per yarn)

a) T-5501,  

b) T-5516
Figure 4.8 Tensile stress-strain curves of PVA yarns (1000 filaments per yarn)
c) T-7901

Figure 4.9 An idealized tensile stress-strain curve of PVA yarns
I: elastic stage, II: yield stage, III: post yield stage
4.2.4 Carbon fibres

4.2.4.1 Carbon fibres

Carbon fibres carry the labels of high cost and high performance properties. The use of carbon fibres in construction materials industry has been limited by its high cost. Generally, there are two type of carbon fibres: PAN (precursor polyacrylonitrile based) carbon fibre and pitch (coal tar pitch or petroleum pitch based) carbon fibre. The elastic modulus of PAN-based carbon fibres can range between 230 to 590 GPa, and the tensile strength between 2400 to 7000 MPa (39), depending on the manufacturing treatments (oxidation, stretching, carbonation and heat treatment) used. The pitch-based carbon fibres have much lower elastic modulus and strength, usually, 30 to 50 GPa and 600 to 900 MPa (40-41), respectively, although their price is lower than that of the PAN-based carbon fibres.

Recently, it is reported by PETOCA Ltd (42) that new types of pitch-based carbon fibres have been developed which would change the conventional image about pitch-based carbon fibres, since the improved physical properties make the material equivalent to PAN-based carbon fibre. Further more, it was said that the price of the improved pitch-based carbon fibres may eventually be similar to that of glass fibres per unit volume (43). Doubtless, the attraction of such fibres to the construction industry is significant.

4.2.4.2 Properties of high performance pitch-based carbon fibres

Two type of continuous pitch-based carbon yarns, HM20-2K-1091 and HM20-2K-1097, were supplied by PETOCA Ltd and a photo of the fibres is shown in Figure 4.10. Table 4.9 contains the main properties of the carbon fibres based on the technical data sheet of the producer (42).

As can be seen from the table, the two types of carbon yarns had same properties and only difference is in the size agent used. The most
Figure 4.10 Continuous pitch-based carbon fibre roving
Table 4.9 Property of HM carbon fibres

<table>
<thead>
<tr>
<th>Property</th>
<th>HM20-2K-1091</th>
<th>HM20-2K-1097</th>
</tr>
</thead>
<tbody>
<tr>
<td>Filament diameter (μm)</td>
<td>10</td>
<td>10</td>
</tr>
<tr>
<td>Filament No per yarn</td>
<td>2000</td>
<td>2000</td>
</tr>
<tr>
<td>Weight per yarn (g/m)</td>
<td>0.31</td>
<td>0.33</td>
</tr>
<tr>
<td>Size agent</td>
<td>Polyether</td>
<td>polyester</td>
</tr>
<tr>
<td>Specific gravity (g/cm³)</td>
<td>1.95</td>
<td>1.95</td>
</tr>
<tr>
<td>Tensile strength (MPa)</td>
<td>2000</td>
<td>2000</td>
</tr>
<tr>
<td>Elastic modulus (GPa)</td>
<td>200</td>
<td>200</td>
</tr>
<tr>
<td>Ultimate strain (%)</td>
<td>1.0</td>
<td>1.0</td>
</tr>
</tbody>
</table>

Impressive values in the table are the order of magnitude of the modulus and the strength of the carbon fibres, 200 GPa and 2000 MPa, respectively.

4.2.4.3 Direct tensile test on the carbon fibre

Direct tensile test on the carbon yarn was carried out using a hydraulic material machine at Broni and the details of the test machine are described in Section 5.2.1.3. Only the type HM20-2K-1091 yarn was tested. The preparation method for the tensile specimens of the carbon yarn was the same as that of the glass strand (Section 4.2.2.3), except that the ends of the sample were glued by epoxy resin (trade name AMIBLOCK, Franchi Vernici spa) on two pieces of cellulose board with thickness about 1 mm, instead of on paper. The same clamps used for fibre-cement samples in the hydraulic machine were used for the tests of carbon fibres, as special clamps for fibres were unavailable for the machine. Compared with the tensile set-up for fibres at Surrey (Figure 4.5) this tensile system at Broni would be
expected to result in greater calculated strains based on the recorded load-displacement curves, as slip of the yarn in the clamps was more likely. Taking this slip effect into account, the whole sample length (250 mm) was used for the strain calculations.

The tensile results of the carbon yarns are given in Table 4.10. Each result was the average of 10 samples. A representative stress-strain curve of the carbon yarn is shown in Figure 4.11.

Table 4.10  Tensile test results on HM20 carbon yarn

<table>
<thead>
<tr>
<th>Tensile strength (MPa)</th>
<th>Elastic modulus (GPa)</th>
<th>Ultimate strain (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1300 (88)</td>
<td>135 (18)</td>
<td>0.94 (0.07)</td>
</tr>
</tbody>
</table>

( ) standard deviation
Figure 4.11 A tensile stress-strain curve of carbon yarn
(2000 filaments per yarn)
5.1 Introduction

In this chapter, the measured tensile stress-strain behaviour of four types of hybrid composites with the fibre combinations listed below are presented:

A) The glass (continuous roving or chopped strand) and polypropylene networks,
B) The glass (continuous roving or chopped strand) and continuous PVA roving,
C) The continuous PVA roving and polypropylene networks,
D) The continuous carbon roving and polypropylene networks.

The major work has been focused on the tensile behaviour of the hybrid composites containing glass and polypropylene networks.

5.2 Glass-polypropylene hybrid composites

5.2.1 Experimental details of composites

The materials and their properties have been reported in detail in Chapter 4. In this section, the experimental programme on the composites is described.

5.2.1.1 Type of composites

Three types of glass-polypropylene hybrid composites were prepared:
a) The chopped glass-polypropylene hybrid composite.
b) The glass roving-polypropylene hybrid composite, and
c) The glass roving-chopped glass-polypropylene hybrid composite.

As control composites, all polypropylene composites and all glass composites were also manufactured. This resulted in a total number of 40 sheets to be designed and manufactured which were classified into seven series. Details of reinforcement and structure of the seven series of composites are given in Table 5.1.

As can be seen in Table 5.1, while Series 5(1) hybrids had the structure of "alternating laminates", that is, the sheet had the structure of superposed layers of cement materials reinforced alternately by the fibrillated polypropylene networks and the chopped glass fibres, Series 5(2) and 5(3) hybrids were designed to have different locations of the glass fibres. The Series 5(2) composite had a sandwich form structure with the chopped glass fibres arranged near the top and the bottom surfaces (glass strand shell) and the central region reinforced with polypropylene networks. For the Series 5(3) composite, the glass fibres were arranged in one side of the neutral axis of the test specimen. The hybrid composites with different dispositions of glass fibres were not expected to result in much difference in tensile properties. However, the locations of the glass would have an effect on their flexural behaviour, since the stress properties in either side of the neutral axis of a flexural specimen differ.

Series 6 and 7 were the hybrids containing continuous glass rovings and the polypropylene networks. A difference between the two series of composites was that series 7 contained a small amount of chopped glass strands which allowed the composite to have a similar constituent structure to the commercial product (Retiver, Fibronit S.R.L.).
Table 5.1 Schematic description of reinforcement positioning in composites

<table>
<thead>
<tr>
<th>Series no.</th>
<th>Reinforcement</th>
<th>Fabrication</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Polypropylene network</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>Chopped glass strand</td>
<td></td>
</tr>
<tr>
<td>3</td>
<td>Continuous glass roving + Chopped strand</td>
<td></td>
</tr>
<tr>
<td>4</td>
<td>Continuous glass roving</td>
<td></td>
</tr>
<tr>
<td>5</td>
<td>Polypropylene network + Chopped glass strand</td>
<td>(1)</td>
</tr>
<tr>
<td></td>
<td></td>
<td>(2)</td>
</tr>
<tr>
<td></td>
<td></td>
<td>(3)</td>
</tr>
<tr>
<td>6</td>
<td>Polypropylene network + Continuous glass roving</td>
<td></td>
</tr>
<tr>
<td>7</td>
<td>Polypropylene network + Continuous glass roving + Chopped glass strand</td>
<td></td>
</tr>
</tbody>
</table>

- Polypropylene network
- Chopped glass
- Continuous glass roving
5.2.1.2 Manufacture of composites

The preparation of the all-polypropylene networks reinforced cement sheets in the laboratory are basically standard hand lay-up procedures and have been described in detail by a few workers in Surrey University (say, Ref. 1). Since majority of the sheets made in the present study contained more than one type of fibres, some differences in fabrication technique were needed, in comparison with one fibre only cement sheets.

All composites used in the study were made in the form of flat sheets of 600x600 mm square with a variable thickness of 5-10 mm, by a hand lay-up method. The lay-up process might vary here and there from one type of sheet to the other but the basic procedure was the same, that is, the sheet was fabricated by placing one layer or sheet of fibres or cement slurry successively on top of the other so that the final product was comprised of a plurality of superposed elementary layers of fibre cement. In this section, the manufacturing procedure is described only for the hybrid composite reinforced with the polypropylene networks, the continuous glass rovings and the chopped glass fibres, that is, the type of Series 7 in Table 5.1, and the manufacturing procedures of other composites can be on the analogy of that.

i) A sheet of polythene film was placed on a smooth wood table to be used as a base, which had been moistened to ensure an intimate contact between the surfaces of the wood and the film.

ii) A pack of fibrillated polypropylene network, with a cut size of 600 mm square, was laid on the polythene film and the mixed cement matrix was then applied onto it and worked by hand until the whole networks were fully impregnated by the matrix. The amount of matrix applied at a time was controlled according to the target volume fractions of fibres, but it was convenient to supply a little more of the matrix for the impregnation of the next layer of fibres especially when they were the rovings or the chopped
iii) After the matrix had been impregnated into the first pack of networks, one by one the continuous glass rovings, orientated in the longitudinal direction and spaced at equal predetermined distances were pressed into the remained cement matrix and then another supply of matrix was placed. Like the polypropylene networks, the impregnation of matrix in the glass rovings was as thorough as possible.

iv) A layer of chopped glass strands was spread as uniformly as possible onto the preceding layer and gently slapped by hand into the remained matrix which was sufficient to absorb fully the applied chopped glass fibres.

v) Another pack of polypropylene networks was pressed against the previous layer, fresh matrix was placed and the second layer of chopped strands then was followed or the step iv) was repeated.

vi) The sheet was completed by impregnating the third pack of polypropylene networks with slurry and a second sheet of polythene film was put on it and trowelled to achieve as uniform thickness over the sheet as possible. It also important to ensure that no air pockets remained trapped in the complete sheet.

vii) The sheet was left in the laboratory overnight and then immersed in water at 20 °C until required for testing, usually at 28 days.

Before testing, the sheet was cut into strips about 300 mm long by 25 mm wide for tensile testing and about 150 mm long by 50 mm wide for flexural testing. Both tensile and flexural specimens were cut in two orthogonal directions from the same sheet.

5.2.1.3 Uniaxial tension testing
Tensile tests were carried out on both longitudinal and orthogonal specimens but only the test results from longitudinal specimens are provided and discussed in the thesis. It should be indicated however that, for commercial cladding and roofing sheets, properties in the orthogonal direction are also important, since in most cases these applications have to resist bidirectionally the applied stresses imposed from edge handling, restrained shrinkage and thermal contraction, etc.

The uniaxial tensile tests were carried out on an Instron 1122 machine equipped with an extensometer and two X-Y-Y recorders shown in Figure 5.1. Prior to placing the specimen in the Instron the width and thickness were measured at three points within the central third of the specimen. The specimen was clamped by the jaws of the Instron with a sheet of lead about 1.5 mm thick and 30 mm wide placed between the sample and the jaws in order to prevent local damage when fixing the specimen.

The strains at each side of the specimen were measured, using two linear variable differential transformers (LVDTs) attached to the clip-on extensometer of 100 mm gauge length (Figure 5.2). The deformation values measured by LVDTs were output in two ways. One was output to an X-Y-Y recorder after electrically averaging the values of two LVDTs and a single load-strain curve was plotted. In order to determine accurately the uncracked composites modulus of elasticity and the cracking strain, a second X-Y-Y recorder was simultaneously used where the strain output from the LVDT on each side of the specimen was recorded up to 0.5% at a larger magnification without being summed and averaged. The load signal was taken from the Instron load cell.

For the composites made with the Broni matrix, their tensile tests at seven days were performed using a hydraulic testing machine equipped with a SE 790 X-Y recorder shown in Figure 5.3. With this testing system, the tensile load and the cross-head displacement can be recorded. One advantage of this testing machine is its clamping system which is capable of fixing the
Figure 5.1  Tensile test set-up of Instron 1122 machine using X-Y-Y recorders
Figure 5.2 Extensometer
Figure 5.3  Tensile set-up of the hydraulic testing machine (Broni)
specimen by "self-tightening" rather than by man-power. Specimen-clamping
by man-power may induce significant strain at the ends of the specimen and
sometimes might cause local failure if care is not performed.

Table 5.2 summarises the test arrangements of the composites made with the
modified Surrey matrix and with the Broni matrix. The differences in testing
age and in the recorded form of the load-deformation curves prevent
comprehensive comparisons between the specimens made with the modified
Surrey matrix and that with the Broni matrix, or between the specimens
tested at one age and that at different ages. However, comparisons made
between the testing results of each series of the specimens or between the
results of different series of specimens (Table 5.1) at a given age tested
on one testing machine are significant, as shown later.

Table 5.2 Test arrangement of composites

<table>
<thead>
<tr>
<th>Composite classification</th>
<th>Test age</th>
<th>Test device</th>
<th>Recorded curve type</th>
</tr>
</thead>
<tbody>
<tr>
<td>Modified Surrey matrix</td>
<td>28 days, 3 years*</td>
<td>1122 Instron machine</td>
<td>Load-strain (%)</td>
</tr>
<tr>
<td>Broni matrix</td>
<td>7 days</td>
<td>Hydraulic machine</td>
<td>Load-extension (mm)</td>
</tr>
<tr>
<td></td>
<td>150 days</td>
<td>1122 Instron machine</td>
<td>Load-strain (%)</td>
</tr>
</tbody>
</table>

* only limited number of specimens were tested.

All the specimens were tested at a constant cross-head displacement speed of
10 mm/min, which was equivalent to a composite strain rate of about 4% per
minute. After testing, the number of cracks over a 100 mm gauge length of the specimen was determined in order to assess the crack spacing. Detection of cracks was made by highlighting them with water on the specimen surface and allowing the surface to dry. Optical microscopes were used for counting the cracks in the case of high fibre volume fractions.

5.2.1.4 Fibre volume fraction determination

The actual fibre volume fractions were determined for each of the specimens either by an acid dissolution technique or by a Known-weight method.

Acid dissolution method

Approximately 50 mm was cut from the test specimen after tensile test for determination of fibre volume fractions by the following method:

(i) Strip weighed in air after oven-drying at 105 °C, \( W_1 \)
(ii) Beaker of water weighed, \( W_2 \)
(iii) Water-saturated strip suspended in beaker of water and combined weight noted, \( W_3 \)
(iv) Strip placed in 50/50 water diluted hydrochloric acid (Assay 35-38). After about ten days the matrix was softened and was washed out of the fibres and then each type of fibres (film, chopped glass or continuous rovings) was dried in an air-ventilated oven at 105 °C and weighed separately to 0.0001 g, \( W_4 \)

The density of the composite \( (D_0) \) was determined by

\[
D_0 = \frac{W_1}{(W_3 - W_2)} \tag{5.1}
\]

and the fibre volume fraction \( (V_f) \) by
\[ V_f = \frac{W_f}{\rho_f (W_a - W_2)} \times 100\% \]  \tag{5.2}

where \( \rho_f \) is the specific gravity of the fibre.

The fibre volume fractions of the samples prepared in the Surrey laboratory were determined by this method.

**Measuring from known-weight**

If weight of a fibre to be put into a unit square sheet is predetermined, and if in the sheet the fibres are able to be uniformly distributed, the fibre volume fractions of the cut strips may then be assessed by

\[ V_f = \frac{W_f}{\rho_f V_c} \times 100\% \]  \tag{5.3}

In this case however, \( W_f \) is the weight of the fibre in the specimen and \( V_c \) is the measured specimen volume.

The fibre volume fractions of the samples prepared in the Fibronit laboratory were determined by this method.

The acid dissolution method could relatively precisely measure the fibre volume fractions in the cut piece and the obtained values were regarded as the fibre volume fractions of the whole tensile specimen. This would be the case if the fibres in the specimen were uniformly distributed.

The known-weight method eliminated the acid dissolution procedure and was advantageous for the immediate results which should be a good evaluation once again when the fibres were uniformly distributed in the whole sheet. In practical specimens the fibre distribution was likely not to be very uniform. Thus, the fibre volume fractions measured by the both methods
5.2.2 Measured tensile stress-strain curves

5.2.2.1 Individual fibre reinforced composites

The typical tensile stress-strain curves for the individual fibre only composites made with the modified Surrey matrix are shown in Figures 5.4 and 5.5. The curves for composites made with the Broni matrix are shown in Figures 5.6 and 5.7. In Figures 5.4 and 5.6, the curves are shown for the specimens tested in the longitudinal direction only and the polypropylene fibre volume aligned in this direction is stated with the additional orthogonal fibre volume in parentheses. In Figure 5.5, the glass fibre volumes are the total two-dimensional random volume, not the effective volume in the stress direction.

As can be seen in Figures 5.4 to 5.7, the shapes of the measured tensile curves of the one fibre only composites are essentially similar to that of the idealised stress-strain curve shown in Figure 2.1 (Section 2.2) predicted by the ACK theory. That is, the tensile stress-strain curve of the one fibre only composites consisted of three increasing-stress stages, the elastic stage, the matrix multiple cracking stage within which the stress-strain curve was either horizontal or slightly ascending depending on the fibre volume fractions, and the post multiple cracking stage within which the curve shape was approximately a linear slope until the strain to failure of the composite (or of the fibre).

Although the general curve shapes (in terms of three stress-increasing ranges) were approximately the same for the polypropylene reinforced composite and for the glass fibre reinforced composites, the stress-strain behaviour between the two types of composites during a loading process can considerably differ. It may be seen from Figures 5.4 to 5.7 that the fibrillated polypropylene network was particularly effective in improving
Figure 5.4 Tensile stress-strain curves of composites reinforced with polypropylene networks only
(orthogonal volume fraction in parenthesis)
Notice the change in X axis scale from Figure 5.4

Figure 5.5 Tensile stress-strain curves of composites reinforced with
a) 2-D random chopped glass strand (24 mm)
b) continuous glass roving and chopped glass strand
Figure 5.6  Tensile stress-elongation curves of individual fibre reinforced composites at 7 days (Broni matrix)

Figure 5.7  Tensile stress-strain curves of individual fibre reinforced composites at 150 days (Broni matrix)
the ductility of the cement based composite which was exhibited by a lengthy region of multiple cracking and a high strain to failure (6-9%, Figure 5.4) on the stress-strain curves. However, the load-carrying capacity of the composite was low immediately after the matrix cracked, since the polypropylene strength was mobilized at large strains.

For the continuous roving glass fibre reinforced cement (curves 2 on Figures 5.6 and 5.7), the glass fibres were more effective in promoting the load-carrying capacity of the composite immediately after the matrix cracked and this was exhibited in the stress-strain curves by a much shorter region of multiple cracking and a stiffer stress-increasing slope following the completion of multiple cracking. The energy absorbing ability (area under the tensile stress-strain curves) of the glass fibre reinforced composite, however, was low since the composites, for the limited glass fibre volumes used, failed at strains of less than 0.5% (Figures 5.5 and 5.7) which was more than ten times smaller than that of the polypropylene reinforced cement. By the combined used of the glass fibres and the fibrillated polypropylene networks in the common matrix, different tensile stress-strain curves resulted showing a favourable balance between the advantages and disadvantages inherent in their parents composite and this "hybrid" behaviour is presented in the following sections.

5.2.2.2 Glass-polypropylene hybrid composites

The typical tensile stress-strain curves for the glass-polypropylene hybrid composites made by the modified Surrey matrix with various volume fractions of the fibres are presented in Figures 5.8 and 5.9. The curves for those composites made from the Broni matrix are shown in Figures 5.10 to 5.12. These tensile stress-strain curves, when compared with those of the parent composites (Figures 5.4 to 5.7), demonstrate how different fibre combinations affected the curve shapes and thus the mechanical properties of the resulting composites.
It should be noted that the apparent differences in shape of the curves in Figures 5.10 and 5.11 are not solely due to the different test ages, but mainly result from the use of different test machines in Broni and Surrey. Thus the X-axis in Figure 5.10 is "elongation" whereas in Figure 5.11 the X-axis is strain.

Figure 5.13 shows schematically the stress-strain curve of the hybrid composite reinforced with the chopped glass strands and the fibrillated polypropylene networks (Figure 5.8). In Figure 5.14 shown schematically is the stress-strain curve of the hybrid composite reinforced with the continuous glass rovings and the polypropylene films (Figures 5.9 to 5.12). Quite obviously, there are good agreements in curve shapes of experiment and theory when we compare that in Figure 5.13 (chopped glass-polypropylene hybrid) with that in Figure 3.4c and that in Figure 5.14 (continuous glass-polypropylene hybrid) with that in Figure 3.4b.

It is notable in Figures 5.9 to 5.12 that there was a sudden reduction after stage 3 (or at point c on Figure 5.14) in the observed load of all the tensile curves. The sudden load reduction was thought to be caused by the fracture of the glass roving which suddenly transferred the additional load onto the polypropylene resulting in its rapid elongation at a faster rate than the cross-head movement. However, it is notable that the extent of load reduction caused by the fracture of glass fibres was reduced when the relative proportion of polypropylene in a hybrid was increased. This might imply that the load reduction in a continuous glass-polypropylene hybrid could become negligible if the relative volume fraction of polypropylene in relevance to that of glass had been sufficiently high but this would only apply for the specific test system used at a given rate of cross-head movement and is thus not a fundamental material parameter.

The deviation of the load-strain curve on the failure of glass fibre was so smooth that there was almost no load reduction on the recorded curves observed for the hybrids reinforced by the chopped glass strand and
Figure 5.8 Tensile stress-strain curves of 2-D chopped glass strand and polypropylene network hybrid composites (Modified Surrey matrix, 28 days)
c) * (curve 2): composite of Series 5(3)

d) Figure 5.8 Tensile stress-strain curves of 2-D chopped glass strand and polypropylene network hybrid composites
(Modified Surrey matrix, 28 days)
Figure 5.9  Tensile stress-strain curves of hybrid composites reinforced with
a) continuous glass roving and polypropylene networks
b) continuous glass roving, 2-D chopped glass strand
   and polypropylene network
(Modified Surrey matrix, 28 days)
Figure 5.10 Tensile stress-elongation curves of continuous glass roving-polypropylene network hybrid composites at 7 days (Broni matrix)

Figure 5.11 Tensile stress-strain curves of continuous glass roving-polypropylene network hybrid composites at 150 days (Broni matrix)
Figure 5.12 A tensile stress-strain curve of the hybrid composite containing 0.63% continuous glass roving and 6.3% polypropylene network (Broni matrix, 28 days)
Figure 5.13 A schematic description of experimental tensile stress-strain curves of chopped glass strand-polypropylene network hybrid composites

Figure 5.14 A schematic description of experimental tensile stress-strain curves of continuous glass roving-polypropylene network hybrid composites
polypropylene networks (Figure 5.8). Actually, the curve shape of the chopped glass-polypropylene hybrids (Figure 5.13) looked to be similar to that predicted in Figure 3.4c assuming equal rate of cross-head movement and stress transfer, and within the region c-d on Figure 5.13, the composites had stresses similar to the first failure stresses at c. The negligible load drop or no load drop in the stress-strain curves of the chopped glass-polypropylene hybrid composites might suggest a pull-out failure mechanism of the chopped glass strand. The chopped glass pull-out rate, compared with the roving glass fracture rate, would be much nearer to the rate of the cross-head movement (10 mm/min).

In Figures 5.13 and 5.14, curve path o-a-b-c-f represented the hybrids in which the volume fractions of polypropylene were insufficient to carry the extra load transferred on it on the failure of the glass fibres (see Figures 5.9 to 5.11). The catastrophic failures shown in Figure 5.11 for the composites tested at 150 days might be also related to the aging effect.

5.2.3 Typical values measured from tensile stress-strain curves

The notations defined below have been assessed for each of the test specimens from the measured tensile stress-strain curves:

$E_c$ - The elastic modulus of the uncracked composite, measured from the initial part of the tensile stress-strain curves up to a strain of 0.5% ($5000 \times 10^{-6}$) at a larger magnification plotted in one of the X-Y-Y recorders.

$\sigma_c$ - The first cracking stress (or the composite cracking stress), defined at the point where there was a marked deviation from linearity in the elastic region of a tensile stress-strain curve.

$\varepsilon_{cu}$ - The first cracking strain (or the matrix failure strain), calculated
according to equation 5.4 with experimentally measured values of $\sigma_c$ and $E_c$.

$\varepsilon_{c,m}$ - The strain at the termination of multiple fracture (or multiple cracking) of the matrix, defined at the point where the tangent lines of multiple fracture region and the post-multiple fracture region intersects.

$E_i$ - The elastic modulus of cracked composite in the range of post multiple fracture of matrix.

$\sigma_i$ - The stress of the hybrid composite at which the stress-strain curve started to deviate or drop due to the failure of glass fibres, referred to "first failure stress" in the present study.

$\varepsilon_i$ - The strain of the hybrid composite at which the stress-strain curve started to deviate or drop due to the failure of glass fibres, referred to "first failure strain" in the present study.

$\sigma_{uu}$ - The ultimate tensile strength of the composite.

$\varepsilon_{cu}$ - The ultimate tensile strain of the composite.

In addition, the length of the second relatively flat part in the stress-strain curves of hybrid composites (this is effectively the section c-d in Figure 5.13), was also assessed, using the same "tangent line" method used for $\varepsilon_{c,m}$ in order to evaluate factors which could affect the length.

The crack width ($w$) at the termination of matrix multiple fracture was approximated by multiplying the measured crack spacing ($C$) by the strain at the end of the multiple fracture ($\varepsilon_{c,m}$).

Where the composites were tested in the hydraulic testing machine without an
extensometer, the deformation under tension was assessed by "extension or elongation (mm)" instead of "strain (%)".

In practice, an accurate determination of some of the above defined values by visual examination of the measured tensile stress-strain curves was not always easy. The difficulties were mainly because the shape of stress-strain curves obtained empirically did not strictly correspond to that predicted by the ACK theory (2) or to that predicted by the current theory (Figure 3.4).

For example, the multiple fracture region of a practical tensile stress-strain curve is often slightly ascending instead of completely horizontal, and the transition from the multiple fracture zone to the post-multiple fracture zone is often a gradually curve-rising process rather than at an obvious salient point. The latter phenomenon was especially true for the polypropylene only composite, resulting from the non-linear elastic behaviour of the polymer fibre. Thus, the values of $\varepsilon_{c.m}$ determined using the "tangent-line intersection" method in this study were affected by personal decisions and should be regarded as an approximation of the parameter rather than an accurate one. The slope of the multiple fracture range varies according to experimental factors such as the fibre volume fractions and the matrix failure strength. Usually, the higher the matrix strength at a given fibre volume, or the lower the fibre volume for a same matrix, would result in a flatter multiple-fracture region.

Whether or not further cracking takes place beyond the matrix multiple cracking region is related not only to the uniformity of the matrix strength but also to the fibre volume fractions. Some cracking was often observed in the post-multiple cracking region of tensile curves of the specimens with low fibre volume fractions. In such cases, the completion of the multiple cracking and thus the strain at this point was still judged in this study by the tangent-line intersection method rather than by the real completion of cracking of the specimen.
After the completion of matrix multiple cracking, the following slope in the stress-strain curves is governed by the elastic modulus of fibres. Ohno (3) determined the modulus of the polypropylene network from the stress-strain curves at a constant composite strain of 5%. In this study, the elastic modulus of polypropylene was determined from a range of stress-strain curves of the composite at a strain range between 5% and 7%. The elastic modulus of the cracked hybrid composite, $E_t$, was determined from the stiffest part of the curve.

The failure strain of the glass in the hybrid was determined to be the point where the load-strain curve started to level off or drop due to the breakage of the glass fibres. In the present work the strain at this yield point is called the first failure strain and the corresponding stress as the first failure stress of the hybrid composite.

Typical data obtained from the tensile stress-strain curves are presented in Tables 5.3 and 5.4 for the composites made with the modified Surrey matrix and in Table 5.5 for the composites made with the Broni matrix. As can be seen in these tables, the two sets of composites had different deformation expressions, either by strain ($\varepsilon$) or by extension (mm), since they were tested in different testing machines. Thus, comparisons in strain properties in the following sections are, in general, made between the composites made from the same matrix using the same test device. Some workers estimate the tensile strain from the crosshead movement or the recorded displacement (mm) of the specimen and the calculated values are likely to be higher or much higher than the actual strains due to the possible slip induced between the specimen and the grip.

In Tables 5.3 to 5.5, generally, each data point was the average of at least four test results.
5.2.4 Discussion of results

5.2.4.1 Composite cracking stress

Bearing in mind the tensile strength level of the unreinforced matrices in this study, i.e. about 3.5 MPa (Section 4.1.4), it is evident from Tables 5.3 to 5.5 that the presence of fibres, especially the combination use of the two fibres can increase the composite cracking stress ($\sigma_c$) and the composite cracking stresses tend to increase with the fibre volume fractions. Basically, the matrix strength ($\sigma_{mu}$) does not change, but the cracking resistance of the composite is enhanced because of fibre reinforcement.

According to the Law of Mixtures, the main effect of the fibre addition is on the composite modulus ($E_c$). Because the modulus of the glass fibres (about 70 GPa, Section 4.2.2) is higher than that of the cement matrices (about 25-28 GPa, Section 4.1.4), the elastic modulus of the all-glass reinforced composites would theoretically be a small amount greater than the unreinforced matrix. On the other hand, the elastic modulus of the all polypropylene cements would theoretically be decreased since the elastic modulus of the polypropylene (say, 4-6 GPa) is much smaller than that of the matrix.

