THE STRENGTH OF
MIXED FIBRE COMPOSITES

by

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This work considers the strength of the carbon fibre in unidirectional composites containing a mixture of glass and carbon fibres, commonly termed Hybrid Composites. It has confirmed earlier reports of enhancement of the failure strain of the carbon fibre in hybrids with respect to similar composites which contain only carbon fibre. The main objective has been to investigate the relationship between the mechanical properties and the ratio of the two reinforcing fibres and their state of dispersion. Hybrid-composites covering a wide range of dispersion and ratio have been fabricated and tested.

The strength of the carbon fibre reinforced plastic phase (cfrp) in the hybrid composites can be explained by a model which considers the statistical probability that the individual (micro) fractures which occur at flaws in the carbon fibres will lead to a gross fracture. This model allows the strength of the cfrp to be related to the strength of single fibres, and predicts the trend of decreasing strength as the volume of the individual cfrp component is increased. Laminates of equivalent size, but containing carbon fibre alone are weaker, and do not follow this trend. This behaviour is not fully understood but is thought to be governed by defects introduced during fabrication.

The work has also investigated the microscopic and macroscopic mechanisms of failure, with an emphasis on the redistribution of load when fractures occur in the cfrp components of the hybrids. This is important in understanding their overall load/extension behaviour and in determining whether failure is progressive and controlled as opposed to catastrophic.
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NOTATION

Abbreviations

AE
grp
cfRP
HMS
HTS
cdf

acoustic emission
glass fibre reinforced plastic
carbon fibre reinforced plastic
high-modulus surface treated (carbon fibre)
high-tensile surface treated (carbon fibre)
cumulative distribution function

Symbols - (specified by subscripts)

E
σ
ε
υ
P
P
\( t \)
L
D
a
s
x
T
Δ
Δt
S
\( \gamma \)
\( \tau \)
\( \beta \)
J
W
G
H
w
\( \sigma_0 \)
Γ
δ

Young's modulus
stress
strain
Poisson's ratio
fraction of volume reinforced with one or other type of fibre
thickness of a ply, or composite as a whole
length
dispersion \( (^{1/\tau}) \)

crack length
depth of defect
distance from a transverse fracture of cfrp component
temperature
difference
separation between glass- and carbon-plies at delaminated interface
stiffness, defined as \( S = Et \)
fracture energy term
shear stress (acting at interface between plies)
stiffness parameter:
\[
\left( \frac{1}{S_c} + \frac{1}{S_g} \right)^{-\frac{1}{2}}
\]
cumulative distribution function for failure of a body
cumulative distribution function for failure of fibres in a composite
cumulative distribution function for failure of a bundle of fibres
cumulative distribution function for failure of a composite
Weibull modulus
Weibull scaling parameter
Gamma function
ineffective length
K stress concentration factor

Subscripts

h of the hybrid composite as a whole (undamaged)
g of the grp component of a hybrid composite
c of the cfrp component of a hybrid composite
u ultimate
l at a specified length
r reference, or repeat
o limiting case, reference, scaling
m number of elements 'chained' in series
n number of fibres, or number of elements 'bundled' in parallel
D of delamination
f frictional
t of thermal origin
R with reduced load

Superscript

- mean
INTRODUCTION

1-1 GENERAL ASPECTS OF MIXED FIBRE COMPOSITES.

1-1-1 What are Hybrid Composites?

The term Hybrid Composites has recently been coined to describe a novel class of fibre-reinforced materials which contain two or more different types of reinforcing fibre in a common matrix. Materials reinforced with a single type of natural or man-made fibre have been used for many years, but with the advent of a greater variety of high-performance fibres, such as glass, carbon, alumina, boron and newer aramid organic fibres, the possibility of mixing fibres is attracting considerable attention. The main objectives of combining fibres are to achieve a balance of mechanical properties which cannot be realised in composites of one fibre type, and to produce structures which make optimal use of the expensive reinforcements. A wide range of structures tailored towards specific applications has been proposed, in which the degree of mixing of the fibres can vary from intimate combination at the fibre level, to selective reinforcement with large discrete volumes of a single type of fibre. The point at which structures containing a number of reinforcements can properly be called hybrid composites is the subject of some debate. A lively discussion at the first International Conference on Composite Materials in Geneva, 1976, is summarised by Phillips (1), and two years later at ICOM II in Toronto (11), the animated panel discussion on hybrid composites could still not produce a definition to better; "A hybrid fibrous composite material is a composite consisting of two or more fibrous materials in one common matrix, or one fibrous material in two or more matrices." However, a consensus of opinion held that; "Uniformity of dispersion of fibrous material(s) in the matrix distinguishes a hybrid material as contrasted with a hybrid structure." This distinction is not without point, because it is this question of dispersion which lies behind the interest in hybrids, and the attempt to classify them separately.

A number of recent studies of hybrid composites have shown that as the mixing of the fibres becomes more intimate, the resulting properties depart significantly from the simple predictions which are adequate for structural hybrids, containing large elements reinforced with a single type of fibre. In many cases composites of hybrid construction
have better properties than might be expected from the sum of their components, and this synergism has been termed the Hybrid Effect. The attraction of hybrid composites lies in the possibility of a bonus in performance, besides the ability to manipulate their properties through the combination of the different types of fibres, and their orientation.

1-1-2 Incentives to use combinations of reinforcing fibres.

The overriding consideration in the selection of materials is ultimately the overall cost effectiveness, and this is the main incentive to explore hybrid composites. The way in which a cost reduction can be realised depends very much upon the application, and generally involves many factors concerned with fabrication, performance, and end use. For example, a carbon and glass fibre hybrid composite might be considered as an alternative material to grp*. The substitution of stiffer and lighter carbon fibre for a proportion of glass fibre could give a component which is more economical in materials, cheaper to fabricate, lighter, stronger, stiffer, more fatigue resistant, smaller, or with other desirable properties, the balance of which is determined by the component's design. The economic benefits of choosing a hybrid composite, or indeed composites at all, depend on the premiums attached to such diverse factors as weight saving, performance, and lifetime of the final product. It is possible to make objective comparisons between materials only by complete analysis of the complex interactions between these factors in a particular application. One of the advantages of composites, and hybrid composites in particular, is that properties may be adjusted between relatively wide limits to meet specific requirements. Many examples of commercial and research applications, and design studies involving hybrid composites have been described in the literature, e.g. (2), and are reviewed by Summerscales and Short, (3). Design considerations for the use of composites in a number of important loading configurations are discussed by Rosen (12), and by the standard works in the Bibliography.

The stiffness, failure mode, and fatigue life are the aspects of the behaviour of hybrid composites which have received most attention, and appear to be the areas which hold the greatest promise for improvements over conventional composites. In this work emphasis will be placed on hybrids with two types of fibre, namely carbon and glass fibre. From Table 1 it can be seen that there are considerable differences between the major high-performance fibres now available, so there is considerable scope for mixing fibres to achieve intermediate properties. E-glass is

* Glass fibre reinforced plastics
by far the cheapest and most widely used reinforcing fibre, but although it has a tensile strength comparable with the other fibres, it is 3-5 times more compliant, and up to twice as dense. Offsetting these disadvantages it is comparatively inexpensive, at current prices about one tenth that of carbon fibre, which is also produced in quantity. In many applications the stiffness of a grp structure designed on a strength criterion is inadequate, and extra material has to be incorporated to increase the rigidity, bringing with it a weight penalty. The substitution of a considerable proportion of the grp with a smaller volume of a stiffer and lighter fibre such as carbon, can give a substantial increase in stiffness, especially if the stiffer fibre can be placed in the regions of the structure which experience highest strain. Typically grp panels or beams would have their bending stiffness increased by the incorporation of higher modulus ribs or skins, see for example (4). Very often stiffening and weight trimming exercises can result in further benefits in other parts of a structure, by reducing vibration loads and power demands on actuating equipment (13,14).

The other major area in which hybridization is of value is in the control of fracture. Many high-modulus, high-strength composites suffer from a tendency towards brittle fracture, which makes them susceptible to impact damage. The most notorious example of this must be the carbon fibre/epoxy fan blades designed for the RB 211 jet engine, which were unable to withstand rain erosion and impacts from bird ingestion. There is a particular problem with carbon fibre because increasing the surface treatment to improve the bond between fibres and typical resin matrices also results in a more brittle mode of failure. The incorporation of strong fibres with a high strain to failure, such as E- and S- glass, and the aramid fibres, into brittle composites generally improves their impact behaviour. This is discussed with specific reference to glass and carbon fibre hybrids by Summerscales and Short (3) who provide a useful review of recent research. A number of mechanisms have been proposed which involve crack-blunting, fibre pull-out, differences in the bond between fibre and matrix, and the higher strain energy at failure of the higher elongation fibres.

Carbon and glass fibre hybrid composites generally compare favourably with grp and cfrp in fatigue, but both improvements and reductions in performance have been reported (3,5). Low modulus composites such as grp can develop high tensile strains of the order 0.03 to 0.05 before their ultimate static strength is exceeded, and if fatigued at high strains suffer degradation of the matrix. By hybridizing with higher modulus fibres the
load bearing capacity is retained, but with reduced strains and prolonged life of the matrix in tensile fatigue, (5). However in transverse tensile and shear loading the higher modulus inclusions can concentrate the strain in the matrix and precipitate its failure.

The engineering of load bearing capabilities may not be the primary reason for mixing fibres. Carbon fibre has a small negative thermal expansion coefficient, while that of glass fibre is larger and positive, so hybrid materials with intermediate expansion coefficients may be produced by combining the fibres in suitable proportions. Electrically conducting fibres may be incorporated into insulating composites to prevent static build-up and to provide electrical screening, or to allow sufficient conduction for electroplating. Conversely insulating fibre may be used to break conduction paths and prevent corrosion in structures. Thermal conductivity and damping properties can also be modified by changing the proportion of fibres. Optical property differences may be exploited by using opaque fibres to screen fibres which are susceptible to degradation by radiation, or skins of relatively transparent composite may facilitate inspection procedures for sub-surface defects such as impact induced damage (15).

1-1-3 Scope of the present work.

Most of the examples given in the previous section of the diverse applications of hybrid composites exploit a unique balance of strength, stiffness, and toughness which can be designed into such a material. But even when a hybrid solution has been chosen on some other design consideration it is still important that the interactions between components should be favourable for these properties. This work is an investigation of simple hybrid composites which aims to identify the constitutional variables, and the interactions between components, which determine their strength and mode of failure. A unidirectional system with glass and carbon fibre in an epoxy resin matrix has been chosen for the following reasons.

i) The majority of work reported in the literature has been on hybrids of this type, and has produced well documented examples of unexpected interactions between reinforcing components.

ii) The system is of commercial interest because
   a) Both types of fibre are already in widespread use, and are easily combined as they are of similar size, and are available in compatible forms such as pre-preg, and tows of similar weight.
b) Properties of composites fabricated with the two types of fibre are very different, particularly in density and modulus.

c) There is a significant difference in the cost of the fibres, but it is not so great as to rule out the use of carbon fibre in grp structures because the contribution of materials costs to the total for the product, tends to be overshadowed by the cost of fabrication.

iii) The system is convenient experimentally because
a) Fabrication is relatively straightforward.
b) There are large differences in fibre properties.
c) The transparency of the glass fibre and resin greatly facilitate the observation of failure mechanisms within the composites.

The forms in which the fibres are available have largely determined the ways in which they have been combined in hybrids. Tows of several thousand filaments each, can be combined directly, or laminated as aligned sheets pre-impregnated with resin Fig. 1, or as a woven fabric (2). Mixing at the fibre level is technologically more difficult, but tows of fibre can be spread into very thin layers which may be combined before impregnating with resin, (7) and Fig. 1. The possible lay-ups illustrated in Fig. 1 are composed of thin layers or small bundles of fibres which are repeated many times to give a structure with a relatively uniform distribution of fibre. In this work composites similar to all these types have been investigated, but to facilitate observation of failure and subsequent analysis, somewhat simplified geometries have also been adopted, in which single layers or bundles of carbon fibre are incorporated into glass fibre composites, Fig. 2. With these specimens it has been possible to study both the characteristic mechanisms of failure, and the apparent enhancement of carbon fibre properties, the Hybrid Effect. The overall aims are:-

(i) To determine the extent of any Hybrid Effect, and its dependence on the constitution and fabrication history of a hybrid composite, in particular the fibre ratio and state of dispersion.

(ii) To understand the origin of differences in strength between hybrid composites with different geometries, and to relate this to theories of failure in composites.
To determine the mechanisms operating at all stages of the development of failure, from microscopic damage to gross fracture, and to understand how they are influenced by the constitution and geometry of the composite. The damage resulting from these mechanisms falls into two categories.

a) Microscopic damage which determines the strength at the first macroscopic failure of part of the hybrid's structure.

b) Secondary damage which is associated with macroscopic failure of the components of a hybrid, and which determines how properties are degraded before ultimate failure.

1-1-4 Unidirectional hybrid fibre composites in the context of composite materials: a perspective.

The primary concern in the drive to develop high performance composite materials has been the improvement of their stiffness and strength with respect to density. The reason composite structures are required to develop these high strengths is a consequence of the nature of the high modulus materials themselves, whose strengths are limited by defects. In general high stiffness and strength are conferred on a material by a high density of strong bonds between its constituent atoms, i.e. metallic, covalent, or ionic. The theoretical maximum strength of these bonds is limited in practice by fracture mechanisms associated with the defects. These mechanisms are brittle crack growth, and in some (predominantly metallic) crystals plastic flow by dislocation movement. The majority of high strength materials are not metallic for this reason, but have structures in which permanent shear deformation by dislocation movement is energetically unfavourable on account of charge displacement in ionic solids, or the highly directional bond in covalent solids. In these covalent materials, e.g. B, C, Al₂O₃, metal carbides etc., the lack of ready slip systems results in a shear strength which is much closer to the tensile strength than is the case in metals, and there is consequently little opportunity for the redistribution of load around defects in the structure, and these create stress inhomogeneities which are substantially above the mean stress level, thus limiting strength. Where the network is 2-dimensional, e.g. graphite, or effectively 1-dimensional, e.g. Kevlar, the slip systems (on c-planes in graphite, or between polymer chains) are perpendicular to the applied load in the maximum strength orientation, and do not therefore operate. For this reason many high strength fibres are highly anisotropic.
reduce the size of defects which can then only cause the failure of a limited volume of material, making a negligible contribution to the general degradation of the structure. The strength limiting defects can be decoupled from the remainder of the structure by dispersing the high-strength material in a number of ways, e.g. precipitation hardening, whiskers, platelets, or fibres. Fibres represent a special case because they are small in only 2 dimensions, and the utilization of their strength in the third direction depends upon maintaining load in the unfractured part of failed fibres.

In any 2- or 3-dimensional structure, other than a cable, load must be transferred between fibres in the directions perpendicular to their axes, and here two functions are accomplished by some form of matrix material. The matrix and fibre matrix bond must transfer useful loads between the reinforcing elements of the composite, but at the same time must be sufficiently weak to allow effective decoupling of fractures in the elements.

A fibre composite, although intrinsically strong in only one direction, is often required to carry loads in 2, or possibly 3 directions, and in these combined stress states the strength of a unidirectional composite in directions off the fibre axis may not be adequate, in which case fibres are combined in several directions. In 1- or 2-dimensional reinforcement with continuous fibres the full theoretical packing efficiency can be approached however, in 3-dimensional reinforcement the limit is 75% (i.e. 25% of the reinforcement in each of 3 directions) such structures are also very difficult to fabricate. Consequently most composite structures are unidirectional or 2-dimensionally reinforced rods or sheets. To analyse such structures it is necessary first to understand how the strength and stiffness of a unidirectional composite depend on those of its constituents, and then to relate these to the behaviour of the structure as a whole. The elastic properties of composite laminates have been reviewed by Kelly (Bib 1), Ashton, Halpin Pettit (Bib 6), Jones (Bib 7). In its most general form a fibre composite may be anisotropic as a result of anisotropy in the fibre or matrix, anisotropy in their disposition, or both. The generalised Hooke's Law for such an anisotropic material in a 3-dimensional stress state relates the (9 element) strain tensor to the stress tensor by an 81 element stiffness matrix, or its inverse, the compliance matrix.

\[
\begin{pmatrix}
e_{ij} 
\end{pmatrix} = \begin{pmatrix} S_{ijkl} \end{pmatrix} \begin{pmatrix} \sigma_{kl} \end{pmatrix}
\]

where S in the compliance matrix.
to the order in which deformations are made, it can be proven that the strain, stress, stiffness, and compliance (e, σ, S, C) matrices are symmetrical, i.e.

\[ \sigma_{12} = \sigma_{21} \text{ etc.} \]
\[ e_{12} = e_{21} \text{ etc.} \]
\[ S_{1122} = S_{2211} \text{ etc.} \]
\[ C_{1122} = C_{2211} \text{ etc.} \]

which reduces the stress and strain tensors to 6 independent elements each, and the stiffness and compliance matrices have 36 elements of which only 21 are independent. A shorter notation may be used to reduce the number of subscripts required as follows:

\[
\begin{align*}
\sigma_{11} &= \sigma_1 & e_{11} &= 1 \\
\sigma_{22} &= \sigma_2 & e_{22} &= 2 \\
\sigma_{33} &= \sigma_3 & e_{33} &= 3 \\
\sigma_{23} &= \sigma_4 & 2e_{23} &= \sigma_4 = \tau_{23} \\
\sigma_{13} &= \sigma_6 & 2e_{13} &= \sigma_6 = \gamma_{13} \\
\sigma_{12} &= \sigma_6 & 2e_{12} &= \sigma_6 = \gamma_{12}
\end{align*}
\]

Then \((\varepsilon_i) = (S_{ij})(\sigma_j)\) in contracted rotation.

In many cases symmetry of the composite allows the generalised Hooke's Law to be further simplified. If the composite is orthotropic, i.e. with 3 mutually perpendicular planes of mirror symmetry, the relation reduces to:-

\[
\begin{bmatrix}
\varepsilon_1 \\
\varepsilon_2 \\
\varepsilon_3 \\
\gamma_{23} \\
\gamma_{31} \\
\gamma_{12}
\end{bmatrix}
= 
\begin{bmatrix}
S_{11} & S_{12} & S_{13} & 0 & 0 & 0 \\
S_{12} & S_{22} & S_{23} & 0 & 0 & 0 \\
S_{13} & S_{23} & S_{33} & 0 & 0 & 0 \\
0 & 0 & 0 & S_{44} & 0 & 0 \\
0 & 0 & 0 & 0 & S_{55} & 0 \\
0 & 0 & 0 & 0 & 0 & S_{66}
\end{bmatrix}
\begin{bmatrix}
\sigma_1 \\
\sigma_2 \\
\sigma_3 \\
\tau_{23} \\
\tau_{31} \\
\tau_{12}
\end{bmatrix}
\]

i.e. 12 non-zero compliance or stiffness matrix elements, only 9 of which are independent in each case. If the composite itself has a plane in which properties are isotropic, for example aligned unidirectional material without planar structure resulting from lamination, the stiffness matrix can be further reduced to
with only 5 independent constants, (l is the fibre direction).

In applications where the composites are utilised as thin sheets in plane stress, through thickness stresses are assumed zero, i.e. $\sigma_3 = 0$ 
\[\tau_{23} = 0\]
\[\tau_{13} = 0\]

and the constitutive equation for a lamina reduces to

\[
\begin{pmatrix}
\varepsilon_1 \\
\varepsilon_2 \\
\gamma_{12}
\end{pmatrix} =
\begin{pmatrix}
S_{11} & S_{12} & 0 \\
S_{12} & S_{22} & 0 \\
0 & 0 & S_{66}
\end{pmatrix}
\begin{pmatrix}
\sigma_1 \\
\sigma_2 \\
\tau_{12}
\end{pmatrix}
\]

Laminated structures are built up from a number of thin laminae whose properties are given by this relation, and so long as they are thin enough to preserve plane stress conditions, consideration of boundary load conditions, and strain compatibility, enables the constitutive equation to be formulated for the laminate as a whole by rotation of the axis systems for each lamina as developed by Ashton, Halpin, Petit (Bib 6), Jones (Bib 7). These four matrix elements are related to the four engineering elastic constants:

\[
S_{11} = \frac{1}{E_{11}}
\]
\[
S_{22} = \frac{1}{E_{22}}
\]
\[
S_{12} = -\nu_{21}/E_{22} = -\nu_{12}/E_{11}
\]
\[
S_{66} = \frac{1}{G_{12}}
\]

When a laminate is composed only of a number of co-aligned laminae, (e.g. the laminates studied in this work), there is obviously no need to consider rotations between laminae as in the more general case, and its elastic behaviour is described by the lamina constitutive relation.
The general problem of relating the elastic constants of a unidirectional lamina to the elastic constants of the matrix and fibre, and their disposition, has been approached by many workers. Elementary bounds on elastic constants can be derived by making the simple assumption of strain or stress equality in the matrix and fibres and this leads to the well known series and parallel (Voigt or Reuss) models. These bounds are quite independent of the cross-sectional geometry. For circular or prismatic fibres randomly dispersed in a uniform matrix, good agreement between model and experiment is achieved only in the case of the parallel rule of mixtures (Voigt) bound for the longitudinal modulus \( E_{11} \). The simple bounds deviate significantly for all the other constants and are least satisfactory in the case of the in plane shear modulus \( G_{12} \).

Increasingly sophisticated bounding approaches, which take into account the structure of the composite, have been developed, and are reviewed in Bib 1, 6, 7. These can be classified into 'self consistent' models, and variational methods. A third exact method is to calculate stress fields numerically for rigidly specified arrays of fibres, from which the effective elastic constants may be obtained.

There are two versions of the self-consistent model. That developed by Hill considers a single fibre embedded in a uniform unbounded 'matrix' material whose elastic properties are macroscopically similar to those of the composite. The stress field around the fibre is calculated for uniform remote loading of the 'matrix' and the resulting expressions give reasonable estimates of the matrix dominated elastic properties \( E_{22} \) etc at low volume fractions. Frohlich, Sach, Kildinski, Flermans, have extended this approach to a fibre, surrounded by a concentric sheath with the elastic properties of the matrix, embedded in a uniform material with the elastic properties of the composite, and these models have been further extended to regular arrays of parallel fibres.

The variational methods employ elastic energy theories to obtain bounds on the elastic properties. The minimum complementary energy theories provides the lower bound (based on an applied stress) and the minimum potential energy theories yields the upper bond (based on an applied strain). Hashin and Rosen have derived expressions for all five elastic constants for transversely isotropic hexagonal and random arrays of fibres, but even these show poor agreement with experiment for \( G_{12} \). Hill has derived expressions for all 5 elastic constants which correspond to the elementary series and parallel models, and has shown in particular that the longitudinal modulus is given exactly by the parallel model when the Poisson's ratios of fibre and matrix are equal, and deviates by only a few percent for
work. Halpin and Tsai have developed equations to interpolate between
the exact numerical calculation for a variety of fibres and arrays,
and provide a semi-empirical link between analytical and experimental
determinations of the elastic constants. These equations provide
relatively good estimates for practical ranges of fibre volume fraction,
but are least accurate in the case of the shear modulus. The curve
showing the variation of Young's modulus with fibre orientation Fig. 83,
has been calculated from experimentally determined elastic constants for
the same type of epoxy resin matrix as has been used in this work.
1-2-1 Young's Modulus.

In any hybrid composite of the type shown in Fig. 1, both fibres contribute to the longitudinal stiffness, in proportion to their modulus, and their relative volume fraction. This is because in tensile loading all fibres have equal longitudinal strains, which leads to the simple parallel rule of mixtures expression for the modulus of the hybrid,

\[ E_h = E_g (1 - P_c) + E_c P_c \]  

where \( E \) is Young's Modulus of the composite in the fibre direction, and \( P \) the fraction of the cross-sectional area reinforced with a particular type of fibre, designated by subscript g or c for glass or carbon. Subscript \( h \) refers to the hybrid composite as a whole.

This expression shows that the modulus should be completely independent of the disposition of the reinforcing components, and this has been confirmed experimentally by Bunsell and Harris (6), Marom et al (8), Phillips (2), and Aveston and Sillwood (7) using carbon and glass fibre, and by Edwards et al (17) using HMS and HTS carbon fibre, and all these results are summarised in Table 2. The rule of mixtures expression (1) can be presented graphically, Fig. 3 giving modulus as a function of the fraction of the cross-section reinforced with each fibre, the two end compositions being grp and cfrp.

Rather small deviations from linear elastic behaviour at low strains have been attributed to changes in the modulus of the carbon fibre, and are discussed in section 2-3-2. In any loading configuration other than uniform tension or compression in the fibre direction, the modulus will be sensitive to the structure of the hybrid, and such a simple approach cannot be adopted. No evidence has been presented that conventional elasticity theory, for example laminated plate analysis (Bib.6), is unable to cope with these cases, (11). However after parts of a hybrid structure have failed the modulus will depend on the way in which load is redistributed within the unfailed portions.

1-2-2 The evidence for a hybrid effect.

The first macroscopic fracture in a hybrid occurs in the fibre with the lowest failure strain, which is in the carbon fibre in a carbon/glass system. This does not necessarily result in complete loss of load bearing ability. The tensile failure strain of the
carbon fibre component is usually greater than that of a similar unit tested in isolation. This enhancement of the carbon fibre failure strain is usually referred to as the hybrid effect, and is first credited to Hayashi (9), but has also been found by Bunsell and Harris (6), Zweben (10), Aveston and Sillwood (7), Marom et al (8), Phillips (2), and Edwards et al (17), whose results are collected and summarised in Table 2, and also in this work, section 2-3. Typical failure strain increases are between 5% and 50% in systems where tows of several thousand fibres, or where layers of roughly tow thickness (~150um) are combined, and up to about 100% where fibres are more intimately mixed. The compilation of Table 2 was frustrated by the omission by several authors of experimental details of the materials and fabrication procedures used (in particular cure temperatures), and of adequate description of their test methods and specimen designs. In several cases it is not clear what criterion for failure has been adopted for the stresses quoted, even when the authors note the progressive nature of failure. Some do not consider thermal strains cured into their composites, and few give any indication of the scatter of their results (probably ~10% when not indicated). All these points are summarised in Table 2.

Not all workers have adopted the same definition of a hybrid effect. That most commonly accepted is based on the common-sense assumption that the carbon fibre reinforced part of a hybrid would behave in the hybrid exactly as it does out of the hybrid. If it fails at a higher strain in the hybrid, the difference can be expressed as a percentage of the failure strain when tested in isolation, and called a hybrid effect, and this is the definition used in section 2-3-4. Other workers have adopted the view that the failure strain (whether of the carbon fibre or of the composite as a whole is not specified) might follow the parallel rule of mixtures in a similar manner to the modulus, as given by equation (1). This approach seems to ignore the fact that the fibres are loaded in parallel, with the same strain, and carry loads which are proportional to their modulus and cross sectional area, which is the assumption made in formulating the rule of mixtures equation (1) in the first place. For example Marom et al (8) note that the failure stresses of their hybrids are slightly below predictions based on the rule of mixtures, and conclude that they show no hybrid effect. Recalculation of their results to give the values of strain listed in Table 2, shows that the failure strains of their hybrids are greater than those of the all-carbon fibre controls by 5%-13%, and further there is a trend towards higher failure strains as the carbon fibre is dispersed as smaller units.
Both these approaches to the hybrid effect can be illustrated in a property/composition diagram. In Fig. 4 (a) the line AB represents the rule of mixtures prediction for failure stress, which is the sum of the stress contributions from glass and carbon components, lines AC and DB. The line XB represents the assumption of a constant carbon fibre failure strain, and is the sum of the carbon fibre contribution to stress, line DB, and that of the glass fibre at the ultimate failure strain of the carbon fibre component, $E_g \varepsilon_{\text{cu}}$, which would be given by a line XC. A similar diagram for strain at failure, Fig. 4 (b) shows the rule of mixtures prediction, line AB, which is again the sum of glass and carbon contributions, lines AC and DB. Line YB represents a constant value of carbon fibre failure strain. The first failure stress and strain of most of the hybrid composites in Table 2 lie in the areas ABX and ABY respectively, and it is a matter of terminology whether this is termed a hybrid effect.