Improvements in modulus and cracking stress of the all glass fibre cements over that of the unreinforced matrix were not obvious in the present study, presumably, due to the small volume fractions of the glass fibres used. The moduli of the all polypropylene composites were generally smaller than that of the unreinforced matrix and that of the all glass composites. However, the composite cracking stresses were quite similar between the two types of composites. According to one equation (Section 2.3.2)

$$\sigma_c = E_c \sigma_{mu} \quad (5.4)$$
Table 5.3 Typical tensile results of composites reinforced with individual fibres (Modified Surrey matrix, 28 days results)

<table>
<thead>
<tr>
<th>Sheet series &amp; number</th>
<th>Volume fraction (%)</th>
<th>$E_c$ (GPa)</th>
<th>$\sigma_c$ (MPa)</th>
<th>$\sigma_{cu}$ (MPa)</th>
<th>$\epsilon_{cu}$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>PP net</td>
<td>Chopped glass</td>
<td>Cont. glass</td>
<td></td>
<td></td>
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<tr>
<td><strong>Series 1</strong></td>
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<td></td>
<td></td>
<td></td>
<td></td>
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<tr>
<td>No. 1</td>
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<td>0</td>
<td>20</td>
<td>4.3</td>
</tr>
<tr>
<td>No. 2</td>
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<td>0</td>
<td>0</td>
<td>22</td>
<td>4.4</td>
</tr>
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<td>No. 3</td>
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</tr>
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<tr>
<td><strong>Series 2</strong></td>
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<td>0.8</td>
<td>0.35</td>
<td>28</td>
<td>5.3</td>
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</table>

the increase in composite cracking stress of the all polypropylene composites (No.5 and No.6 composites, Table 5.3) over that of the unreinforced cement matrix and that of the all glass composites could be explained by an increase in the matrix failure strain ($\epsilon_{uu}$) of the composites in the presence of the polypropylene networks. However, this is only theoretically possible using a fracture mechanics approach and is not predicted by the Laws of Mixtures.

Figures 5.15 and 5.16 show the effectiveness on the enhancement of the matrix failure strain of the composites by incorporating fibres for the individual fibre reinforced composites and for the glass-polypropylene hybrid composites (28 days results), respectively. In the figures, the results of the individual test specimens were plotted and the fibre volume for the all glass strand reinforced composite was the total volume of the glass strands in random two dimensions and for the all polypropylene...
Table 5.4 Typical tensile results of glass-polypropylene hybrid composites
(Modified Surrey matrix, 28 days results)

<table>
<thead>
<tr>
<th>Sheet series &amp; number</th>
<th>Volume fraction (%)</th>
<th>E&lt;sub&gt;s&lt;/sub&gt; (GPa)</th>
<th>σ&lt;sub&gt;c&lt;/sub&gt; (MPa)</th>
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<th>ε&lt;sub&gt;1&lt;/sub&gt; (%)</th>
<th>σ&lt;sub&gt;cu&lt;/sub&gt; (MPa)</th>
<th>ε&lt;sub&gt;cu&lt;/sub&gt; (%)</th>
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- specimen failed at ε<sub>1</sub>
Table 5.5 Typical tensile results of composites made with Broni matrix

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<th>$E_c$ (GPa)</th>
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<th>$\sigma_1$ (MPa)</th>
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<th>$\sigma_{cu}$ (MPa)</th>
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- specimen failed at $\epsilon_1$

No. 6+ average of two test results
Figure 5.15 Effect of fibres on the enhancement of matrix failure strain in individual fibre reinforced cement composites

Figure 5.16 Effect of fibres on the enhancement of matrix strain in glass-polypropylene hybrid composites
(Fibre volume fraction shown in the figure was the sum of that of individual fibres. PP: polypropylene network, S: chopped glass strand, R: continuous glass roving)
reinforced composite was that in the stress direction only. The fibre volume shown for the hybrid composites in Figure 5.16 was the sum of that of the individual fibres, that is, the polypropylene and glass in the specimen, without considering the efficiency factor (\(\eta\)) of glass strands.

A general trend in Figures 5.15 and 5.16 is obvious: the matrix failure strain was enhanced with fibre volume fractions. The enhanced matrix failure strain may be explained by the "energy-release" models (4-6) which assumed that the Griffith-type flaws are inherent in any brittle matrix and the Griffith energy criterion for failure is to control the failure strain of the matrix. The models indicate that the presence of closely spaced fibres are capable of interfering with the propagation of microcracks that initiate from the inherent flaws in the material, by altering the strain distribution and decreasing the rate of energy release in the system necessary to catastrophically open the Griffith flaw. This delays the formation of an unstable crack system and, thus, increases the matrix failure strain of the composite.

In comparison with the individual fibre reinforced composites, it has been noted in the study that a slight increase in composite cracking stress of the hybrids over that of their parent composites resulted, as can be seen in Tables 5.3 to 5.5. For example, the composite cracking stress was about 6 MPa for an all polypropylene composite with a fibre volume of 4% while the value for a hybrid with volume fractions of 2.3% polypropylene and 1.9% chopped glass was about 7 MPa. Further, the composite cracking stress for the hybrid with volume fractions of 0.8% polypropylene and 1.4% continuous glass was 6.8 MPa at the age of 7 days, to be compared with about 6 MPa for an all polypropylene composite with \(V_f\) of 3.8%, or 6.1 MPa for an all continuous glass composite with \(V_f\) of 1.32% (Table 5.5). Of course, these comparisons may not be very fair, since the fibre volume fractions in these composites were not completely the same. Further, such improvement might not have been achieved for the hybrids when compared with the all glass composites, if the glass fibre (especially the continuous glass) content in
the glass composite had been sufficiently high. Nevertheless, results in the study appeared to show that the combined use of two fibres in a composite would have greater control of the matrix cracking property, especially of the composite cracking stress (Tables 5.3 to 5.5). One explanation for this is that the combined use of the densely spaced fibrillated polypropylene networks and the glass fibres was more effective in arresting or suppressing the unstable propagation of microcracks in the hybrids than a single fibre type and therefore made more efficient utilization of the stiffer glass fibres in terms of cracking stress in the pre-cracking stage.

When comparing the Series 7 and Series 5 hybrids, shown in Table 5.4, it can be seen the Series 7 hybrids resulted in higher composite cracking stress than the chopped strand hybrids presumably due to the inclusion of the continuous glass roving in Series 7 which behave, theoretically, with an efficiency of 1.0.

In Table 5.5, the hybrids No.3 to No.6 in Series 6 were tested at 7 days and 150 days respectively. The higher composite cracking stress exhibited by the 150 days specimens were attributed to the greater hydration of the cement matrix.

5.2.4.2 Multiple fracture of matrix in individual fibre reinforced composites

The process of multiple cracking is important, not only because it contributes to the composite a quasi-ductile behaviour but also because the crack spacing controls the crack width at this stage which has a considerable influence on the serviceability of the material.

It can be seen from Figures 5.4 and 5.7 that the length of the multiple cracking region, represented by the nearly horizontal jagged part of the curve, is much larger for the polypropylene reinforced cement than the glass fibre reinforced cement, due to the much lower stiffness and greater
extensibility of the polypropylene fibre. The strains at the completion of matrix multiple fracture were observed to be in the range of 2.0% to 4.8% corresponding to the polypropylene volume fractions in the all polypropylene composites of 4.0% to 1.5%. This large multiple cracking strain was beneficial to toughness but non-favourable in utilising effectively the load-carrying capacity of the fibre at low deformations after cracking.

Due to the short strand length and the limited volume level of the chopped glass used, stress-strain curves with quite poor performance for the chopped glass strand reinforced composites have been obtained. As can be seen in Figure 5.5a, below a total fibre volume 3% of the chopped glass, multiple fracture of the matrix, in a poorly controlled form, continued until the failure of the composite. This poor result may give strong support to the benefits of composites manufactured using the hybrid concept, since the cracking or stress-strain performance was improved dramatically in the hybrids using similar volume fractions of the glass fibres. However, comparisons of the strain at the completion of multiple cracking for hybrid and non-hybrid composites were difficult at these small failure strains. The multiple cracking performance of the specimen reinforced with chopped glass and one continuous glass roving (Figure 5.5b) was similar to that of the chopped glass reinforced specimens due to the fibre volume being about the critical fibre volume in tension.

Comparative results of multiple cracking were obtained by the continuous glass roving reinforced composite which had a fibre volume about 1.4%. For the specimens tested at 7 days, the length of the multiple cracking region for the all glass composite was about 1.2 mm to be compared with 4.7 mm of that for the all polypropylene composite with a fibre volume of 3.8%. For the specimens tested at 150 days, the strain at the completion of multiple cracking for the all glass composite was about 0.33% while that of the all polypropylene composite with 3.8% fibre content was about 2.4% (see Figures 5.6 and 5.7).
5.2.4.3 Multiple fracture of matrix in glass-polypropylene hybrid composites

As can be seen from Figures 5.4 to 5.12, through the combined use of the polypropylene networks and the glass fibres in the common matrices, the hybrid composites exhibit noticeably different multiple fracture behaviour under tensile loading in comparison with the all polypropylene composites and the all glass composites.

Firstly, the most obvious effect observed due to the addition of the stiffer glass fibres was the much reduced length of multiple cracking in the stress-strain curves of the hybrid composites, compared with that of the all polypropylene composites. The estimated values for the strain at the completion of multiple cracking plotted against the fibre volume fraction of polypropylene are shown in Figures 5.17 (modified Surrey matrix) and 5.18 (Broni matrix) for individual specimens of the composites used in the study. The volume fractions of the glass fibres in the hybrids were not plotted in the figures but were indicated in the subtitles of the figures. As can be seen in Figure 5.17, the strains of at completion of multiple fracture were in the range of 0.4% to 0.75% for all the hybrid composites compared with the range of 2% to 4.8% for the all polypropylene composites. For example, the strain at the termination of multiple cracking was about 2% for an all polypropylene composite with a fibre volume of 4% whereas the value was about 0.45% for the hybrid with 3.7% polypropylene and 2.0% of random 2-D glass strands. All polypropylene composites with fibre volumes over 4% were not manufactured in the present study but Ohno’s experimental results (3) had demonstrated that, for achieving a strain at completion of multiple cracking of 0.5%, a polypropylene fibre volume of 14% at least had to be incorporated.

Figure 5.17 also shows that the strain at the end of matrix multiple fracture could be further reduced if the glass fibres in the hybrids were used in a continuous form. In other words, similar strain at the completion
of multiple cracking could be achieved by using the continuous glass in the hybrid with smaller volume fractions than that of the chopped glass. For instance, the strain at the completion of multiple cracking was about 0.6% for a hybrid (No.8, Table 5.4) containing volume fractions of 2% chopped glass and of 3% polypropylene whereas the same strain value could be achieved for a hybrid which contained about 0.43% by volume of the continuous glass roving, 0.7% of the chopped strands and a same volume of the polypropylene net. The strain at the end of multiple cracking could be reduced to about 0.35% when the volume fraction of the continuous glass was over 1% in the hybrid which contained about 3% by volume of the polypropylene (Figure 5.18).

From Figures 5.17 and 5.18, one may note that the strain level at the completion of matrix multiple cracking in the hybrid was not very sensitive to the volume fraction of polypropylene but controlled by the glass fibres. Thus the reduction of the strain at the completion of matrix multiple cracking in the hybrid by increasing the volume of polypropylene may be regarded as less significant, in comparison with the effect of glass addition. This result was not surprising, since polypropylene had much a lower elastic modulus than the glass fibres. However, experimental results in the study did show that the strain at the end of multiple cracking decreased with the increased polypropylene volumes at a given content of glass. This effect is demonstrated in Figure 5.19 with an enlarged Y-axis scale in strain using the strain data of four hybrid composites in Figure 5.17. From Figure 5.19, one may expect that the effect of polypropylene volume fractions on the strain at completion of matrix multiple cracking could become significant if the inclusion of the fibre is sufficiently high. This is in agreement with equation 3.13 in Chapter 3 which predicts reduced strain at the end of matrix multiple fracture in a hybrid with the increasing volume fractions of the low elastic modulus fibres. Detailed comparisons between the experiment and the theory are made in Chapter 6.

Additional results in support of those shown in Figure 5.19 were also
Figure 5.17 Strain at termination of matrix multiple cracking in composites. Volume fraction of glass in "PP+S": 1.8% ~ 3.3%; in "PP+R": 0.4% ~ 0.8%; in "PP+S+R": 0.4% ~ 0.6% roving, 0.6% ~ 0.8% 2-D chopped strand. (Modified Surrey matrix, 28 days) S denotes chopped strand, R denotes continuous roving.

Figure 5.18 Strain at termination of matrix multiple cracking in composites. Volume fraction of glass in 28-day "PP+R": 0.6% ~ 0.7%, in 150-day "PP+R": 1.3% ~ 1.8%. (Broni matrix)
Figure 5.19 Effect of polypropylene content on the strain at termination of matrix multiple cracking in glass-polypropylene hybrid composites (28 days)
obtained by the hybrids containing continuous glass fibre rovings at greater
volume fractions. For the specimens tested at 7 days (Table 5.5) with
similar volume fractions of glass rovings, the average (5 specimens) length
of the multiple cracking region in the load-extension curves of the all
glass composite was 1.2 mm while the multiple cracking extensions for the
hybrids containing 0.85%, 1.5%, 2.7% and 4.2% fibre volume fractions of
polypropylene were 1.1 mm, 0.91 mm, 0.8 mm and 0.63 mm, respectively.
However, at 150 days the effect of polypropylene fibre volume was not so
noticeable, i.e. for the specimens tested at 150 days, the strain at the
completion of multiple cracking was about 0.33% for the all glass composite
while for the hybrids containing 0.85%, 2%, 2.7% and 4.1% fibre volume
fractions of polypropylene the strains at the completion of multiple
cracking were 0.36%, 0.37%, 0.36% and 0.25%, respectively. The reasons for
this are not clear.

It may be seen in some of the stress-strain curves in Figures 5.4 and 5.8
that even a small addition of glass fibres can affect the matrix multiple
cracking behaviour (i.e. curve length, shape and jagged extent) of a
polypropylene based cement. Characteristic cracking of the matrix,
represented by the jagged part of curve 4 in Figure 5.4, could continue to a
strain about 3% for the polypropylene composite with 3.4% fibre content. By
an inclusion of 0.8% by volume of the chopped glass strands in a
polypropylene composite with 3.6% polypropylene volume fraction (curve 1 in
Figure 5.8c), real termination of matrix cracking could be reached at a
strain within 2% and the strain at the end of multiple cracking determined
by the tangent-line method was within 1%. The curve shape was also changed
with a much stiffer increasing-stress curve for the hybrid instead of a
nearly horizontal one for the polypropylene composite in the multiple
cracking region. However, it is obvious in the curves in Figures 5.8 of
hybrid composites with a small addition of chopped glass (less than 1%) that
cracking in the hybrids has continued well beyond the multiple cracking
region determined by the tangent-line method, indicating that the
inclusions of the chopped glass with volume fractions below 1% was too small
to effectively control the multiple cracking behaviour of the composite.

The phenomenon of poorly controlled matrix multiple cracking was also observed in those hybrids which contained approximately equal volumes of polypropylene and chopped glass with a total fibre volume in the hybrid of about 4% (curves 2 and 3, Figure 5.8a). However, the phenomenon was not observed in a hybrid containing equal volumes of polypropylene and chopped glass with a total fibre volume in the hybrid about 6% (curve 4, Figure 5.8b). These results may suggest that, in glass-polypropylene hybrids, to achieve the desired properties through the interaction of the two companion fibres, both the ratio and the total volume of the fibres in the hybrid have to be considered. The importance of this point is further evidenced in the following sections.

The strain at the end of matrix multiple fracture in the stress-strain curves of the three sandwich hybrids (with chopped glass shell) was found to be larger than that of the interlaminated hybrids (with polypropylene shell) of similar fibre volume fractions, and the post multiple cracking stiffness also appeared smaller for the sandwich hybrids, presumably, due to the local crumbling of the glass-cement phase which had been observed during the tensile tests of the sandwich composites.

Figure 5.20 shows the changes of crack spacing (C) of types of composites used in this study. It is apparent in Figure 5.20a that the values of crack spacing of the glass-polypropylene hybrids were smaller than that for the polypropylene fibre only cements. According to the ACK theory (2), no further cracking will take place beyond the multiple cracking region. If the matrix strain is neglected, the crack width (w) at the end of multiple cracking (strain εₘₙ) may be approximated by multiplying the crack spacing (C) by the strain at the end of multiple cracking (εₘₙ). The crack widths, corresponding to the crack spacings in Figures 5.20a and 5.20b, are plotted in Figures 5.21a and 5.21b, respectively. As shown in Figure 5.21a, the crack width was 46 μm for a polypropylene cement with 4% fibre content,
Figure 5.20 Crack spacing of fibre-cement composites.

a) Modified Surrey matrix, 28 days

b) Broni matrix, volume fraction of glass roving in hybrids:
   7 & 150 days: 1.3% ~ 1.8%; 28 days: 0.6% ~ 0.7%

For the glass roving only specimens, volume fractions of the roving were plotted on the horizontal axis.
Figure 5.21 Crack width of fibre-cement composites at termination of matrix multiple cracking.

a) Modified Surrey matrix, 28 days

b) Broni matrix, volume fraction of glass roving in hybrids:
   150 days: 1.4% ~ 1.8%; 28 days: 0.6% ~ 0.7%.

For the glass roving only specimens, volume fractions of the roving were plotted on the horizontal axis.
whereas the value was 6 μm for the hybrid containing 3.7% polypropylene and 2.1% chopped glass. In Figure 5.21b, the crack width was 77 μm for the polypropylene cement with 3.8% fibre content and 37 μm for the continuous glass roving reinforced cement with 1.4% fibre volume. However, the value was 3.5 μm for a hybrid containing 4.1% polypropylene and 1.4% continuous glass (150-day results) and the crack width was 2.4 μm for the hybrid containing 6.3% polypropylene and 0.63% continuous glass (Figure 5.21b, 28-day result). The effect of controlling and stabilising cracking by means of the combined reinforcing of glass and polypropylene was significant and could be important for practical roofing applications.

An explanation is required as to why the hybrid composites, with sufficient volume fraction of polypropylene, did not appear to continue cracking beyond the matrix multiple cracking region or within the range 3 (curve b-c, on Figure 5.13 or 5.14). The length of multiple cracking region in the stress-strain curve of one fibre only composite is given by equation 5.5 according to the ACK theory (2, 7)

\[ \Delta \varepsilon_{c-m} = \frac{1}{2} \cdot \frac{E_m V_m}{E_f V_f} \varepsilon_{mu} \]  

(5.5)

Thus, one might expect a short multiple cracking length \( \Delta \varepsilon_{c-m} \) for the glass fibre-cement component where \( E_f \) is large followed by a larger \( \Delta \varepsilon_{c-m} \) for the polypropylene-cement component where \( E_f \) is small. This second \( \Delta \varepsilon_{c-m} \) did not appear to occur from the stress-strain plots. Thus, either the final crack spacing has already been achieved by interaction of both fibres at a low strain and no further stress is transferred between the polypropylene and the matrix at higher strains or there is minor cracking of the matrix still occurring at the higher strains which is not visible on the X-Y-Y recorder. The visibility of cracking on the chart plot is dependent on the interaction between straining rate, Instron stiffness and crack opening rate. For polypropylene composites, the cracks open more rapidly than the straining rate and hence a reduction in load in the stress-strain plots is observed.
If the crack opening is able to be slowed, it might become less than the straining rate and no load reduction (or no matrix cracking) would be visible on the chart recorder.

In comparison with those tested at 7 days (Figure 5.20b), the composites except the polypropylene cement tested at 150 days age showed more or less reductions in crack spacing. According to equation 3.10 (8), the crack spacing \( C \) is mainly affected by the average bond strength \( \tau \) and the specific surface area \( P_f / A_f \), assuming that bond is frictional in character between the fibre and cement in a one fibre only composite. Thus, the reductions in crack spacing of the all glass reinforced composite and the glass-polypropylene hybrids suggest an increase of the average bond strength and the bond area between the glass fibres and the matrix, and the little changed crack spacing of the all polypropylene reinforced composite may imply the constancy of \( \tau \) value and \( P_f / A_f \) value in the fibrillated polypropylene film reinforced cement. This is in agreement with previous work on individual polypropylene or glass reinforced composites (8-11). The introduction of the polypropylene film will be particularly beneficial to retain composite ductility or toughness when the glass-cement phase in the hybrid becomes more brittle. Also, the interaction between the glass fibres and the polypropylene film is particularly effective in reducing the crack width in the multiple cracking stage and further results in a much greater load-carrying capacity in the post multiple cracking region, in comparison with the all polypropylene composite.

5.2.4.4 Post multiple fracture stiffness in single fibre reinforced composites

After the matrix multiple cracking stage, the fibres alone take any further increase in load and the post multiple cracking slope \( (E_f V_f) \) of the composite is determined by the fibre modulus and the volume fraction. In
other words, at a given $V_r$, the degree of this slope will depend on the fibre modulus $E_f$, or $E_{f2}$, the secondary elastic modulus in the cases of polymer fibres.

It is known that the elastic modulus of polypropylene decreases with increased tensile strain. Thus the fibre modulus values are very much dependent on the composite strain or the strain range at which the fibre modulus is measured. For the all polypropylene reinforced cements with 2-4% fibre volume fractions, the steepest slopes of post multiple cracking were observed to usually occur at strains between 3-4% in the stress-strain curves, i.e. shortly after the completion of multiple cracking, and the elastic modulus of the polypropylene ($E_{f2}$) measured within this strain range were found between 3 to 4 GPa, slightly higher than that of Ohno’s (3) about 2.3 GPa. In his study, Ohno used a slightly stronger matrix and the fibrillated film modulus was determined at a given composite strain of 5%. It is recognised that the values of $E_{f2}$ obtained from tensile test of the composite are much lower than the values obtained from tensile tests of unfibrillated film alone. Ohno (3) has reported an initial film modulus about 12 GPa corresponding to a 0.5% strain, and a secondary film modulus about 8 GPa at an equivalent strain of 4% for the unfibrillated polypropylene film. $E_{f2}$ values of the fibrillated film obtained from the all-polypropylene reinforced composites indicates only about 40 to 50% of the $E_{f2}$ value of the unfibrillated film at 4% strain.

The reduction in elastic modulus of the fibrillated polypropylene film may be caused by two effects, i.e. the reduction of fibre alignment due to fibrillation and the reduction of the effective fibre volume caused by the damage of some of the fine fibrils or hairs held in the matrix near the edge of the main fibres during the tensile process (3). Improved bond between the matrix and the hydrophobic polypropylene film is attributed to the brilliant idea of fibrillation which has changed the poor image of the bond of the polypropylene to the hydrated cement and made the industrial production of the polypropylene network reinforced cements become a reality. The
reduction of the film modulus after fibrillation is however the unfortunate by-product which makes the elastic modulus of the film even lower.

The effective $E_r$ values of the chopped glass fibres were unable to be measured from the all glass composites because of the low fibre volume fractions used. Two all-glass specimens containing about 1.3-1.5% by volume of the continuous glass roving were tested at 150 days giving effective $E_r$ values of 63 GPa and 66 GPa, which were slightly smaller than the modulus values of the AR-glass fibres tested in the form of continuous long strands alone of 70 GPa measured in the present study and of 70 to 76 GPa given by the glass manufacturers and reported in the literature (12-13). The reduction in elastic modulus of glass fibres in the two 150-day specimens might be related to the mechanism of alkali attack on the surface of the glass fibres which damages the Si-O-Si bonds in the glass networks. If alkali attack on the glass fibres more or less always takes place with aging, the elastic modulus of the glass fibres measured from glass fibre reinforced cements would be inevitably more or less lower than the directly measured values using glass fibre strands only.

In the present study, what we are more interested in is the post multiple cracking performance of the hybrid composites. It is most evident from this study that, by incorporating glass fibres, the post multiple cracking stiffness or the load carrying capacity within, say, 1.5% strain of the polypropylene based composite has been dramatically improved (Figures 5.4 to 5.12) since the elastic modulus of the glass fibres is much higher than that of the polypropylene networks. However, an additional question is, whether the incorporation of the low modulus polypropylene fibre into an all glass fibre composite would increase or decrease the stiffness of the post-multiple cracking region. This point is examined in the next section.

5.2.4.5 Load sharing in glass-polypropylene hybrids after multiple fracture of matrix
It has been shown in a previous section that, by combining reinforcement of the polypropylene networks and glass fibres, the resulting composites exhibit much reduced strain at the completion of multiple cracking compared with the all polypropylene reinforced cement. The polypropylene fibres, on the other hand, also have a small additional effect to reduce the strain at completion of multiple cracking, compared with the glass alone component. This would imply that the additional load transferred from the matrix at a crack has been somewhat shared by the two fibres in the region. Thus, one might predict that, in the post multiple cracking region, i.e. in the stress-strain curve range b-c (stage 3) on Figure 5.13 or 5.14, the dominant role would be provided by the glass fibres but the total applied load would be shared by the two fibres. Experimental composite stiffness values (\(E_f V_f\)) are shown in Figures 5.22 and 5.23 for the chopped glass-polypropylene hybrids and the continuous glass-polypropylene hybrids, respectively, compared with the theoretical prediction lines of \(\eta E_f V_{f,\text{L}}\) assuming that the glass fibres were wholly responsible for the composite stiffness. In the theoretical calculations, the elastic modulus (\(E_{f,\text{L}}\)) of glass was assumed to be 76 GPa which would give the maximum possible contribution to the composite modulus by the glass fibres and therefore lowest contribution if any, from the polypropylene fibres, when compared with the experimental values. The efficiency factor (\(\eta\)) of the continuous glass roving (Figure 5.23) is assumed to be 1.0. However, the efficiency factor of the 2-D chopped strand (Figure 5.22) with a cut length of 24 mm with respect to post-multiple cracking stiffness is not known, but Oakley and Proctor (12) have published factors between 0.16 and 0.26 for 38 mm long glass strands in spray-suction composites. In the present study, the glass fibre will be less efficient because of its shorter length (24 mm) and the weaker cement matrix used. Theoretical prediction lines for 2-D glass strand using efficiencies of 0.15, 0.20 and 0.26 are plotted in Figure 5.22. It can be seen that the experimental \(E_f V_f\) values of the chopped glass-polypropylene hybrids lies close to the theoretical line calculated using a factor of 0.26. Bearing in mind the shorter fibre length (24 mm) and the high value of glass modulus (76 GPa) assumed for the theoretical calculations, it is reasonable to say
that the true experimental $E_{f,L}V_{f,L}$ values representing the chopped glass fibres alone would be displaced towards the curves for $\eta=0.2$ or $\eta=0.15$ while the additional stiffness contribution from the polypropylene networks has increased the $E_fV_f$ slope in the hybrid composites although this stiffness is controlled mainly by the glass fibres.

This conclusion that the polypropylene fibre in the chopped glass-polypropylene hybrid composites provided a stiffness contribution in stage 3 (curve b-c on Figures 5.13 and 5.14) was further confirmed by the results obtained from the hybrids containing continuous glass rovings and polypropylene networks (Series 7 in Table 5.4 and 28 days results of Series 6 in Table 5.5), which are shown in Figure 5.23. In plotting the volume fractions of continuous glass in the horizontal axis in Figure 5.23, an efficiency 0.26 is assumed for the small amount of 24 mm long chopped glass strand (about 0.7%) incorporated in the hybrids which meanwhile contained 0.4%-0.6% continuous glass rovings and 2.8%-3.0% polypropylene networks. An elastic modulus 76 GPa again is used for the glass fibres to calculate the theoretical line. These parameters ($\eta = 0.26, E_{f,L} = 76$ GPa) would allow minimum stiffness contribution from the polypropylene networks when comparing the theoretical line with the experimental values. It is obvious in Figure 5.23 that the experimental stiffness values ($E_fV_f$) obtained from the stress-strain curves of the two groups of specimens are almost all above the theoretical prediction line ($E_{f,L}V_{f,L}$). For the hybrid containing 0.63% volume fraction (average value) of the continuous glass rovings and 6.3% of the polypropylene networks, the measured $E_fV_f$ values of the hybrid composite were about one and half times of the theoretically predicted $E_{f,L}V_{f,L}$ values using a glass modulus 76 GPa. The polypropylene fibres must therefore provide a stiffness ($E_{f,H}V_{f,H}$) contribution (about 34%, in the case of the hybrid with 6.3% by volume of the polypropylene) in the post matrix multiple cracking stage in the hybrid to explain the difference between experiment and prediction.

Figure 5.24 shows the measured $E_fV_f$ values from the hybrids containing
Figure 5.22 Comparison between experimental $E_fV_f$ values and a theoretical prediction, $\eta E_{fL}V_{fL}$, assuming that only glass fibres contribute to the composite stiffness at stage 3 (b-c, Figure 5.13) of the chopped glass-polypropylene composites. Volume fractions of polypropylene which were not plotted in the figure range from 3% ~ 3.8%. (28 days)

Figure 5.23 Comparison between experimental $E_fV_f$ values measured from continuous glass roving-polypropylene hybrid composites and a theoretical prediction, $E_{fL}V_{fL}$. (28 days)
Figure 5.24 Comparison between experimental $E_f V_f$ values measured from continuous glass roving-polypropylene hybrid composites at 150 days and a theoretical line of $E_{f,L} V_{f,L}$. 

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Theoretical, $E_{f,L} = 76$ GPa.
polypropylene and continuous glass rovings tested at 150 days. Having been exposed to the high alkali environment for about 5 months, it is likely that the glass fibres had been subjected to a certain degree of damage due to alkali attack and consequently the effective volume fractions of the glass could have been reduced. Also, increased bond would have reduced slip in the chopped strand. Even allowing for those effects, the experimental $E_f V_f$ values were still generally above the theoretical prediction line using a glass modulus 76 GPa, as shown in Figure 5.24.

The experimental evidence shown in Figures 5.22 to 5.24 support equation 3.17 proposed in Chapter 3 that at stage 3 up to the failure strain of the low elongation fibres the composite stiffness is shared by the two fibres. Detailed comparison between the predicted $E_1$ from equation 3.17 and the experimental $E_1$ is made in Chapter 6.