The predictions from these diagrams relate to the first failure of the lower elongation carbon fibre component of a hybrid, but some simple predictions may also be made about the ultimate strength and mode of failure, see for example Aveston and Kelly (22), and Edwards et al (17). When the carbon first fails, its load is placed on the glass fibre in the cross-section containing the fracture. If this increased load exceeds the ultimate failure stress of the glass it too will fracture, and the failure will be catastrophic. The stress concentration arising from the carbon fractures might also cause catastrophic propagation of the crack through the glass fibre at a stress below its ultimate strength, but this is not simply predicted. The point at which the ultimate carbon fibre load equals the excess load bearing capacity of the glass fibre is represented by the intersection of lines AC and XB at Z in Fig. 4 (c), and by the condition

$$\sigma_{\text{gu}} (1-P_c) - \frac{E_g}{E_c} \sigma_{\text{cu}} - \sigma_{\text{cu}} P_c = 0$$

from which $P_c \text{ crit.} = \frac{E_c \sigma_{\text{gu}} - E_g \sigma_{\text{cu}}}{E_c (\sigma_{\text{gu}} + \sigma_{\text{cu}})}$ (2)

Below this critical fraction of carbon, $P_c \text{ crit.}$ multiple fracture of the carbon is possible with the glass fibre carrying the load where it bridges fractures in the carbon, and in this case the ultimate strength of the hybrid will be that of the glass fibre component. Ultimate, as opposed to
Relating the experimental results in Table 2 to diagrams of this type is not as straightforward as it might appear, because in most of the geometries studied alteration of the proportion of glass to carbon fibre has changed their state of dispersion and introduced an additional variable. In fact Edwards et al (17) base their analysis on the fact that the sizes of groups of like fibres changes as their proportions are altered. Marom et al (8) have tested 3 glass/carbon hybrids with the same 1:1 proportion of fibres, but with 1, 2 and 5 layers of similar fibres together, and their results indicate hybrid effects of 13%, 11%, and 5% respectively, though they make no comment on this. A variation of the hybrid effect with dispersion has also been found in the results presented in section 2-3-4. A further factor affecting the magnitude of the hybrid effect is the dependence of the failure strain of the carbon fibre control on specimen size. Theoretical and experimental work shows that the failure strain of composite materials should be described by a statistical distribution which takes account of the volume of material stressed, and its aspect ratio, rather than by a unique value which is intrinsic to the material. The underlying reason for this is the distribution of strength-determining flaws along the length of the fibres, and is discussed further in following sections. Zweben (10) and Edwards et al (17) consider this theoretically for the carbon contained in the hybrids but none of the works whose results appear in Table 2 have approached the problem experimentally for the case of the all-carbon control. Rather than comparing the failure strains of hybrid composites with all-carbon fibre composites whose failure strain is not well defined, it is probably better, where data for several hybrids are available, to compare hybrid with hybrid, and concentrate on the variations which result from differences in geometry. The comparison is then not sensitive to the value of failure strain adopted for the all-carbon fibre control. For example, Aveston and Sillwood (7) have noted that Bunsell and Harris (6) found large values of hybrid effect which probably result from their uncharacteristically low all-carbon failure strain of 0.003. In addition, where hybrid composites have been cured at elevated temperatures thermally induced strains must be considered. Aveston and Sillwood (7) avoided the problem by using a room temperature curing resin matrix. Bunsell and Harris (6) can account for only 10% of their 40% - 80% hybrid effect in terms of the residual compressive strain in the carbon fibre as noted by Zweben (10), but they offer no further explanation for the failure strain enhancement. The Kevlar 49/Thornel 300 system used by Zweben (10) results in tensile residual strains in the carbon fibre, which would then be expected to fail at a lower strain in the hybrid giving a negative hybrid effect: however, a
positive effect was found. The remaining authors listed in Table 2 taken no account of residual thermal strains, and no thermal expansion information is given for their systems. In summary, the results in Table 2 from glass/carbon, Kevlar 49/carbon, and HTS/HMS carbon fibre hybrid composites show:-

(i) in all cases, an enhancement of the failure strain of the carbon fibre reinforced component in the hybrid with respect to an all-carbon fibre composite tested in isolation.

(ii) where residual thermal strains have been considered they are too small, (or of opposite sign) to account for the enhancement in carbon fibre failure strain.

(iii) where hybrids have been fabricated with the same ratio of fibre types, but different degrees of dispersion, the hybrid effect is greater at higher dispersion, and cannot be accounted for by thermal strains which only vary with the ratio of fibre types.

From these published results it must be concluded that the hybrid effect is real, and not entirely of thermal origin, but there is a need for better data from experiments designed to systematically explore or eliminate constitutional, structural, and processing variables.

1-2-3 Load/extension behaviour.

The hybrid composites listed in Table 2 have been found to fail in a variety of modes which may be catastrophic or progressive depending on the ratio and types of fibres, and their disposition. A catastrophic failure occurs when the first fracture of the low elongation component causes an excessive load to be transferred to the remaining high elongation fibre, or when the stress concentration at the fracture of the low elongation fibre(s) is sufficient to propagate a crack right through the section. When multiple fracture of the carbon component occurs there is a progressive loss of stiffness. The failure modes, where described, are summarised in Table 2. The carbon/carbon fibre hybrids generally failed catastrophically in a 'brittle' manner with limited fibre pull-out, whereas the glass/carbon hybrids mostly showed multiple cracking. With progressive multiple fracture there is often debonding between glass and carbon plies as observed by Bunsell and Harris (6) or between fibre and matrix, Aveston and Sillwood (7). Phillips' (2) load strain curve also indicated progressive failure, but he does not describe the process in detail. In contrast, the similar system used by Zweben (10) containing
yarns of fibre failed catastrophically with no apparent pre-cracking of the carbon fibre. Perhaps the best example of multiple cracking is the photomicrograph in Aveston and Sillwood's paper (7) in which carbon fibre fractures are shown to be spaced at about 1mm intervals throughout the specimen, with areas of debonding either side of the fractures. The HMS/HTS carbon fibre hybrids (8) and (17) failed catastrophically in a similar manner to composites fabricated with a single type of carbon fibre.

For multiple cracking to occur load must be transferred back into the lower elongation component away from the fractures. This failure mode has been studied extensively for the case of a brittle matrix containing strong fibres by Aveston and Sillwood (7), Cooper and Sillwood (18), Aveston, Cooper and Kelly (19), Aveston and Kelly (20), Aveston, Mercer and Sillwood (21), McColl and Morley (27) and for laminated composites in which one lamina, usually a 90° layer, fails at a lower strain, by Aveston and Kelly (22) Garrett and Bailey (23,24) Parvizi (63) Parvizi, Garrett and Bailey (25), and Parvizi and Bailey (26). The essential feature in all these systems is the build-up of load around fractures in the low elongation component by elastic, plastic, or frictional means, or a combination of these mechanisms, so that some distance away loads and strains are again essentially uniform. This is shown schematically in Fig. 5 where the fractured component could be a brittle matrix bounded by strong fibres, a low elongation fibre or tow in a more extensible matrix, or a low elongation ply in a laminated composite. The form of the load build-up in the specific case of fractured fibres in a matrix is considered later, but the general behaviour with each load transfer mechanism is shown in Fig.5. A constant plastic or frictional load transfer mechanism results in a linear load increase (broken line), while an elastic shear lag or variable frictional force (depending on the normal force at the interface) gives a decreasing load transfer rate away from the fracture (full line). If the load transfer is plastic there will be a region of elastic load transfer at the end of the plastic zone. The spacing of the cracks during multiple fracture depends on the form of the load transfer, and the variability of the strength of the low elongation component. Fig. 6 illustrates the development of multiple cracking for a shear lag non-linear load transfer following the approach of Garrett, Bailey and Parvizi (23-26), and Aveston and Kelly (22), and the linear case is very similar. When the cracks are widely spaced there is a large region of relatively uniform loading between them, although strictly the maximum loading occurs midway between adjacent cracks. Fig. 6 (a). The precise position of another crack forming in this region in practice depends on the distribution of weak points in the low
elongation component, although the theories assume it will be midway between two pre-existing cracks. As this process continues to reduce the crack spacing the load on the low elongation component is reduced in regions adjacent to the cracks. Successively higher applied loads are thus required to initiate further cracking. The multiple cracking theory of Garrett and Bailey (23), subsequently modified by Parvizi and Bailey (26), shows that the crack spacing at a given strain may differ by a factor of two depending on the specimen length and position of the first crack, but in practice because of strength variations it is unlikely that cracks will occur precisely midway between existing cracks, which means that when the number of cracks is large the specimen will have a range of crack spacings, but the mean spacing will be midway between the bounds.

The effect of multiple cracking and load transfer between components is best illustrated with reference to a schematic load extension curve for a unidirectional hybrid, Fig. 7. Load and extension increase with a linear relation until the first fracture, which in a displacement controlled test causes a load drop because of the stiffness decrease. With continued extension the load rises at a reduced rate until a second fracture occurs at a load equal to or above that for the first. The shape of the curve between a load drop and the next fracture is basically determined by the reduction in stiffness, which may not be constant if there is propagation of any debonding which unloads the low elongation component. The magnitude of the load drop, and the number of fractures vary widely between hybrids, compare for example the load extension curves of Bunsell and Harris (6) which contain load drops ~ 8% and that of Aveston and Sillwood (7) in which the load drops cannot be distinguished, and also the curves in Chapter 2.

The effect of residual strains on the load/strain curve of a hybrid is shown in Fig. 8, which is the model used by Bunsell and Harris (6) to explain their hybrid effect. When a hybrid composite cured at elevated temperature is subsequently cooled a strain mismatch between the parts reinforced with each type of fibre will be locked in if their thermal expansions differ. The effective temperature at which the resin becomes solid may not be the maximum cure temperature, and must generally be determined experimentally. The proportion of the mismatch which appears as thermal strains in each component is a function of their proportions and moduli.
1-3-1 General, and thermodynamic aspects of strength.

Unidirectional fibre composites are a rather special class of materials because their tensile load bearing capacity is distributed over a large number of discrete load bearing elements, (the fibres), while at the same time their ultimate strength is often determined by a failure mechanism in which a crack passes through the material as if it were homogeneous. The problem of understanding the strength of these materials in terms of that of their constituents is a difficult one, and has been approached from two directions. The macroscopic fracture mechanics approach treats a composite as a homogeneous material, and applies linear elastic analysis of the body and loading mechanism to obtain a fracture condition in terms of intrinsic materials properties, which satisfies overall thermodynamic considerations. The microscopic approach recognises that the heterogeneous structure of the material can give rise to a mode of failure in which discrete fractures may occur in the fibres or within the matrix without causing gross fracture, and considers the (statistical) probability that a sequence of microscopic fractures will result in total failure at any given stress. Since both approaches attempt to explain the same phenomenon they should be compatible, and it is the overall thermodynamic considerations that link the two. The fracture mechanics approach indicates when fracture is energetically possible, and does not take account of mechanisms at the microscopic level, while the microscopic fracture processes can be considered to be the 'kinetic' factors which determine whether or not fracture will proceed along a thermodynamically favourable path. In theory the thermodynamic fracture mechanics approach could be extended to the microscopic level, but in practice the (statistical) variability of the position and strength of weak points in fibres becomes an overriding factor. In a sense statistical treatments can embody elements of a fracture mechanics approach by considering local load concentrations around fractured fibres, but this does not usually extend beyond nearest neighbours, and the models are insensitive to the larger scale structure, for example the arrangements of plies or free surfaces. Because the hybrid composites considered in this study have a range of structures which bridge the 'macroscopic' and 'microscopic', neither of these approaches alone is suitable. A further consequence of this is that the concept of tensile failure becomes imprecise when the composites suffer a progressive degradation of strength, brought about by the increase in size and number of small fractures which are neither large enough to precipitate complete separation, nor small enough not to interact with the structure of the hybrid.
The aim of this section is to present the energetic and statistical approaches to composite strength and their limitations, with emphasis on those aspects which are relevant to the discussion of the failure strain enhancement of the carbon fibre in hybrids, and the variation of the strength of all carbon fibre composites with specimen size.

Linear Elastic Fracture Mechanics (LEFM), which was developed from the concepts of Griffith on a thermodynamic argument to account for the influence of defects on the strength of 'brittle' materials, has been extended to cover the behaviour of cracks and other stress concentrations in materials which are normally considered to be ductile, or at least not ideally or completely brittle. For an introduction to the subject, and a development from Griffith's original theory, see Knott (Bib 5). Briefly summarised, the Griffith argument considers the total energy changes when a pre-existing crack loaded in tension extends by an infinitessimal amount $\Delta a$. Considering the body containing the crack and the loading mechanism as a closed system, there must be a free energy balance for an equilibrium extension of the crack. Equating the decrease in elastic strain energy and the work done by the loading system, to the surface energy required to create the new areas of crack, gives the fracture stress

$$\sigma_u = \left( \frac{2E \gamma}{\pi a} \right)^\frac{1}{2} \quad \text{or} \quad \left( \frac{2E \gamma}{\pi (1-\nu^2)a} \right)^\frac{1}{2} \quad (3), (4)$$

in plane stress and plane strain respectively. $E$ is Young's Modulus, $\gamma$ is the surface energy (single surface), $\nu$ is Poisson's ratio, and $a$ is the crack half-length. Both expressions give the well-known dependence

$$\sigma_u \propto a^{-\frac{1}{2}}$$

The Griffith argument considers only initial and final states when the crack advances a distance $\Delta a$, and is not concerned with the details of the fracture process, so it is a necessary, though possibly not sufficient condition for fracture. While the Griffith approach works well for brittle solids, it does not explain the strength of materials in which there is a zone of plastic flow, or other energy absorbing microstructural damage at the crack-tip. To take into account work done in this zone, Orowan re-wrote the Griffith equation in terms of a more general energy term, the critical energy release rate. When the energy release for unit extension of the crack, the energy release rate, equals the critical energy
The critical energy release rate, the crack can grow. The critical energy release rate is assumed to be a property of the material, but in practice may depend on factors such as temperature and rate of loading. This is a useful concept and can be applied to loading in tension (Mode I), and in shear perpendicular (Mode II) or parallel (Mode III) to the crack front. Having measured the critical energy release rate, LEFM would be used in a typical application to predict the minimum strength of a body containing cracks no larger than some particular maximum size, perhaps the smallest that can be detected by non-destructive inspection techniques. Direct application to unidirectional composite test specimens in this fashion is not generally possible for two reasons. Firstly, it is hoped that there are no significant pre-existing flaws in laboratory specimens. Instead, defects of increasing size develop by the successive fracture of fibres during loading. The energy release rate is a function of both applied stress, and the defect size, which is itself stress dependent. To predict when the energy release rate becomes critical requires a model for the growth of the largest defect with applied stress, which must involve the statistics of fibre-strength. Rosen (28) and Rosen and Zweben (29) present the fracture condition in a stress vs. defect size graph similar to Fig. 9, in which the intersection of the maximum defect size, and critical defect size curves, represents the point of failure. This concept is also discussed by Argon (34). In addition to the applied stress/defect size relation, the dependence of the failure stress on defect size must be known, and in general this does not follow precisely the $a^{-1}$ Griffith behaviour. A basic assumption of LEFM is that the crack tip zone is small in comparison with the crack, whereas in practice a rather large region of damage is often associated with a transverse crack in most unidirectional carbon fibre composites, and splitting at notch-tips is common (59). The critical energy release rate, or its equivalent expressed as a stress intensity, for Mode I crack opening, is usually estimated from the strength of specimens containing notches, or from the work of fracture in specimens specially designed to confine the crack to a single plane, such as the Tattersall-Tappin test piece, as described in (40,41).

The statistical models and experimental evidence presented later indicate that the critical defect is very small, probably a group of less than five adjacent fibre fractures, so the type of notch typically studied in laboratory specimens is not representative of the type of defect which determines the strength of undamaged composites, particularly hybrid composites in which the dimension of the carbon reinforced component can be of the same order as the defect size. Similarly the work of fracture is unrepresentative because it involves fibre pull-out on a scale which does
Although the LEM approach has been used with some success to explain the strength of composites, in particular angle-plied laminates, containing cut-outs or other stress concentrations its application to undamaged unidirectional specimens does not lead to realistic estimates of intrinsic defect size. For example, substitution in eq. (3) or eq. (4) of typical fracture energies for CFRP from Marom (8) and Beaumont and Harris (42) in the range 5-25 KJm^{-2}, with a fracture stress 1GPa, gives a critical defect length in the range 0.4-2mm. Cracks of this size would be clearly visible, and would be larger than the cross-section of the carbon fibre reinforced elements in many of the hybrids considered in this study. The largest defects to have been detected are an order of magnitude smaller. A possible explanation for this is that the crack-tip damage zone increases in extent as the crack grows, resulting in an increasing critical energy release rate. Microscopic evidence is presented later in support of this. There seems to have been no attempt to put experimental data into an integrated model embodying statistical and fracture mechanical aspects of failure along the lines of Fig. 9.

Aveston and Sillwood (7), Aveston and Kelly (20) Garrett and Bailey (25), and Bader and Manders (App.2) have used a very much simplified energetic argument to derive what is effectively a lower bound on the strength of the low-elongation component in hybrid composites, and cross-ply laminates, which fail by multiple cracking. Rather than considering a small extension of a crack, they take a full-size crack as the final state, and the unflawed composite as the initial condition, and calculate energy changes for the formation of the crack as a single event. Fig. 10 summarises the model in which a constant load is applied to the composite. Point A represents the condition just prior to fracture of the low elongation component, and B represents the final state after fracture. Just before fracture the stored energy of the specimen is given by the area OAD, but after fracture it has greater compliance, and greater stored energy, given by OBC at the failure load. The loading mechanism does an amount of work ABCD during failure, exactly half of which goes to increasing the stored energy of the specimen by an amount OAB. By simple geometry: OAB is half ABCD which leaves energy OAB free to meet the fracture energy requirements, to which it is equated to give the lower bound failure load. Equating these two energies effectively assumes that at all stages the propagation of the crack occurs in thermodynamic equilibrium. If the fracture does occur at the predicted load it implies that there is no energetic barrier to initiation of the crack, and that the critical energy release rate changes
In exactly the same way as the energy release rate as the crack grows. In practice this will invariably not happen, and the fracture will occur at some higher load determined by the mechanism of crack growth. The energetic argument can however give some insight into the possible influence of the ply-structure on the failure load. At an early stage of this work such an approach was developed, (App.2), but was found to be unsatisfactory, for example it is unable to account for the strength variations of non-hybrid specimens.
The requirement for statistical models of composite strength arises from the variability of the strength of fibres resulting from the random distribution of flaws along their length (44,45). The strength of a fibre is determined by its weakest cross-section containing the most severe flaw, and obviously the longer it is the greater the probability of it containing a flaw of given severity Fig. 55. In many cases the variability of fibres of a given length can be represented by an equation of the type proposed by Weibull.

\[ W(\sigma)_{1r} = 1 - \exp - (\sigma/\sigma_o)^w \]  

where \( W(\sigma)_{1r} \) is the cumulative distribution function (cdf) for the failure of fibres of some reference length 1r, \( \sigma \) is the fibre stress, \( \sigma_o \) is a scaling parameter which depends on the units employed, and \( w \) is a dimensionless constant termed the Weibull modulus. A fibre L reference units long can be considered as a chain of L units, and its probability of survival is (probability of survival of reference length)^L which is \( (1 - W(\sigma)_{1r})^L \). Thus the probability of failure, the cdf for fibres of length L is

\[ W(\sigma)_L = 1 - (1 - W(\sigma)_{1r})^L = 1 - \exp - L(\sigma/\sigma_o)^w \]  

The mean fibre strength from the Weibull distribution is

\[ \frac{\bar{\sigma}}{\sigma_o} = L^{-1} \Gamma(1 + 1/w) \]  

where \( \Gamma \) is the gamma function.

The Weibull modulus \( w \) is an inverse measure of the variability, and the coefficient of variation is given by

\[ \left( \frac{\Gamma(1-2/w)}{\Gamma(1-1/w)} - 1 \right) ^{1/2} \]  

which for large values of \( w \) can be approximated closely by \( \frac{1.2}{w} \).

By taking logarithms of eq.\((\gamma)\) to give

\[ \log \bar{\sigma} - \log \sigma_o = -\frac{1}{w} \log L + \log \Gamma(1+1) \]  

it is seen that \( \log \bar{\sigma} \) should plot as a straight line of gradient \( -\frac{1}{w} \) against \( \log L \), and this can be used to check if fibre strength data fits the Weibull distribution. The two Weibull parameters \( w \) and \( \sigma_o \) can be determined from the gradient and intercepts of this type of graph, or
Studies of the strength/gauge length dependence of glass and carbon fibres show imperfect agreement with the Weibull distribution. Metcalfe and Schmitz (43) tested various glass fibres of at least 12 different gauge lengths ranging from 0.25 -300 mm, and provide evidence for two flaw populations which give rise to two straight line segments of different gradient in the log strength vs. log length plot. Moreton measured the strength of carbon fibres prepared under clean-room conditions at three gauge lengths from 5 - 100 mm and produced results which do not plot as a straight line (45), and also showed that the strength at the longer gauge lengths does not behave as that for a chain of the shorter gauge lengths (44). Although the Weibull distribution is not a perfect description of fibre strength, it is convenient mathematically, and can be used for a limited range of gauge lengths. In the case of carbon fibres no better expression is available, and the inaccuracies introduced by its adoption in statistical strength models is small in comparison with other approximations.

This is a convenient point at which to introduce the Weibull probability graph as a means of displaying the behaviour of chains which contain weak links. The vertical axis is linear in \( \ln(-\ln(1-p(\sigma))) \), where \( p \) is the probability of failure, and the horizontal axis is linear in \( \ln \sigma^2 \), the dimensionless load parameter in eq. \((6)\). On these axes the Weibull distribution plots as a straight line, Fig. 9. The presentation can be used for the cdf's for the failure of single fibres, and bundles and chains of bundles of fibres in composite bodies. It has two features which make it particularly useful for this purpose. If \( m \) similar elements with a cdf \( J(\sigma) \) are loaded in series as a chain, the strength of the chain is given by the cdf \( J_m(\sigma) = 1-(1-J(\sigma))^m \) in a similar manner to eq. \((6)\). The vertical axis is linear in \( \ln(-\ln(1-p(\sigma))) \), and \( \ln(-\ln(1-J_m(\sigma))) = \ln(n) + \ln(-\ln(1-J(\sigma))) \) so \( J_m(\sigma) \) is \( J(\sigma) \) translated vertically by an amount \( \ln(m) \) without any change in shape. The vertical axis is therefore linear in log (number of elements chained). Size effects which occur when the strength of a body is determined by the weakest link in a chain of elements are easily accommodated by vertical translations of the cdf for the elements, and this is effectively what was done for single fibres between eq. \((5)\) and eq. \((6)\). The second useful feature is that by rotating the graph \( 90^\circ \) anti-clockwise, and relabelling
the probability axis to be linear in \( \ln(m) \), the graph can give the
median chain strength vs. chain length. By putting the origin of the
\( \ln(m) \) axis at the position of 0.5 probability of failure the graph becomes
one of \( \ln \) (median strength) vs. \( \ln \) (length) like Fig. 57.

The statistical models of failure presented in the literature
consider a unidirectional composite as an array of parallel fibres carrying
equal loads in a matrix material which bears negligible load. The first
failure event as load is applied is fracture of a fibre, after which there
are a number of possibilities for the progress of failure, of which three
special cases are considered by Zweben and Rosen (28,29,30,32). They identify
a 'weakest link' fracture mode in which the matrix transfers the load
which was carried by a broken fibre to adjacent fibres which fail under
the increased load. Fracture of the composite therefore follows from
the first fibre fracture which represents the 'weakest link', as considered
by Guceler and Gurland (37). In practice, for cfrp this criterion is over
conservative, and has been extended to fracture of the first overstressed
fibre as a lower bound to composite strength by Zweben (31,32) and Zweben
and Rosen (29,30). If the first fibre fracture is contained, further
fractures will occur at scattered positions throughout the composite
causing a general weakening, and ultimate failure may then occur when
the weakest cross-section is no longer able to support the applied load.
This is termed the 'cumulative weakening' mode by Zweben and Rosen
(29,30). The failure mode which is most relevant to this study is the
'fracture propagation' mode in which groups of adjacent fibre fractures
grow by successive fracture of neighbouring overstressed fibres until
one group of cracks becomes unstable and precipitates total failure. This
is the model considered by Zweben and Rosen (28,29,30,32) and Argon (34), and
more recently by Harlow and Phoenix (39) for fibre composites, and by
Scop and Argon (35,36) for laminates.

The key to fracture in this mode is the redistribution of load
between fibres when a fibre breaks. In the extreme case a fibre could
become completely decoupled from the matrix by debonding over its whole
length under the action of interface shear stresses, and the load it
carried would be evenly distributed between the fibres remaining intact.
The composite would then behave as a bundle of uncoupled fibres as
considered by Coleman (38). It is more likely that the load in the broken
fibre builds up away from the plane of fracture, so that a considerable
distance away it can effectively support the average fibre load. This
is analogous to the situation of a fractured ply in a laminate, Fig. 5.
Various models have been proposed for the redistribution of load at
fibre breaks, using an elastic shear-lag analysis, by Rosen (33) for a 2-D array of fibres, and by Hedgepeth and Van Dyke (47) for 3-D square and hexagonal arrays with and without plastic flow in the matrix, and by Van Dyke and Hedgepeth (48) for a special behaviour in which the matrix becomes completely debonded from the fibre at a prescribed stress level. More recently Ko (49) has performed a finite element analysis of a single fibre in a matrix material. The results of these analyses are usually incorporated into the statistical strength model in an approximate way by defining an 'ineffective length' (Rosen (33)) in which load in a broken fibre builds up to its average value. Generally the load concentration is only significant in close neighbours, but the exact value and form of load build-up depends on the matrix elastic and plastic properties, the modulus of matrix and fibre, and the fibre-matrix interface bond strength. Hedgepeth (46) has calculated a dynamic overshoot for the load concentration from 1.15 to a limiting value of 1.27 for the case of a 2-D array. No experimental information is available on the load redistribution in close packed 3-D arrays of carbon fibres, but Zweben (32) has found ineffective lengths in the range 0.78 - 2.2mm (to one side of the fracture) in a photoelastic study of a single layer of glass fibres in two different epoxy resins.

In essence the weakest link, cumulative weakening, and fracture propagation failure sequences belong to a single model which takes into account the redistribution of load, and the differences between them arise from the different load sharing rules adopted in each instance. This model, which is developed with a number of simplifying assumptions by refs. (28-30, 32-37, 39) considers a composite body as a chain of thin slices Fig. 11. The approach takes the cdf \( W(\tau) \) for the fibres at a length equal to the thickness of the slices, (usually the ineffective length), and applies, an appropriate load sharing rule, considering each slice as a bundle of fibres to derive the cdf for the bundles \( G(\sigma) \), and then derives the cdf for failure of a chain of these bundles \( H(\tau) \), which is the composite cdf. This last step is easily accomplished as demonstrated between eq. (5) and eq. (6). Generation of \( G(\tau) \) from \( W(\tau) \) is a problem of considerable mathematical complexity for all but trivial load sharing rules. It is also the stage in which the behaviour of the matrix as a coupling medium is introduced through the load sharing rule. Simplifying assumptions made at this stage are responsible for some of the conflicting results produced in the references listed above. The first approximation is made in adopting a thickness for the slices. The load sharing rule assumes that within a slice the broken fibres carry no load, and that other fibres are overstressed.
uniformly. This is in conflict with the analysis of load distribution cited above, but the incorporation of a more detailed stress field would make subsequent analysis completely intractable. Each slice is assumed to behave independently.

Harlow and Phoenix (39) have reviewed earlier workers statistical models, and have raised important questions about their assumptions, in particular the load sharing rules adopted. They choose a local load sharing rule, which is defined for all combinations of broken fibres in a slice, and which is compatible with the observations in Chapter 2. Their approach is followed in this work. By conducting exact computer simulations for bundles of \( n = 4 \) to \( 9 \) fibres they have provided powerful evidence that behaviour converges to a limit for \( 9 \) or more fibres, but this has not been rigorously proven. When a uniform load is applied to all fibres in the bundle, failure will occur if the strengths of the fibres permit a progression through some sequence of the \( 2^n \) possible configurations of broken fibres, with the proviso that broken fibres cannot be repaired. The mathematical complexity arises from the need to consider all sequences of fibre fractures, and to assign probabilities to each, applying the load sharing rule of each stage of the sequence. The probabilities of each sequence are summed to give \( G(\sigma) \). It is only practicable to write analytic expressions for \( G(\sigma) \) for small numbers of fibres (\( n \sim 5 \)), and in any case these are cumbersome and offer little insight into the nature of \( G(\sigma) \). Harlow and Phoenix (39) generate all \( 2^n \) states of broken fibres with associated load factors according to their local load sharing rule. This gives the load concentration on an unbroken fibre as \( K_r = 1 + \frac{r}{2} \) where \( r \) is the number of broken fibres immediately adjacent to it. The rule has two important features, the loads on unbroken fibres sum to the total load on the bundle, and the rule specifies load concentration factors for every configuration of broken and surviving fibres. The computation then generates all possible failure sequences and their probabilities at given load levels, and sums them to give \( G(\sigma) \) for the bundle. Their major finding is that the bundle strength behaves as if determined by a weakest link when the number of fibres is large. They assume weakest link behaviour for \( G_n(\sigma) \), and apply weakest link scaling back to single fibre size to obtain \( W_n(\sigma) \) according to

\[
W_n(\sigma) = 1 - (1 - G_n(\sigma))^{1/n} \tag{9}
\]

which is just a \( \ln(n) \) downward translation on the Weibull probability graph. \( W_n(\sigma) \) is the cdf for a single fibre when it is part of a bundle of \( n \) fibres. \( W_n(\sigma) \) for \( n = 1 \) to \( 9 \) is presented in Fig. 12(a) and (b) for \( w = 5 \) and 10 respectively, and apparently converges to a limit close to
of \( w \) from 3 to 50 is shown in Fig. 12 (c). This implies, but does not prove, that large bundles behave in a weak link manner, and \( G_n(\sigma) \) is given by

\[
G_n(\sigma) = 1 - (1 - W_n(\sigma))^n
\]

where \( W_n(\sigma) \) changes little for \( n > 9 \).