5.2.4.6 First failure stress and synergistic interaction

Since the glass fibres have much lower strain to failure than the polypropylene film, the glass failed first, well before the ultimate strain of the hybrid was reached, provided that there was sufficient polypropylene in the hybrid.

The first failure stresses $\sigma_1$ measured from the tensile stress-strain curves of the glass-polypropylene hybrids are included in Tables 5.4 and 5.5. In order to check any additional contribution from the polypropylene fibres to the first failure stress, the measured $\sigma_1$ values of individual specimens of the hybrids in Table 5.4 (chopped glass-polypropylene hybrid composites) are plotted in Figure 5.25. In Figure 5.26, similar plots are shown for hybrids containing continuous glass rovings (Tables 5.4 and 5.5), assuming that the glass fibres alone sustained the load. Hence, $\sigma_1 = \eta \sigma_{fu-L} V_{f-L}$, where $\sigma_{fu-L}$ and $V_{f-L}$ are the ultimate tensile strength and volume of the glass fibres respectively.
Figure 5.25 Comparison between experimental values of first failure stress of chopped glass strand-polypropylene hybrid composites and theoretical lines of $\eta\sigma_{fu,L}V_{f,L}$ (Modified Surrey matrix, 28 days)

Figure 5.26 Comparison between experimental values of first failure stress of continuous glass roving-polypropylene hybrid composites and theoretical lines of $\sigma_{fu,L}V_{f,L}$
It is known that the position of the theoretical line predicted by \( \eta \sigma_{fu-L} V_{f-u-L} \), to a great degree, depends on the selected \( \eta \) and \( \sigma_{fu-L} \) values. For the strand efficiency \( \eta \), \( \eta = 0.2 \) might be a more correct approximation (Section 4.2.4.5) with respect to the strand length of 24 mm. However, \( \eta = 0.26 \) is chosen as the strand efficiency which would give the lowest effect of the polypropylene on the first failure stress. Regarding the strand tensile strength \( (\sigma_{fu-L}) \), the manufacturers have given a value of 1700 MPa (Cem-FIL) and 1400 MPa (Nippon Glass) in their technical data sheets. However, it appears that the "as produced strengths" are rarely able to be reached in practical composites produced either in the laboratories or in the industrial processes. Oakley and Proctor (12) estimated the glass strand strength after 24 hours in a pure cement paste environment and reported a value of 1245 MPa. Ali et al (13) reported the tensile stress-strain curves of glass fibre reinforced cement paste composites containing 30 mm long strand with different volume fractions at 28 days. If the strand efficiency is assumed to be 0.26, the strand tensile strength or pull-out stress in their air-stored composites could be evaluated from the ultimate composite strength of the stress-strain curves. Values of 1075 MPa, 982 MPa and 929 MPa of the strand tensile strength were obtained for the fibre volume fractions of 4.4%, 6.3% and 8.2%, respectively. In the present work, the tensile strengths of the glass fibres were determined directly using long strand length samples and static loading. The strengths were found to be 935 MPa for the Nippon glass and 981 MPa for the Cem-FIL glass (Section 3.2.2.3) which is in agreement with the fibre strength values above and in agreement with the strengths used by Allen in his theoretical calculation work (14).

The reasons for the lower strengths obtained in these tests, in comparison with the data given by the producers, are considered to be related to the following facts. The glass filaments are brittle in nature and vulnerable to breakage. Additional damage might take place at every subsequent operation (say, winding, handling, incorporation into cement, etc) after the filaments are formed in the manufacturing processes. Thus, the stage at which the long glass strands are sampled and the method by which they are
tested might affect the measured results. Secondly, the problem of alkali attack of glass fibres has still not been completely resolved and hence the age at which the strand strength in an alkali environment is measured could affect the test results. Partly due to these reasons, strand strengths of 1200 MPa has been broadly adopted in the literature (15-16) for theoretical calculations, instead of the values of the technical data sheets given by the manufacturers.

In Figures 5.25 and 5.26, theoretical lines, $\sigma_{fu-L V_f}$, for glass strand solely using tensile strength values given by the manufacturers (Cem-FIL 1700 MPa, Nippon 1400 MPa), in the literature (1200 MPa) and in the study (Cem-FIL 981 MPa, Nippon 935 MPa, Section 4.2.2.3) are plotted against the glass volume fractions using a strand efficiency $\eta=0.26$ for the 24 mm long chopped strands (Figure 5.25). It can be seen from Figure 5.25 that almost all the experimental values, $\sigma_i$, obtained from the chopped glass-polypropylene hybrids fall above the theoretical lines including the line using the less conservative strength data of 1700 MPa. Bearing in mind the much smaller tensile strengths (≤ 6.1 MPa) of the chopped glass only composites in Table 5.3, one might conclude that the polypropylene networks have provided a contribution to the first failure stress at which the chopped glass fibre in the hybrids failed or pulled out. Experimental evidence shown in Figure 5.26 obtained from the hybrids containing average volume fractions of 0.63% continuous glass rovings and 6.3% polypropylene networks would also support the conclusion drawn from the test results of the hybrids using chopped glass fibres. In the figure, the theoretical lines, $\sigma_{fu-L V_f}$, were plotted using glass strengths of 1400 MPa, 1200 MPa and 935 MPa, respectively, against the volume fractions of continuous glass rovings assuming a roving efficiency of 1.0 and an efficiency of 0.26 for the small amount of 24 mm long chopped glass (0.7% by volume) contained in the hybrids of Series 7 in Table 5.4, although it is thought that this high efficiency ($\eta=0.26$) was unlikely to be achieved for the 24 mm long strands.

It is most obvious in Figure 5.26 that the first failure stress in the
hybrids depended not only on the total volume fractions of the two fibres but also, critically, on the relative proportions of the two fibres. For instance, if we assume a strand strength of 1200 MPa and, further, this strength to be fully developed at the first failure strains (about 1-1.5% for the 28 days glass-polypropylene hybrids, Tables 5.4 and 5.5) of the hybrids, the first failure stresses ($\sigma_1$) of the hybrids containing 2.6-4.2% polypropylene and 1.4-1.8% continuous glass (indicating a polypropylene-glass ratio about 2 to 3) shown in Figure 5.26 would not demonstrate any contribution to $\sigma_1$ from the polypropylene fibres. However, when the relative proportion of the polypropylene increased, the first failure stresses of the hybrids containing about 0.5-0.7% continuous glass and 2-3% polypropylene (indicating a polypropylene-glass ratio about 3 to 6) would be about 28-90% greater than the theoretical prediction (1200 MPa line), and the average first failure stress of the hybrid with the volume fractions of 0.63% continuous glass roving and 6.3% polypropylene network (i.e. the polypropylene-glass ratio was 10) would indicate a 110% increase compared to the 1200 MPa theoretical prediction line (Figure 5.26). Actually, all the first failure stress values obtained from the latter hybrid were above the 1400 MPa theoretical line indicating the great importance of the polypropylene in the development of glass strength and as a direct stress contributor in the glass-polypropylene hybrids. Similar effects of the relative proportions of the two fibres can also be found in the hybrids with the chopped glass-polypropylene system. Obviously, to efficiently utilize the potential strength capacity of the two fibres, especially the glass fibres, there must be sufficient volume fraction of the polypropylene present in the hybrids.

The experimental evidence provided in this section supports the load-sharing theory proposed in the study (Chapter 3) and comparisons between the experiment and the proposed theory for the first failure stress are given in Chapter 6.

One fact which has to be recognised is that, when the glass fibres are used
alone in cement composites, their potential strain capacity is rarely able to be effectively utilised. In the present study, the all glass composites failed within 0.5\% tensile strain (Tables 5.3 and 5.5) mainly due to the low volume fractions of the fibres used. Even for the glass fibre composites with higher fibre volumes shown in the literature, the failure strain results have not been found to be much greater. The tensile results provided by Ali etc (13) have shown that the failure strains of the 30 mm long glass in the air-stored cement composites at 28 days were only about 0.7\%, 0.85\% and 1.1\% for fibre volume fractions of 3.4\%, 6.3\% and 8.2\%, respectively, and the results were found to be lower for the water-cured composites with same volume fractions of the glass. The reduction of the strain to failure of glass in composites has been explained by the ACK theory (2) or in the literature (say, 15, 17) as the result of load sharing by the cement matrix or the effect of matrix restraint. The explanation appears not to be able to fully explain the larger or much larger scale reduction of the glass failure strain in composites. In the glass-polypropylene hybrids, the strain to failure of the glass would also be dependent on the arrangement of the polypropylene networks or on their relative volume fractions.

Bearing in mind the failure strains of the all glass composites, the results of the first failure stress of most of the hybrids in Tables 5.4 and 5.5 would indicate a synergistic interaction between the polypropylene and the glass fibres. For instance, for the hybrid with 1.9\% chopped glass and 3.5\% polypropylene networks, the glass in the hybrid failed at a strain about 1.2\% (Table 5.4) and resulted a first failure stress 11.3 MPa. However, at 1.2\% strain the strength of the one fibre only composite with 1.9\% chopped glass would be zero because it would fail at much smaller strains (say, 0.35\%) and with 3.5\% polypropylene on its own the strength was about 6.5 MPa (Figure 5.4). Therefore, the direct mathematical addition of the two fibres (0+6.5 MPa) is only about 58\% of the measured stress (11.3 MPa). An explanation may be that the polypropylene nets enable the glass fibres to sustain their maximum load (i.e. 0.26x1700MPax1.2\% = 5.3MPa) at higher strains than would be possible on their own and hence the combination of the
two fibres at a given strain (5.3+6.5 MPa)) is more efficient than would be expected from the addition of the individual results (0+6.5 MPa).

5.2.4.7 First failure strain and hybrid effect

It can be seen in Tables 5.3 to 5.5 that the first failure strains ($\varepsilon_1$) measured from the hybrids were greater than that of the composites reinforced only with glass fibres in the form of either chopped strands or continuous rovings. For instance, the failure of the glass fibres ($\varepsilon_1$) which occurred in the composites with the chopped glass fibre alone at less than 0.5% strain (Table 5.3) appeared to be stabilised to strains of over 1% in the hybrids. This phenomenon is termed as the "hybrid effect". One manifestation of the hybrid effect is that the failure strain of the low elongation fibre (glass) appears to be greater in the hybrid than in an all glass fibre composite structure.

The mechanism of the hybrid effect is not well understood. It might be that the presence of the high elongation polypropylene fibres reduces the probability of catastrophic crack propagation initiated by the failure of the weakest glass fibres, since the glass fibres are "constrained" by the polypropylene-cement "matrix". The polypropylene fibres might redistribute the fracture energy of the weakest glass while it could be fatal to an all glass composite. Thus, the effectiveness of the stabilising effect imposed by the polypropylene would be dependent not only on the total fibre volumes but, to a great extent, on the relative proportions of the two fibres. This point is confirmed by the experimental evidence in Tables 5.4 and 5.5 and the graphical evidence in Figures 5.27 and 5.28 for individual specimens of the hybrids containing either chopped glass or the continuous glass. Figure 5.27 shows that the first failure strain $\varepsilon_1$ of the hybrids caused by the failure of the chopped glass increases as the volume fraction of the polypropylene increases. Figure 5.28, where comparison is made between the two group of hybrids containing glass fibres in the form of continuous rovings only (Series 6, Tables 5.4 and 5.5), clearly demonstrates the
Figure 5.27 Effect of polypropylene volume on the first failure strain of chopped glass strand-polypropylene hybrid composites (28 days)

Figure 5.28 Effect of polypropylene volume on the first failure strain of continuous glass roving-polypropylene hybrid composites. Volume fractions of glass shown in the figure were the average values (28 days)
effectiveness by increasing the polypropylene volumes on the first failure strains.

In Table 5.5, the experimental data concerning first failure strain (\(\varepsilon\)) of the hybrids at 150 days exhibited similar trend as the above discussed. That is, the first failure strains in the hybrids increase with the increased volume fractions of the polypropylene at a given volume of the glass.

5.2.4.8 Stress transfer between glass and polypropylene

After the failure of the glass fibres at \(\varepsilon_1\), there was another nearly horizontal part in the recorded stress-strain curves of the hybrids, as can be seen in Figures 5.8 to 5.12 and schematically in Figures 5.13 and 5.14. The stress-strain curves of this second horizontal part were found to be relatively smooth for the hybrids containing chopped glass fibres (Figure 5.8) and to be jagged after the first failure event of the glass fibres accompanied by a sudden load reduction at \(\varepsilon_1\) for the hybrids containing continuous glass rovings (Figures 5.9 to 5.12). It was noted that a few noises associated with fibre breakage had been heard during a loading-to-failure test of the hybrid specimens containing continuous glass fibres and each of the noises was observed to be associated with a load drop in the second horizontal part of the load-strain chart. It is considered that the noise sources were from fractures of the continuous glass strands rather than from the high elongation polypropylene networks in the hybrid, although the fine secondary fibrils or hairs held in the matrix may tend to break near the edge of the main polypropylene fibres (3). Thus, it would appear that there was another multiple fracture stage in the hybrid specimens containing glass fibres during a loading-to-failure process after the first failure event of the glass and that this stage was characterized by the multiple fracture of the long glass fibres, instead of the multiple fracture of the cement matrix observed at an earlier stage. This experimental finding concerning the multiple fracture phenomenon of the glass fibres is in agreement with the theoretical prediction made in an earlier Chapter.
As mentioned earlier, no sudden load reductions at $\varepsilon_1$ were observed for the hybrid specimens containing chopped glass fibres. The composite stresses of the hybrids with the chopped glass fibres within region c-d on Figure 5.13 were found to be more or less the same as the first failure stresses ($\sigma_1$). These observations suggest that the loss of load bearing capacity of the chopped glass in the hybrids occurs as a large number of smaller events and is a more progressive process rather than a few large events. Bearing in mind the 24 mm long cut length of the glass strands (approximately 200 filaments per strand) with an average pull-out length of 6 mm and the post yielding tails of the stress-strain curves of the all glass strand composites after their peak load (Figure 5.5a), the main failure mode of the chopped glass-cement component in the hybrid might be by strand pull-out rather than by fracture. When the glass fibres were added in the form of continuous rovings (approximately 6400 filaments per roving), the energy released on the fracture of the strands in the roving(s) was large and often was a sudden event. A load drop in the load-strain curve of the hybrid for the cross-head speed practically used thus was inevitable, unless, presumably, the size of the glass roving incorporated had been considerably smaller than 6400 filaments per roving, and the volume fraction of the high elongation (HE) fibre in the hybrid had been considerably greater. The effects of HE fibre volume fractions and bundle sizes of LE fibres on the resultant stress-strain curves have been demonstrated for the carbon-glass-epoxy hybrids (say, 18-19) but have not been proved for fibre-cement hybrid composites since there has been no published experimental data. It is thought that, by increasing the volume fraction of polypropylene and meanwhile reducing the bundle size of the glass roving, each glass fibre roving would amount to a smaller proportion of the cross-section and the additional load resulting from individual fractures of the glass bundles would cause negligible load reduction, reduced extension of the bridging polypropylene fibre in relation to the cross-head movement and, probably, reduced damage of the interfacial bond between glass and cement at either side of the crack. As mentioned earlier, a 10 times greater volume fraction of polypropylene than glass roving was used for one hybrid and the resulting
stress-strain curves (Figure 5.12) showed a reduced extent of load reduction after the failure of the glass roving fibres compared to that of similar types of hybrids with smaller polypropylene contents (Figures 5.9 to 5.10). This experimental evidence is considered to support the points made above concerning the volume fractions of HE fibres and the bundle sizes of LE fibres.

In Chapter 3, two possibilities are suggested to explain the stress-strain curve shape of the hybrid at strains exceeding $\varepsilon_i$:

1) simultaneous fracture of all glass filaments, and
2) progressive fracture of glass filaments.

Obviously, the "simultaneous fracture of all glass filaments" pattern (Figure 3.4a) was not supported by the experimental results (Figures 5.8 to 5.12, or Figures 5.13 and 5.14) and therefore is invalid. However, the "progressive fracture of glass filaments" theory would appear to explain the experimental phenomenon occurred in region c-d or c'-d on the curves on Figure 5.13 and 5.14. The nearly horizontal part (region c-d or c'-d) in the stress-strain curve of the hybrid must result because the compatibility between the stress, strain and modulus in the polypropylene fibre along the gauge length of the test specimen have to be satisfied before further loading of the HE fibre along path d-e, and because of the progressive or multiple fracture of the glass filaments in this region.

In order to examine factors which could affect the strain at point "d" on Figure 5.14, the lengths of the horizontal part (c'-d) within which the continuous glass progressively fractured were measured from the experimental stress-strain curves of the individual specimens for both the continuous glass roving-polypropylene hybrids and the chopped glass-polypropylene hybrids at 28 days and plotted in Figure 5.29 against $E_{f,L}V_{f,L}/E_{f,H}V_{f,H}$, the ratio of modulus and fibre volume fraction of the glass to the polypropylene. Figure 5.29 would appear to show that the length of
the horizontal part (c'-d) in the stress-strain curve of the glass-polypropylene hybrid was affected by the moduli and volume fractions of the two fibres and was smaller for greater relative volume fractions of polypropylene. For instance, for a hybrid containing 6.3% polypropylene and 0.63% continuous glass rovings, its length c'-d was measured to be about 1.3%. Instead, the length was approximately 2.9% for the hybrid containing about 3% polypropylene and 0.55% continuous glass roving fibres. In Figure 5.29, the inputs used for \( E_{f,L}V_{f,L}/E_{f,H}V_{f,H} \) are: \( E_{f,L} = 70 \) GPa, \( E_{f,H} = 5 \) GPa and \( \eta = 0.2 \) for the 24 mm two-dimensional random chopped glass strands. However, as discussed in Chapter 3, there are difficulties for theories to be derived to quantitatively predict this length. This is because at this loading stage the hybrid composites have become extensively cracked and the exact stress transfer mechanism between the components is complicated and unknown. Also the effects of machine stiffness are important in this region.

5.2.4.9 Polypropylene stress in region c'-d on stress-strain curve on Figure 5.14

It can be seen in Figures 5.8 to 5.12 that the polypropylene could sustain much higher stresses at low average composite strains (say, at strains of 1-2%) in the hybrids compared with that in all polypropylene cements (Figure 5.4). This is quantitatively demonstrated in Figures 5.30 and 5.31 for the individual specimens of these composites. In Figure 5.30, the polypropylene stresses are plotted for the hybrids containing chopped glass fibres and in Figure 5.31 for the hybrids containing continuous glass. In the two figures, the polypropylene stresses both in the hybrids and in the all polypropylene composites, \( \sigma_{r,H} \), were determined at a given arbitrary composite strain \( \epsilon_c \) (i.e. 1.2% strain for the composites in Figure 5.30 and 2% strain in Figure 5.31) from the experimental tensile stress-strain curves and calculated by

\[
\sigma_{r,H} = \frac{\sigma_c}{V_{r-H}}
\]  

(5.6)
Figure 5.29 Measured strain lengths of c'-d on Figures 5.14 (or c-d on Figure 5.13) from tensile stress-strain curves of glass-polypropylene hybrid composites at 28 days against the ratio of modulus and volume fraction of the glass \( \left( \frac{E_{f-L} V_{f-L}}{E_{f-H} V_{f-H}} \right) \) to the polypropylene (\( E_{f-H} V_{f-H} \))
Figure 5.30 Measured polypropylene stresses at 1.2% strain (in the range of c-d on Figure 5.13) from tensile stress-strain curves of chopped glass strand-polypropylene hybrid composites, in comparison with that of all polypropylene cements at the same composite strain.

Figure 5.31 Measured polypropylene stresses at 2% strain (in the range of c'-d on Figure 5.14) from tensile stress-strain curves of continuous glass roving-polypropylene hybrid composites, in comparison with that of all polypropylene cements at the same composite strain.
Figure 5.32 Polypropylene apparently extends less in glass-polypropylene hybrid than in all polypropylene composite at a given composite stress due to load sharing and strain localization.
where $\sigma_c$ is the composite stress measured from the hybrids and from the all polypropylene composites at the given arbitrary composite strain, $\varepsilon_c$, and $V_{f.H}$ is the volume fraction of polypropylene in the composites. When using equation 5.6 to calculate the polypropylene stress in the hybrids, it was assumed that the polypropylene film in the hybrids carried all the applied load at a crack where the glass had failed at the arbitrary strains. In Figure 5.31, the composite stresses for the hybrids tested at 7 days were taken within the nearly horizontal region after the first failure stress in the stress-extension curves because the strain gauge had not been used for this group of specimens.

A explanation is given in Chapter 3 as to why the HE fibre stress at a given average strain in the composite will apparently be greater for a hybrid composite than for a HE single fibre composite. If we have a local glass failure at one crack at C ($\varepsilon_i$) in Figure 5.32, the polypropylene would extend to a high strain at that crack to develop an appropriate polypropylene stress to meet the load balance requirement. This would result in a real local strain (point D, Figure 5.32) in the polypropylene fibres which is much greater than the average composite strain at the same stress. However, we do not measure the local strain at the crack but only the average strain over many cracks in the 100 mm gauge length, some of which will still contain unbroken glass fibres and small crack widths. This will result in average strains which are insufficient to develop the high polypropylene stress at a crack whereas the real strain at a crack is sufficient to do so. Eventually, most of the glass is no longer carrying stress and the hybrid curve and polypropylene only curve converge.

5.2.4.10 Elastic modulus of polypropylene in hybrids

After the glass multiple fracture stage, any increase in load was thought to be carried, theoretically, only by the polypropylene fibres and the stage 5 (range d-e, Figures 5.13 and 5.14) or $E_x$ slope (Figure 3.4) resulted, although in practice there must be a continuing or gradual loss of stress in
the glass due to frictional forces reducing and ends pulling out at cracks. It would appear that the $E_2$ slopes on the stress-strain curves of the hybrids were heavily affected by the level of volume fractions of glass fibres and as a general trend the greater the relative proportion of the glass in relation to that of polypropylene, the flatter was the $E_2$ slope of the composite. Therefore, the $E_2$ curve slope of a glass-polypropylene hybrid was invariably less steep compared to that of a polypropylene composite, at a given composite strain and given polypropylene content. Thus, the polypropylene modulus $E_{f,H}$ in stage 5, determined by $E_2/V_{f,H}$, would be expected to be smaller than that measured from the all polypropylene composite alone, as shown in Figure 5.33. The composite modulus in stage 5 ($E_2$), for the all polypropylene cements and the hybrids, was determined from a part of stress-strain curve at a strain range between 5% and 7%. This strain range has been taken simply because for most of the glass-polypropylene hybrids, for the small volume fraction of polypropylene used, the stress-strain curves following the first failure event of the glass were essentially horizontal prior to a strain of 4%.

It is thought that the apparent reduction in elastic modulus of the hybrid composites ($E_2$) at the post multiple fracture (of glass fibres) stage, compared to the elastic modulus ($E_{f,V}$) of the polypropylene only composites, may be due to the residual frictional effect of the glass fibres produced during their pull-out through the matrix.

5.2.4.11 Ultimate strength, ultimate strain and critical volume content of polypropylene

For an all polypropylene fibre composite where all the fibres break at the failure of the composite, the ultimate tensile strength of the composite ($\sigma_{cu}$) increases linearly with increasing fibre volume. The strength of the polypropylene fibres ($\sigma_{fu}$) at the failure strain of the composite can be assessed by (2)
Figure 5.33 Comparison in elastic modulus of polypropylene calculated from $E_f V_f$ slopes in tensile stress-strain curves of glass-polypropylene hybrids and in curves of all polypropylene cements, at 5% - 7% composite strain (28 days)
where $V_f$ is the volume fraction of the polypropylene fibres. However, for a glass-polypropylene hybrid composite a different approach must be used to determine the composite and fibre strength because of the different strains at which the fibres reach their maximum loads.

It was observed in the study that there were two types of composite failure modes present for the hybrids containing polypropylene and glass fibres and the failure modes were very much dependent on the relative proportions of the two fibres.

If there was not sufficient polypropylene fibre present in the hybrid, the polypropylene fibres would suffer immediate fracture after the failure of the glass fibres and the failure mode of the hybrid would be catastrophic, as shown for some of the stress-strain curves 1, 2 and 3 in Figures 5.10 and 5.11. In these cases, the first failure stress, $\sigma_f$, was also the ultimate strength of the hybrid, $\sigma_{cu}$, and both of the fibres contributed to the composite strength. At $\sigma_f$ the glass fibre was expected to reach its tensile strength and would be expected to be the major contributor to $\sigma_{cu}$. However, it has been demonstrated in Section 5.2.4.6 that even when the first failure stress is also the ultimate composite strength the volume fractions of polypropylene have an effect. If there was sufficient polypropylene present to support the additional load thrown upon it after the failure of the glass fibres, the ultimate strength of the hybrid ($\sigma_{cu}$) would depend on the polypropylene fibres alone and could be theoretically predicted by equation 5.7. However, because the polypropylene was stretched to failure in a "matrix" containing failed glass fibres, a check could therefore be made on the magnitude of the residual effect, if any, of the glass fibres on the composite strength. Based on equation 5.7, the polypropylene strength was calculated from the tensile stress-strain curves of glass-polypropylene
hybrids compared with that obtained from the all polypropylene composites alone and this is shown in Figure 5.34. As may be seen in the figure the hybrids containing 2% to 3% of 2-D chopped glass fibres appears to result in higher ultimate composite strengths and thus higher calculated tensile strengths of the polypropylene, compared with the all polypropylene reinforced composites of similar fibre volume fractions. This result might suggest a small additional effect from the 2-D chopped glass fibres presumably due to the small residual interfacial shear stress as the glass fibres slide out of the matrix long after their maximum load capacity has been reached.

The ultimate tensile strength values of the hybrids containing continuous glass rovings were found to be scattered (Figure 5.34). A general trend which may be seen in Figure 5.34 is that the ultimate composite strength and thus its calculated tensile strength of the polypropylene were higher for the hybrids containing higher volumes of polypropylene networks and this is in agreement with the findings in polypropylene reinforced composites alone in which the polypropylene strengths were found to increase with increasing volume of the networks.

It is believed from the experimental results that there is a critical volume content of polypropylene at which the ultimate failure mode of the hybrids changes from being dominated by the glass fibre to by the polypropylene fibre. Although this critical fibre volume is unable to be assessed by the available theories (2) in the literature, the results shown in Tables 5.4 and 5.5 and Figure 5.35 give an empirical idea how of the critical polypropylene volume might be determined for the glass-polypropylene hybrids used in the study.

Figure 5.35 shows the variation of the ultimate tensile strain of the hybrid specimens with the proportion of polypropylene. The polypropylene proportion is defined in Figures 5.35a and 5.35b as the volume fraction of polypropylene to the total volume of the polypropylene and the two-
Figure 5.34  Tensile strength of polypropylene network measured from tensile stress-strain curves of composites with or without glass fibres
Figure 5.35 Effect of polypropylene proportions on ultimate failure strain of glass-polypropylene hybrid composites.

a) Hybrids with 1.8% - 3.3% chopped glass strand and 1.9% - 3.8% polypropylene;

b) Hybrids with 0.5% - 1.8% continuous glass roving and 2% - 6.3% polypropylene.
dimensional random glass (Figure 5.35a), or to the total volume of the polypropylene and the continuous aligned glass (Figure 5.35b) in a hybrid, respectively. It should be indicated that using the term "polypropylene proportion" in the present study is subjected to the condition that both the volume fractions of the two fibres in the hybrid are close to or above their theoretical critical fibre volumes when they are used alone.

As can be seen in Figures 5.35 a) and b), at small volume fractions of polypropylene the failure strain of the hybrids was essentially that of glass fibres and so there was little further extension of polypropylene or multiple fracture of the glass after $\sigma_1$. When the polypropylene proportion in the hybrids increased to certain levels, about 0.5 for the hybrids containing chopped glass strand (24 mm) in Figure 5.35a, and about 0.75% for the hybrids containing continuous glass roving in Figure 5.35b, there were sudden increases in ultimate failure strain of the hybrid. The strain increases depended both on the total fibre volumes and on the relative proportions of polypropylene. For instance, the specimens with about 2% chopped glass and 2% polypropylene resulted in failure strains about 2% while the other group of specimens containing 3% chopped glass and 3% polypropylene increased the strain to 5%-6% at 28 days. The two groups of specimens had same polypropylene proportion of 0.5 but the total content of fibres of the former specimens was 33% less than the latter. Regarding the much reduced polypropylene failure strain in the former specimens, it is considered that the presence of the glass fibres had increased the critical fibre volume of polypropylene necessary for toughening the composite. This critical volume of polypropylene is variable mainly depending on the volume fractions of the glass which controls $\sigma_1$. In order to achieve a failure mode governed by the polypropylene instead of by the glass there should always be sufficient polypropylene presence to support the additional load after the failure of the glass fibres.

In Table 5.4, the hybrids (No.15 and No.16) of Series 5 (2), which had been constructed in a sandwich form (with glass strand shell), appeared to have
slightly lower first failure stresses and ultimate composite strengths, compared with the interlaminated hybrids (Series 5 (1)) at similar volume fractions.

The ageing effect on the composite failure strains, when this failure strain was dominated by the glass, was considerable. As can be seen in Table 5.5, the ultimate failure strains of the hybrids No.3, No.4, No.5, and No.6 tested at 150 days ranged between 0.6-0.9% which was considerably less than the first failure strains of other hybrids (Series 5 (1)) tested at 28 days (Tables 5.4).

The all polypropylene composite tested at 150 days had showed no reductions in either strength or strain to failure.

5.2.4.12 Typical stress-strain data for fibre-cement hybrids at three years

At the age of 3 years, several hybrid specimens which had been immersed in water and had representative volume fractions of fibres were tested. The tensile results are given in Table 5.6 and their stress-strain curves are shown in Figure 5.36. In the table and the figure, the fibre volume fractions of polypropylene were calculated for each specimen by the Known-weight method (Section 5.2.1.4) and the calculated results were found to be close to those measured by the acid dissolution method at 28 days. The reason for this similarity was considered to be due to the small variation in sheet thickness. Because of the similarity in thickness and width of specimens, the fibre volume fractions of the glass fibres were directly taken from those at 28 days in Table 5.4. Because only a few specimens were used in the test, it is not possible to draw conclusions from the limited results and thus only general discussions are made below.