Two further points of interest are that the variability of the bundle strength is less than that of the fibre for large bundles, and that the \( G(\sigma) \) is not a Weibull distribution.

The approximate weak link behaviour means that for composites containing large numbers of fibres, combining bundles in parallel, i.e. making the composite wider, decreases its strength in the same way as making it longer by increasing the number of hypothetical slices. Combining equations (10) and (6) for \( n \) fibres loaded in parallel in each slice, and \( m \) slices chained in series gives the composite cdf.

\[
H_{mn}(\sigma) = 1 - (1 - W_n(\sigma))^{mn}
\]

\( H_{mn}(\sigma) \) is obtained graphically by translating \( W_n(\sigma) \) vertically by an amount \( \ln(mn) \). This is an important result because it indicated that the median strength of large composites is volume dependent, and decreases as the volume is increased, however this is accomplished. The analyses of Argon (34) and Scop and Argon (36) are in conflict as they predict a maximum in composite strength at some bundle size, and Harlow and Phoenix (39) attribute this to the load sharing rule adopted. (When \( n \) is small and \( m \) is large Harlow and Phoenix do predict an increase in strength, but for composites of interest here both \( n \) and \( m \) are large). Weakest link behaviour and the convergence of \( W_n(\sigma) \) near \( n = 9 \) imply that local accumulation of fibre fractures dominates any longer range interactions more than 9 diameters apart.

Experimental studies of the size dependence of the strength of composites have produced conflicting results. Scop and Argon (36) have shown an increase in strength with the number of parallel elements for 2-D arrays of up to 10 vapour-deposited boron films on polyimide sheets, and similar behaviour for 2-D arrays of glass fibres reinforcing packaging tape. Argon (34) compares this with the monotonic decrease of composite strength with volume of glass fibre (shown by Kies) for a range
Structures from a single fibre, to the Polaris 1st stage, and concluded strength passes through a maximum. Hitchon, McCausland and Phillips, and Hitchon and Phillips (50, 52) show that the Weibull equation adequately describes the difference in strength between tensile and hoop-burst CFRP specimens arising from differences in volume and stress distribution, but not in the case of flexural specimens, and they attribute this discrepancy to a change in failure mode. Hughes and Jackson (53) and Hughes Morley and Jackson (54) have shown that fabrication and testing procedures can have a very significant influence on apparent strength. In particular improving the alignment of fibres in the impregnated tow test by moulding inside shrink-fit plastic tubing can increase strength by a factor of \( \sim 1.5 \) (53), and the importance of packing and alignment of fibres is stressed in (54).
Before discussing the models which have been proposed to account for hybrid effects it is worth emphasising the nature of the strain enhancement which has been found in this work, and which the discussion in Chapter 3 will attempt to explain. In all the hybrid composites tested the first macroscopic failure event was a fracture of the low-elongation component, in this case reinforced with carbon fibre, with no indication, even at the microscopic level, of fracture in the high-elongation components (reinforced with glass fibre). The primary concern is to understand why the first and subsequent fractures of the low-elongation component occurred at a higher strain in the hybrid composites than is expected from its behaviour when tested separately.

Bunsell and Harris (6) have suggested residual thermal strain, but this cannot account for the full magnitude of the observed enhancement. Zweben (10) has proposed a model for a hybrid containing equal numbers of high- and low-elongation elements, which may be fibres or yarns of fibres, and predicts the strain at the first failure of a high-elongation element to give a lower bound for strength. Essentially the argument follows his previous bounding approach (31), but the difference lies in there being two types of fibre present with different strength distributions (assumed to be Weibull). The model is not directly applicable to the hybrid composites in this study because the failure criterion involves fracture of the high-elongation elements, whereas the strain enhancement phenomenon does not.

Aveston and Sillwood (7) consider the relaxations and release of strain energy when a fibre breaks for the two cases, where it remains bonded to the matrix, and where it debonds at a limiting value of interface shear stress. The energy which becomes available as a result of these relaxations is equated to the fracture energy of the carbon fibre to obtain a lower bound on the failure strain for the fibre. Using a fracture energy value for graphite predicts a failure strain for carbon fibres in a glass reinforced matrix of \( \sim 0.01 \), which agrees well with experiment. No prediction is made for the case of a carbon fibre in a carbon fibre reinforced matrix. In these analyses the resin and high-elongation fibres are lumped together as 'matrix'. The failure strain in the bonded case is proportional to the 'matrix' shear modulus to the power 0.25, and in the debonded case is proportional to the 'matrix' Young's Modulus to the power 0.5. Surrounding the carbon fibres by more compliant glass fibres should result in a lower failure strain for individual carbon fibres than when they are surrounded by stiffer carbon fibres.
The energetic argument does have application to laminates, and predicts a failure strain enhancement as proposed in Appendix 2. If a lamina of low-elongation material is tested in isolation, in theory, all the strain energy stored within the specimen is available to meet fracture energy requirements, but if the lamina is combined with high-elongation material which does not break, the relaxations and energy release at the fracture will be reduced, whether the high-elongation material has a higher or lower modulus. In this case a 'free' element is being compared with a similar one whose relaxations are 'constrained', and the reduction in energy release in the constrained condition may require a higher applied strain to meet the thermodynamic conditions for fracture.
EXPERIMENTAL WORK

2-1 INTRODUCTION

This chapter consists of a detailed account of the experimental techniques and the results obtained, together with a preliminary discussion of their significance. The detailed discussion is developed in Chapter 3.

The main elements of the experimental work were the fabrication of composite laminates, mechanical testing, and structural studies to determine the mechanisms of damage initiation and development.

One of the principal aims of the research was to study the effects of the level of dispersion of the two types of fibre in the laminate on its properties. For the initial experiments, it was convenient to fabricate the laminate from sheets of unidirectional fibres impregnated with a B-staged resin ("pre-preg"). These were available with three fibres, and a range of sandwich laminates was prepared from these materials. This technique was limited in that the ultimate level of dispersion was controlled by the thickness of the pre-preg (0.125 mm), and the dispersion was variable only through the thickness of the laminate. For this reason further laminates were prepared from separated and dispersed tows of carbon fibres using wet resin impregnation techniques.

Using the pre-preg system, specimens could be fabricated containing between $10^5$ and $10^4$ carbon fibres in a discrete layer, but to achieve greater dispersion of the carbon fibre, two further techniques were developed to reduce the minimum size of carbon fibre bundles in the composite. In the first of these, hybrid composites were filament-wound to include discrete bundles of between $10^4$ to $10^3$ fibres, obtained by subdividing the $10^4$ filament tow in which the carbon fibre was supplied.

The second of these techniques was to spread the carbon fibre tow into a 'veil', on average only 1.5 fibre diameters thick, but containing bundles of between 1 and $10^2$ fibres. Between all three fabrication techniques the range of bundle sizes covered is from 1 to $\sim 10^5$ fibres. Difficulties in handling bundles of less than a tow ($10^4$ fibres) meant
that work with divided tows and spread tows concentrated on obtaining
carbon fibre bundles of the required size within a glass-reinforced
test coupon of convenient standard dimensions, rather than covering a
range of proportions of the two fibres.

Basic evaluation of these hybrids in tensile tests showed that
the strength realized by the carbon fibre component could be increased as
the fibre was more highly dispersed and the aim of subsequent mechanical
testing was then to investigate in detail the various processes which
lead up to failure, and the way in which they are influenced by the
arrangement of fibres in a hybrid. Although similar mechanisms of failure
were found in all the hybrids, the range of scale involved is very large,
and the significance of each mechanism in the failure process as a whole
varies greatly between hybrids of different dispersion. For example,
cracking of fibres and matrix, normal to the tensile axis, was found in
all of the hybrids, but ranges in scale from breaks in individual fibres,
$\sim 10^{-10}$ m$^2$ in area, to fractures of entire laminae $\sim 10^{-5}$ m$^2$ in area.

Optical microscopy was successfully used to observe damage features over
this range of scale, both on polished sections, and in intact specimens,
and has been extended with an electrical etching technique to enable the
identification and location of individual fibre fractures. In addition,
the microscopy was supplemented with acoustic emission monitoring which
gave semi-quantitative measure of the very large number of microscopic
failure events. At the macroscopic scale the propagation of transverse,
and delamination cracks is of interest, and surface strains were measured
in the failure zones by using a laser moiré technique.

Overall, these techniques have been developed to provide the
qualitative and quantitative information from single fibres and composites
needed to assess models of the failure of hybrids with a wide range of
mixing of the fibres.
2-2-1 Materials

All the laminated composites were unidirectional and were fabricated from fibre pre-preg sheets supplied by Fothergill & Harvey Ltd. Two types of carbon fibre were used, high-strength carbon (HTS), and high-modulus carbon (HMS), manufactured by Courtaulds Ltd. Both fibres were surface treated to improve adhesion to the matrix. This treatment is proprietary and full details have not been released, but it is known to consist of an electrolytic oxidation treatment. The extent of this treatment has been subject to considerable variation by the manufacturer and of the batches of fibre supplied for this work, the HTS fibre had been treated to a "lower" and the HMS to a "higher" level of treatment. However the HMS fibre has lower intrinsic adhesion to the resin than the HTS so that direct comparisons are very difficult.

The third fibre was a Silenka 1200 TEX E-glass roving carrying an epoxy – compatible size. Fibre and resins are summarised in Table 3. The impregnating resin was a proprietary epoxide system consisting of a B-staged pre-condensate with a BF$_3$ – complex as an activator, (Fothergill & Harvey Ltd., Code 69). It had been applied to the fibres with a ketone solvent which was subsequently removed by evaporation.

The pre-preg sheets were designed to give a lamina thickness of 0.125 mm at 0.6 volume fraction of fibres.

The remaining wet lay-up composites were also unidirectional, and were fabricated from a single batch of HTS fibre, but not the same as that used for the pre-preg laminates. An anhydride-cured bisphenol -A based epoxy resin system was chosen for these hand-lay-up composites, and was formulated from Shell Chemicals, Epikote 828 (resin), Epikure NMA (nadic methyl anhydride curing agent), and BDMA (benzyl dimethyl amide) accelerator, in the proportions 100 : 80 : 1. A number of batches of these materials were used, which although nominally identical were, in the case of the resin, of slightly different colour.

2-2-2 Fabrication of laminates from "pre-preg".

Laminates were prepared by laying up the appropriate sequence of 100 x 200 mm sheets of pre-preg (fibres parallel to 200 mm dimension) in an open ended trough-mould, and hot pressing in a hand-operated screw press. The press was instrumented with a strain-gauge load-cell between the screw and upper platen to allow accurate control of pressure on the
Before assembly the inner surfaces of the mould were sprayed with a proprietary PTFE based silicone-free mould release agent (Rocol MRS), the pre-preg pack was inserted, and the mould was placed in the press under minimal contact pressure at 130°C for 10 minutes, to melt the resin, Fig. (13). It was then removed to a pre-heated vacuum oven and degassed for 15 minutes at 130°C, after which it was returned to the hot press at 130°C. The platen temperature was then raised to 170°C over a period of 10 minutes whilst the press was gently screwed down to close the mould onto metal stops which established the final laminate thickness. A pressure of 0.5 MPa was then maintained which ensured the production of flat laminates of uniform thickness. The excess resin and any remaining voids (bubbles) were expelled through the open ends of the trough mould, and the resin then gelled rapidly and cured over a period of 1 hour to give a void free composite. After this pre-cure the pressure was released and the mould removed from the press and allowed to air-cool to room temperature. The laminate was extracted, and then post-cured under light pressure for a further 3 hours at 175°C in an air circulation oven. Throughout this work, the term 'lamina' is used to describe the basic layer of reinforcement, for example a single cured layer of pre-preg, from which a 'ply' is built up. A 'ply' is a number of laminae with the same fibre orientation, and a number of plies with, for example, different fibre types, make up a laminate.

The vacuum degassing of the pre-preg pack was essential to obtain void free composites, since pressing without degassing inevitably resulted in foaming of the resin as the residual solvent in the resin boiled, although the manufacturer kept pre-preg solvent content to below 1.5% by weight. This procedure differs from that recommended by the manufacturer in that resin gel occurs after consolidation pressure is applied, so it is not critically important to reach the gel point and apply pressure before any solvent in the resin boils.

An unavoidable problem when simultaneously pressing a number of pre-pregs into a composite is that as the resin melts, the individual tows, which have been rolled flat during manufacture of the pre-preg, relax back to a more circular cross-section, and intermingle with the tows in adjacent laminae. The result is laminae of non-uniform thickness, although the overall dimension of the final plate is unaffected. When all pre-pregs are of the same fibre type this is not important, but when glass-fibre and
may become discontinuous, and effectively an assemblage of individual tows. The poor control of lamina geometry can be seen in the micrographs of laminate cross-sections, e.g. Figs. 75, 76. In the case of the laminates produced for mechanical property evaluation, it was considered essential to cure the hybrid laminates in one operation so as to ensure homogeneity and consistency of any cured-in stress, and of the properties of the interface between laminae. However, for the laser moiré work on delaminations between plies, the variations in ply-thickness obscured most of the rather small strain variations in the delaminated areas, and a two-stage pressing procedure was used to achieve greater uniformity of ply thickness. These three-layer hybrid laminates were produced by first pressing a single carbon-ply from pre-preg as outlined above, but without a post-cure, and then removing all traces of mould release agent by abrading with 400 grit wet carborundum paper until the surface would hold a water film. This ply was dried and cleaned with acetone, and assembled between outer plies of uncured glass fibre pre-preg for pressing in the normal way, followed by the standard post-cure, Fig. 42.

2-2-3 Fabrication of divided-tow hybrid composites by wet-lay-up

Glass fibre composites containing bundles of carbon fibre as either single tows or portions of a tow, were fabricated by a wet-lay-up procedure in the form of flat plates 200 mm x 200 mm x 2 mm thickness. Each plate contained 5 equally spaced bundles sandwiched between two 1 mm thick glass-fibre reinforced plies and provided 5 specimens approximately 20 mm x 200 mm.