Significant features of Figure 5.36 for which the typical data are listed in
Table 5.6 Typical tensile results of composites tested at three years

<table>
<thead>
<tr>
<th>Specimen series &amp; number</th>
<th>Volume fraction (%)</th>
<th>$E_c$ (GPa)</th>
<th>$\sigma_c$ (MPa)</th>
<th>$\sigma_t$ (MPa)</th>
<th>$\epsilon_1$ (%)</th>
<th>$\sigma_{cu}$ (MPa)</th>
<th>$\epsilon_{cu}$ (%)</th>
<th>Crack spacing (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>PP net</td>
<td>Chopped glass</td>
<td>Cont. glass</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Series 1</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>* No. 6-7</td>
<td>4.1(2.0)</td>
<td>0</td>
<td>0</td>
<td>37</td>
<td>9.3</td>
<td>16.3</td>
<td>7.0</td>
<td>4.76</td>
</tr>
<tr>
<td>Series 5/1</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 8-8</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 11-8</td>
<td>3.0(1.5)</td>
<td>1.9</td>
<td>0</td>
<td>39</td>
<td>10.7</td>
<td>11.7</td>
<td>1.25</td>
<td>12.9</td>
</tr>
<tr>
<td>No. 11-9</td>
<td>3.1(1.5)</td>
<td>3.0</td>
<td>0</td>
<td>33</td>
<td>11.3</td>
<td>13.6</td>
<td>0.82</td>
<td>12.6</td>
</tr>
<tr>
<td>No. 12-6</td>
<td>3.1(1.5)</td>
<td>3.0</td>
<td>0</td>
<td>33</td>
<td>10.8</td>
<td>13.2</td>
<td>0.75</td>
<td>13.1</td>
</tr>
<tr>
<td></td>
<td>3.6(1.8)</td>
<td>2.1</td>
<td>0</td>
<td>35</td>
<td>10.5</td>
<td>12.1</td>
<td>0.92</td>
<td>15.4</td>
</tr>
<tr>
<td>Series 7</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 21-6</td>
<td>3.2(1.3)</td>
<td>0.7</td>
<td>0.43</td>
<td>35</td>
<td>9.9</td>
<td>11.3</td>
<td>1.25</td>
<td>13.5</td>
</tr>
<tr>
<td>No. 21-12</td>
<td>3.9(1.6)</td>
<td>0.7</td>
<td>0.57</td>
<td>29</td>
<td>9.3</td>
<td>14.0</td>
<td>1.30</td>
<td>16.0</td>
</tr>
</tbody>
</table>

* No. 6-7: No. 7 specimen from the No. 6 sheet in Table 5.3, other specimens from the composites in Table 5.4.
Figure 5.36 Tensile stress-strain curves of composites at three years (Modified Surrey matrix)
Table 5.6 are that the first cracking stress of the composite had increased from 6-8 MPa at 28 days (Tables 5.3 and 5.4) to 10-11 MPa at 3 years, and the elastic modulus of the composites had increased from 20-25 GPa at 28 days to 30-39 GPa at 3 years. Needless to say, these increases in matrix strength and modulus were related to the continuing availability of water for hydration.

For the specimen (No. 6-7, Table 5.6) containing solely polypropylene networks with about 4% fibre volume fraction, the matrix continued to crack up to 3.5% strain with an almost straight linear slope to failure after the matrix cracked. Such cracking phenomenon was different from that at 28 days where the strain to completion of multiple cracking was smaller, about 2%, and a nearly horizontal portion of matrix multiple cracking on the stress-strain curves was distinguishable.

All the four hybrid specimens (Table 5.6) containing chopped glass fibres at 3 years showed about 1 MPa increase in the first failure stress ($\sigma_1$) over that at 28 days (Table 5.4). This was surprising because it is generally assumed that glass fibres deteriorate with time in an alkali environment. However, if "pull-out" is the main failure mechanism, an increase in bond strength with time could increase $\sigma_1$. Reductions in the first failure strain ($\varepsilon_1$) were observed to be about 20-30% for three of the four specimens containing chopped glass with the exception of No.8-8 specimen, and about 15% for the two No.21 specimens which contained continuous glass combined with a small amount of chopped glass, compared to the first failure strain of same group of specimens at 28 days (Table 5.4).

The average crack spacings ($C$) of the hybrid specimens at 3 years (Table 5.6) were similar to or slightly smaller than that at 28 days (Figure 5.20a).

The ultimate strength of the 3-year specimens, when it was determined by the polypropylene, showed no any sign of reduction compared with that at 28 days.
(Tables 5.3 and 5.4). This result is in agreement with that of Hannant (8-9) suggesting the excellence in strength retention of the polypropylene networks. There was little change at 3 years in the failure strain of the hybrid specimens (No.21-6, No.21-12, Table 5.6) containing continuous glass and small fraction of chopped glass and the specimens (No.8-8 and No.12-6). However, two of the hybrid specimens (No.11-8 and No.11-9) which had about same volume fractions of polypropylene and 2-D chopped glass strand failed soon after the first failure event of the glass resulting in a significant reduction in the composite failure strain, compared with the 28 days specimens of same sheet (Table 5.4). Bearing in mind that the two specimens (No.11-8, No.11-9) at 3 years had first failure stresses about 1 MPa higher than at 28 days (No.11 composite, Table 5.4), it is possible that the slightly increased first failure stress had increased the critical fibre volumes of the polypropylene necessary for further tensile strengthening after point $\sigma_f$. The two specimens had polypropylene volume fractions about 3.1% and 2-D random glass strands (24 mm) of about 3%. Based on equation 3.27 proposed in Chapter 3 and using assumed value of $\sigma_{fu,PP}$=400 MPa, we obtain

$$V_{crit-PP} = \frac{\sigma_1}{\sigma_{fu,PP}} = \frac{13.4}{400} = 3.3\%$$

Hence polypropylene failure following glass failure was inevitable.

The limited results obtained at 3 years emphasize again that if polypropylene fibres are added to increase the toughness of the hybrids, the critical volume fraction of the polypropylene corresponding to the first failure stress ($\sigma_f$) must be exceeded so that the polypropylene can support the load when the glass fibre fails, thus giving a high failure strain.

5.2.4.13 Energy absorption evaluation

The energy-absorbing ability or toughness of a fibre cement composite is
frequently expressed as the area under the load-displacement curve and such curves may be obtained either from direct tensile tests or, more often, from the simpler flexural tests. For structural members, building codes often postulate limits on deflections and hence the area under the load-displacement curve should be cut off beyond a certain point. Only its "effective" part should be converted into the toughness quantity. Generally the criteria for "cutting" the area are based either on a specified load or on a specified displacement. Using different criteria (say, 20-22), toughness has been defined from the area up to a given displacement, expressed in a dimensionless form as the "toughness index", which is the ratio of the total energy absorption up to that deflection to the area up to the limit of proportionality.

For non-structural thin sheeting materials where toughness is usually required to provide protection from the consequences of collapse following transient overloads, large deflections may be acceptable particularly in certain applications such as roofing sheet. Under these circumstances, toughness may be assessed from the total area under the flexural or tensile curves and expressed as energy per unit area of fracture surface (23), or per unit volume of test specimen (24-25).

Based on the ACK theory and using an average crack spacing of 1.364x, the toughness, \( U \), of a single fibre reinforced cement may be calculated by (25)

\[
U = 0.5\sigma_{fu}e_{tu}V_x + 0.159\alpha E_0 e_{wu}^2
\]  

(5.8)

where the first term represents the fibre strain energy and the second term the contribution from the multiple cracking of the matrix. Though equation 5.8 has its limitations due to the simplifying assumptions on which it is based, it enables comparisons to be drawn between different type of fibres and the calculated values were in good agreement with those achieved in practice for the aligned polypropylene reinforced cement (25). By comparing
the toughness results for polypropylene fibre cement with those estimated from published stress-strain curves for other fibre cement sheet materials on the basis of energy absorbed per unit volume, Hibbert and Hannant (25) demonstrate that polypropylene fibre cement can offer a material with considerably higher energy absorption capacity than other fibre cements currently available.

In this study, the composite toughness was estimated using Hibbert and Hannant's approach (24-25) by measuring the area under the stress-strain curves of direct tension expressed as energy absorbed per unit volume of the material.

In Figure 5.37a, energies absorbed by individual specimens from five all polypropylene cement sheets and five chopped glass-polypropylene cement sheets at 28 days are plotted against the fibre volume fraction of polypropylene, compared with a theoretical line of equation 5.8 for all polypropylene composites only. Values of polypropylene strength ($\sigma_u$) and modulus were assumed to be 400 MPa and 3.5 GPa respectively. A matrix cracking strength of 25 GPa is assumed for both the unreinforced matrix ($E_u$) and the reinforced matrix ($E_c$) in the theoretical calculation. As can be seen in Figure 5.37a, the toughness values for the all polypropylene cement are in good agreement with the theoretical predictions, but are generally lower than those for the chopped glass-polypropylene hybrids. This is despite the fact that experimental evidence (Table 5.4) has shown that the ultimate failure strain of the polypropylene composites generally decreased in the chopped glass-polypropylene hybrids. We may attribute this increase in toughness of the hybrids to the result of synergistic interaction between the two fibres which gave higher composite stresses at given strains than the individual fibre cements alone. As a result, the toughness at an arbitrary composite strain was higher for the hybrids, as demonstrated in Figure 5.37b in which at two arbitrary strains of 1.5% and 3% the toughness values of the same number of composites in Figure 5.37a are plotted. It can be seen in the figure that at 3% strain toughness for the hybrid composites
Figure 5.37 Comparisons in energy absorbing ability of glass-polypropylene hybrid composites and all polypropylene composites.

a) Overall energy absorptions of fibre cement composites, in comparison with a theoretical prediction (by Ref. 31) for all polypropylene fibre reinforced cements;

b) Energy absorptions determined at arbitrary composite strains of 1.5% and 3.0%. S: chopped glass strand
was generally approximately 100 kJ/m³ greater than that of the polypropylene only composites.

It must be emphasized, however, that the energy absorbing capacity of the hybrid materials is largely dependent on the volume fractions of the polypropylene networks in relation to the first failure stress $\sigma_1$. The energy absorbing capacity of a hybrid would not much differ from that of the glass alone if the polypropylene fibres could not sustain the additional load following the failure of the glass fibres. The effect of polypropylene proportion on toughness of the hybrids is shown in Table 5.7. In the table, toughness of representative hybrids containing chopped glass or continuous glass or both are included and each value was generally the average of 3-4 test results. The results in Table 5.7 emphasize once again the importance of the volume proportions of the polypropylene in relation to that of the glass fibres which control $\sigma_1$. The energy absorbing capacity could be 10 times greater for the hybrid containing 4.1% polypropylene and 1.4% continuous glass (polypropylene proportion 0.75), compared with the hybrid with same volume fraction of polypropylene but 1.8% volume fraction of the continuous glass fibres (polypropylene proportion 0.69). The latter case had increased the critical volume fraction of polypropylene at $\sigma_1$ to more than the included polypropylene volume of 4.1%.

The energy absorbing value for a typical hybrid containing 6.3% polypropylene networks and 0.63% continuous glass rovings at 28 days was 1700 kJ/m³ (Table 5.7). This value was an average of seven test results with a coefficient of variation of 5% only.
Table 5.7 Effect of polypropylene proportion on energy absorbing capacity of hybrids

<table>
<thead>
<tr>
<th>Sheets series &amp; number</th>
<th>Fibre volume fraction (%)</th>
<th>Energy absorption (kJ/m³)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>PP net</td>
<td>Chopped glass</td>
</tr>
<tr>
<td>Series 5(1) (M-Surrey matrix, 28 days)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 7</td>
<td>1.9(0.9)</td>
<td>3.3</td>
</tr>
<tr>
<td>No. 11</td>
<td>3.0(1.5)</td>
<td>3.0</td>
</tr>
<tr>
<td>No. 8</td>
<td>3.0(1.5)</td>
<td>1.9</td>
</tr>
<tr>
<td>No. 12</td>
<td>3.7(1.9)</td>
<td>2.1</td>
</tr>
<tr>
<td>Series 6 &amp; 7 (M-Surrey matrix, 28 days)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 19</td>
<td>2.0(1.0)</td>
<td>0</td>
</tr>
<tr>
<td>No. 21</td>
<td>3.0(1.2)</td>
<td>0.7</td>
</tr>
<tr>
<td>Series 6 (Broni matrix)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>150 days</td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 6+</td>
<td>4.1(2.0)</td>
<td>0</td>
</tr>
<tr>
<td>No. 6</td>
<td>4.1(2.0)</td>
<td>0</td>
</tr>
<tr>
<td>28 days</td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 7</td>
<td>6.3(3.2)</td>
<td>0</td>
</tr>
</tbody>
</table>

5.2.5 Summary

Glass-polypropylene cement hybrid composites yield tensile stress-strain curves consisting of five ranges instead of three ranges for their parents composites. Up to the first failure strain of the hybrid, load is shared by
the two fibres. This agrees with the behaviour predicted by the theory in Chapter 3.

Glass-polypropylene cement hybrid composites yield substantial improvements in composite properties compared with their individual fibre composites. In particular, the strain at the completion of matrix multiple cracking and the crack widths within the multiple cracking region are much smaller for the hybrids compared to those of the all polypropylene composites. The strain at which the glass fibres maintain their maximum stress can be considerably increased by the presence of polypropylene fibres leading to a much higher load bearing capacity within 1.5% strain than either of the individual systems alone or of the sum of their components. The presence of polypropylene therefore enables the reinforcing potential of glass fibres to be more efficiently utilized.

Reductions in the ultimate failure strain of the polypropylene in the hybrid are sometimes observed. These reductions are found to be dependent on the relative volume levels of glass and polypropylene fibres. No reductions in ultimate strength of polypropylene or of the composite due to the presence of glass are observed. Toughness for the glass-polypropylene cements is considerably greater than that of glass reinforced cement alone. At a given polypropylene volume fraction, the toughness of the hybrid is greater than that of the all polypropylene reinforced cement alone, particularly at strains within 3%.

It is of fundamental importance that, in glass-polypropylene hybrid composites, the volume fraction of polypropylene must always be above or well above its critical fibre volume in the hybrids determined from the first failure stress $\sigma_i$. 
5.3 Glass-polyvinyl alcohol (PVA) hybrid composites

5.3.1 Introduction

In Section 5.2, hybrids containing fibrillated polypropylene film and glass fibres have shown superior tensile properties over their individual composites. One practical problem however is that in most countries, fibrillated polypropylene film in the form of open networks is not available. In fact, the Retiflex company of the Montedison group in Italy has been the only producer of such materials on an industrial scale as far as the author knows. Also, the tensile strength and elastic modulus of polypropylene are considered to be low compared to some of the other synthetic fibres. The durability problem of glass fibres in an alkali environment remains unresolved and therefore there is still a need to find alternatives to the polypropylene networks or to the AR-glass fibres in the glass-polypropylene hybrid product.

One of the alternative to polypropylene in glass-polypropylene hybrid composites is continuous polyvinyl alcohol (PVA) fibres in the form of multifilament yarns or rovings. As discussed in Section 4.2.3, PVA fibre has a broad spectrum of properties including its hydrophillic nature and good chemical resistance which could benefit fibre cements. Table 5.8 shows the specific strengths and moduli of the types of PVA yarns used in this study compared with that of polypropylene and glass fibres. It may be seen in the table that the specific moduli (modulus divided by specific gravity) for PVA ranged from 20 to 30 GPa are similar to that for glass about 27 GPa but much higher than that for polypropylene which is less than 6 GPa. The specific strengths (tensile strength divided by specific gravity) of the PVA however are approximately two or three times that of the glass and polypropylene fibres.

In order to exploit the undoubtably attractive properties of the continuous
### Table 5.8 Comparisons in specific modulus and specific strength of fibres

<table>
<thead>
<tr>
<th></th>
<th></th>
<th></th>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>T-5501 PVA</td>
<td>T-5516 PVA</td>
<td>T-7901 PVA</td>
<td>Glass strand</td>
<td>PP net</td>
</tr>
<tr>
<td>Tensile strength (MPa)</td>
<td>1300 940</td>
<td>1490 1090</td>
<td>1900 1330</td>
<td>1400 940</td>
<td>550 400</td>
</tr>
<tr>
<td>Elastic modulus (GPa)</td>
<td>25 29</td>
<td>32 34</td>
<td>38 39</td>
<td>74 70</td>
<td>14 3-5</td>
</tr>
<tr>
<td>Strain to failure (%)</td>
<td>6.7 5.5</td>
<td>6.3 5.0</td>
<td>5.4 4.1</td>
<td>2.0 1.5</td>
<td>8.1 10</td>
</tr>
<tr>
<td>Specific gravity (g/m²)</td>
<td>1.3 1.3</td>
<td>1.3 1.3</td>
<td>2.7</td>
<td>0.93</td>
<td></td>
</tr>
<tr>
<td>Specific strength (MPa)</td>
<td>1000 720</td>
<td>1140 830</td>
<td>1460 1020</td>
<td>520 350</td>
<td>590 430</td>
</tr>
<tr>
<td>Specific modulus (GPa)</td>
<td>19 22</td>
<td>25 26</td>
<td>29 30</td>
<td>27 26</td>
<td>15 3-5</td>
</tr>
</tbody>
</table>

PVA fibres in the hybrids based on polypropylene and glass fibres, PVA rovings were fabricated either with glass fibres or with polypropylene networks in a cement matrix to form hybrid composites and their tensile properties were investigated. The tensile stress-strain behaviour of the glass-PVA hybrids is presented in Section 5.3 and that of the PVA-polypropylene hybrids is in Section 5.4.

#### 5.3.2 Experimental programme

Four types of glass-PVA hybrid composites consisting of a total of six sheets were prepared in the experiment, as shown in Table 5.9. PVA rovings used were the T-7901 type supplied by Kuraray Co. Ltd. Each PVA roving
Table 5.9 Sheet type and fibre combination

<table>
<thead>
<tr>
<th>Sheet type</th>
<th>Fibre combination</th>
<th>Number of sheets fabricated</th>
</tr>
</thead>
<tbody>
<tr>
<td>I</td>
<td>PVA roving only</td>
<td>1</td>
</tr>
<tr>
<td>II</td>
<td>PVA roving + Chopped glass</td>
<td>3</td>
</tr>
<tr>
<td>III</td>
<td>PVA roving + Glass roving</td>
<td>1</td>
</tr>
<tr>
<td>IV</td>
<td>PVA roving + Glass roving + Chopped glass</td>
<td>1</td>
</tr>
</tbody>
</table>

NOTES: PVA roving: T-7901 type, Kuraray
Glass roving: AR-2500H-200, Nippon
Chopped glass: 38 mm length, Nippon

consisted of five strands with 1000 filaments per strand. The continuous glass roving, provided by Nippon Electric Glass Co. Ltd, consisted of 32 strands with about 200 filaments for each strand. The chopped glass strands with a length of 38 mm were cut from the continuous Nippon glass. Only the Broni matrix was used. Full details of physical and mechanical properties of the raw materials are described in Chapter 4.

Fabrication technique of the PVA-glass hybrids was essentially the same as that used for the interlaminated glass-polypropylene hybrids (Section 5.2.1) but with the PVA rovings in place of the layer of polypropylene net. Great care had been taken to align these continuous rovings in the stress direction and to spread the rovings to achieve approximately same width 7 mm during the hand lay-up process. The manufactured sheets were cured under water at 20 °C and cut afterwards into strips of about 300 mm long and 25 mm wide for tensile testing. Specimens of each sheet were arranged to be tested at 14 days, using the hydraulic testing machine in the Fibronit laboratory, as well as at one month using the Instron 1122 machine in Surrey University.
The crosshead movement adopted for both of the tensile machines was 10 mm/min.

Tensile load-displacement curves were obtained for the 14-day specimens. With the installation of a strain gauge, tensile load-strain curves for the one month specimens were recorded. After testing to failure, the number of cracks in a 100 mm distance of each specimen was counted either by the naked eye or by means of optical microscopes in order to calculate the crack spacing. The fibre volume fractions were determined for each specimen from the known weight of fibres. More details of the test instrumentation and methods are given in Section 5.2.1.

5.3.3 Measured tensile stress-strain curves and typical stress and strain data

Some differences were observed between the shapes of the tensile curves at 14 days and at 1 month. These differences were considered to be due to the use of two different test machines rather than due to material changes. Thus, it was decided that, for the tensile curves, those at 14 days would not be shown and only the one month curves are presented.

Table 5.10 shows typical data measured from the tensile curves at 14 days and at one month. Each entry in the table is the average of 3-5 test specimens. Some representative stress-strain curves for each type of composite at one month are shown in Figures 5.38 to 5.40.

5.3.4 Discussion of results

5.3.4.1 All PVA roving reinforced composite

As may be seen in Figure 5.38, the tensile stress-strain curve of the all-PVA roving reinforced cement, as other single fibre reinforced composites, essentially consisted of three parts: elastic range, multiple fracture
<table>
<thead>
<tr>
<th>Sheet number</th>
<th>Fibre volume fraction (%)</th>
<th>Composite elastic modulus (GPa)</th>
<th>Composite cracking stress (MPa)</th>
<th>First failure stress (MPa)</th>
<th>First failure strain (mm or %)</th>
<th>Composite ultimate strength (MPa)</th>
<th>Composite ultimate strain (mm or %)</th>
</tr>
</thead>
<tbody>
<tr>
<td>14 days</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 1</td>
<td>2.4</td>
<td>0</td>
<td>0</td>
<td>5.5</td>
<td>16.3</td>
<td>8.9mm</td>
<td></td>
</tr>
<tr>
<td>No. 2</td>
<td>1.8</td>
<td>1.3</td>
<td>0</td>
<td>5.5</td>
<td>17.9</td>
<td>10.1mm</td>
<td></td>
</tr>
<tr>
<td>No. 3</td>
<td>2.4</td>
<td>1.3</td>
<td>0</td>
<td>6.8</td>
<td>23.1</td>
<td>11.0mm</td>
<td></td>
</tr>
<tr>
<td>No. 4</td>
<td>3.3</td>
<td>1.9</td>
<td>0</td>
<td>8.4</td>
<td>32.0</td>
<td>10.7mm</td>
<td></td>
</tr>
<tr>
<td>No. 5</td>
<td>2.2</td>
<td>0</td>
<td>0.66</td>
<td>7.0</td>
<td>13.2</td>
<td>3.8mm</td>
<td>21.6</td>
</tr>
<tr>
<td>No. 6</td>
<td>1.4</td>
<td>0.7</td>
<td>0.42</td>
<td>5.8</td>
<td>9.3</td>
<td>3.7mm</td>
<td>9.8</td>
</tr>
<tr>
<td>one month</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No. 1</td>
<td>2.1</td>
<td>0</td>
<td>0</td>
<td>31</td>
<td>5.9</td>
<td>14.9</td>
<td>2.4%</td>
</tr>
<tr>
<td>No. 2</td>
<td>2.0</td>
<td>1.3</td>
<td>0</td>
<td>22</td>
<td>6.1</td>
<td>19.0</td>
<td>3.1%</td>
</tr>
<tr>
<td>No. 3</td>
<td>2.3</td>
<td>1.3</td>
<td>0</td>
<td>27</td>
<td>6.9</td>
<td>22.8</td>
<td>3.2%</td>
</tr>
<tr>
<td>No. 4</td>
<td>3.5</td>
<td>2.0</td>
<td>0</td>
<td>26</td>
<td>8.5</td>
<td>34.5</td>
<td>3.1%</td>
</tr>
<tr>
<td>No. 5</td>
<td>2.2</td>
<td>0</td>
<td>0.76</td>
<td>30</td>
<td>7.4</td>
<td>14.0</td>
<td>1.1%</td>
</tr>
<tr>
<td>No. 6</td>
<td>1.4</td>
<td>0.78</td>
<td>0.49</td>
<td>28</td>
<td>6.1</td>
<td>11.2</td>
<td>1.1%</td>
</tr>
</tbody>
</table>
Figure 5.38  A tensile stress-strain curve of composite reinforced with 2.1% Vol. continuous PVA roving only (Broni matrix, one month)

Figure 5.39  Tensile stress-strain curves of chopped glass strand-continuous PVA roving hybrid composites (Broni matrix, one month)
Figure 5.40 Tensile stress-strain curves of glass-PVA hybrid composites.

a) 0.76% continuous glass roving and 2.2% PVA roving;
b) 0.49% continuous glass roving, 0.78% chopped glass strand and 1.4% PVA roving
(Broni matrix, one month)
range and post multiple fracture range. One major difference in curve shape between the all PVA fibre composite and the all polypropylene net composite appeared to be that the slope steepness (and thus secondary modulus of the fibre) of the PVA composite at the post multiple cracking stage increased with tensile strain while that of the polypropylene composite decreased with strain increase. This was in agreement with the observations on the load-displacement curves of the two fibres when they were tested in direct tension alone.

Table 5.11 summarises the typical properties of the two type of composites produced in the study. Compared with an all-polypropylene cement at a given fibre volume, the PVA roving reinforced composite resulted in a higher composite cracking stress, smaller strain at completion of multiple cracking, stiffer curve slope in the post-multiple cracking stage and greater composite strength which, however, was accompanied by a reduced failure strain of the composite. Obviously, the PVA roving is a more efficient material in reinforcement due to its higher tensile strength and elastic modulus. The PVA composite failed at a strain less than half of that for the polypropylene composite and this resulted in a reduced energy absorption capacity for the PVA composite. However, within the strain of 2.5% the energy absorbing ability of the polypropylene composite was only about 60% of that of the PVA composite.

It is notable in Table 5.11 that the tensile properties of the PVA rovings (T-7901) measured from the all PVA composite were well below the values of the fibre alone either given by the technical literature or measured by the author, although the values of the latter were much closer (Table 5.8). For example, the tensile failure stress and strain of the T-7901 PVA roving obtained from the all PVA composite were about 700 MPa and 2.4%, compared with 1900 MPa and 5.4% of the manufacturer data, respectively. If 1900 MPa is the real tensile strength of the PVA strand, then only about two fifths of the potential strength was utilised in the composite. We may attribute this inefficiency to several factors. If the frictional shear bond between
Table 5.11 Comparison in properties between all-PVA composite and all-polypropylene composite at one month

<table>
<thead>
<tr>
<th>Composite type</th>
<th>PVA roving composite</th>
<th>PP net composite</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fibre volume (%)</td>
<td>2.1</td>
<td>2.4</td>
</tr>
<tr>
<td>Composite cracking stress (MPa)</td>
<td>5.9</td>
<td>4.3</td>
</tr>
<tr>
<td>Strain at end of multiple cracking (%)</td>
<td>1.2</td>
<td>3.0</td>
</tr>
<tr>
<td>Cracking spacing (mm)</td>
<td>3.3</td>
<td>3.1</td>
</tr>
<tr>
<td>Secondary modulus of fibre (GPa)</td>
<td>28</td>
<td>3.5</td>
</tr>
<tr>
<td>Ultimate strength of composite (MPa)</td>
<td>14.9</td>
<td>8.6</td>
</tr>
<tr>
<td>Ultimate strain of composite (%)</td>
<td>2.4</td>
<td>6.4</td>
</tr>
<tr>
<td>Energy absorption (kJ/m³)</td>
<td>220</td>
<td>400</td>
</tr>
</tbody>
</table>

the fibres and the matrix is not high enough, the high PVA strength will not be fully developed. There is likely to be no mechanical anchorage between the PVA fibre used and the cement matrix due to its smooth and circular geometry. Another factor which could have caused apparent reductions of tensile strength and strain to failure of the PVA fibre might be the splitting delamination along the stress direction or PVA alignment direction which occurred in the specimens during the tensile testing. The splitting could result from the imperfect fibre alignment in the hand-fabricated sheet. In addition, the manufacturer’s data were measured using a single "twisted" strand (26) consisting of 1000 filaments whereas the roving used in the composite of the present study consisted of five untwisted strands of 5000 filaments. It was observed during the tensile test that only small number of PVA filaments fractured at failure and the major failure mode of the all PVA composite was characterized by matrix splitting and roving pull-out. This would suggest an insufficient fibre-matrix bond.
No bond strength ($r$) data was available in the literature regarding the continuous PVA rovings in cement. Hannant (8) has suggested that the average frictional bond strength ($r$) may best be calculated from the final average spacing ($C$) in the composite by using equation 3.10. Difficulty arose, however, regarding the perimeter ($P_r$) of filaments in a PVA roving in contact with the matrix which was assessed by a microscope examination in the study. It was almost impossible to precisely measure the $P_r$ in contact with the matrix in a PVA roving which consisted of about 5000 filaments. Furthermore, the contact area of the PVA rovings with the matrix probably varied from roving to roving for the fabrication method used. In this study, all of the PVA rovings had been thinly spread out by hand (6-8 mm width) in the fresh cement matrix during fabrication in order to get more bonding area between the filaments and the matrix. The hand lay-up method was not perfect in obtaining uniformity of cement penetration and in the spread-out width for each roving. Nevertheless, it appeared under the microscope that approximately one third of the filaments in any PVA roving were in contact with the matrix. This was in agreement with 35% given by Oakley and Proctor (12) for a glass strand consisting of 204 filaments. Using the following values: fibre volume fraction $V_f = 2.1\%$, matrix cracking stress $\sigma_{mc} = 5.9$ MPa, taken to be equal to composite cracking stress, average cracking spacing $C = 3.3$ mm, calculated cross-section area of roving $A_r = 0.7697$ mm$^2$ (14 $\mu$m diameter, 5000 filaments), and perimeter of filaments in contact with cement $P_r = 73.3$ mm (one third of the perimeter of total filaments in one roving), the average frictional bond strength of PVA rovings was calculated to be 1.19 MPa, based on equation 5.6. This value is similar to that of glass and polypropylene fibres given by some other investigators (3, 12). A more precise method of measuring the perimeter ($P_r$) of the PVA roving in contact with the matrix is still needed, but the contact area will also vary with time due to continuing cement hydration.

5.3.4.2 Hybrid composite containing PVA roving and glass fibres

It was expected that by the combined use of the low elongation glass fibres
and the high elongation PVA rovings in the cement matrix the first failure strain, which should be similar to the ultimate strain of the glass, and the ultimate composite strain which should be similar to the failure strain of the PVA, would be clearly distinguishable on the experimental tensile stress-strain curve, as already demonstrated for the glass-polypropylene hybrids. This didn’t happen, however, for the three sheets of the chopped glass-PVA roving hybrid composites and the resulting tensile stress-strain curves of the hybrids (Figure 5.39) consisted only of the traditional "three stages": the elastic stage, matrix multiple fracture stage and the post-multiple cracking slope. This may have been due to the limited volume fractions of the chopped glass which may have started to pull-out at about 1% strain without showing a significant stress reduction in the stress-strain curve. When the effective volume fraction of the glass fibres was increased, the characteristic sudden failure and load reduction caused by the glass fibres occurred on the tensile stress-strain curves of the hybrids (Figure 5.40).

A comparison at 1.2% strain of glass-PVA and glass-polypropylene hybrids at similar fibre volumes (curve 2 in Figure 5.39 and sheet 9 in Table 5.4) gave stresses of 15.9 MPa and 11.3 MPa for the PVA and the polypropylene hybrids respectively. However, the ultimate strengths of the glass-PVA hybrids were at least as twice as great as those of the glass-polypropylene hybrids, based on similar fibre volume fractions. Also, the ultimate strains of the glass-PVA hybrids, although less than half of the glass-polypropylene hybrids, were still at least twice as great as the failure strains of the glass component. When the volume fraction of the high extension fibres (i.e. polypropylene or PVA) in a hybrid was low, the advantages using the continuous PVA fibres instead of the polypropylene were particularly obvious. For instance, at approximately the same volume fractions of 2% high elongation fibres and 0.7% glass rovings, the first failure stress, ultimate strength and strain were 14 MPa, 21 MPa and 3.6% respectively for the PVA-glass hybrid (No.5 sheet, Table 5.10), compared with that of 11 MPa, 11 MPa and 1.1% respectively for the polypropylene-glass hybrid (No.19 sheet,
Table 5.4). It appeared that the properties of the PVA-glass hybrids, particularly, the ultimate strain to failure, were not so sensitive to the relative proportions of the two fibres as were the polypropylene-glass composites, because of the much higher tensile strength of the PVA fibres. The practical significance of these results is that, if continuous PVA fibres are used in place of the polypropylene in a glass-polypropylene hybrid, then the volume fraction of the PVA could be less than half that of the polypropylene which in certain circumstances could be cost-effective. There would be no strength penalty and an increased load-carrying capacity of the hybrid at small strains. In 1993, the CIF price of the fibrillated polypropylene was about 6000 Italian lire/kg, and that of the PVA fibres ranged approximately between 5000 - 9000 lire/kg, depending on the sources and quality. If same unit weight price for the two fibres is considered, the cost of PVA could only about 50-70% of that of the polypropylene for achieving a given tensile strength, since the polypropylene volume needed would be two or three times greater than that of the PVA to achieve the same strength.

It can be seen in Table 5.10 that the measured ultimate strength and strain of the glass-PVA hybrids were higher compared to that of the all PVA composite. It appeared that the chopped glass fibre, as a cracking stabilizer, had stabilized the delamination or splitting of the PVA rovings from the matrix and as a results the potential strain and strength capacities of the PVA fibres were better utilized.

The smaller failure strains of the glass-PVA hybrids, compared with those of the glass-polypropylene hybrids, are not necessarily considered to be a disadvantage of the glass-PVA hybrid. Because the glass-PVA hybrid failed at a higher strength, the energy absorbing ability of the hybrid was still quite high and could be regarded as of a similar order of magnitude to that of the polypropylene-glass hybrids. The energy absorption capacity of the chopped glass-PVA hybrids is shown in Table 5.12.
Table 5.12 Additional tensile data from glass-PVA roving (T-7901) hybrid composites

<table>
<thead>
<tr>
<th>Sheet No.</th>
<th>Volume fraction (%)</th>
<th>Crack spacing (mm)</th>
<th>Crack width* (μm)</th>
<th>PVA roving modulus (GPa)</th>
<th>PVA roving failure stress (MPa)</th>
<th>Energy (kJ/m³)</th>
</tr>
</thead>
<tbody>
<tr>
<td>No. 1</td>
<td>2.1</td>
<td>3.3</td>
<td>40</td>
<td>28</td>
<td>700</td>
<td>220</td>
</tr>
<tr>
<td>No. 2</td>
<td>2.0</td>
<td>2.7</td>
<td>23</td>
<td>26</td>
<td>920</td>
<td>440</td>
</tr>
<tr>
<td>No. 3</td>
<td>2.3</td>
<td>2.5</td>
<td>16</td>
<td>28</td>
<td>980</td>
<td>460</td>
</tr>
<tr>
<td>No. 4</td>
<td>3.5</td>
<td>2.5</td>
<td>9</td>
<td>30</td>
<td>980</td>
<td>630</td>
</tr>
<tr>
<td>No. 5</td>
<td>2.2</td>
<td>3.1</td>
<td>10</td>
<td>30</td>
<td>950</td>
<td>500</td>
</tr>
<tr>
<td>No. 6</td>
<td>1.4</td>
<td>3.6</td>
<td>17</td>
<td></td>
<td>770</td>
<td>360</td>
</tr>
</tbody>
</table>

* at the end of matrix multiple cracking

5.3.4.3 Crack spacing, crack width and secondary modulus of PVA fibres

It has been shown in Section 5.2.4.3 that, the combined use of the glass and polypropylene resulted in reduced crack widths at the completion of matrix multiple cracking mainly due to the reduced strain at that point. Similar results were observed for the glass-PVA hybrid specimens, as shown in Table 5.12. The incorporation of glass fibres has effectively reduced the crack widths of the glass-PVA hybrid composites at the end of matrix multiple cracking. However, the crack spacings were not much affected by the limited increase in volume fractions of either PVA or glass fibres. In comparison with the glass-polypropylene hybrids, it would appear that the crack spacings and widths were smaller for the glass-PVA hybrids when the fibre volume fractions in both of the hybrids were low (say, 2% or so of high-
elongation fibres). However, at higher fibre volume fractions (say, 3.5% high-elongation fibre of either polypropylene or PVA and 2% chopped glass), the PVA-glass hybrid showed greater values of crack spacing although the crack widths were similar due to the smaller strain at the end of multiple cracking when PVA fibres were used. The strains at the completion of matrix multiple cracking of the glass-PVA hybrids are shown in Figure 5.41 and each point was the average 3-5 test results.

If the residual effect (if any) of the glass fibres is ignored at the strains greater than 2% in the tensile curves of glass-PVA hybrids, the secondary modulus of the PVA fibres \( (E_f) \) may be approximated on the stress-strain curves shown in Figures 5.39 and 5.40a, by the conventional equation: \( E_r = E_x/V_f \), where \( E_x \) is the measured composite modulus in strains of 2% to 4% and \( V_f \) is volume fraction of PVA. The calculated modulus values of the PVA roving are included in Table 5.12 which are much greater than that of the polypropylene fibres but smaller than the value of 38 GPa obtained when the PVA yarn was tested alone (Section 4.2.3.6) at dry condition.

5.3.5 Summary

The strain and strength properties of PVA and glass fibres can be more fully utilized in the hybrid composites than in their individual composites. Glass-PVA hybrids have superior properties over the glass-polypropylene hybrids in that the PVA-glass hybrid system allows similar or more favourable composite properties to be achieved at a much lower addition of the HE fibres. This could be cost-effective.

The glass-PVA cement hybrid composites yield tensile stress-strain curves consisting of five increasing-stress stages when sufficient continuous glass rovings are used.
Figure 5.41  Average strain at termination of matrix multiple cracking in continuous PVA roving and glass fibre hybrid composites (Broni matrix, one month results)
5.4 PVA-polypropylene hybrid composites

5.4.1 Experimental programme

Four complete sheets with the combination of fibres shown in Table 5.13 were fabricated by hand using the Broni matrix. The specimens were cured under water and tested at 14 days on the hydraulic testing machine in Broni and at other ages shown in the table on the 1122 Instron machine at Surrey. The methods for instrumentation and testing were the same as described in Section 5.2.1.3.

5.4.2 Measured tensile stress-strain curves and typical data

Table 5.14 shows typical data measured from the tensile curves at 14 days and at other ages. Each entry in the table is the average of 4-7 test results. Figures 5.42 and 5.43 show some representative stress-strain curves.

Table 5.13 Sheet type and test specimens

<table>
<thead>
<tr>
<th>Sheet type</th>
<th>Fibre combination</th>
<th>Number of specimens</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>14 days</td>
</tr>
<tr>
<td>I</td>
<td>Polypropylene net only</td>
<td>4</td>
</tr>
<tr>
<td>II</td>
<td>Polypropylene+PVA roving(T-5501)</td>
<td>5</td>
</tr>
<tr>
<td>III</td>
<td>Polypropylene+PVA roving(T-5516)</td>
<td>7</td>
</tr>
<tr>
<td>IV</td>
<td>Polypropylene+PVA roving(T-7901)</td>
<td>5</td>
</tr>
</tbody>
</table>
### Table 5.14 Typical tensile results of PVA-polypropylene fibre hybrid composites

<table>
<thead>
<tr>
<th>Composite type &amp; test age</th>
<th>Volume fraction (%)</th>
<th>$E_c$ (GPa)</th>
<th>$\sigma_c$ (MPa)</th>
<th>$\sigma_{cu}$ (MPa)</th>
<th>$\epsilon_{cu}$ (mm or %)</th>
<th>Energy (kJ/m$^3$)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>PP net</td>
<td>PVA roving</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td><strong>14 days</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>I (PP only)</td>
<td>2.6</td>
<td>0</td>
<td>5.5</td>
<td>9.5</td>
<td>13.8 mm</td>
<td></td>
</tr>
<tr>
<td>II (T-5501)</td>
<td>2.5</td>
<td>1.2</td>
<td>6.3</td>
<td>15.7</td>
<td>14.8 mm</td>
<td></td>
</tr>
<tr>
<td>III (T-5516)</td>
<td>3.2</td>
<td>2.3</td>
<td>8.0</td>
<td>31.1</td>
<td>13.9 mm</td>
<td></td>
</tr>
<tr>
<td>IV (T-7901)</td>
<td>3.2</td>
<td>2.3</td>
<td>8.0</td>
<td>39.2</td>
<td>14.8 mm</td>
<td></td>
</tr>
<tr>
<td><strong>8 months</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>I</td>
<td>2.6</td>
<td>0</td>
<td>27</td>
<td>7.7</td>
<td>9.8</td>
<td>520</td>
</tr>
<tr>
<td>II</td>
<td>2.7</td>
<td>1.3</td>
<td>29</td>
<td>8.3</td>
<td>19.9</td>
<td>628</td>
</tr>
<tr>
<td><strong>80 days</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>III</td>
<td>3.2</td>
<td>2.2</td>
<td>31</td>
<td>8.1</td>
<td>29.9</td>
<td>4.5%</td>
</tr>
<tr>
<td>IV</td>
<td>3.2</td>
<td>2.4</td>
<td>25</td>
<td>8.2</td>
<td>&gt; 40.3</td>
<td>&gt; 4.1%</td>
</tr>
</tbody>
</table>

Note: Specimens of composite IV at 80 days didn't fail because the load exceeded the 5 kN tensile machine capacity.

$E_c$ - elastic modulus of uncracked composite,

$\sigma_c$ - composite cracking stress,

$\sigma_{cu}$ - ultimate tensile strength of composite,

$\epsilon_{cu}$ - ultimate tensile strain of composite.

5.4.3 Discussion of results

5.4.3.1 Shape of stress-strain curves

As can be seen in Figures 5.42 and 5.43, the stress-strain curve of the PVA-polypropylene hybrid composites in the study essentially consisted of three stages, compared with five stages of the stress-strain curves for the glass-polypropylene hybrid (Figures 5.8 to 5.14) and the glass-PVA hybrid (Figure
Figure 5.42 Tensile stress-strain curve of PVA roving (T-5501) and polypropylene network hybrid composites, compared with that of an all polypropylene composite of similar volume fraction of polypropylene (Broni matrix, 8 moths).

Figure 5.43 Tensile stress-strain curves of continuous PVA roving-polypropylene hybrid composites. PVA roving type: T-5516 for curve 1 and T-7901 for curve 2. (Broni matrix, 80 days)
5.40a). The reason for this is considered to be that the PVA fibres in the hybrids were incorporated at high volume fractions and as a result the volume fractions of the polypropylene were well below its critical fibre volume for preventing the catastrophic failure of the composites on the failure of the PVA fibres.

As both PVA and polypropylene were essentially high elongation, low/medium modulus fibres, the PVA-polypropylene hybrid system appeared to be less effective in controlling matrix crack opening, compared to the hybrid systems with glass fibres discussed in the previous sections. As a result, the strain for matrix multiple fracture in the tensile curves was larger and load carrying capacity at strains of 1% or so was smaller for the PVA-polypropylene hybrid. This is considered to be the main deficiency of the PVA-polypropylene hybrid from the viewpoint of material properties preferred in the engineering. However, as may be seen in Figures 5.42 and 5.43 the PVA-polypropylene hybrid would still be a superior material when it is compared with the composite reinforced with the polypropylene or the PVA alone. This point is discussed further in the following sections.

Figure 5.43 shows that neither of the fibres appeared to fail until the final failure of the hybrid. This phenomenon suggests that the applied load was shared by the two fibres all the way up to the composite failure strain. The curves in the post multiple cracking stage of the PVA-polypropylene hybrids had the characteristics of that of the PVA only composite, that is, the post multiple cracking slope was not linear but increased with tensile strain.

5.4.3.2 Effect of PVA fibre properties

In the present investigation, three types of PVA rovings, T-5501, T-5516 and T-7901, with different moduli, strength and elongation (Table 5.8) were used showing distinctive effects on the properties of the resulting hybrids in terms of composite strength, energy absorbing capacity (Table 5.14) and
other related properties shown in the tensile stress-strain curves (Figures 5.42 and 5.43). Generally speaking, by using the higher performance PVA fibres, a smaller strain at the completion of multiple cracking ($\varepsilon_{c,m}$), stiffer slope ($E_t$) at the post multiple cracking stage, and a greater ultimate composite strength ($\sigma_{cu}$) resulted. The measured strains at the termination of matrix multiple cracking ($\varepsilon_{c,m}$) are included in Table 5.15 compared with that of the all polypropylene composite. The best composite properties among these three PVA-polypropylene hybrids were obtained from the hybrid containing T-7901 rovings which had the greatest elastic modulus and strength.

As can be seen in Table 5.14, a tensile composite strength about 40 MPa has resulted for a hybrid containing 3.2% polypropylene and 2.3 continuous PVA fibres (T-7901 type). This demonstrates a synergistic effect because the stress of the all polypropylene composite with fibre volume 3.2% at the failure strain of the PVA (about 4% for T-7901) was about 9 MPa (Figure 5.4) and the strength for an all PVA roving (T-7901 type) composite was about 16 MPa (Table 5.10). The direct mathematical addition of the two fibres (9+16 MPa) was much less than the measured strength (40 MPa). It is clear that the presence of one fibre enables the strengthening capacity of another fibre to be better utilized, as already demonstrated in the glass-polypropylene hybrids (Section 5.2) and in the glass-PVA hybrids (Section 5.3).

5.4.3.3 PVA Strength and secondary modulus/crack spacing and crack width

By assuming equal stress carried by polypropylene at a given strain in the hybrid and in the all polypropylene cement, the polypropylene stress in the hybrid at the failure strain of the PVA may be approximated from all polypropylene composite at the same strain and fibre volume. Then the PVA roving strength ($\sigma_{fu,PVA}$) developed in the PVA-polypropylene hybrid may be assessed by equation 3.18 but in a rearranged form which assumes complete composite failure at $\sigma_1$: 
\[
\sigma_{fu-L} = \frac{\sigma_{cu} - \sigma_{f-H} V_{f-H}}{V_{f-L}} \tag{5.9}
\]

where \(\sigma_{f-H}\) is the polypropylene stress developed at the failure strain of the PVA, and \(\sigma_{cu}\) is the first failure stress as well the ultimate strength of the hybrid.

Similarly, the secondary modulus \((E_{f-L})\) of the PVA rovings may be approximated from the measured stress-strain curves using equation 3.17 which suggests that the post matrix multiple cracking stiffness \((E_1)\) in hybrids is proportionally sustained by the two fibres according to their moduli and volume fractions. Equation 3.17 may be rewritten as

\[
E_{f-L} = \frac{E_1 - E_{f-H} V_{f-H}}{V_{f-L}} \tag{5.10}
\]

then the secondary modulus of the PVA roving may be assessed if the secondary modulus \((E_{f-H})\) of the high elongation polypropylene in the equation is given.

The PVA roving failure stresses and moduli calculated from the tensile stress-strain curves of the PVA-polypropylene hybrids by using equations 5.9 and 5.10 are given in Table 5.15 and each data point in the table was the average of 4-7 test results. In the calculation, the polypropylene modulus \((E_{f-H})\) used was 3.5 GPa and polypropylene stresses \((\sigma_{f-H})\) were 280 (for 2.5% \(V_{f-H}\)) and 290 MPa (3.2% \(V_{f-H}\)), assessed from the tensile curves of the all polypropylene composites with similar fibre volume fractions at a given strain of 4% (28 days). Comparing the data in Table 5.15 with that in Table 5.12, it is obvious that the PVA roving (T-7901) failure stress measured in the PVA-polypropylene hybrid is higher or much higher than that measured in the glass-PVA hybrids. The T-7901 roving modulus (42 GPa) and tensile failure stress (1300 MPa) measured from the PVA-polypropylene hybrid are higher than that (26-30 MPa and 920-980 MPa) from the PVA-glass.
Table 5.15 Additional typical data of PVA-polypropylene hybrid composites

<table>
<thead>
<tr>
<th>Sheet type &amp; test age</th>
<th>$\epsilon_{e.m}$ (%$\pm$)</th>
<th>$E_{f.L}$ (GPa)</th>
<th>$\sigma_{fu.L}$ (MPa)</th>
<th>C (mm)</th>
<th>W ($\mu$m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>I (PP only)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>14 days</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>8 months</td>
<td>3.5</td>
<td>3.5*</td>
<td>365*</td>
<td>6.25</td>
<td>238</td>
</tr>
<tr>
<td>II (T-5501)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>14 days</td>
<td>1.7</td>
<td>21</td>
<td>725</td>
<td>2.65</td>
<td>33</td>
</tr>
<tr>
<td>8 months</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>III (T-5516)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>14 days</td>
<td>1.2</td>
<td>25</td>
<td>948</td>
<td>1.29</td>
<td>17</td>
</tr>
<tr>
<td>80 days</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>IV (T-7901)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>14 days</td>
<td>0.8</td>
<td>42</td>
<td>&gt;1292</td>
<td>1.16</td>
<td>10</td>
</tr>
<tr>
<td>80 days</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

$\epsilon_{e.m}$ strain at completion of multiple cracking,

$E_{f.L}$ secondary modulus of PVA rovings,

$\sigma_{fu.L}$ failure stress of PVA rovings,

C crack spacing,

W crack width at end of multiple cracking.

* same fibre volume fractions as in Table 5.14
* measured polypropylene properties from the all-polypropylene composite

hybrids as well as higher than the values from the all-PVA cement alone (28 GPa and 700 MPa). This suggests that either the modulus and stress of the polypropylene used (3.5 GPa and 290 MPa) in the calculation are too low (unlikely) or the utilization of the PVA fibre in the polypropylene-cement environment is more efficient. There was no splitting cracking observed during the loading process of the PVA-polypropylene test specimens indicating that the fibrillated polypropylene film was an excellent crack...
arrestor when it was used together with other fibres of higher performance.

No significant change in crack spacing \( (C) \) of the PVA-polypropylene hybrids or the all polypropylene composite with ageing had been observed, as shown in Table 5.15. In the table the calculated values of crack width \( (w) \) are obtained by multiplying the crack spacing \( (C) \) by the strain at completion of multiple cracking \( (\epsilon_{cm}) \). The crack spacings of the sheets III and IV appeared to have only increased by about 0.1 mm from 14 days to 80 days. The test results provided by Hannant (9) show that there has been little changes in the average crack spacing of the polypropylene reinforced cements for a observed period of ten years and thus the ratio of \( \sigma_{cm}/\tau \) in equation 3.10 has remained constant with time for PVA and polypropylene.

5.4.4 Summary

The introduction of PVA rovings into a polypropylene based cement composite resulted in superior properties over the polypropylene only composite. The load-bearing ability after cracking and the ultimate strength of the hybrid composite were greatly improved. Splitting cracking of the specimens was eliminated by the inclusion of polypropylene. It is possible to replace part of polypropylene networks with a less amount of PVA roving to achieve similar or better composite properties.

In comparison with the hybrids containing glass fibres, the PVA-polypropylene hybrids would be less efficient at strains up to about 1% but thereafter the load-bearing capacity for the PVA-polypropylene hybrid would eventually exceed that of the hybrids with glass fibres at similar volume fractions. The PVA-polypropylene hybrid is advantageous in achieving high ultimate strengths because of the synergistic interaction between the two reinforcing fibres and because of the greater contribution made by the polypropylene networks.
5.5 Carbon-polypropylene hybrid composite

5.5.1 Introduction

In the following sections, the tensile behaviour of carbon and polypropylene fibre hybrids is discussed. Although only limited specimens were prepared it is worth making some comments on the test results because it is not a combination which would normally be expected to produce good results due to the incompatibility in properties of the two fibres. This is because carbon and polypropylene fibres have the largest differences in breaking strains and moduli, and even in strengths of most fibres in the market.

5.5.2 Experimental programme

Two types of pitch based high performance carbon rovings with similar properties but different surface treatments were used in the experiment. According to the manufacturer data (Petoca Ltd), the high performance pitch based carbon rovings have a tensile modulus 200 GPa, tensile strength 2000 MPa and elongation 1%, which are equivalent to some of the PAN based carbon fibres. The directly measured tensile strength on the carbon roving alone by the Author (Section 4.2.4) using the hydraulic testing machine was however smaller about 1300 MPa and the reason for the big difference in strength is considered due to different test methods and devices used. More details concerning the physical properties of the carbon fibres are located in Section 4.2.4.

Four carbon-polypropylene hybrid sheets were fabricated by hand using the modified Surrey matrix. In order to achieve precise positioning, the carbon rovings, which have a weight of 0.31 g per meter, were prepared by sticking the ends of pre-cut lengths onto sheets of paper which were placed as one unit onto the matrix before removing the paper, as shown in Figure 5.44. The specimens were eventually to be cut to 300 mm length by 25 mm width and the number of carbon rovings in this width was adjusted according to the desired
fibre volume. The position of the reinforcing elements and the dimensions of
the specimens are shown in Figure 5.45 the fabrication procedure being
essentially the same as that in Section 5.2.1.2. In addition, two glass-
polypropylene hybrid sheets which had similar volume fractions of fibres to
those of the carbon-polypropylene hybrids (Figure 5.45) and one all
polypropylene composite were prepared. The specimens were cured under water
and tested at 7 days and at 28 days using the 1122 Instron machine.

5.5.3 Measured tensile stress-strain curves and typical data

Figures 5.46 and 5.47 show representative tensile stress-strain curves of
the carbon-polypropylene hybrids at 7 days and at 28 days, respectively,
compared with those of the glass-polypropylene hybrids and the all
polypropylene composite. In Table 5.16, typical data measured from the
tensile curves are included. Each data point in the table is the average of
three test specimens, except that the ultimate tensile strength and strain
value of the 1091C-1 hybrid at 7 days was the average of two specimens. It
was observed during the tensile tests that majority of the carbon-
polypropylene specimens with about 0.75% carbon fibres by volume failed
instantly following the failure of the carbon fibres. A few specimens
however didn’t fail after the failure of the carbon fibres and this result
will be discussed separately from Table 5.16 and Figures 5.46 and 5.47.

5.5.4 Discussion of results

5.5.4.1 Stress at 0.5% strain

It is most obvious from Figures 5.46 and 5.47 that there was a considerable
gain in tensile stress at the stage of earlier loading for the carbon-
polypropylene hybrid specimens, compared with other composites in the same
figures. An approximately 50% or more increase in tensile stress at a given
strain of 0.5% were obtained for the carbon-polypropylene 1091C-2 and 1097C-
2 specimens compared with the glass-polypropylene and the all polypropylene
Figure 5.44  A piece of carbon roving sheeting
a) 1091C-2 (or 1097C-2)  
~ 0.8% by volume

b) 1091C-1 (or 1097C-1)  
~ 0.4% by volume

c) Glass-2  
Glass rovings were spread out by hand  
~ 0.8% by volume

d) Glass-1  
One roving was split into 2 halves  
~ 0.4% by volume

Figure 5.45 Description of reinforcement in carbon (or glass) roving-polypropylene hybrid composites
Figure 5.46 Tensile stress-strain curves of continuous carbon roving and polypropylene network hybrid composites at 7 days, compared with other fibre cement composites (Modified Surrey matrix)
At 28 days

Figure 5.47 Tensile stress-strain curves of continuous carbon roving and polypropylene network hybrid composites at 28 days, compared with other fibre cement composites (Modified Surrey matrix)
Table 5.16 Typical tensile data of carbon-polypropylene hybrid composites

<table>
<thead>
<tr>
<th>Composite type</th>
<th>Volume fraction (%)</th>
<th>Elastic modulus (GPa)</th>
<th>Composite cracking stress (MPa)</th>
<th>First failure stress (MPa)</th>
<th>First failure strain (%)</th>
<th>Ultimate strength (MPa)</th>
<th>Ultimate strain (%)</th>
<th>Roving pull-out stress (MPa)</th>
<th>Crack spacing (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1091C-1</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
</tr>
<tr>
<td>7 days</td>
<td>0.35 2.0</td>
<td>25</td>
<td>4.8</td>
<td>8.0</td>
<td>0.58</td>
<td>7.6</td>
<td>4.3</td>
<td>857</td>
<td>5.00</td>
</tr>
<tr>
<td>28 days</td>
<td>0.34 1.8</td>
<td>25</td>
<td>6.9</td>
<td>8.7</td>
<td>0.53</td>
<td>~</td>
<td>~</td>
<td>1088</td>
<td>10.72</td>
</tr>
<tr>
<td>1091C-2</td>
<td></td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>7 days</td>
<td>0.76 2.2</td>
<td>24</td>
<td>6.3</td>
<td>11.9</td>
<td>0.55</td>
<td>~</td>
<td>~</td>
<td>908</td>
<td>4.65</td>
</tr>
<tr>
<td>28 days</td>
<td>0.70 1.9</td>
<td>25</td>
<td>6.8</td>
<td>12.4</td>
<td>0.55</td>
<td>~</td>
<td>~</td>
<td>1057</td>
<td>6.12</td>
</tr>
<tr>
<td>1097C-1</td>
<td></td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>7 days</td>
<td>0.42 2.2</td>
<td>21</td>
<td>5.8</td>
<td>9.1</td>
<td>0.65</td>
<td>~</td>
<td>~</td>
<td>976</td>
<td>5.88</td>
</tr>
<tr>
<td>28 days</td>
<td>0.39 2.0</td>
<td>30</td>
<td>6.5</td>
<td>9.4</td>
<td>0.68</td>
<td>~</td>
<td>~</td>
<td>1128</td>
<td>5.26</td>
</tr>
<tr>
<td>1097C-2</td>
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<td></td>
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<tr>
<td>7 days</td>
<td>0.79 2.1</td>
<td>26</td>
<td>7.1</td>
<td>12.2</td>
<td>0.63</td>
<td>~</td>
<td>~</td>
<td>911</td>
<td>6.00</td>
</tr>
<tr>
<td>28 days</td>
<td>0.84 2.2</td>
<td>31</td>
<td>8.2</td>
<td>13.9</td>
<td>0.75</td>
<td>~</td>
<td>~</td>
<td>988</td>
<td>3.41</td>
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<td>Glass-1</td>
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<td></td>
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<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>7 days</td>
<td>0.36 1.9</td>
<td>23</td>
<td>4.5</td>
<td>7.4</td>
<td>1.32</td>
<td>7.2</td>
<td>3.6</td>
<td>667</td>
<td>7.14</td>
</tr>
<tr>
<td>28 days</td>
<td>0.40 2.2</td>
<td>24</td>
<td>5.8</td>
<td>8.8</td>
<td>1.10</td>
<td>7.8</td>
<td>2.6</td>
<td>950</td>
<td>7.14</td>
</tr>
<tr>
<td>Glass-2</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>7 days</td>
<td>0.75 1.9</td>
<td>31</td>
<td>5.1</td>
<td>10.4</td>
<td>1.0</td>
<td>~</td>
<td>~</td>
<td>720</td>
<td>7.14</td>
</tr>
<tr>
<td>28 days</td>
<td>0.72 2.0</td>
<td>30</td>
<td>7.1</td>
<td>11.0</td>
<td>1.1</td>
<td>~</td>
<td>~</td>
<td>833</td>
<td>5.00</td>
</tr>
<tr>
<td>PP cement</td>
<td></td>
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</tr>
<tr>
<td>7 days</td>
<td>0 2.0</td>
<td>19</td>
<td>4.9</td>
<td></td>
<td>7.8</td>
<td>6.5</td>
<td></td>
<td></td>
<td>6.38</td>
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<tr>
<td>28 days</td>
<td>0 1.9</td>
<td>22</td>
<td>4.4</td>
<td></td>
<td>6.7</td>
<td>5.9</td>
<td></td>
<td></td>
<td>7.32</td>
</tr>
</tbody>
</table>

~ Composite failed on the failure of glass fibres and each data was the average of 2-3 specimens (a few specimens didn't fail on the failure of glass fibres)
specimens and this is quantitatively demonstrated in Figures 5.48b (7 days) and 5.49b (28 days). When the carbon content in the hybrids was 0.4% or less, the increases were not so significantly pronounced (Figures 5.48a and 5.49a). The carbon fibre was superior over the glass in promoting the load-bearing capacity of the hybrid within 0.5% strain.