A carbon fibre bundle of required size was gently separated as a metre length from an unsized tow, cleaned of loose fibres, and cut into 5 lengths which were attached under light tension with adhesive tape to a rectangular metal frame just large enough to fit over the trough mould. The bundle size was determined by weighing to an accuracy of ± 0.1%, and the practical difficulties of separating a metre of tow containing an equal number of fibres along its length restricted the range covered to just over a decade, from \(10^4\) to \(0.6 \times 10^5\) fibres.

Tows of glass fibre were wound by hand to the required thickness on two similar metal frames, and vacuum-impregnated with the pre-mixed resin. For impregnation, a light tray of aluminium foil was crimped around the two frames of glass-fibre, and about 100 ml of resin at room temperature were spread over the upper fibres. This
Assemblies were then placed in a vacuum oven pre-heated to 70°C, and rapidly evacuated before the resin melted and soaked into the fibres. In this way foaming of the resin was avoided, and complete wetting of the fibres took about 15 minutes. The impregnated frames were stacked in the trough mould with the carbon fibres sandwiched in the middle, (Fig. 13) and were manipulated with a blunt-edged tool to expel large air bubbles, and to work resin into the carbon fibres. The mould was assembled with stops as for the laminates, and heated in the hot-press as it was screwed down to expel excess resin with a pressure not exceeding 0.1 MPa. The temperature was held at 100°C, and as the resin gelled the pressure was increased to 0.5 MPa. After 1 hour pre-cure at 100°C the mould was cooled, and the composite was extraced and post-cured for 3 hours at 150°C under light pressure to prevent distortion. The resulting laminates were transparent and virtually void-free, see for example Figs. 63 and 77.

2-2-4 Fabrication of highly dispersed hybrids

Although the technique of simply dividing tows by hand could cover a decade of bundle size, it was limited to a minimum bundle size of ~600 fibres, and the composite produced offered no opportunity to study the interactions between bundles. Mechanical tests on the laminated hybrids indicated that strength improvements might be obtained with higher dispersion of the carbon fibre, and this led to the developments of a process to spread a single tow into a 35 mm wide tape which could be incorporated into a composite as a discrete lamina, with a mean thickness of about 1.5 fibre diameters. The tapes produced were not uniform, but contained collimated groups of between 1 and approximately 200 fibres, with a spread in alignment of ± 15%. The general appearance of a tape can be seen in Fig. 49.

The tow was spread into a tape by leading it through a divergent flow water jet. Collimation of the fibres was maintained by ensuring acceleration of water flow past the fibres which were fed through the nozzle at constant velocity. Fig. 14 is a schematic view of the apparatus with details of the nozzle design, and Fig. 15 shows the implementation of this. The fibre and water enter the nozzle through a circular hole 4 mm diameter and pass into a square channel of equal area. Then over a length of 70 mm the cross-section changes from square to rectangular, with the width of the channel increasing linearly to 35 mm at the exit of the nozzle. The cross-sectional area decreases linearly by a factor of three over the length of the nozzle to ensure a linear increase
The area decrease is accomplished by a hyperbolic decrease in thickness of the channel. The whole nozzle assembly was made from a transparent acrylic plastic for easy viewing of the flow pattern, and the hyperbolic profile was generated from a number of stepped cuts made with a milling cutter, and was finished with a template. Considerable care was needed to ensure accuracy of the exit slot dimensions (0.3 mm x 35 mm), which are critical for producing an even distribution of fibres across the width of the tape.

A water head of 5 m is required to generate stable laminar flow through the nozzle, and this is held in a vertical tube above the nozzle. Water is fed into the apparatus through two opposed jets just above the nozzle, and which helps to spread the tow. The major part of the flow is through the nozzle, but a proportion flows up the tube past the incoming tow to remove loose surface fibre which would clog the nozzle, and overflows at the top of the tube maintaining a constant head.

At the exit of the nozzle the spread-tow is drawn off horizontally over a curved-edged plate (of radius equal to the nozzle length) allowing the water to drain downwards. To prevent the wet tape being pulled together by surface tension, it is collected on a strip of gauze bandage, which passes under the radiused plate, and is reeled under light tension, together with the gauze as a carrier, on a motor driven drum. The water flow in the nozzle and the reeling tension pull the spread-tow through the apparatus as it is unspooled from a motor driven reel at the top of the water head, at about 100 mm min⁻¹. The reel of tape, still on the gauze, is then dried.

This tape was incorporated into composites in a similar manner to the divided tows. The tape, held onto the gauze carrier strip by loose fibres and the light size on the gauze, was pressed onto resin-impregnated glass-fibres in the trough mould. After a few seconds the fibres wetted and stuck to the resin, and the backing gauze could be peeled away leaving the fibres in place. Thereafter the process was the same as for the divided-tow composites, with fibre alignment maintained during pressing by flow of excess resin out of the ends of the trough mould.

2-2-5 Range of composites fabricated

For evaluation of strength and failure mode a basic 3-layer construction was adopted for the hybrid laminates with the carbon-fibre component sandwiched between two glass-fibre reinforced plies of equal thickness. This simplified fabrication, and facilitated observation of
and also, since the carbon fibre failed before the glass fibre, ensured that the specimens remained balanced after failure. The aim was to investigate the effects of the ratio and dispersion of the two fibre types and this was done within the limitations imposed by the choice of fabrication technique. The pre-preg system allowed ply-thickness to be varied, only in quanta of one pre-preg layer, (0.125 mm) and dispersion could be varied therefore only in one dimension, through the thickness of the laminate. Similarly the spread-tow laminates extended dispersion of the carbon-fibre only in one dimension. In the divided-tow hybrids the carbon-fibre bundle assumed an approximately oval cross section.

With the pre-preg system 3-layer hybrids were produced with between 3 and 25 laminae, using HTS-carbon and E-glass, HMS-carbon and E-glass$ and HMS-carbon and HTS-carbon fibre. In addition, control laminates of a single fibre type were fabricated in a number of thicknesses. Divided-tow hybrids were made with four bundle sizes, and spread-tow hybrids with two thicknesses of glass-ply, the thinner being more convenient for microscopy through the outer layers. The spread-tow and divided-tow hybrids utilized only HTS fibre. All these composites are summarised in Table 4.

Tensile specimens were prepared from the composite plates by diamond-sawing into parallel strips, 10 mm x 200 mm long in the case of pre-preg laminates, otherwise 20 mm x 200 mm. Aluminium alloy end tags were bonded on to these strips as shown in Fig. 2. Aluminium alloy was found to be the most suitable material for end tags as it reduced grip-generated noise which interfered with acoustic emission monitoring.

Such specimens are special in that the scale of dispersion is of the same order as the specimen dimensions, because in the main they are comprised of only 3-layers. However, these 3-layer hybrids can be visualised as a basic unit from which a more extensive hybrid component could be made, consisting of many repetitions of the basic unit as in Fig. 16. This concept is used as the basis for a definition of dispersion. When comparing the properties of hybrid composites two fundamental constitutional parameters need to be considered, the hybrid ratio, and the state of dispersion. The hybrid ratio is expressed as the proportion of the total fibre volume represented by each reinforcing fibre. (E g. for a 1 : 1 : 1 glass-carbon-glass laminated hybrid the carbon fraction $P_C = 0.33$, and the glass fraction $P_g = 0.67$). For laminated hybrids dispersion through the thickness is defined as the reciprocal of the thickness (in metres) of the smallest
\[ D = \frac{1}{t_r} \]

In the case of simple 3-layer sandwich laminates \( t_r \) is the total laminate thickness Fig. 16. With this definition \( D \) increases as the laminae are more intimately mixed. This concept can be extended to 2- or 3-dimensions along the lines of the reciprocal lattice used in crystallography, though, of course, 3-layer hybrids effectively have no dispersion across their width. These definitions of \( D \) and \( P_c \) have the merit that they are independent variables, and may be used to represent the range of laminated hybrids graphically.

Fig. 17 is a 2-dimensional representation of the range of laminates studied. The left-hand vertical axes show dispersion, and the number of pre-preg laminae in the thickness \( t_r \) respectively, whilst the horizontal axis is the carbon fraction \( P_c \). The sets of hyperbolae represent the two restrictions, that all laminates be constructed from integral numbers of laminae, and that for a balanced 3-layer construction there must be an even number of glass laminae. Laminate geometries satisfying these twin conditions are found at the intersections of the hyperbolae, and those fabricated are indicated by solid circles. The open circles represent multiple-layer-laminates.

An alternative set of axes, the numbers of carbon and glass laminae, could equally well be used to represent the data, but this would map lines of constant dispersion and carbon fraction into sets of hyperbolae. This is inconvenient as the density, cost, and modulus of the composite depend linearly on \( P_c \), so it is desirable to keep this variable as one of the axes, this convention is also used in the presentation of mechanical properties.

The divided-tow hybrids are such a special case of a more general type of hybrid in which many tows or smaller bundles of each fibre are mixed in 2-dimensions, that a rigorous application of the dispersion definition is of little value, and bundle size is a better parameter against which to present test results. However to see where these hybrids lie in relation to hybrids laminated from pre-preg and spread-tows, it should be recalled that single carbon-fibre pre-pregs split into tows during moulding, and that the divided-tow hybrids cover a decade of dispersion from single tows upwards towards that of spread-tows. Additional pre-preg laminates containing pre-fractured carbon plies were fabricated for the laser moiré work.
2-3-1 Tensile Testing, strain measurement, and acoustic emission monitoring

All tensile specimens were strained in a model TTD/Instron universal testing machine, using a constant crosshead speed of 16.7 m/s which gives a strain rate of approx. $170 \times 10^{-6} \text{ s}^{-1}$. Strain was measured with an electrical resistance strain gauge bonded directly to the specimen gauge length.

The gauge was connected into a bridge completion and balancing network whose output was fed into the Instron strain amplifier to drive the paper feed of the machine's chart recorder. Before each test a calibration resistance was temporarily connected across one of the bridge arms to simulate 0.02 strain. The value of this resistance was calculated from manufacturer's data for the gauge factor, and was not changed throughout the experimental programme. The strain measurements were not calibrated against an external standard, but are self-consistent. Errors due to variations in gauge factor and the value of the calibration resistance should be less than 2%. Both wire (10 mm and 30 mm) and foil (6 mm) strain gauges were used, the former up to 0.02 strain, and the latter up to at least 0.03 strain. All gauges were bonded to specimens, after lightly abrading and degreasing, with a cold-cure epoxy adhesive (Araldite).

Acoustic emissions were monitored during straining of the specimens. These were received by a piezoelectric transducer held in contact with the specimen, and acoustically coupled to it with a thin film of petroleum jelly. The transducer has a resonance in the region of 50 KHZ, and the response of the specimen-transducer combination to a train of impulses is a series of damped oscillations (66) called ring downs, Fig. 18. In the monitoring system used, Fig. 19, the direct output of the transducer is amplified (by a constant factor) and fed to a level detector which outputs a pulse each time a pre-set threshold is exceeded. The number of pulses in a gating period of 2 s is counted, and converted to an analogue signal for output to a chart recorder. The quantity displayed is the count rate, or number of counts in the 2 s gating period, but this is not the same as the rate at which impulses are received by the transducer. Because of the ring-down response of the transducer and specimen, more than one pulse, (or no pulse at all) may be output by the level detector, depending upon the amplitude of the acoustic impulse. The recorded count rate therefore depends upon the amplitude of the acoustic impulses, which depends on the proximity of acoustic activity and the attenuation properties of the
sensitive channel of the equipment was used. The next least sensitive channel has a 20 dB higher detection level, and registered a count rate lower by a factor of roughly $10^4$.

In the earlier laminate tests load/strain, and acoustic emission (A-E)/time traces were obtained on separate recorders, but for the divided-tow and spread-tow hybrids both A-E rate and load were recorded against strain on a two pen chart recorder.

2-3-2 Elastic properties of composites before failure

The Young's modulus of the composites was calculated from the gradient of the load/strain curves which were linear over the greater part of their range up to failure. At low strains ($\sim 0.001$) the modulus of all-carbon fibre composites was about 10% less than at strains nearer failure, and this non-linearity has been attributed by Reynolds (61) to realignment of fibrils within the carbon fibre. This non-linearity is found in the hybrids in proportion to the amount of carbon fibre present. Moduli for all-carbon fibre, all-glass fibre, and hybrid composites are presented in Table 4, and are closely in agreement with that predicted by the parallel rule of mixtures. The modulus of a hybrid,

$$E_h = E_c P_c + E_g (1 - P_c)$$

where $E_c$ and $E_g$ are the moduli of the carbon and glass reinforced components, and $P_c$ and $(1 - P_c)$ are respectively the fractions of the volume reinforced with each fibre.

2-3-3 Failure strain and mode of failure of composites with a single type of fibre

The glass-fibre and carbon-fibre composites tested presented two very distinct modes of ultimate failure which reflect differences in the failure strain and surface properties of the two fibres. The carbon-fibre composites failed by the propagation of a macroscopic crack perpendicular to the tensile axis with limited longitudinal splitting, Fig. 20. Although this type of fracture is often termed 'brittle', the fracture surfaces reveal substantial fibre pull-out, and in this instance brittleness refers to the fact that the failure is sudden and catastrophic, rather than progressive.

This failure is typical of relatively high modulus carbon-fibres with good adhesion to the matrix. Glass-fibre composites failed
at approximately three-fold higher failure strains, in a more progressive mode which involved much splitting parallel to the fibres, Fig. 21. Typically the splitting first occurred at bundles of misaligned fibres, and rapidly accumulated throughout the gauge length leading to ultimate failure. Such failures are characteristic of relatively high elongation fibres with low adhesion to the matrix.

The failure strains and stresses of all the single fibre type composites are included in Table 4, and it is seen that both carbon-fibre and glass-fibre composites have comparable failure stresses, (~1.2 GPa) since the higher modulus of the carbon-fibre compensates for its lower elongation.

The most significant result from these tests is that the failure strain of the carbon-fibre composites is a function of laminate thickness, whereas that of the glass-fibre laminates is apparently constant over the same thickness range. To present this result more clearly, the failure strain of a 12-ply carbon-fibre laminate has been taken as a standard, and the differences of the failure strains of the other laminates from this standard are presented as percentages in Table 4, and plotted against D and P_c in Fig. 22. The failure strain falls approximately exponentially by about 20% for each halving of the laminate thickness. This behaviour is termed the Size Effect.

The A-E rate traces for the two fibre types, Figs. 23, 24 have similar forms and rise to comparable levels at failure, but in the case of the carbon-fibre composite the whole trace (0 - 0.01 strain) is accommodated within one third of the strain to failure of the glass-fibre composite (0.03). Carbon-fibre failures, and matrix cracking are the two major sources of acoustic emissions, and are discussed further in section 2-3-11. Damage to the specimen in the Instron wedge grips, (the major external source of A-E) has been virtually eliminated by the use of soft Al alloy end tags. In the range 0 - 0.01 strain the low level of A-E from the glass fibre specimens is mainly from matrix failure, so over the same strain range the much higher A-E from the carbon-fibre specimens must be associated with the carbon fibres or different interface behaviour. There is substantial evidence from microscopy that this A-E originates in the failure and subsequent debonding of the carbon fibres.

2-3-4 Failure strain and mode of failure of laminated hybrids

As the strain applied to the laminated hybrids was increased,
The first macroscopic failure was fracture of the carbon-ply. This event is shown schematically in Fig. 25 in which a transverse crack extends the full width and thickness of the carbon-ply, normal to the tensile axis. Where this transverse crack meets the glass plies there are delaminations along the interface in both directions for a distance $L_D$.

The length of these delaminations depends critically upon the sandwich geometry, as can be seen in Table 4, and ranges from about 3 mm for one lamina-thick carbon-plies to over 100 mm (greater than the gauge length) for thicker carbon plies. The delamination length decreases as $D$ is increased and $P_C$ reduced. The factors determining the extent of delamination are considered in detail in section 2-3-6.

The transverse crack has a cruciform cross-section, Fig. 26, which suggests that the crack front initiates near the central plane of the carbon-ply, and propagates simultaneously through the thickness and across the width of the ply. A consequence of the cruciform crack is that small volumes of carbon-fibre composite, triangular in cross section, remain attached to the glass plies, and trace the path of the transverse fracture. This is seen clearly in Fig. 25 and the scanning electron micrograph Fig. 27, which shows the short lengths of carbon-fibre adhering to a delaminated glass-ply. The transverse crack also branches to a limited extend across the width of the carbon-ply, Fig. 28, indicating the probable direction of the fracture.

The first failure of the carbon-ply was detected visually by the appearance of delaminations, which appeared lighter than surrounding undamaged laminate, and by the audible 'ping' which accompanied the fracture. The 'ping' was also detected by the A-E monitoring system, but made an insignificant contribution to the count in the most sensitive channel, and could not be reliably distinguished from occasional pulses arising from grip slippage, and end-tag debonding in the less sensitive channels.

The strain at which the first carbon-ply failure occurred was not constant, but was found to be dependent on laminate geometry. Details for HTS- and HMS-carbon/glass hybrids are given in Table 4. The more highly dispersed, and lower carbon fraction hybrids have higher carbon-ply failure strains, and this variation has been termed the Hybrid Effect. The Hybrid Effect can be expressed as a percentage increase in failure strain in comparison with a standard carbon-fibre laminate consisting of 12 laminae (as was done for the Size Effect), and is listed in Table 4,
The measured range of delamination lengths is rather large, from several millimetres to greater than the gauge length, and in this latter case continued loading of the completely delaminated specimen after the carbon-ply failure resulted only in failure of the remaining glass plies at their ultimate strength, and the carbon-ply had no further influence on the load carrying capacity of the laminate. When the delamination length was less than the gauge length, further straining after the initial carbon-ply failure resulted in only limited growth of the delamination, but further fractures occurred at random positions in the remaining bonded portions of the specimen. These multiple fractures of the carbon-ply and their associated delaminations are clearly seen in Fig. 29, and the development of multiple cracking can be followed in the
load/strain curves, Fig. 30 (a) - (d); which show a load drop at each successive fracture. In the region of the carbon-ply fractures the glass-plies carry increased load, and have higher strain, but because the specimen length is fixed at any instant during the test, the increased glass-ply elongation at carbon-ply fracture causes a load drop. The actual form of the load/strain trace at the load drops depends on the position of the strain gauge in relation to the delaminated areas. As \( P_c \) is reduced the magnitude of the load drop is decreased, and \( L_D \) also decreases, permitting a greater number of carbon-ply failures. Successive carbon-ply failures generally occurred at successively, but only marginally, higher loads. Occasionally a later fracture occurred at a lower strain than those preceding it, implying that earlier fractures had damaged the carbon-ply. Ultimately most of the hybrid became delaminated, and its strength and stiffness approached that of the glass plies alone. During continued straining the lengths of the delaminated areas either remained constant, or increased slightly to eventually stabilize at some value which appears to be characteristic of the laminate geometry. As multiple cracking developed the bonded areas were reduced, and as they became comparable in length to the delaminated areas it was noticeable that the carbon-ply failed preferentially midway between delaminated areas. When this occurred the characteristic delamination length was often somewhat reduced, and the delaminations rarely ran into each other and coalesced. There is typically 50% scatter in measurements of \( L_D \), and the values in Table 4 are the average of the many delaminations in a multiply-cracked specimen.

The practical significance of these observations is that the initial fracture in the carbon-ply does not propagate through the glass-plies, but is contained by the delamination at the carbon/glass-ply interface. The delaminated zone is also restricted, and this implies that load can be progressively diffused back into the carbon-ply, so that away from the fracture zone the carbon-ply remains capable of bearing load. This transfer of load from the glass-plies back into the carbon-ply, in and around the delaminated region, is essential to the progressive multiple-cracking type of failure which could be an attractive feature of hybrids. The delamination between plies is of practical importance because it drastically reduces the strength of a laminate in shear, or through-thickness tensile loading, and may propagate rapidly in these conditions resulting in complete separation of the plies. For these reasons interface forces have been studied in detail in selected hybrids using laser moiré strain measurement, and have been estimated from delamination
2-3-5 Model for the mechanism of load transfer between plies in the failure zone of laminated hybrids

The previous section has highlighted the importance of load transfer between plies around the carbon-ply fracture. The purpose of the next two sections is to present the models for the mechanisms of load transfer and delamination which have largely determined the direction of subsequent experimental work.

At this stage it is appropriate to make some preliminary deductions about load transfer in the delaminated zone. Consider a hybrid-laminate which fails by the mechanism depicted in Fig. 25. When the carbon-ply fails, the shear stress concentration on the interface between carbon and glass plies causes the initiation of delaminations which are effectively shear cracks in Mode II loading. Now, if there is no shear force acting at the delaminated interface, the loads in the glass- and carbon-plies will be equal at all positions in the delaminated regions, so, with a constant load on the hybrid, the stress intensity at the delamination crack tip will be the same whatever the length of the crack. (This is only true when the crack tip strain field is small compared with the length of delamination, and this is so for most laminated hybrids where the ply thickness is much less than the delamination length.) The consequence of a crack tip stress intensity which does not decrease as the crack length increases is that delaminations would be unstable and propagate continuously at some critical applied load. In contrast to this, the debond length stabilises at some value characteristic of the laminate geometry. This indicates that as the delamination grows, some mechanism operates to decrease the crack tip stress intensity to below some lower critical value when the delamination becomes stable. The most obvious mechanism is some frictional (or other) shear-stress acting across the interface to transfer load from the glass-plies back into the carbon-ply. This reduces the difference in strain between the inner and outer plies, and reduces the crack tip shear stress intensity. In deducing this the applied load has been assumed constant, whereas in practice there is a load drop when the carbon-ply fails. However, this assumption is justifiable because the delaminations have a stable length even when the load is increased again to the value attained immediately before fracture occurred.

In addition to the shear forces at the delaminations, there is evidence for shear forces acting in the bonded region beyond the crack tip. The fact that the carbon-ply failures rarely occur close to the delamination...
Friction is an obvious mechanism by which load could be transferred between delaminated plies, but requires that the two surfaces remain in physical contact. Whether this is so depends on the relative Poisson contractions through the thickness of each ply before and after carbon-ply fracture. Using a simple model for a 3-layer hybrid, with a thin central carbon-ply, the interface separation can be calculated as a function of the strain in the bonded portion of the hybrid, and the relative Young's moduli of the plies. The calculation assumes that the centre-planes of the glass-plies remain plane and parallel for the whole of the gauge length. (The high tensile loading limits bending of plies.) Fig. 31 illustrates the through thickness Poisson contractions for the bonded and delaminated areas in relation to an unloaded specimen. If it is assumed there is no friction, the carbon-ply in the delaminated area carries no load. All cross-sections carry equal load, so this defines the load in each ply. The Poisson contractions are calculated from the longitudinal strains of each ply, and give the separation $\Delta t$ of the plies at each delamination as

$$\Delta t = \frac{\varepsilon_h t_c \gamma_c}{2} \left( \frac{\gamma_g E_c}{2 \gamma_c E_g} - 1 \right)$$

(12)

Where $E_g$, $E_c$ are longitudinal Young's Moduli, $\gamma_g$ and $\gamma_c$ are the longitudinal Poissons' ratios for glass and carbon-plies respectively, and where $t_c$ is the carbon-ply thickness, and $\varepsilon_h$ the strain in the bonded portion of the hybrid. This predicts that, for plies of equal Poissons' ratio, the interface separates if the inner to outer-ply modulus ratio exceeds 2.

For the laminated hybrids where:

$$E_c = 144 \, \text{GPa} \, (\text{HTS fibre}), \quad 192 \, \text{GPa} \, (\text{HHS}) \quad E_g = 45 \, \text{GPa}$$

$$E_c > 3 E_g$$

and thus $\Delta t$ is positive and the interface should separate.
gives an interface separation of 0.1μm for every pre-preg in the carbon-ply. Even for the thickest carbon-plies tested, this separation is an order of magnitude less than both the fibre diameter, and the typical roughness of the delamination surface which is about 10μm, so significant friction between the plies is still expected. Unfortunately, because of this interfacial separation it is not possible to calculate simply a normal force from which to estimate the frictional force, and in this situation the friction will depend on the fracture debris and unbroken fibres which bridge the delamination.

2-3-6 Factors affecting the extent of delamination in laminated hybrids

In this section a model is developed for the energy changes during delamination, based on simple assumptions about the load transfer between plies. The model predicts the extent of delamination, and is fitted to experimental data for laminated hybrids to provide estimates of the interface friction and delamination energy. The analysis shows these two parameters to be major factors controlling delamination. Because this is an important practical consideration in the use of laminated hybrids, it has been investigated more fully using the laser moiré technique for strain visualisation, which also allows the load transfer across the delaminated interface to be estimated.

The basic assumption of the model is that load is transferred from the glass-plies back into the carbon-ply by a uniform frictional stress acting at the delaminated interface. This gives rise to a linear increase in load in the carbon-ply with distance from the transverse fracture, shown schematically in Fig. 32. As this load increases the load in the glass-ply must decrease to maintain a constant load on all cross-sections.

At the tip of the delamination crack the shear stress at the bonded interface is assumed to be significantly higher than the frictional stress on account of the smaller amount of material supporting the shear, and this gives rise to the load step.

The energy changes are considered for an infinitesimal increase in the delamination length dx. The carbon-ply load at any point, x distant from the transverse fracture, is 2τf; where τf is the frictional stress acting on each side. The sum of the carbon-ply and glass-ply loads on any section of the delaminated region equals the total load on the hybrid.
2 \varepsilon_f x + \varepsilon_g S_g = \varepsilon_h S_h

(where stiffness S is defined as the product of the modulus and thickness of the respective components, S = Et). From this:

\varepsilon_g = \frac{\varepsilon_h S_h}{S_g} - 2 \frac{\varepsilon_f x}{S_g} \tag{13}

The stiffnesses of all three plies act in parallel

S_h = S_g + S_c \tag{14}

Because the extension is infinitesimal there is no load drop, and the change in carbon-ply strain energy for the length dx is

\frac{1}{2} \varepsilon_h^2 S_c \, dx - \frac{1}{2} \varepsilon_c^2 S_c \, dx \tag{a}

Similarly for the glass-ply the strain energy change is

\frac{1}{2} \varepsilon_h^2 S_g \, dx - \frac{1}{2} \varepsilon_g^2 S_g \, dx \tag{b}

The work of delaminating length dx is

- 2 \gamma_D \, dx \tag{c}

where \gamma_D is the energy to delaminate unit area of interface.

As the delamination extends dx there is a relative displacement of glass- and carbon-plies of (\varepsilon_g - \varepsilon_c)dx against the total frictional force transferred to the carbon-ply, 2 \varepsilon_f x. The frictional work done is

- 2 \varepsilon_f x (\varepsilon_g - \varepsilon_c) \, dx \tag{d}

When the delamination extends dx, there is an overall extension of the hybrid of (\varepsilon_g - \varepsilon_h)dx, at constant load, which does work.

(\varepsilon_g - \varepsilon_h) \varepsilon_h S_h \, dx \tag{e}
For a delamination which is long in comparison with the ply-thickness, the changes in shear deformation are small, and the changes in shear strain energy will be negligible in comparison with those in longitudinal strain energy, and may be ignored. For static equilibrium the total energy change is zero. Therefore

\[(a) + (b) + (c) + (d) + (e) = 0,\]

and substituting of eq. (13) and eq. (14) for \(\varepsilon_g\) and \(S_h\) throughout yields

\[
\frac{2\gamma_c}{\beta^2} - 2\gamma_f\frac{\varepsilon_g S_c}{\beta^2} + \frac{\varepsilon_h^2 S_c^2}{2\beta^2} - 2\sigma_d = 0
\]

where \(\beta = \left(\frac{L}{S_c} + \frac{L}{S_g}\right)^{-\frac{1}{2}}\) (square root of series stiffness of carbon and glass-plies).

The appropriate solution for the stable delamination length is

\[
L_D = \left(\frac{\varepsilon_h S_c - 2\beta \sigma_d^{\frac{1}{2}}}{2\gamma_f}\right)
\]

Delamination is possible only when \(L_D\) is positive, a negative value implies that additional energy is required to propagate the fracture. The extent of delamination is inversely proportional to the frictional shear stress, and is expected to be completely suppressed when \(\varepsilon_h S_c < 2\beta \sigma_d^{\frac{1}{2}}\). The term \(2\beta \sigma_d^{\frac{1}{2}}\) represents the minimum load (per unit width), in the carbon-ply in the bonded portion of a hybrid, for which delamination is energetically possible.