5.5.4.2 Strain at completion of matrix multiple cracking/Crack spacing and width/Bond behaviour

Due to the small volume of the two fibres, the tensile stress-strain curves of the hybrids were not very smooth. Some cracking was still observed following the "completion" of matrix multiple cracking defined by the "tangent-line intersection" method. Nevertheless, the strain ($\varepsilon_{cm}$) at completion of multiple cracking was determined to be about 0.4% and 0.2% for the "C-1" and "C-2" carbon-polypropylene specimens, respectively, which was much smaller than about 0.6-0.7% and 0.6% for the Glass-1 and Glass-2 specimens, respectively. The crack spacings of the carbon-polypropylene were similar to or slightly smaller than for the glass-polypropylene specimens and crack widths of the former would be much smaller because of the reduced strains at the end of multiple cracking. For example, the average crack width was only about 11 $\mu$m for the hybrid with 0.7% carbon fibre and 1.9% polypropylene (1091C-2, 28 day) while that with similar volume fractions of glass roving and polypropylene was approximately 30 $\mu$m.

It was observed during the tensile tests that the carbon fibres, regardless their sizing difference on the bundle surfaces, appeared to show poor dynamic bonding with the cement matrix. Figure 5.50 is the photograph of failed tensile specimens in which the carbon rovings were pulled out by hand without much resistance. However, even though the dynamic bonding was poor, superior properties were demonstrated by the carbon-polypropylene hybrids.

The explanation of this phenomenon may be that the cores of the fibre bundles which were not in contact with cement were being pulled out while
Figure 5.48 Effect of roving type and content on composite tensile stress at 0.5% strain - 7 days
PP: polypropylene network, G: glass fibre, C: carbon fibre)
Figure 5.49 Effect of roving type and content on composite tensile stress at 0.5% strain - 28 days
PP: polypropylene network, G: glass fibre, C: carbon fibre)
Figure 5.50  Failed carbon-polypropylene hybrid specimens showing carbon roving pull-out
the outer fibres which might have an adequate static bond with the cement remained fixed in the matrix. It is apparent from the high tensile stresses sustained at 0.5% strain that a considerable amount of stress transfer between cement and carbon fibre was achieved before pull-out failure occurred. It may be that the high tensile strength of the fibres made pull-out more likely than fibre fracture.

The static bond strength is closely associated with the crack spacing. Generally speaking, higher the static bond strength, smaller the crack spacing. As can be seen in Table 5.16, the carbon-polypropylene hybrids appeared to have slightly smaller crack spacings, compared with the other composites. This might indicate a possibly stronger static bond between the carbon fibres and the matrix. A great deal more work is required before valid comments can be made regarding the bond mechanism and bond strength.

5.5.4.3 First failure stress and ultimate strength

Although the first failure stresses of the carbon-polypropylene hybrids were only 1-2 MPa greater than those of the hybrids containing glass rovings (Table 5.16), the higher stresses for the carbon-polypropylene were obtained at smaller first failure strains, i.e. 0.52%-0.75%, instead of 1.0-1.3%. This result suggests that the carbon is a superior low-elongation fibre over the glass up to the first failure strains of the hybrid composites.

Assuming that the polypropylene (about 2% by volume) carried 5 MPa composite stress (Figures 5.46 and 5.47) at the first failure strains of the hybrids and the rovings carried the extra stress. The calculation based on equation 3.18 in Chapter 3 gives the roving pull-out stress at the first failure strain in Table 5.16. It can be seen that, although the calculated roving stresses were generally higher for the carbon rovings compared with the glass rovings, the values were still much smaller than the theoretical strength of 2000 MPa given by the manufacturer. A major cause of this is
that most of the carbon fibre specimens failed by fibre pull-out (Figure 5.50) and hence the strength of the fibre could not be assessed. Additionally, the volume fraction of the polypropylene networks in the carbon-polypropylene hybrid was not high. Results from the glass-polypropylene hybrids already showed that the potential strength of the LE fibres could be optimally utilized only if there were sufficient amount of HE fibres present in the hybrids.

A few specimens, i.e. two 1091C-1, one 1097C-1 and one 1091C-2 at 7 days, and each of the 1097C-1, 1097C-2 and 1091C-1 at 28 days, didn't fail at the first failure strain. These samples were clearly distinguishable by their stress-strain curves in Figure 5.51. The energy absorbing ability of these specimens was therefore greatly increased. It was as expected that the specimens with approximately 0.75% glass roving and 2% polypropylene would fail instantly on the failure of the glass rovings but it was a little surprising that those specimens with similar or greater fibre volume fractions of carbon and polypropylene didn't fail in a similar way. The reason may have been due to the poorly bonded carbon fibres which generally pulled out through the matrix (Figure 5.50) rather than fracturing, thus allowing a slow stress transfer to the polypropylene at a reduced load.

It is notable in Table 5.16 that the carbon-polypropylene hybrids had slightly greater first failure stresses at 28 days compared to at 7 days. A possible enhanced interfacial bond between the carbon fibres and the matrix might be responsible for this.
Figure 5.51 Tensile stress-strain curves of specimens which didn’t fail on the failure of carbon roving fibres in carbon-polypropylene hybrid composites. a) 7 days,  b) 28 days
The major benefits from the use of carbon fibre rovings compared with glass fibre rovings in the polypropylene based composites are the increase in tensile stress at 0.5% strain. Correspondingly, the strain at the completion of multiple cracking ($\varepsilon_{cm}$) as well as crack width at $\varepsilon_{cm}$ were reduced. From the limited test results, there are reasons to speculate that the optimum balance between the composite stiffness after cracking and toughness could be achieved by the appropriate use of the carbon fibres and the polypropylene networks.

The carbon-polypropylene hybrid composite yields tensile stress-strain curves consisting of five increasing-stress stages when the relative volume fractions of the two fibres are appropriate.

Due to fibre pull-out, rather than breakage, strength of the carbon fibres could not be assessed from tests on the composite.

5.6 Implications from the present work for the design of fibre-cement hybrid composites

Glass, carbon, PVA and polypropylene fibres have been used in order to form hybrid composites. The test results have shown that the hybrid composites are superior in properties over their individual fibre composites, mainly, in four aspects:

1. Increased ability to suppress and stabilise cracking;
2. Increased load capacity after cracking;
3. Increased failure strain and stress of the low elongation fibres;
4. Energy absorbing capacity.
It is recognized from the experimental results of various hybrid composites that, to obtain the optimum balance between the properties of strength, stiffness and toughness the choice of fibres and their relative proportions are critical. It appears that the best combination of fibres consists of low-elongation and high-elongation fibres with the largest difference in failure strains but with moduli as close as possible. There must be enough high-elongation fibres present in the hybrid to ensure that the mode of failure is by fracture of the high-elongation fibres rather than by the low-elongation fibres otherwise toughness will be greatly reduced.
CHAPTER 6

APPLICABILITY OF THE PROPOSED THEORY TO FIBRE-CEMENT HYBRID COMPOSITES

In Chapter 5, there was strong experimental evidence exhibited by the hybrid composites, particularly, by the glass-polypropylene cement hybrids, which would appear to support the theory proposed in Chapter 3. Four supporting features of the stress-strain curves are as follows:

1. Although the strain at termination of matrix multiple cracking in the glass-polypropylene hybrid composite was mainly controlled by the low elongation, high modulus glass fibres, at the limited volume fractions of the fibres used the high elongation polypropylene fibre was also providing additional effect on reducing the strain (see Figure 5.19).

2. In Figure 5.23, the theoretical composite stiffness $E_t$ obtained from $E_{f,L}V_{f,L}$ is compared with the experimental $E_t$ of the continuous glass-polypropylene hybrids. An $E_{f,L}$ value of 76 GPa was used which assumed a minimum contribution to the $E_t$ from the polypropylene. Most of the experimental points fall above the theoretical line indicating that the high elongation polypropylene fibre makes a contribution to the composite stiffness. Similar experimental evidence was also obtained for the chopped glass-polypropylene hybrids (Figure 5.22).

3. In Figure 5.26, theoretical lines $(\sigma_{fu,L}V_{f,L})$ assuming that the glass fibres are wholly responsible for the first failure stress $(\sigma_t)$ are compared with the experimental $\sigma_t$ of the continuous glass-polypropylene hybrids. The results show that the experimental $\sigma_t$ values were strongly affected by the total volume of the two fibres as well as by their relative volume fractions. The majority of the experimental points fall above the theoretical lines of $\sigma_{fu,L}$ values of 1400 MPa and 1200 MPa. For a
polypropylene fibre volume of 6.3%, an experimental first failure stress which was 80% greater than that predicted by $\sigma_{fu,L} V_{f,L}$ using a less conservative $\sigma_{fu,L}$ value of 1400 MPa for the glass fibres (from Product data sheet, Ref. 1) resulted (Figure 5.26). The AR-glass fibres are unlikely to be able to develop a strength of 2520 MPa ($\sigma_{fu,L} = \sigma_s / V_{f,L} = 15.9 / 0.63 = 2524$ MPa) in composites when it is incorporated in the form of large bundles or rovings (6400 filaments per roving). Therefore, the high elongation fibre must contribute to the first failure stress of the hybrid composite.

4. The essential features of the theory which implies a five-stage curve are all reflected in the experimental stress-strain curves of the hybrids.

In this Chapter, quantitative comparisons are made of the experimental values, particularly, that shown for the glass-polypropylene hybrid composites in Chapter 5 with the values predicted by the model proposed in Chapter 3.

6.1 Glass-polypropylene hybrid composites

6.1.1 Strain at termination of matrix multiple fracture

According to the proposed theory, the strain at the completion of matrix multiple cracking ($\epsilon_{c,m}$) shall lie between the limits of equations 3.15 and 3.16 in Section 3.2.1. Based on these equations, strains at the termination of matrix multiple cracking were calculated and plotted against $\alpha'$ in Figures 6.1a and 6.1b (solid and broken lines), compared with experimental $\epsilon_{c,m}$ values measured from individual specimens of the continuous glass roving-polypropylene hybrids and the chopped glass-polypropylene hybrids, where $\alpha'$ is defined by equation 3.8, i.e.

$$\alpha' = \frac{E_m V_m}{E_{e,L} V_{e,L} + E_{e-H} V_{e-H}}$$
Figure 6.1 Comparison between experimental data and proposed equations for the strain at the termination of matrix multiple cracking in
a) continuous glass roving and polypropylene hybrid composites
   ○ v 28 days, △ 150 days
b) chopped glass strand and polypropylene hybrid composites
   * 28 days
The inputs for theoretical calculations were: 1) $E_{r,L}$ value of 76 GPa, which has been the maximum value of elastic modulus of AR-glass fibres used in the literature and by the glass producers (2-3), 2) $E_{r,H}$ value of 6 GPa, the maximum polypropylene modulus obtained from the all polypropylene cement composites in the study, 3) $E_m$ value of 25 GPa, obtained from the tensile and flexural tests of matrices, Section 4.1.4, 4) $\eta$ value of 3/8, which is the assumed strand orientation and bond efficiency factor in the matrix multiple cracking stage. For instance, Oakley and Proctor (4) obtained $\eta$ values ranging from 0.3 to 1, and Aveston et al (5) suggested an $\eta$ factor of 0.5. By taking into account the strand length (24 mm) and the matrix strengths, a value of 3/8 proposed by Cox (6) was adopted. The efficiency of the continuous glass roving was assumed to be 1.

It can be seen in Figures 6.1a and 6.1b that there is good agreement between strain at the completion of multiple cracking calculated using the proposed model and the experimental data for the glass-polypropylene cement hybrid composites. The two figures show that the strain at the completion of matrix multiple cracking decreases with increasing volume of the two fibres and that, combined with Figure 5.19, the high elongation fibre participates in restricting crack opening at the matrix multiple fracture stage. Otherwise the glass would require an $E_{r,L}$ value greater than 76 GPa to explain the experimental results.

6.1.2 Post matrix multiple-fracture stiffness and fibre modulus

Predictions of the post multiple-cracking stiffness ($E_i$) based on the proposed equation 3.17 in Section 3.2.2 are compared with typical experimental data for the continuous glass-polypropylene cement hybrids and the chopped glass-polypropylene cement hybrids in Figures 6.2a and 6.2b, respectively.

Experimental $E_i$ values for individual specimens are included in Figure 6.2a.
Figure 6.2 Comparison between experimental data and Eq. 3.17 (solid and broken lines) for post multiple cracking stiffness of
a) continuous glass roving and polypropylene hybrid composites,
b) chopped glass strand and polypropylene hybrid composites.
Polypropylene modulus used for Eq. 3.17 in Figure 6.2 b) was 5 GPa and a glass modulus of 76 GPa was adopted for all theoretical lines.
One hybrid contained an average polypropylene volume fraction of 6.3% and the other hybrid had a polypropylene content of 2.93% which was the average of 9 test specimens with a standard deviation 0.12%. Using polypropylene volume fractions 6.3% and 2.93%, the corresponding theoretical lines are plotted respectively. Thus, comparisons between the predictions and the experimental values in Figure 6.2a should only be made for each set of experimental data with their corresponding theoretical lines.

The hybrids with an average polypropylene content of 2.93% contained approximately 0.4-0.6% continuous glass and 0.7% 2-D random chopped glass strand. An efficiency factor 0.2 is assumed for the chopped glass strands to obtain the total continuous glass volume plotted on the horizontal axis in Figure 6.2a. The actual efficiency factor for the 24 mm chopped strands is unknown but Oakley and Proctor (4) have published factors between 0.16 and 0.26 for 34 mm strands in cracked spray-suction composites. In the present study, the chopped glass strands will be less efficient because of the 24 mm length and thus an efficiency of 0.2 is assumed. Other inputs for the theoretical lines shown on Figure 6.2a were: 1) an \( E_{r,L} \) value of 76 GPa and 2) \( E_{r,H} \) value of 5 GPa and 2 GPa, which were less than the 6 GPa used in the matrix multiple fracture stage, because the elastic modulus of the polypropylene fibre decreases with increasing tensile strain.

In Figure 6.2b, the experimental values were obtained from 25 individual specimens of six chopped glass-polypropylene hybrid composites. The polypropylene volume fractions in the six composites ranged between 3.0% and 3.8%. Their average polypropylene volume fraction 3.35% (25 specimens, standard deviation 0.30%) however is used for the convenience of plotting theoretical lines. The other inputs for theoretical calculations in Figure 6.2b were: 1) \( E_{r,L} \) value of 76 GPa, 2) \( \eta \) values of 0.15 and 0.26, which were similar to those used by Oakley and Proctor (4) at the post multiple cracking stage, 3) \( E_{r,H} \) value of 5 GPa.

As may be seen in Figures 6.2a and 6.2b, there is good agreement between the
composite stiffness values calculated using equation 3.17 and the experimental data. The stiffness contribution by the high elongation polypropylene fibre to the hybrid composite can be obtained by rearranging equation 3.17 as follows:

\[ E_{f-H} = \frac{E_f - E_{f-L}V_{f-L}}{V_{f-H}} \]  

(6.1)

Substituting the experimental values of one of the continuous glass-polypropylene hybrids (No. 7, Table 5.5, Section 5.2.3), i.e. \( V_{f-L} = 0.63\% \), \( V_{f-H} = 6.3\% \), \( E_1 = 721 \text{ MPa} \), and the assumed glass modulus value of 76 GPa in equation 6.1, an average secondary modulus 3.84 GPa of the high elongation polypropylene fibre results. This value is in good agreement with measured values for polypropylene alone.

6.1.3 Composite failure stress (\( \sigma_f \)) and strain (\( \varepsilon_f \)) at which the glass fibres fail in hybrids

Figures 6.3a and 6.3b show the theoretical predictions for the first failure stress \( \sigma_1 \) in two ways. In one case, \( \sigma_{f-H} \) obtained from equation 3.20 is substituted in equation 3.18 to give \( \sigma_1 \). In the second case \( \sigma_{f-H} \) is obtained from equation 3.21 and placed in equation 3.18. These are compared with the experimental values of one continuous glass roving-polypropylene hybrid and six chopped glass strand-polypropylene hybrids respectively.

In Figure 6.3a, the inputs for theoretical calculations for the continuous glass-polypropylene hybrids were: 1) \( \sigma_{f-L} \) value of 1400 MPa, the ultimate tensile strength of the Nippon glass strand (200 filaments per strand) given by the producer. In the present study, the Nippon glass was incorporated in the form of continuous glass rovings (6400 filaments per roving), the ultimate tensile strength of the glass would be expected to be smaller than when it was used in the strand form according to statistical theories (say, 7). Thus, 1400 MPa may be regarded to be the upper bound of tensile strength.
Figure 6.3 Comparison between experimental data and theoretical predictions in first failure stress $\sigma_1$ of
a) continuous glass roving and polypropylene hybrid composites, 
$\sigma_{fu-L} = 1400$ MPa was used for the theoretical calculations.
b) chopped glass strand and polypropylene hybrid composites, 
$\sigma_{fu-L} = 1700$ MPa was used for the theoretical calculations.
of the glass rovings in the composite. 2) $E_{r,H}$ value of 5 GPa. 3) $\epsilon_{fu,L}$ values of 1% and 2%, the higher limit of $\epsilon_{fu,L}$ (2%) corresponds to the strain value specified by the producer (1) and the lower one (1%) was the approximate maximum failure strain value to be found in the all glass (chopped) reinforced cements in the literature, 4) $V_{f,H}$ value of 6.3%, the polypropylene volume fraction of the experimental hybrid plotted in Figure 6.3a, 5) $\sigma_{r,H}$ values of 100 MPa and 200 MPa, the experimental lower and upper tensile stress values developed in the polypropylene fibre in the all polypropylene cements at the strains (of 1% to 1.4%) where the glass fibre failed in the glass-polypropylene hybrids, calculated from the stress-strain curves of the all polypropylene cement composites used by the author and by Ohno (8) using equation 3.21.

In Figure 6.3b, the inputs for theoretical calculations based on equations 3.18 for chopped glass-polypropylene hybrids were similar to that used above for the continuous glass-polypropylene hybrids, except that: 1) A $\sigma_{fu,L}$ value of 1700 MPa was used, the ultimate tensile strength of the Cem-FIL glass strand (200 filaments per strand, Ref. 9). However, Oakley and Proctor (4) have shown that the "as produced strength" (up to 1700 MPa) of the glass strand is unable to be practically achieved in composites. Thus, 1700 MPa may be regarded to be the upper bound of tensile strength of the Cem-FIL glass in the cement composites, 2) $\eta$ values of 0.15 and 0.26 (4). 3) $V_{f,H}$ value of 3.35%, corresponding to the average polypropylene volume fraction (standard deviation 0.3%) of 25 specimens of the six hybrid composites plotted in Figure 6.3b.

As can be seen in Figures 6.3a and 6.3b, in general, there is good agreement between the first failure stress values predicted by equation 3.18 and those obtained from the experiments for both the continuous glass-polypropylene hybrid and the chopped glass-polypropylene hybrids. In Figure 6.3a, all the experimental points of the specimens fall on the theoretical lower bound of broken lines (Eq. 3.18 combined with Eq. 3.21) which corresponds however to the upper bound of the solid lines (Eq. 3.18 combined with Eq. 3.20). This
indicates that the stress developed in the polypropylene fibres intersecting the crack opening where the first failure event of glass fibres occurred was possibly greater than the average stress calculated using equation 3.18 combined with equation 3.20 of the composite materials approach. If 1400 MPa given by the manufacturer is the maximum stress of continuous glass strand, it is reasonable to believe that the continuous glass roving in the cement composite would sustain a fibre stress less than 1400 MPa. Thus, all the theoretical lines predicting the first failure stress of the hybrid in Figure 6.3a should be displaced towards the horizontal axis. As a result, the experimental points would fall between the theoretical broken lines, and the predictions (solid lines) made using the composite materials approach would generally give a lower limit to the first failure stress. By substituting the experimental first failure stress \( \sigma_f \) value of 15.9 MPa into equation 3.18 and assuming glass roving stresses \( \sigma_{fu,L} \) 1000 MPa and 1400 MPa, the calculated polypropylene stress \( \sigma_{f,H} \) in the hybrid (Figure 6.3a) when the glass failed (at 1.36% strain, Table 5.5) would range between 112 MPa and 152 MPa, compared to about 160 MPa of the polypropylene stress at the same strain in an all polypropylene cement composite with 6.2% fibre content studied by Ohno (8). Bearing in mind that Ohno (8) used a higher strength matrix and thus a higher polypropylene stress at the given strain would result, the agreement between the values of polypropylene stress above calculated by equation 3.18 and measured from the individual polypropylene composite is quite good.

Similarly, in Figure 6.3b, if 1700 MPa given by the manufacturer (9) is the maximum stress of the glass strand, it is reasonable to expect that the chopped glass strand in the cement composite would sustain a fibre stress less than 1700 MPa. Thus, all the theoretical lines in Figure 6.3b predicting the first failure stress of the hybrids would be displaced towards the horizontal axis. Again, good agreement between the proposed theory and experiment is obtained.

It should be noted that the real strength efficiency of glass strand in the
glass-polypropylene hybrids is unknown. Although Oakley and Proctor (4) have given values between 0.16 and 0.26 for their 38 mm strands in glass fibre reinforced composites, the actual efficiency could be greater if they had been used in the glass-polypropylene hybrid system because of the hybrid effect. If this is true, it may partly explain the results shown in Figure 6.3b that the experimental points nearly fall between the bounds of theoretical lines (broken lines, equations 3.18 and 3.21). Another reason for this result is that a greater stress might have been developed in the polypropylene fibres at lower volume fractions compared with at higher volume fractions in the glass-polypropylene hybrids at the glass failure strains. It has been shown by the all polypropylene cements in Chapter 5 that, within the matrix multiple cracking stage, greater fibre stress was developed in the composites with smaller fibre volume fractions.

Thus, it would appear that equation 3.18 combined with equations 3.20 or 3.21 is able to predict the first failure stresses in glass-polypropylene hybrid composites, although it is limited by the assumptions which have to be made regarding the ultimate strength and strain of the continuous glass fibres and the efficiency of the chopped glass strand. Further, it is known from Chapter 5 that the glass failure stresses and strains in the hybrids are not constant but are dependent on both the total fibre volume and the relative proportions of two fibres. Equation 3.18 may predict with relative accuracy the first failure stress if both the total fibre volume and the polypropylene proportion related to the glass in the hybrid are well chosen. Otherwise, an overestimation of the first failure stress may result, as the strength capacity of the glass may not be fully developed in the practical hybrids. Such an example is given in Figure 6.4 in which the experimental first failure stress values of four individual specimens from a 7-day hybrid (No. 6, Table 5.5) with about 1.4% continuous glass roving and 4.2% polypropylene networks are plotted compared with the theoretical predictions. All the experimental values fall below the theoretical lines. Substituting the experimental first failure stresses ($\sigma_f$) in equation 3.18 and assuming $\sigma_{f-H}$ value of 150 MPa, the glass roving failure strength ($\sigma_{fu-L}$)
Figure 6.4 Experimental values at 7 days were well below theoretical predictions due to relative small volume fraction of polypropylene incorporated in the continuous glass roving-polypropylene hybrid composite. $\sigma_{fb-L} = 1400$ MPa was assumed in the theoretical calculations.
in the hybrid was calculated to be only about 840 MPa which was very much smaller than the assumed 1400 MPa. Calculations using equation 3.27 show that a critical polypropylene volume fraction of approximately 6.8% in the hybrid (Figure 6.4) would be needed to theoretically sustain a glass roving failure stress 1400 MPa.

In Chapter 5, it has been shown that the average failure strain of the composite at which glass fails in the hybrids can be considerably increased compared with that in the glass fibre reinforced composites alone and this effect has been termed as the "hybrid effect". It is proposed in Chapter 3 that the lower and upper bounds of the first failure strain of the composite \( \varepsilon_1 \) may be predicted by equations 3.22 and 3.23 (Section 3.2.2) which suggest that the average first failure strain of the composite \( \varepsilon_1 \) increases with the volume fraction of polypropylene at a given content of glass and this appears to be the case shown by the experimental evidences in Chapter 5.

In Figures 6.5a and 6.5b, the theoretical average first composite failure strain \( \varepsilon_1 \) values predicted by equations 3.22 (solid lines) and 3.23 (broken lines) are plotted against \( a' \) (equation 3.8). These are compared with the experimental \( \varepsilon_1 \) values of the continuous glass-polypropylene hybrids and the chopped glass-polypropylene hybrids, respectively. Although the theoretical predictions are dependent on the values (especially \( \varepsilon_{fu} \) value, the failure strain of glass fibre) chosen for the calculations, good agreement between the theoretical lines and the experimental results are observed in Figures 6.5a and 6.5b. It is interesting to see in Figure 6.5a that the continuous glass-polypropylene hybrids containing a small amount of chopped glass fibres (about 0.7%) resulted in higher first failure strains although the absolute volume fraction of polypropylene for the hybrids was smaller than the other hybrid shown in the same Figure. This may suggest that the glass roving-chopped glass-polypropylene film hybrid might be a more efficient hybrid system for increasing the average first failure strains of the composite.
Figure 6.5 Comparison between experimental data and theoretical predictions in average first failure strain $\varepsilon_1$ of
a) hybrids containing continuous glass roving, and
b) chopped glass strand-polypropylene hybrid composites
The inputs for theoretical calculations in Figures 6.5a and 6.5b were: 1) \( \epsilon_{\text{u}, \mu} \) value of 0.03\%, 2) \( E_{\text{f}, \mu} \) value of 76 GPa, 3) \( E_{\text{f}, \mu} \) value of 5 GPa, 4) \( \epsilon_{\text{f}, \mu} \) values of 1400 MPa/76 GPa = 1.84\%, and 5) \( \eta \) value of 0.2, the assumed efficiency of the 24 mm chopped glass strand, based on Oakley and Proctor's results (4).

6.1.4 Critical polypropylene volume fraction and ultimate composite failure strain

In Section 3.2.3 of Chapter 3, it is suggested that the critical HE fibre volume fraction, \( V_{\text{crit}, \mu} \), to avoid instant failure of the hybrid composite on the failure of the LE fibres may be predicted by equation 3.27. The effect of achieving the critical fibre volume of polypropylene on the ultimate composite failure strain is clearly shown in Figures 6.6a and 6.6b. In these figures the vertical dotted lines for the critical polypropylene volume were obtained from equation 3.27 using the ultimate tensile strength of glass \( \sigma_{\text{f}, \mu} \) of 1400 MPa, the ultimate tensile strength of polypropylene \( \sigma_{\text{f}, \mu} \) of 430 MPa and the polypropylene stress at \( \epsilon_{\text{f}, \mu} \) of 150 MPa.

As can be seen in the two figures, at small volume fractions of polypropylene the failure strain of the hybrid is essentially that of the glass (i.e. < 1\%). At volume fractions of polypropylene close to the critical fibre volume (or critical lines in the figures) given by equation 3.27 there is a sudden increase in the ultimate composite failure strain and beyond the transition point the ultimate failure strain of hybrids is determined by the high elongation polypropylene networks. As a result, significant increases of the composite failure strain can be obtained. The ultimate failure strain of polypropylene in the hybrid is also affected by the volume content of glass fibres because some residual restraint to crack opening is provided by fibre pull-out. Thus the greater the volume fraction of the glass, the larger the reduction of the failure strain of the hybrid composite. This may partly explain why the experimental values of failure strain were scattered for the hybrids with similar proportions of
Figure 6.6 Ultimate composite strains.

a) continuous glass roving-polypropylene hybrids with 0.5% - 1.8% glass and 2% - 6.3% polypropylene network,
b) chopped glass strand-polypropylene hybrids with 1.8% - 3.2% 2-D random glass and 1.9% - 3.8% polypropylene network.
6.1.5 Comparison of theoretical and experimental tensile stress-strain curves up to the first failure stress $\sigma_1$

In the previous sections, the proposed theory has been examined using the test results of the glass-polypropylene hybrid composites. It was shown that there is, in general, good agreement between the theory and the experiment. To be complete, the experimental stress-strain curves for typical glass-polypropylene hybrids are plotted in Figure 6.7 (solid lines) compared with those predicted by the present theory (broken lines). Multiple fracture strains of the matrix ($\epsilon_{c,a}$) were calculated using equation 3.15 for the case of 2x crack spacing only. The first failure stresses ($\sigma_1$) were calculated using equations 3.18 and 3.20, except for the hybrid containing 0.63% continuous glass rovings and 6.3% polypropylene networks (Figure 6.7), for which $\sigma_1$ was predicted using equation 3.18 combined either with equation 3.20 (theoretical curve 3) or with equation 3.21 (theoretical curve 4). The values used for the theoretical calculations were similar to that in the previous sections, i.e. $E_m = 25$ GPa, $\epsilon_{m} = 0.03\%$, $E_{f,L} = 76$ GPa, $\sigma_{fu,L} = 1400$ MPa (Nippon glass roving) or 1700 MPa (Cem-FIL glass strand), $\epsilon_{fu,L} = 1.84\%$, $\eta = 3/8$ (before completion of matrix multiple cracking) or 0.2 (after matrix multiple cracking), $E_{f,H} = 6$ MPa (before completion of matrix multiple cracking) or 5 MPa (after matrix multiple cracking), $\sigma_{fu,H} = 440$ MPa (for higher polypropylene volumes) or 420 MPa (for lower volumes) and $\epsilon_{fu,L} = 10\%$. The reasons for the selecting values are given in the previous sections.

As may be seen in Figure 6.7, the essential features of the proposed theory described in Chapter 3 up to $\sigma_1$ are reflected in the experimental stress-strain curves of the hybrids. At similar volume fractions of polypropylene (curves 1 & 2), the higher the volume fraction of glass, the smaller is the strain at the end of matrix multiple cracking, the stiffer is the slope of $E_{f,L}V_{f,L} + E_{f,H}V_{f,H}$ and the greater is the first failure stress. As explained in Section 3.2 the length of the relatively horizontal part following the first
Figure 6.7  Tensile stress-strain curves (up to first failure stress $\sigma_i$) for glass-polypropylene hybrid composites predicted by the proposed theory (broken lines), compared with experimental curves (solid lines). The theoretical ultimate stress $\sigma_{cu}$ and strain $\varepsilon_{cu}$ of the hybrids are shown in the figure by curve markers alone.
failure stress on the experimental curves cannot be accurately predicted because of the extensively cracked nature of the composites and because the exact stress transfer mechanism between the components at this loading stage is unknown. Although experimental results in Chapter 5 (Figure 5.29) showed that this length could be partly affected by the moduli and volume fractions of the two fibres. The ultimate stresses and strains of the composites (shown as separate markers in Figure 6.7) are predicted from the polypropylene fibre alone assuming single composite theory. The difference between these ultimate values and the experimental results are presumably due to residual pull-out effects of the glass fibres.

6.2 Other fibre-cement hybrid composites

6.2.1 Carbon-polypropylene hybrid composite

Figure 6.8 shows the predicted stress-strain curves of the continuous carbon and polypropylene hybrid composites, using the values of $E_{f-L} = 200$ GPa, $\sigma_{fu-L} = 2000$ MPa and $\epsilon_{fu-L} = 1\%$ (10). The input values for the properties of the polypropylene and the cement matrix were the same as that in Section 6.1.5. In the figure, an experimental stress-strain curve of a hybrid (7 days) containing same volume fractions of carbon and polypropylene as that of the theoretical curve 1 is included. The shape of the theoretical stress-strain curve (curve 1) indicates that the polypropylene volume fraction in the hybrid is about the critical fibre volume in agreement with the experimental stress-strain curve. The predicted composite failure strain is greater than the experimental strain because the assumed $\epsilon_{fu-H}$ value of 10\% in the calculation is not normally achieved for polypropylene composites with fibre volumes only 2 or 3 percent.

Similar to the predictions in Figure 6.7 for the glass-polypropylene hybrids, curves 2 & 3 in Figure 6.8 show that, with increasing volume fractions of carbon fibres at a given volume fraction of polypropylene, the
strain at the termination of matrix multiple fracture decreases. The first failure strain of the hybrid increases with the volume fractions of the two fibres.

6.2.2 Glass-PVA hybrid composite

Figure 6.9 shows the predicted stress-strain curves of three glass-PVA fibre cement hybrid composites. The inputs of the PVA (T-7901 type) roving properties in the theoretical calculation were: 1) $E_{f-H}$ values of 15 MPa, the approximated PVA modulus from one all PVA roving reinforced cement at strains less than the glass failure strain of 1.5%. 2) $\sigma_{fu-H}$ value of 1000 MPa, the assumed tensile strength of the PVA roving based on the result of one all PVA roving cement composite. 3) $\epsilon_{fu-H}$ value of 5.4%, the PVA roving failure strain given by the producer (11). The input values for the continuous glass roving and the cement matrix were the same as that in Section 6.1.5. In Figure 6.9, an experimental stress-strain curve of a hybrid (one month) with the same fibre volume fractions as that of the theoretical curve 1 is plotted. As shown in the figure, the stress-strain curves of the PVA-glass hybrids predicted by the theory have the characteristics of shorter length of multiple fracture of the matrix, and stiffer composite stiffness after multiple matrix fracture, when compared with that of the polypropylene-glass hybrid shown in Figure 6.7. This is in agreement with the experiments.

Quantitatively, there is a problem, however, in that the theoretical predictions using the tensile PVA strength values of 1300 MPa (measured from PVA roving alone in the study, Section 4.3.3.6) or 1900 MPa (given by the PVA producer, Ref. 11) for the ultimate strength of the glass-PVA hybrid would give higher or much higher values than practically obtained in the experiments. This might suggest that either the utilization of the reinforcing potential of the PVA fibres needs to be optimized (say, by reducing the roving bundle size, by using stronger cement matrix, etc.) or
Figure 6.8 Tensile stress-strain curves (up to first failure stress $\sigma_f$) for continuous carbon roving-polypropylene network hybrid composites predicted by the proposed theory (broken lines). Theoretical curve 1 is compared with an experimental curve (solid line) of same fibre volume fractions.

Figure 6.9 Tensile stress-strain curves (up to first failure stress $\sigma_f$) for continuous rovings of glass-PVA hybrid composites predicted by the proposed theory (broken lines). Theoretical curve 1 is compared with an experimental curve (solid line) of same fibre volume fractions.
the response of the high performance PVA fibres in the fibre-cement hybrid to aging needs to be further investigated.

6.2.3 PVA-polypropylene hybrid composite

Predictions based on the proposed theory for the tensile stress-strain curves of a few PVA-polypropylene hybrids are shown in Figure 6.10. The inputting values of the fibres and matrix properties for the theoretical calculations are the same as that in the previous sections, except that a PVA roving tensile strength of 1300 MPa instead of 1000 MPa was used, since it is thought that the PVA could develop its tensile stress more efficiently in an environment of higher elongation fibre (PP) instead of in that of lower elongation fibre (glass). An experimental stress-strain curve of a hybrid with same fibre volume fractions as that of the theoretical curve 2 is also plotted in the figure. For the inputting values used, the predicted curve 2 gives good agreement with the experimental curve which consisted of three increasing-stress stages. For curve 2, the low polypropylene volume incorporated (3.2%) was well below its critical volume fraction (approximately 20%) in the particular hybrid. However, this is not considered to be a problem in practice because the first failure strain (around 4%) of the hybrid is already high enough to meet the general toughness requirement and the main purpose for adding the polypropylene in the PVA roving based composite is thus to achieve more efficient utilization of the reinforcing efficiency of the PVA rovings rather than to prevent catastrophic failure when the PVA fails in the hybrid. It is obvious by comparing the curves in Figure 6.10 with those in Figures 6.7 to 6.9 that the hybrid system of PVA-polypropylene is not a very efficient one if it is required to limit the strain at the end of multiple cracking because of the low elastic moduli of both of the two fibres. For example, the controlling factor ($\alpha'$, defined by equation 3.8) for the matrix multiple cracking length is predicted to be 41.0 for the PVA-polypropylene (PVA-PP) hybrid containing 0.63% PVA and 6.3% polypropylene, compared with the theoretical $\alpha'$ values of
27.1 (glass-PP), 14.2 (carbon-PP), 9.8 (glass-PVA), 7.4 (carbon-PVA) and 3.8 (carbon-glass), with exactly the same volume fractions of the LE and HE fibres in these hybrids. From the above data, the most efficient hybrid system for reducing the strain at the end of multiple cracking in the matrix appears to be the carbon-glass combination which has the greatest average modulus value of the two fibres. However, the failure strain of the carbon-glass fibre cement hybrid would be small which would result in a low toughness for this hybrid. Nevertheless, as shown in Chapter 5, the PVA-polypropylene hybrid is still a superior material when compared with the single fibre composites.

6.2.4 Other fibre-cement hybrid systems

Figure 6.11 shows the theoretical stress-strain curves predicted by the present theory for the carbon-PVA fibre cement hybrid and the carbon-glass fibre cement hybrid which were not experimentally prepared in the study. All the fibres are used in the form of continuous rovings. The predictions further indicate how the properties of both the low elongation and high elongation fibres determine the properties of the resulting hybrids and thus the shape of stress-strain curves. From these theoretical curves, one may see that, to achieve the most favourable properties up to the first failure strain, the elastic moduli of the two fibres in a hybrid system should be as high and as close together as possible. Also, to achieve an optimum balance between the stiffness and the toughness in the hybrid, then the best choice of fibres should be those with the largest difference in breaking strains.
Figure 6.10  Tensile stress-strain curves (up to first failure stress $\sigma_1$) for continuous PVA roving-polypropylene hybrid composites predicted by the proposed theory (broken lines). Theoretical curve 2 is compared with an experimental curve (solid line) of composite with same fibre volume fractions.

Figure 6.11  Predicted tensile stress-strain curves (up to first failure stress $\sigma_1$) for continuous carbon roving-PVA roving hybrid composites and for continuous carbon roving-glass roving hybrid composites.
The main equations of the proposed theory have been examined using the tensile test results of glass-polypropylene hybrid composites and the limited test results of other fibre-cement hybrid composites. It is shown that there is good agreement between the values predicted by these equations and those obtained from the experiments. The essential features of the proposed theory are all reflected in the experimental stress-strain curves of the hybrids.

The theory predicts that, with increasing volume fraction of the low elongation (LE) fibres at a given volume fraction of the high elongation (HE) fibres, the strain at the termination of matrix multiple fracture decreases and the post matrix multiple cracking stiffness of the hybrid increases. On the other hand, the theory predicts that, with increasing volume fractions of HE fibres at a given volume fraction of LE fibres, the strain at the termination of matrix multiple fracture decreases and the composite stiffness after the multiple fracture of the matrix increases. These predictions are in good agreement with the experimental results.

The theory predicts that the first failure stress of the hybrid increases with the volume fraction of the HE fibres at a given content of the LE fibres and this is in agreement with the experimental results.

The theory also predicts the first failure strains and the predictions are in good agreement with the experimental results. The failure strains of the LE fibres (i.e. glass) in the hybrids are found to be much greater than the failure strains of the fibre in the LE fibre only composites. This difference is called the hybrid effect.

The prediction of the critical volume fraction of the HE fibres to prevent catastrophic failure of hybrid composites after the failure of LE fibres has been confirmed by experiment.
7.1 Introduction

The research so far reported concerns the performance of hybrid composites in uniaxial tension. This is a much easier system to analyse than flexural systems and hence the correlation between theory and experiment can more easily be established. However, in practical applications, flexural stresses are commonly involved and it is therefore necessary to carry out some tests on the flexural performance of hybrid composites to determine whether the advantages obtained in uniaxial tension can be transferred to practical flexural situations.

The characteristics under uniaxial tension of several types of hybrid reinforcement in cement using glass and fibrillated polypropylene film, glass and PVA, PVA and polypropylene, and carbon and polypropylene, respectively, have been described in Chapter 5. In this Chapter, the typical engineering properties of such hybrid reinforcement, notably with glass and polypropylene, and carbon and polypropylene under flexure are reported. The focus of the research programme was to determine load-deflection curve shapes, to assess data on strength, deflection and cracking behaviour of both small scale specimens and full-size corrugated sheets. The data presented here is intended to show that the use of hybrid fibres in cement can provide more efficient reinforcement for thin sheet applications in flexure as well as in tension as established previously.

7.2 Test details

7.2.1 Test specimens
The specimens used in the flexural test programme consisted of small scale flat specimens and large-size corrugated sheets (profile: 177/51, Ref. 1). The testing on flat specimens consisted of two different hybrids: glass-polypropylene hybrids using the composite types shown in Table 5.1 and detailed in Tables 5.3 and 5.4, and carbon-polypropylene hybrids using the composite types shown in Figure 5.45 or in Table 5.16. The flexural sample size cut from the composite sheets in Table 5.1 was 150mm x 50mm, and was 150mm x 25 mm for the composites shown in Table 5.16. Thicknesses of the flat specimens were generally ranged between 6-9 mm. Eight specimens of each composite sheet in Table 5.1 were tested, with 4 of them tested in the longitudinal direction and the remained 4 in the transverse direction, at the age of 28 days. Thus there were a total number of 240 flexural specimens tested in the glass-polypropylene composite test programme. Only the test results of the longitudinal specimens are presented in this thesis. Some of the test results in the transverse direction of the specimens may be located somewhere else (2, 3). The specimens prepared for the carbon-polypropylene composites programme were tested at two ages of 7 and 28 days. Two samples for each age were tested for each composite type (Table 5.16) in the longitudinal direction. Experimental details concerning the forms and locations of the fibres, fabrication procedures and preparation of the two types of flat specimens has been described in Sections 5.2.1 and 5.5.2, respectively. As control, flat specimens reinforced with individual fibres of glass or polypropylene were also tested.

The full-size corrugated roofing sheets tested were commercially available sheets with the trade names of Retiver and Reticem respectively, manufactured by Fibronit S.R.L. in Italy. Reticem was a cement-based sheet reinforced with continuous fibrillated polypropylene networks only. Hereafter it is called the "PP sheet". Retiver was a cement-based sheet with hybrid reinforcement of continuous polypropylene networks, continuous glass rovings and chopped glass strands. Hereafter this is called the "hybrid sheet". The dimensions and section properties of the full-size sheets are given in Table 7.1. More details concerning the hybrid sheet (Retiver) can
be found in the technical literature (4).

Table 7.1 Dimensions and section property of corrugated sheets (Profile 177/51)

<table>
<thead>
<tr>
<th>Sheet type &amp; number</th>
<th>Test span (mm)</th>
<th>Overall length (mm)</th>
<th>Overall width (mm)</th>
<th>Average height (mm)</th>
<th>Average thickness (mm)</th>
<th>Section modulus (mm$^3$ per 920 mm width)</th>
</tr>
</thead>
<tbody>
<tr>
<td>PP sheet 01</td>
<td>1380</td>
<td>1640</td>
<td>920</td>
<td>49.3</td>
<td>6.93</td>
<td>76329</td>
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<td>1100</td>
<td>1640</td>
<td>920</td>
<td>49.5</td>
<td>7.33</td>
<td>80649</td>
</tr>
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<td>1100</td>
<td>1640</td>
<td>920</td>
<td>48.6</td>
<td>7.43</td>
<td>77390</td>
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<tr>
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<td>7.11</td>
<td>77434</td>
</tr>
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<td>920</td>
<td>48.9</td>
<td>7.07</td>
<td>77422</td>
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<td>920</td>
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<td>6.92</td>
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<td>920</td>
<td>49.0</td>
<td>7.04</td>
<td>76837</td>
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<td>920</td>
<td>48.7</td>
<td>7.30</td>
<td>78995</td>
</tr>
</tbody>
</table>

Note: two test spans were used for the corrugated sheets in the study: 1380 mm and 1100 mm, the most common spans for general roofing used in European countries.

7.2.2 Test facilities and procedures

7.2.2.1 Flat specimens

The flexural tests on flat specimens were conducted in four-point loading in the Instron 1122 machine with a test span of 135 mm. The flexural test rig used is shown in Figure 7.1. The test specimen was supported by two rollers.
of the test rig which are free to rotate, 45 mm apart, and a cross-head movement rate of 10 mm/min was adopted throughout the flexural tests. The cross-head movement and the output from the load cell of the Instron were monitored and recorded on the chart paper. With the test system, the load-deflection curves recorded were for deflection at the third points. Thus in the foregoing text deflections are referred to be that at the third points. Centre point deflection was approximately 1.15 times the third point deflection.

All the specimens, except for the all polypropylene reinforced cement specimens, were loaded up to their maximum load and then unloaded when the load-deflection curves were obviously descending. Because of the high extension capacity, it was proved to be impossible to fracture the polypropylene (Figure 7.2) in the specimens, thus load was removed from the polypropylene only composite specimens when they were loaded to deflections of 15 mm or so (135 mm test span), although at this load level load-deflection curves of most of the polypropylene only specimens were still ascending but at a much reduced rate. This would introduce an error in the calculated notional "ultimate flexural strength" for the polypropylene-only specimens as they had not been loaded to their maximum strength or ultimate strain. It was however thought that this error was unlikely to be very significant if one observes the load-deflection relationship in Figure 7.2. In the figure, the nominal flexural stress for the 5 mm thick specimen containing 3.2 vol% polypropylene was about 21 MPa at 15 mm deflection. When the specimen was deflected to 25 mm, corresponding flexural stress was only increased to 23.6 MPa (2.6 MPa difference only). This deflection (25 mm), was the maximum deflection able to be reached for the 5 mm specimen under the loading system. At this load the maximum notional strain at the tensile beam surface was approximately 3.7%, if the simple beam theory is assumed to apply. In fact the real maximum tensile strain was probably very much greater than this, as the final position of the neutral axis would certainly not be in the neutral position but probably be very close to the compression surface at such a high deflection.
Figure 7.1  Four point loading rig in Instron 1122 testing machine
Figure 7.2 A load-deflection curve showing the ductility of a polypropylene network reinforced cement (3.2% fibre volume, four point loading, 135 mm test span)
After testing, the crack spacing was measured by counting the number of cracks in a length of 100 mm in the centre of each specimen. Measurement of fibre content was based on the acid dissolution technique.

7.2.2.2 Full-size corrugated sheets

A test rig (Figure 7.3) was built for the flexural tests of the full-size corrugated sheets loaded under four-point loading with two-line loads applied across the sheet width at quarter span positions. Two test spans (L), 1380 mm and 1100 mm, were adopted. With this loading regime, the total load applied on the sheet, W, produces the same maximum value of bending moment (WL/8) as would be produced by the total load W uniformly distributed over the sheet. For this reason the two-line load system was sometimes adopted to simulate the bending moment and shear force distributions imposed by snow and wind loads on roofing sheets in service (5). In the foregoing text, loads are also expressed as equivalent loads in kN/m², defined as the total load (kN) divided by the area of sheet between supports, in addition to the more conventional expression of kilonewtons per metre sheet width (kN/m).

Deflections were measured at midspan and a quarter span under three corrugations by Linear variable differential transducers (LVDTs, Figure 7.4). Figure 7.5 shows the positions of the displacement transducers on the underside corrugations. The voltage outputs from the displacement transducers and the load cell (20 kN capacity) were monitored by a data acquisition instrument (ADU system) controlled by a microcomputer (Figure 7.3). This ADU system has 24 independent input channels which allow multiple test tasks to run simultaneously and efficiently and data (i.e. load (N), deflection (mm), etc) are output in readily useable form. Further details of the data acquisition device and instrumentation for tests can be found in the ADU System Manual (6).

The test sheet was placed on adjustable supports on the end. The adjustable
Figure 7.3  Flexural loading rig monitored by Data Acquisition System
Figure 7.4 Linear Variable Differential Transducer (LVDT)
Figure 7.5 Layout of displacement transducers (LVDTs) on tensile surface corrugated sheet under Four point loading over 1100 mm test span
supports were to ensure that each corrugation was in contact with the support before loading was applied. This overcame the necessity to pack beneath corrugations in cases where sheets were warped and rested on only a few corrugations.

All the corrugated hybrid sheets were loaded in excess of their maximum load and then unloaded. It proved impossible to break the corrugated PP sheets and loading had to be removed when displacement of some LVDTs reached their limit. The maximum load was thus taken from the recorded unloading point, although the ultimate load-bearing capacity of the PP sheets was not completely reached. The configuration of fibre reinforcement in the corrugated hybrid sheet was similar to that of Series 7 in Table 5.1, consisting of 3 packs of fibrillated polypropylene networks, two layers of chopped glass strands randomly distributed in the two dimensional planes and continuous glass rovings which were arranged longitudinally in tensile region around the trough area with six rovings per single corrugation. The PP sheet contained four packs of polypropylene networks only and the content of polypropylene networks in the longitudinal direction was found to be about 3.3% using the acid dissolution method. The polypropylene volume fraction in the transverse direction of the sheets was about half of that in the longitudinal direction. The actual fibre volume content in the PP sheet however would be slightly greater than measured, since the non-structural sand in the sheet surface was included in the sheet volume used for calculating the fibre content during the measurement.

7.3 Results and discussion

7.3.1 Glass-polypropylene hybrid composites
   - flat specimens

7.3.1.1 Shape of load-deflection curves and typical data

Representative notional stress-deflection curves of glass-polypropylene
hybrid composites and composites reinforced with their individual fibres are shown in Figures 7.6 to 7.10. The term "notional flexural stress is used because the stresses are calculated from normal flexural beam theory which assumes that the neutral axis remains at the centre of the section and that stress and strain are proportional each other. This is an incorrect assumption in the post-cracking region but is nevertheless common practice in the industrial use of such materials. It is known that shape of load-deflection curves are affected by many factors, notably by fibre volume, matrix properties, sample thickness and test machine characteristics. Specimens with lower fibre volumes clearly show the stages of multiple cracking and post multiple cracking on the load-strain curves under both tension and flexure. For specimens with higher fibre volumes, e.g. 5%, multiple cracking characteristics observed on the tensile load-strain curves may not be reflected on their flexural load-deflection curves. For these specimens load-deflection curves are often observed to be rather smooth, compared with the jagged tensile load-strain curves in the multiple cracking region of similar tensile specimens. At a given deflection, specimens of greater thickness will experience greater tensile strains on the surface than in smaller thickness samples, and therefore an earlier failure of a thick sample does not necessarily mean a lower material ductility than for a thin sample. These observations were also applicable for the fibre-cement hybrid specimens for which the shape of load-deflection curves was additionally influenced by the type of fibres used, the relative fibre proportions and positions of each of the fibres.

As can be seen from Figures 7.2, 7.6 and 7.7, the shape of load-deflection curves of the individual fibre (glass or polypropylene) reinforced cements was roughly similar except for their failure mode. The 2-D chopped glass cement samples failed at their peak load accompanied by subsequent pull out of the fibres. Instead, the polypropylene cement could reach its peak load only at exceptionally large deflections and showed remarkable recovery of deformation when load was removed at its maximum load.
Figure 7.6 Flexural stress-deflection curves of all polypropylene network reinforced cements (Table 7.2, Series 1)
Figure 7.7 Flexural stress-deflection curves of all glass reinforced cements
a) 2-D chopped glass strand (24 mm) reinforced composites (Table 7.2, Series 2);
b) Composite reinforced with continuous glass roving and 2-D chopped glass strand (Table 7.2, Series 3).
Figure 7.8 Flexural stress-deflection curves of chopped glass-polypropylene network hybrid composites (Table 7.3, Series 5(1))
Figure 7.9 Flexural stress-deflection curves of hybrid composites with different positions of chopped glass strand
a) Glass located near the tensile surface below the neutral axis
   (Table 7.3, Series 5(3));
b) Sandwich form, i.e. glass located in the extreme surfaces of specimen
   (Table 7.3, Series 5(2))
Figure 7.10  Flexural stress-deflection curves of glass-polypropylene hybrid composites containing continuous glass rovings (Table 7.3, Series 7)
Fibre volume fractions of curve 1:  2.9% polypropylene network, 0.4% continuous glass roving and 0.9% chopped glass strand.
Fibre volume fractions of curve 2:  3.8% polypropylene network, 0.51% continuous glass roving and 1% chopped glass strand.
The shape of load-deflection curves of the hybrid composites were different from that of the individual fibre cements. A schematic description of load-deflection curve shapes is shown in Figures 7.11 a) and b) for the chopped glass-polypropylene hybrid composite and for the hybrid containing continuous glass rovings, respectively. As may be seen in Figure 7.11 a), the load-deflection curve of chopped glass-polypropylene hybrid composites (Series 5, Table 5.1) may be roughly divided into four ranges:

(1) **Stage 1**: Elastic stage, up to the bend over point "a". Elastic theory may be applicable up to point "a".

(2) **Stage 2**: Elastic-plastic stage I, point "a" to "b". Multiple cracking occurred with neutral axis moving towards to the compression face. Glass and polypropylene shared the applied stress at this stage.

(3) **Stage 3**: Elastic-plastic stage II, point "b" to "c". The first failure event of the glass fibres near the tensile surface occurred. Additional load would be transferred to the neighbouring fibres and the failure of the glass fibres would cause a net loss in composite stiffness at this stage. At point "c", the glass fibres ceased to act as tensile reinforcement and an obvious load drop on the load-deflection curve at point "c" occurred.

(4) **Stage 4**: Elastic-plastic stage III, after point "d". The polypropylene mainly sustained the load and further applied load caused further extension of the composite. This considerably increased the post peak load ductility of the composite compared with glass fibre alone although as a hybrid material its primary function has been lost.

The load-deflection curve shape, schematically shown in Figure 7.11 b), for the glass-polypropylene hybrid composites containing large-bundled continuous glass rovings (Series 6 & 7, Table 5.1) was essentially similar to that in Figure 7.11 a), except that Stage 3 (Elastic-plastic II in Figure 7.11 a)) did not occur on the load-deflection curve of the continuous glass-polypropylene hybrids which consisted of three stages only. This is because the stress transfer between the continuous glass roving and the
Figure 7.11 Schematic description of load-deflection curve shapes of hybrid composites reinforced with
a) chopped glass strand and polypropylene network;
b) continuous glass roving and polypropylene network.
polypropylene resulted from a sudden event on the breakage of the continuous glass fibres rather than slow stress transfer due to slip of short glass strands.

Figure 7.9 and 7.10 show that when load was removed after the peak load had been exceeded, approximately 40% to 50% of the total deformation was recovered for these hybrid specimens. Such high level recovery of elasticity would not be expected for the all glass reinforced composite. The amount of residual deformation of the specimens containing polypropylene when load was removed was dependent on the overall deflection achieved, that is, greater the overall deflection, greater the residual deflection resulted. On the other hand however, it was found that the percentage recovery of the overall deformation appeared to not very much be influenced by the deformation levels. The latter observation seems particularly true for the specimens containing polypropylene networks only. For instance, for the same polypropylene content and thickness, two samples were deflected to 14 mm and 25 mm, respectively, and then unloaded. The residual deflections were 6 mm and 11.5 mm, respectively, whilst the recovery percentages of total deformation for the two specimens were quite similar, at 56% and 54% respectively.

Typical data measured from the experimental load-deflection curves are given in Tables 7.2 and 7.3 for the composites reinforced with individual fibres and with combination of fibres, respectively. In the tables, the limit of proportionality (LOP) is defined as the point of deviation of the load-deflection curve from a straight line on the Instron chart. All the flexural stresses shown in the tables were calculated assuming a linear elastic stress distribution with the neutral axis at the beam centre line. Although this assumption is obviously unjustified theoretically for the composites used in the study which exhibited pseudo-ductile behaviour after cracking, it allows comparisons of flexural properties to be made among the types of composites. In the two tables, each data point was generally the average of four test specimens.
Table 7.2 Typical flexural test results of composites reinforced with individual fibres

<table>
<thead>
<tr>
<th>Sheet series &amp; number</th>
<th>Volume fraction (%)</th>
<th>LOP (MPa)</th>
<th>Flexural stress at 2mm deflection (MPa)</th>
<th>Ultimate flexural strength (MPa)</th>
<th>Crack spacing (mm)</th>
</tr>
</thead>
<tbody>
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<td></td>
<td>PP net</td>
<td>Chopped glass</td>
<td>Cont. glass</td>
<td></td>
<td></td>
</tr>
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<td>10.3</td>
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<td>12.0</td>
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<td>11.6</td>
<td>12.2</td>
</tr>
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<td>0</td>
<td>9.3</td>
<td>12.6</td>
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<td>9.5</td>
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Table 7.3 Typical flexural test data of glass-polypropylene hybrid composites

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<tr>
<th>Sheet series &amp; number</th>
<th>Volume fraction (%)</th>
<th>LOP (MPa)</th>
<th>Flexural stress at 2mm deflection (MPa)</th>
<th>Ultimate flexural strength (MPa)</th>
<th>Crack spacing (mm)</th>
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<td></td>
<td></td>
</tr>
<tr>
<td>No. 20</td>
<td>2.9(1.0)</td>
<td>0.9</td>
<td>0.40</td>
<td>12.1</td>
<td>18.5</td>
</tr>
<tr>
<td>No. 21</td>
<td>3.8(1.4)</td>
<td>1.0</td>
<td>0.51</td>
<td>14.0</td>
<td>21.9</td>
</tr>
</tbody>
</table>
7.3.1.2 Notional flexural stress at 2 mm deflection

For a practical roofing material, producers and clients will generally be concerned with the avoidance of cracking. Thus the limit of proportionality would be taken as the critical working stress, rather than the ultimate flexural strength. However, for intrinsically brittle matrix, cement-based non-asbestos fibre sheeting such as polypropylene reinforced cement and glass reinforced cement, it is extremely difficult to resist the forces caused by inappropriate workmanship during handling and installation without a small amount of cracking. Therefore, sufficient attention should also be given to the after-cracking performance of such composites in order to satisfy the serviceability and the durability requirements of national standards.

As can be seen in Figures 7.6 to 7.10, composites containing polypropylene networks are highly pseudo-ductile material, the deflections at maximum load are likely to be much greater than those permitted in practical applications. A comparison of notional flexural stress at a lower deflection may therefore give more realistic assessment for the sheets in practical situations. A deflection of 2 mm was chosen for this comparison.

From Table 7.2 it appeared that, up to fibre volumes of 2%, there was little change in the notional stress up to 2 mm deflection compared with the value of LOP, when polypropylene fibres or the chopped glass fibres were used alone. The flexural stresses of the all polypropylene cements at 2 mm deflection were no more than 30% greater than the LOPs, for the range of fibre volumes used. However, more than 40% (approximately ranging from 40% to 60%, depending the fibre volume fractions) increases in notional stress at 2 mm deflection over the LOPs were achieved for all the chopped glass-polypropylene hybrid composites (Series 5, Table 7.3) with glass content 2% or so. Obviously, the introduction of stiffer glass fibres has increased the composite stiffness and thus the load-bearing capacity at a given deflection after cracking was greater for the hybrid compared to that for the all
polypropylene cement. For instance, at a fibre volume 4% and at a given
deflection 2 mm, the flexural stress was about 15 MPa for the all
polypropylene cement and 18-19 MPa for the hybrid which contained about 2%
chopped glass strand and 2% polypropylene networks.

The above comparison in flexural stress at 2 mm deflection between various
composites may not provide completely fair and useful data, as the tensile
strain at a given deflection varies with the specimen thickness. The effect
of variations of specimen thickness on the tensile strain at the surface is
demonstrated in Table 7.4. In the table, the maximum strain values on the
tensile surface of different thickness beams at 2 mm 1/3 point deflection
were calculated using the following equation assuming simple beam theory to
apply:

\[ e = \frac{108 \delta}{23L^2} \]  

(7.1)

where \( \delta \) = central deflection = 1.15 x 1/3 point deflection
L = test span
d = beam thickness

Thus, an alternative approach, probably the more reasonable approach, may be
to compare the notional stresses of specimens at a given surface tensile
strain estimated from the specimen deflection. The notional flexural
stresses of glass-polypropylene hybrid specimens measured at a specific
strain of 0.5%, compared with that of all polypropylene reinforced composite
specimens, are shown in Figure 7.12. The beneficial effect of glass fibre
addition (approximately 2-2.5% in the hybrids shown in the figure) on
increasing the nominal flexural stress of the polypropylene based composite
is obvious. Figure 7.12 also shows that, at a given volume fraction of glass
fibres, the notional flexural stress in the chopped glass-polypropylene
hybrid composite increases with the increased volume fraction of
polypropylene.
Table 7.4 Maximum tensile strains on the beam surface assuming neutral axis at mid-depth (one third point loading, 135 mm test span)

<table>
<thead>
<tr>
<th>Beam thickness (mm)</th>
<th>5</th>
<th>6</th>
<th>7</th>
<th>8</th>
<th>9</th>
<th>10</th>
</tr>
</thead>
<tbody>
<tr>
<td>Microstrain at 2mm deflection</td>
<td>2963</td>
<td>3556</td>
<td>4148</td>
<td>4741</td>
<td>5333</td>
<td>5926</td>
</tr>
</tbody>
</table>

7.3.1.3 Synergistic effect and ultimate flexural strength

It has been demonstrated in Chapter 5 that there is a synergistic effect in reinforcement under tension when two fibres with different modulus and elongation are used in hybrid form. This effect was also exhibited in the flexural tests. For example, with average volume fractions of 3.5% polypropylene and 2.1% chopped glass, the stresses at 4 mm deflection and at ultimate (about 9 mm) were 27 MPa and 33 MPa respectively for the hybrid composite. The corresponding stresses, however, were about 18 MPa (4 mm) and 23 MPa (9 mm) for the 3.5% all polypropylene cement, and were 0 MPa (4 mm) and 0 MPa (9 mm) for the 2.1% all chopped glass cement because the composite shown in Figure 7.7 with 2.4% chopped glass failed at 3.6 mm deflection with an ultimate stress of 16 MPa. Thus the direct mathematical addition of the two fibres (18+0 MPa at 4 mm, 23+0 MPa at 9 mm) was only about 67% and 70% of the combined effect.

A greater synergistic effect was demonstrated by the glass-polypropylene hybrid composite containing continuous glass rovings. With one such composite (No. 21 composite in Table 7.3) which had average volume fractions of 3.8% polypropylene, 0.51% continuous glass roving and 1% chopped glass (with the continuous glass arranged near the tensile surface, Series 7, Table 5.1), the stresses at 4 mm deflection and at the deflection at
Figure 7.12 Effect of glass fibres on notional flexural stress at 0.5% strain in chopped glass strand-polypropylene network hybrid composites, compared with all polypropylene cements. (The volume fraction of chopped glass in the hybrids, ranging from about 2% to 2.5%, was not plotted on the horizontal axis)
ultimate (about 8 mm) were 33.4 MPa and 44.5 MPa, respectively. The corresponding stresses, however, were about 18 MPa (4 mm) and 23 MPa (8 mm) for the 3.8% all polypropylene cement. For a 1.3% all-chopped glass composite and for a composite with the volume fractions of 0.5% chopped strand and 0.4% continuous roving, the stresses at 4 mm and ultimate were zero since the former composite failed at 2.8 mm deflection with a maximum strength of 11 MPa, and the latter one failed at 3.6 mm deflection with a maximum strength of 15.5 MPa (Figure 7.7). In an approximation, the mathematical addition of the three fibres (18+0+0 MPa at 4 mm deflection, 23+0+0 MPa at 8 mm deflection) was only about 54% and 52% of the combined effects, respectively.

A possible explanation for this synergistic effect has been described in Chapter 5. That is, the polypropylene networks, which have much greater strain capacity, enable the glass fibres to sustain their maximum load at higher deflections than would be possible on their own, and hence the hybridisation of the two fibres at a given deflection is more efficient than would be expected from the addition of the individual results.

Comparing the test results of Series 7 hybrids with those of other composites shown in Tables 7.2 and 7.3, it would appear that the composite performance can be significantly improved by the hybrid use of polypropylene networks, continuous glass rovings and chopped glass strands. Table 7.5 lists the flexural results of all eight samples cut from two sheets of Series 7 which quantitatively show the effect of roving volume fractions on the stresses. It would appear that even a small increment in the volume fraction of the continuous glass can result in an obvious increase in load bearing capacity of the composite. This effect was particularly noticeable at the ultimate stress in the last column of the table. The test results in the transverse direction, however, show that the aligned glass rovings in large-bundles do not benefit the transverse properties of the composite and may make them worse. The reason for this may be that the interfacial area between the roving and the matrix weakens the composite in the orthogonal
Table 7.5 Typical flexural results of single hybrid specimens

(Table 7.3, Series 7)

<table>
<thead>
<tr>
<th>Specimen No.</th>
<th>Volume fraction (%) of PP net Chopped net glass Cont. glass</th>
<th>Notional stress (MPa) at LOP 1 mm 4 mm Ultimate</th>
</tr>
</thead>
<tbody>
<tr>
<td>20-1</td>
<td>2.9(1.0) 0.8 0.51</td>
<td>13.9 17.2 29.5 43.5</td>
</tr>
<tr>
<td>20-2</td>
<td>2.8(0.9) 0.9 0.40</td>
<td>13.0 15.3 25.5 36.0</td>
</tr>
<tr>
<td>20-3</td>
<td>3.0(1.0) 0.9 0.33</td>
<td>12.2 15.4 25.2 35.6</td>
</tr>
<tr>
<td>20-4</td>
<td>2.9(1.0) 1.1 0.35</td>
<td>13.0 15.3 25.5 35.9</td>
</tr>
<tr>
<td>21-1</td>
<td>4.0(1.3) 1.0 0.58</td>
<td>14.1 17.9 34.0 47.6</td>
</tr>
<tr>
<td>21-2</td>
<td>3.3(1.4) 1.0 0.52</td>
<td>13.6 16.7 31.9 46.3</td>
</tr>
<tr>
<td>21-3</td>
<td>4.3(1.4) 1.0 0.41</td>
<td>14.5 17.7 30.2 40.9</td>
</tr>
<tr>
<td>21-4</td>
<td>3.6(1.4) 1.0 0.54</td>
<td>13.8 18.2 34.2 47.2</td>
</tr>
</tbody>
</table>

7.3.1.4 Cracking behaviour

Closely spaced multiple cracks were observed in all the glass-polypropylene hybrid specimens and the average crack spacings are given in Table 7.3, to be compared with that of the individual fibre composites in Table 7.2. It may be seen in Table 7.3 that, at similar volume fractions of fibres, crack spacing was greater for the "sandwich" hybrid (Series 5(2)), compared with other types of hybrids. This is because that there were some obviously wider cracks (one to four in number) which occurred in the multiple cracking zone of the sandwich specimens during the flexural process, probably, due to non completely uniform distribution of the 2-D chopped glass fibres which had been arranged in the extreme surface. Instead, for the hybrid specimens of Series 5(1) and 5(3) in which polypropylene networks were in the position of extreme surface, no particularly wide cracks were observed among the
densely spaced cracks. Polypropylene networks are effective crack arrestors in fibre cement sheets in that they can efficiently redistribute load, especially suddenly applied loads, and hence reduce the probability of composite failure at one or few major cracks caused by load localization.

7.3.2 Carbon-polypropylene hybrid composite
- flat specimens

Typical data measured from the experimental load-deflection curves of the carbon roving-polypropylene hybrid composite, of the glass roving-polypropylene hybrid composite and of an all polypropylene composite, both at 7 days and 28 days, are included in Table 7.6. Representative load-deflection curves of the composites at 28 days are shown in Figure 7.13. As all the composites shown in Table 7.6 or in Figure 7.13 contained about 2% polypropylene networks, the effect of variations (i.e. roving type, roving volume fraction) on composite properties can be easily observed.

There were no particular differences in the shape of the load-deflection curves between the carbon roving-polypropylene hybrid and the glass roving-polypropylene hybrid, except that the carbon-polypropylene hybrid had higher load carrying ability after cracking at a given deflection, and greater ultimate flexural strength achieved at a smaller deflection, compared with the glass roving-polypropylene hybrid (Figure 7.13). Needless to say, this was the result of greater modulus and tensile strength for the carbon rovings. The curve shape of the hybrid composites containing carbon rovings may be categorised into that shown in Figure 7.11 b). It was apparent from the carbon fibre pull-out length (Figure 5.50) that the bond with cement was poor and an improved performance would be expected if better bond could be achieved.

Figure 7.14 shows the notional flexural stresses of the composites in Table
Table 7.6 Typical flexural test results of continuous carbon-polypropylene hybrid composites and continuous glass-polypropylene hybrid composites tested at 7 and 28 days (Modified Surrey matrix)

<table>
<thead>
<tr>
<th>Composite Type</th>
<th>Volume fraction (%)</th>
<th>Limit of proportionality (MPa)</th>
<th>Flexural stress at 2mm deflection (MPa)</th>
<th>Ultimate flexural strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Polypropylene</td>
<td>Continuous roving</td>
<td>7-day 28-day</td>
<td>7-day 28-day</td>
</tr>
<tr>
<td></td>
<td>Networks 7-day</td>
<td>28-day</td>
<td>7-day</td>
<td>28-day</td>
</tr>
<tr>
<td>Carbon-PP hybrid</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1091C-1</td>
<td>1.9</td>
<td>2.2</td>
<td>0.35</td>
<td>0.39</td>
</tr>
<tr>
<td>1091C-2</td>
<td>2.0</td>
<td>2.2</td>
<td>0.72</td>
<td>0.74</td>
</tr>
<tr>
<td>1097C-1</td>
<td>2.0</td>
<td>2.1</td>
<td>0.40</td>
<td>0.34</td>
</tr>
<tr>
<td>1097C-2</td>
<td>2.2</td>
<td>2.3</td>
<td>0.82</td>
<td>0.85</td>
</tr>
<tr>
<td>Glass-PP hybrid</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Glass-1</td>
<td>2.0</td>
<td>2.1</td>
<td>0.37</td>
<td>0.40</td>
</tr>
<tr>
<td>Glass-2</td>
<td>2.0</td>
<td>2.1</td>
<td>0.71</td>
<td>0.70</td>
</tr>
<tr>
<td>PP only composite</td>
<td>2.0</td>
<td>2.1</td>
<td>0.0</td>
<td>0.0</td>
</tr>
</tbody>
</table>

* Series 6 in Table 7.3, continuous glass roving and polypropylene hybrid composite
Figure 7.13 Effect of fibre types and roving volume fractions on composite properties (Table 7.6, all the composites in the figure contained about 2% polypropylene network)
Figure 7.14 Effect of roving type and content on notional flexural stress at 0.5% strain. The volume fraction of polypropylene in all hybrid composites shown in the figure was about 2%
7.6 at a given tensile strain of 0.5% calculated using equation 7.1. At same volume fractions of 0.4% roving and 2% polypropylene, the notional stress for the carbon-polypropylene hybrid at 28 days was 19 MPa, and for the glass-polypropylene hybrid was 15.2 MPa, compared with 10.5 MPa for the all polypropylene composite. The effect on flexural stress by incorporating rovings especially the carbon roving was obvious.

7.3.3 Glass-polypropylene hybrid corrugated sheets

Representative load-deflection curves at midspan and quarter span are shown in Figures 7.15 and 7.16 for the corrugated PP sheets and hybrid sheets over test spans of 1380 mm and 1100 mm, respectively. Cyclic loading-unloading curves for the two types of corrugated sheets are illustrated in Figures 7.17 and 7.18, respectively. The deflection was the average of three readings of LVDTs either at midspan or at quarter span. The load was expressed as "kilonewton per metre sheet width (kN/m)", as well as "equivalent uniformly distributed (U.D.) load per unit area of sheet between supports (kN/m²)".

Important load-deflection data are summarized in Table 7.7 and a comparison of the polypropylene reinforced sheets and the hybrid sheets is given below.

1) The limit of proportionality (LOP) was greater for the hybrid sheets, indicating that the hybrid sheet could sustain higher applied load before cracking.

2) At given deflections, the hybrid sheet had greater load bearing capacity. For a 1100 mm span the loads at midspan were about 1.3 - 1.5 times, 1.6 - 1.7 times and 1.5 - 1.7 times that of the PP sheet at
Figure 7.15 Load-deflection curves at midspan and at quarter span of polypropylene network reinforced cement corrugated sheets tested under four point loading over test spans of 1380 mm and 1100 mm
Figure 7.16 Load-deflection curves at midspan and at quarter span of glass-polypropylene hybrid corrugated sheets tested under four point loading over test spans of 1380 mm and 1100 mm.
Figure 7.17 Cyclic load-deflection curves at quarter span and at midspan of all polypropylene reinforced cement corrugated sheet under four point loading over 1100 mm test span (7.43 mm sheet thickness)
Figure 7.18  Cyclic load-deflection curves at quarter span and at midspan of glass-polypropylene hybrid corrugated sheets under four point loading over 1100 mm test span (7.30 mm sheet thickness)
Table 7.7 Comparison in typical values from load-deflection curves for PP sheet and hybrid sheet (four point loading)

<table>
<thead>
<tr>
<th>Sheet type &amp; Number</th>
<th>Thickness (mm)</th>
<th>Test span (mm)</th>
<th>Load at LOP (kN/m)</th>
<th>Load at 30 mm deflection (kN/m)</th>
<th>'Ultimate flexural load or stress</th>
<th>Equivalent ultimate deflection recorded (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>PP sheet</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>01</td>
<td>6.93</td>
<td>1380</td>
<td>2.0</td>
<td>5.0</td>
<td>4.17</td>
<td>5.75</td>
</tr>
<tr>
<td>02</td>
<td>7.33</td>
<td>1100</td>
<td>2.5</td>
<td>5.9</td>
<td>6.46</td>
<td>7.10</td>
</tr>
<tr>
<td>03</td>
<td>7.43</td>
<td>1100</td>
<td>3.0</td>
<td>6.4</td>
<td>7.37</td>
<td>8.10</td>
</tr>
<tr>
<td>Hybrid sheet</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>01</td>
<td>7.11</td>
<td>1380</td>
<td>3.2</td>
<td>7.1</td>
<td>7.22</td>
<td>9.96</td>
</tr>
<tr>
<td>02</td>
<td>7.07</td>
<td>1380</td>
<td>2.9</td>
<td>6.4</td>
<td>7.28</td>
<td>10.05</td>
</tr>
<tr>
<td>03</td>
<td>6.92</td>
<td>1100</td>
<td>3.7</td>
<td>9.7</td>
<td>9.73</td>
<td>10.70</td>
</tr>
<tr>
<td>04</td>
<td>7.04</td>
<td>1100</td>
<td>4.4</td>
<td>10.3</td>
<td>10.68</td>
<td>11.75</td>
</tr>
<tr>
<td>05</td>
<td>7.30</td>
<td>1100</td>
<td>3.6</td>
<td>10.3</td>
<td>11.13</td>
<td>12.24</td>
</tr>
</tbody>
</table>

* For hybrid sheets, yes. For PP sheets the value represents the load at recorded ultimate deflection only, as the real ultimate deflection and equivalent load would be greater than that recorded in the load-deflection chart.

* U.D. - equivalent uniformly distributed load.
similar sheet thickness when deflection was increased to 10 mm, 30 mm and 50 mm, respectively. At 30 mm midspan deflection for 1100 mm test span, the average load was about 10.1 kN/m for the hybrid sheet, and about 6.2 kN/m for the PP sheet.

(3) For the hybrid sheets with thickness about 7 mm, a load of 10 kN/m was reached or exceeded at deflections between 40 and 60 mm at midspan under both test spans of 1100 mm and 1380 mm. This load would correspond to a flexural tensile stress in the sheets of about 20 MPa, calculated using the simple elastic theory. Instead, for the PP sheets, loads at midspan were about 7.6 kN/m at 57 mm deflection (1100 mm span) and 6 kN at 80 mm deflection (1380 mm span) which correspond to flexural tensile stresses in the sheets of about 12 MPa. As already discussed in Chapter 5 for the tensile results, the glass fibres in the hybrid sheets do not solely contribute to the load bearing capacity of the composite but also make more efficient utilization of its companion fibre. The same is true for the polypropylene.

(4) When unloaded at their "maximum" load, both PP and hybrid sheets showed remarkable recovery in deformation: about 60% or more of the total deformation was recovered at midspan deflections up to 60 mm (1100 mm span) or 70 mm (1380 mm span) for both types of sheets. This "post-elastic recovery" ability was attributed to the reinforcement by the polypropylene networks and continuous glass rovings and would benefit the sheets in dynamic properties and serviceability.

(5) Closely spaced multiple cracking, pseudo-ductile behaviour and great toughness was demonstrated by both types of sheets containing polypropylene networks. Soft body impact tests (7) showed that the glass-polypropylene hybrid sheets could resist a 50-kg-weight sand bag falling from a two metre height without crushing, under a test span of 1380 mm. Instead, using the same test method and sheet dimensions,
the impact body passed through cellulose-PVA sheets at about 0.8 m or less and through asbestos sheets at 0.4 m or less. Details concerning the soft body impact test is described in a French standard (8).

Regarding the cyclic load-unload curves in Figures 7.17 and 7.18, it is seen that the envelopes of the cyclic curves roughly correspond to those curves obtained from the normal load-deflection tests (i.e. Figures 7.15 and 7.16), implying that the short term load-unload cycles didn't yield extra plastic deformation in the sheets. The residual deflection increased with the increment of deflection after each cycle and the slope of each loading curve decreased with every cycle. It was observed from the cyclic load-deflection curves that the deformation recovery capacity when unloading was greater (say, 10% or more) for the hybrid sheet. The residual deformation was about 16 mm for the PP sheet and about 12 mm for the hybrid sheet when they were unloaded at common deflection of 40 mm.

Results presented in this Section for the corrugated sheets demonstrate once again that hybridisation of glass and polypropylene fibres in a common matrix produces superior engineering properties over one-type fibre only composite.

Durability is always a factor of great importance with roofing materials. Hannant's up to 10 year results (9) on polypropylene network reinforced cement have shown that there has been little change in the strength of the film during this time scale under conditions of both natural weathering and water immersion. Eight year results (10) on the corrugated PP sheets (Reticem) show that, after 8 years of exposure to Italian weather, the maximum flexural strength or the modulus of rupture (MOR) of the PP sheet remains unchanged and the ductility of the sheet has been little affected.

The weathering characteristics of AR-glass fibre-cement composites are much more complex because they depend on matrix alkalinity, on surface treatment of the fibres, on interfacial property between the two phases, and on the
weather conditions to which they are exposed. Four year results (10) on the corrugated hybrid sheets (Retiver) show that, after four year exposure in Broni (Italy) weather, there has been no reduction in modulus of rupture of the hybrid sheets, instead, there has been some increase, compared to the MOR at one month of same product. For instance, hybrid sheets of one of the trial productions at one month had an average maximum load about 7.5 kN/m (about 23 MPa) corresponding to an average deflection of 41 mm (1100 mm span, 3 point loading). It should be noted that in the early productions of the hybrid sheets the bond between the glass fibres and the matrix was not as good as at present and several innovative modifications have been made to improve the bond between the two phases since then. Nevertheless, the breaking load (MOR) in the young composite was still much higher than 4.25 kN/m, the minimum value required by the CEN standard (11). After four year weathering in Broni weather, the MOR of the trial hybrid sheets had increased to about 9 kN/m (about 27 MPa) at an equivalent deflection of 40 mm. The LOP of the weathered sheets was about 3.2 kN/m, slightly greater than 3.1 kN/m at one month. This result was rather surprising, as it is known that AR-glass fibres may deteriorate in a high alkali environment. It appears that the ageing for four years of the hybrid sheet containing glass fibres was not sufficient for alkali attack of the glass fibres by the highly alkaline Portland cement matrix to be come apparent. Additionally, microstructural changes between the glass filaments and the cement matrix will be taking place. It is possible that the improved bond between the large-bundled glass roving had increased the load bearing capacity of the hybrid sheets and this effect might be greater than any deterioration effect caused by alkali attack and hydrates deposition. The large-bundled rovings contained 6400 filaments per roving and it might take longer for the filaments in the roving to be filled by the hydration products, compared to single glass strands (200 filaments per strand). This could delay the alkali attack and embrittlement deterioration mechanisms. Four years of weathering is too short to draw a conclusion and long-term weathering tests on the hybrid sheets are in process.
7.4 Summary

The combined use of polypropylene networks and glass fibres can produce composite sheets (flat and corrugated) with higher LOP and improved flexural performance in the post-cracking stage compared with their individual composites, and can result in a synergistic interaction between the two fibres in a flexural system. Continuous glass rovings are more effective in improving the load bearing ability of the hybrids in the aligned roving direction, compared with 2-D chopped glass strands. Optimum composite properties may be obtained by properly arranging the polypropylene networks, continuous glass rovings and 2-D chopped glass strands in the common matrix. Upon removal of loads, a remarkable recovery of deformation is exhibited by both the all polypropylene composites and the hybrid composites. The improved stiffness and great toughness of the hybrid composite are important assets in its use as roofing or cladding material.

The major benefit from the use of carbon fibre rovings compared with glass fibre rovings in the polypropylene-based hybrid composite is the increased load carrying ability at given deflections. It is likely that this increase could be improved if better bond with the carbon fibres could be achieved.
8.1 General tensile stress-strain behaviour of fibre-cement hybrid composites

The stress-strain behaviour of four types of hybrid composites, with particular emphasis on glass-polypropylene hybrid composites, have been studied under uniaxial tensile loading. Results indicate that hybrid composites based on glass-polypropylene, carbon-polypropylene and glass-PVA yield stress-strain curves different from their parents' composites. The overall mechanical behaviour of the hybrid composites may be described in terms of five stages in the tensile stress-strain curves shown in Figures 5.13 and 5.14:

(1) Elastic stage, up to the point of first crack, where both the low elongation (LE) fibres and the high elongation (HE) fibres and the cement matrix are in their linear elastic stage. With the addition of LE fibres, the first matrix cracking stress of the HE fibre composite is increased. HE fibres also increase composite cracking stress.

(2) Matrix multiple fracture stage, in which the composite strain has exceeded the ultimate strain of the cement matrix and the matrix breaks into a series of parallel blocks. The strain at the completion of matrix multiple fracture and the crack width within the multiple cracking range can be considerably reduced by the inclusion of LE fibres compared with that for the composite made with HE fibres only. The effect on controlling and stabilising cracking by means of the LE fibres at the matrix multiple fracture stage can be important for practical roofing applications.

(3) Post matrix multiple-fracture stage, in which the fibres alone are carrying load and load is shared by the two fibres. This stage is terminated at the onset of the LE fibre failure. The apparent failure
strain of the LE fibres in the composite can be considerably increased by the presence of the HE fibres and this is termed the hybrid effect. This effect leads to a much higher load bearing capacity at the failure strain of the LE fibre in the hybrid than in either of the individual systems alone or of the sum of their components. Thus there is some synergistic interaction in which the presence of one of the fibres enables the reinforcing potential of another fibre to be more efficiently utilized.

(4) LE fibre multiple fracture stage, in which the composite strain has exceeded the ultimate strain of the LE fibres and the LE fibre may break in a number of locations. Transfer of load to the HE fibres is achieved upon failure of the LE fibres.

(5) Post LE fibre multiple-fracture stage, in which the HE fibre solely carries load and is stretched until ultimate strength of the HE fibre is reached.

No changes in the ultimate strength of the HE fibres due to the presence of the LE fibres have been observed. However, reductions in the ultimate failure strain of hybrid composites containing HE fibres compared with HE fibre only composites have been measured. The range of strain between the onset of the LE fibre failure and of ultimate composite failure depends on the relative volume fractions and moduli of the two fibre species. To achieve a favourable balance in properties between the two fibre species, the volume fraction of the HE fibre incorporated must be always above its appropriate critical volume in that composite to avoid instant failure of the hybrid on the failure of the LE fibres. This critical HE fibre volume in hybrids may be approximated by a proposed equation (equation 5.27) in the study.

A simplified theory, based on the conventional ACK theory for single fibre composites assuming linear frictional stress transfer, has been developed to describe the tensile behaviour of fibre-cement hybrid composites. The theory predicts the properties of hybrid composites in relation to the
properties of the two fibres and the matrix, and their volume fractions. It was shown that by substituting appropriate values for these parameters in the equations, the predicted tensile stress-strain curves agreed fairly well with the experimental curves or values, and many of the essential features of the proposed theory are reflected in the experimental stress-strain curves of the hybrids and especially in those curves for the glass-polypropylene hybrid composites. However, in areas where the test machine system significantly affects the stress-strain curves it is not possible to predict accurately the curve shape.

8.2 LE fibre bundle length, size and failure mode

When the LE fibres are used in continuous large-sized bundles or rovings, the failure mode of the fibre in the hybrid is a sudden event accompanied by a load drop in the observed load-strain curve. This load reduction is partly related to the loading rate and test machine stiffness. Such load drop is reduced when the LE fibre rovings are present at low volume fractions and when used in smaller bundle sizes. No sudden load reductions on the failure of glass fibres are observed in glass-polypropylene hybrid composites in which glass fibres are used in chopped strand form, presumably, due to relatively slow fibre pull-out when compared with sudden fibre fracture. We conclude that the failure mode for LE fibres in fibre-cement hybrid may be controlled through proper design as regards the volume fractions and geometries of the two fibres.

The reinforcing potential of the continuous glass and carbon fibre rovings are much better utilized in the presence of polypropylene networks than in single fibre composites. Similarly, when HE fibres, say, PVA, are used in the form of continuous rovings, the maximum stress of the rovings in the hybrid in the presence of chopped glass strands can be considerably greater than that in all PVA roving composite. This is because the presence of the chopped glass strands reduces the probability of failure of the composite caused by splitting cracking.
8.3 LE fibre positions

When the hybrids were produced from alternate layers of glass and polypropylene fibres the tensile performance, for the same fibre volume, was better than when all the polypropylene was placed in the core and the glass fibres were concentrated on the outside surface. In the latter case, local crumpling and debonding failure accelerated the damage process.

8.4 Flexural load-deflection behaviour of glass-polypropylene hybrid composites

Similarly to the results obtained in the uniaxial tensile tests, the combined use of LE and HE fibres yielded composite materials with higher LOP and improved flexural performance in the post cracking stage compared with their individual composites. Hence a similar synergistic interaction between the two fibres was found in a flexural system. Upon removal of loads, a remarkable recovery of deformation was exhibited by both the all polypropylene composite and the glass-polypropylene hybrid. The improved stiffness and great toughness of the hybrid composite are important assets in use as roofing or cladding materials.

8.5 Implications from the present work for the design of fibre-cement hybrid composites

Extensive data established in the current work have clearly demonstrated the superior properties of the fibre-cement hybrid composites over their parent composites, mainly in the following aspects: increased ability to suppress and stabilize cracking; improved load bearing ability both before and, particularly, after matrix cracking; increased apparent failure stress and strain of the LE fibres and thus better utilization of the reinforcing potential of the LE fibres; significantly improved toughness over that of LE fibre only composites.
Four types of fibre-cement hybrid systems have been studied. The major benefits from the use of continuous carbon rovings compared with glass rovings in the polypropylene based composites are the increase in tensile stress at 0.5% strain. It is expected that further improvements could be achieved with improved carbon fibre-cement bonding. The PVA-glass hybrid has superior properties over the polypropylene-glass hybrid in that the PVA-glass hybrid system allows similar or better composite properties to be achieved at much lower addition of the HE fibres which could be cost-effective. The PVA-polypropylene hybrid is advantageous among the four types of hybrids in achieving high ultimate composite strength.

It has been established by the research reported in this thesis that, to obtain the optimal balance between the properties of strengthening and toughening, the choice of fibres and their relative volume fractions are critical, since they are the dominant factors determining the shape of the stress-strain curves and governing the amount by which each of the fibres may contribute. It appears that the best choice of fibres consists of low elongation and high elongation fibres with the largest difference in failure strains but with elastic moduli as close as possible. To better utilize the strength and strain potential of the low elongation fibres, and to obtain any benefit of toughening from the high elongation fibres there must be sufficient of the latter present in the hybrid to ensure that the mode of failure is controlled by the fracture of the high elongation fibres rather than by the low elongation fibres. As long as this condition is satisfied the volume fractions of the two fibres may be chosen by taking into account the durability aspects of the components to give the desired engineering properties.
CHAPTER 9

RECOMMENDATIONS FOR FURTHER WORK

The work described in the thesis is an initial investigation of the stress-strain behaviour of fibre-cement hybrid composites with particular emphasis on glass-polypropylene hybrid composites for roofing and cladding. The focus of the research programme has been on the determination of the stress-strain curve shape and stress-strain relationships. A simplified theory is derived based on the assumptions of equal fibre strains in the two fibres at a crack to predict the tensile behaviour of fibre-cement hybrid composites. Although the predicted stress-strain curves were in good agreement with the experimental curves or values, the hybrid theory has room for improvement. Firstly, work is needed to identify and characterize the stress transfer mechanisms at each of the stage of the tensile stress-strain curve after the matrix cracks. Scanning electronic microscope and acoustic emission may be useful tools for such studies. Until the true stress transfer mechanisms are clearly understood, precise modelling of the hybrid composites is not possible. Also, an alternative hybrid model should be derived by assuming different strains at a crack for the two fibres in the hybrid.

Different failure modes of the low elongation (LE) fibres in the glass-polypropylene hybrids are observed when the glass is used in different forms (chopped or continuous). Limited results from the hybrids containing glass or carbon rovings also suggest that the bundle size of LE fibres affects the tensile behaviour of the hybrid composites. The effect of LE fibre bundle size and bond on the failure mode, failure strain and strength of the LE fibres in the hybrids is an important research area for further research. Further work may also be useful to investigate the effects of high elongation (HE) fibre type and different interfacial bond strength on the failure strain and strength of the LE fibres.
Concerning the use of glass fibres, it is worth while optimizing the cement matrix to increase the durability of the glass fibres in the hybrid composite.

With so widely dispersed fields in the building and construction industry for using fibre cement and concrete composites, and with so many choices for the fibres and the matrices, there are almost an unlimited number of potential hybrid composites. Truly this is a rich and extensive research area with great potential for the future.
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CHAPTER 1


CHAPTER 2


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CHAPTER 3


CHAPTER 4


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CHAPTER 6


CHAPTER 7


APPENDIX I

Portland cement 425 (Italcementi)

Physical test data

Setting time:
- Initial setting: 148 mins
- Final setting: 191 mins

Standard consistency water: 28.5%

Fineness specific surface (Blaine): 3910 cm²/g

Soundness: 0 mm

Compressive strength (cement:sand:water = 1 : 3 : 0.5):
- 3 days: 27.3 N/mm²
- 7 days: 37.3 N/mm²
- 28 days: 57.7 N/mm²

Flexural strength (cement:sand:water = 1 : 3 : 0.5):
- 3 days: 5.0 N/mm²
- 7 days: 6.1 N/mm²
- 28 days: 7.6 N/mm²

Density: 3.07 g/cm³