From Table 4 it is seen that only hybrids with a thin carbon-ply consisting of a single pre-preg consistently contain delaminations shorter than the gauge length, but six of these laminates have been tested, and the results fitted to expression (15). The values of \(L_D\) measured are the equilibrium delamination lengths at the carbon-ply failure strain \(\varepsilon_{cu}\). Replacing \(\varepsilon_h\) by \(\varepsilon_{cu}\) in eq. (15) and differentiating yields:

\[
\frac{dL_D}{d\beta} = \left(\frac{d\varepsilon_{cu}}{d\beta} \cdot \frac{S_c}{\gamma_f}\right) - \frac{\sigma_d^{\frac{1}{2}}}{\gamma_f}
\]
\[ \frac{\gamma_D}{\gamma_f} = \left( \frac{dE_{cu}}{d\beta} \cdot \frac{S_c}{\gamma_f} \right) - \left( \frac{dL_D}{d\beta} \right) \]  

Over the range of \( \beta \) considered \( \frac{dE_{cu}}{d\beta} \) and \( \frac{dL_D}{d\beta} \) are approximately constant, and have been evaluated graphically from Fig. 33 and Fig. 34. The values of \( E_{cu} \) and \( L_D \) at \( \beta = 0 \) have also been taken from these figures. When \( \beta = 0 \) expression (15) reduces to \( L_D = \frac{E_{cu} S_c}{2 \gamma_f} \), and substitution of \( E_{cu} \) and \( L_D \) gives

\[ \gamma_f = 0.59 \text{ MPa} \]

Substitution of these values into expression (16) yields

\[ \gamma_D = 4.99 \text{ kJ m}^{-2} \]

These results are examined more critically in Chapter 3, but in passing it should be noted that the \( \gamma_f \) value is very much less than typical laminate interlaminar shear strengths obtained in the short-beam shear test, 40 - 60 MPa.

2-3-7 Laser moiré analysis of laminated hybrids

The analysis outlined in the previous section raised a number of questions which it was hoped could be answered by high resolution measurement of strain at the delamination. The laser moiré technique, developed at the Materials Department of the Royal Aircraft Establishment, provides this capability, and with their co-operation two series of hybrid laminates were examined. The first series was three hybrids with a single pre-preg carbon-ply, selected to cover a range of \( L_D \), but the intrinsic variability of their failures prompted further work on a second series of hybrids. These contained discontinuous carbon-plies in order to model carbon-ply transverse fractures.

The aim of the laser moiré analysis was to test the validity of assumptions made in the model in section 2-3-6. Namely:-

i) Is the model correct in assuming a load step at the delamination crack-tip, and how does the shear stress here compare with the frictional stress?

ii) The model assumes shear deformations are negligible for delaminations which are long in comparison with the ply-thickness. For how short a delamination is this justified?

iii) \( \gamma_f \) has been determined indirectly over a limited range of dispersion
iv) Is the large scatter in $L_D$ primarily a consequence of poor control of ply-thickness?

v) Is there significant 'stick-slip' behaviour in the propagation of delaminations, and if so, how is $L_D$ influenced by dynamic effects at the time of carbon-ply fracture? How does the strain distribution change between the arrest and re-propagation of delaminations?

These questions are discussed further in Chapter 3.

The laser moire technique is well described in refs. (56-59), and is only summarised here. The major strains in the tensile test are longitudinal, and because delamination extends across the width of the specimens transverse strains were of secondary interest. Consequently the technique could be considerably simplified by measuring only strains in the fibre direction, which by symmetry is a principal axis. The laser moire set-up is shown schematically in Fig. 55. The specimen is strained in a Hounsfield tensometer mounted on the optical bench which carries the other optical components in order to minimise relative movements. Two collimated laser beams (~50 mm diameter) are generated from the filtered output of an argon-ion laser. These converge onto the specimen creating Young's fringes in the volume of their overlap. The fringes are recorded by a blue-sensitive photo-resist coating on the specimen surface, which is developed to remove resist in proportion to the exposure received. This creates undulations which act as a phase grating. In practice a small pre-strain is applied to the specimen to ensure its correct alignment during the exposure. When the specimen is replaced in the tensometer and illuminated by the laser beams coincident diffracted orders are generated. Specifically, the first order diffraction of the secondary incident beam coincides with the zero order diffraction (straight reflection) of the primary incident beam, and these two beams are captured by the camera, Fig. 55. When the specimen is strained, the phase of the first order diffraction is changed by the change in dimension of the grating on the specimen, but the zero order diffraction remains unchanged. Interference between these two beams creates moire fringes which are recorded by the camera. The maximum fringe contrast is obtained when the two interfering orders are of equal intensity, and this is achieved by careful monitoring of the grating depth during development. The moire fringes represent contours of equal in-plane displacement in a direction normal to the grating lines. The difference in displacement between adjacent fringes is the grating spacing, so the strain normal to the grating is given by
For both series of specimens the grid spacing was 2.53 μm. It should be remembered that the fringe patterns show surface strains, and it is valid to assume uniformity through the outer-ply thickness only when the strain gradients are small.

The lay-up sequences of all laminates used in the laser moiré work are summarised in Table 5. Series 1 and 2 specimens were of similar design to Fig. 2 (a) (b) respectively. Series 1 is considered first. During loading the fringes were straight, parallel, and evenly spaced normal to the tensile axis showing uniformity of strain. Figs. 36 - 38 show the fringe patterns immediately after failure, and at two lower loads during unloading after failure, for specimens A, B, C, respectively. The longitudinal strain derived from the fringe frequency along the specimen centre line is shown in Figs. 39 - 41. The boundaries of the visibly lighter delaminated area are indicated, but the carbon-ply fracture could not be precisely located. In the bonded areas of the specimens the surface strain is uniform, rising over about 5 mm in the region of the delamination crack-tip, to a higher value, which is approximately constant over the delaminated zone. On unloading after multiple cracking the strain distribution remained similar, being reduced in proportion to the applied load. When unloaded to the pre-strain at which the grating was recorded, no residual strain could be observed.

The expected ratio of maximum glass-ply strain, (which is found at the carbon-ply transverse fracture), to the applied strain in the bonded hybrid can be easily calculated since all cross-sections must carry equal load. The load is \( \varepsilon_h S_h = \varepsilon_h (S_g + S_c) \) per unit width. The glass-ply strain is a maximum when it carries the full load, and is

\[
\varepsilon_{g\text{ max}} = \varepsilon_h \left( \frac{S_g + S_c}{S_g} \right)
\]

Thus

\[
\frac{\varepsilon_{g\text{ max}}}{\varepsilon_h} = 1 + \frac{S_c}{S_g}
\]

(17)

Measured strains in the hybrid and maximum glass-ply strains are tabulated in Table 5 with their observed and predicted ratios. Agreement is reasonable, though maximum glass-ply strain is marginally higher than expected. This result confirms that the carbon-ply carries very little load over the delamination. In the case of laminates A and C the ratio \( \varepsilon_{g}/\varepsilon_h \)
An average interface shear stress can be calculated from the rate of increase of longitudinal strain in the glass-ply. If a shear stress \( \tau \) acts at the delaminated interface, the rate of increase in carbon-ply load \( P_c \) is

\[
\frac{dP_c}{dx} = 2 \tau \]

where \( x \) is the length ordinate measured along the tensile axis.

The glass-ply load \( P_g \) at any point is the total load minus the carbon-ply load.

\[
P_g(x) = P_h - P_c(x)
\]

or in terms of strain

\[
\varepsilon_g(x) S_g = \varepsilon_h S_h - P_c(x)
\]

From this

\[
\frac{d\varepsilon_g}{dx} = \frac{dP_c}{dx} \frac{1}{S_g} = \frac{2 \tau}{S_g}
\]

and

\[
\tau = \frac{1}{2} S_g \frac{d\varepsilon_g}{dx} \text{ (Assuming uniform strain through the thickness of each ply)}
\]

Glass-ply strain gradients at the delamination crack-tip have been measured from Figs. 39 - 41, and the derived values for \( \tau \) are listed in Table 5. Strain gradients at the interface are diffused by the thickness of the outer plies, and the calculated values of \( \tau \) are averages for the 3.5 mm length over which the strain gradient was measured. Over the delaminated area the glass-ply strain gradient is too small to reliably estimate a value for the postulated frictional shear stress \( \tau_f \). However it is possible to set an upper bound on its magnitude, and this is included in Table 5.

A number of practical difficulties complicated the application of the laser moiré technique to series 1 hybrids. Random failure of the carbon-ply meant that several specimens had to be tested before a failure occurred in a suitable position within the area illuminated by the laser, and adequately separated from neighbouring fractures. For the first series there was no provision to photograph the specimens in normal light to record the extent of delamination without the fringe pattern. The two were related in the processed prints by using dust spots in the photo-resist as reference points. The most serious problem was that delaminations were rarely uniform
thickness of the carbon-ply, Fig. 29, complicating the measurement of strain gradients. The second series of specimens was fabricated in an attempt to reduce these difficulties. These hybrids contained a half-length carbon-ply, Fig. 42, and thus modelled one half of a series 1 specimen after failure of the carbon-ply. The important difference was that delamination could be initiated at the end of the carbon-ply, and propagated as the load was increased. The site of delamination initiation could be positioned conveniently in the illuminated area, and electronic flash was used to photograph the delaminations without fringes. The uniform thickness of the pre-cured carbon-ply gave more uniform delaminations.

Fringe patterns and photographs of the delaminations in series 2 specimens are shown in Figs. 43 - 46. For these photographs the specimens were rotated a small amount about the incident beam axis in order to tilt the fringes causing them to cross the end of the carbon-ply at a convenient angle. The spacing of the fringes across the width of the specimen is constant, and is determined by the rotation (re-adjusted in some cases during loading to maintain the angle), while the spacing along the length depends on the longitudinal strain as previously, and is essentially unchanged by small rotations (60). The curvature of the fringes is therefore a sensitive indication of small strain gradients.

Delamination initiated as diffuse white areas at the end of the carbon-ply in all specimens except F,(which underwent multiple cracking,) see for example Fig. 43 (d). With continued loading the delamination grew in a controlled manner to a length of about 5 mm before becoming unstable at a load about 40% higher than that for initiation. At this stage the strain in the delaminated region approximates to the average of that in both halves of the specimen. The delamination would then jump about 10 mm with a load drop, and thereafter propagated in similar or larger jumps at nearly constant load. This is shown in Figs. 43 - 46 which were taken at the stable positions between jumps. In these figures some areas appear dark because the specimen surface has developed local curvature and so no longer reflects the laser beams into the camera. By re-aligning the camera it was possible to photograph these areas at the expense of those which were formerly light, and in several cases two fringe patterns have been included.

When specimens with delaminations of about 10 mm long were unloaded to approximately one third of the load at initiation, the strain
In a 1 - 3 mm zone behind the crack tip became greater than that in the delaminated area, and remained so when load was removed completely, Fig 44 (f), (g), 45 (d). There is then a rather small residual strain, less than 5% of the strain at propagation, over the crack tip zone. The length of this zone is 3 - 5 times the length of the transition zone between high and low strain at the end of the carbon-ply before delamination, and is similar in size to the largest stable delaminations.

Unfortunately the apparatus was not suitable for study of the stable propagation to 'stick-slip' transition because

i) Loading was intermittent, and at a variable rate (by hand)
ii) There was no provision to record a load extension curve
iii) The Hounsfield tensometer was operated at its rated capacity, and was therefore rather 'soft'.

2-3-8 The failure strain and mode of failure of divided-tow hybrids

In many respects failure of the divided-tow hybrids was similar to that of the laminated hybrids with a single pre-preg carbon-ply. The first visible failure event was fracture of the carbon-fibre bundle, at a random position, with delaminations either side at the interface between the carbon-bundle and the glass-plies surrounding it. This is shown schematically in Fig. 47. The delaminated areas appear comparatively lighter, as seen in Fig. 48, and are very much shorter than the gauge length. When strained beyond the first carbon-bundle failure, successive fractures occurred initially at random positions. But as multiple cracking progressed, fractures were spaced at least two delamination lengths away from previous fractures. Ultimately this resulted in a regular array of fractures and delaminations, Fig. 48, with no delaminations meeting. Four bundle sizes were tested, 10,000, 3,630, 1,620 and 630 fibres, each of which has a characteristic delamination length, Table 4. This was measured by travelling microscope focussed through the transparent glass-plies, and ranged from 15 mm (10,000 fibres) to 1.5 mm (1,620 fibres). Typically the coefficient of variation was 30%, somewhat less than for laminated hybrids (~ 50%). The delamination length decreased rapidly with bundle size and can effectively be considered to have been inhibited below 1,620 fibres since the delamination could not be distinguished from irregularities in the carbon fracture path, Fig. 50.

Load strain curves for these hybrids are very nearly linear up to ultimate failure which occurred by longitudinal splitting of the glass-plies, Fig. 58 (c) to (f). The carbon bundle represents such a small percentage of the total specimen stiffness, between 0.5% and 5%, that
individual failures cause a negligible load drop or change in stiffness. Down to 1,620 fibres, failure of the bundle could be reliably detected by the audible 'ping' which accompanied the fracture, and each event was indicated on the load/strain trace with a 'pip' marker. Failures of the 630 fibre bundles could not be heard above general machine noise. Neither could they be detected by the acoustic emission monitor, because their acoustic output was low compared with the level before the first fracture, and because the period between fractures was less than the 2 s gating period used in rate counting. Instead fractures were counted under the microscope in three specimens taken to different strains. (Unfortunately the intermediate one split longitudinally, and the crack total may be in error.)

Initially, carbon-bundle failures occurred at an increasing rate as the strain was increased, as illustrated in Fig. 51 by the gradient of the crack density/strain curves. Crack density is simply the average number of cracks in a 1 m length of specimen. Beyond about 0.018 strain the rate of cracking shows an apparent decrease, and unexpectedly the curves also intersect at this point. The rate of increase in measured crack density at any strain is not a good measure of the probability of fracture at that strain, because the effective length of undamaged carbon-bundle decreases as multiple cracking develops. Each side of a fracture the carbon-bundle supports reduced load over some distance approximately to the delamination length. At any strain the fraction of the original gauge length not delaminated is:

\[ 1 - (\text{crack density} \times \text{mean } L_D) \]

If it is assumed that all the carbon-bundle outside delaminations is uniformly loaded, the expression above may be used to convert the measured crack density curves to curves of crack density in the bonded portion of the gauge length, Fig. 52. The low strain portions of the curves are little changed, but at higher strains the crack density is significantly increased and there is less fall-off in the rate of cracking. Only the three curves for the smaller bundle sizes now intersect, and at higher strain. For reasons discussed in Chapter 1 it is most unlikely that the probability of carbon-bundle failure decreases at higher strains, as is implied by the low rate of increase in crack density at higher strains seen in both Figs. 57 and 52. For this reason, and because there appears to be a minimum separation between fractures, which is greater than two delamination lengths, it is probable that the carbon-bundle supports reduced load over a distance greater than \( L_D \) each side of a fracture. The
mean separation of the most closely spaced 10% of fractures, was taken as an approximation to the effective length over which the carbon-bundle load is reduced, \( L_R \) (to one side of a fracture). Both \( L_D \) and \( L_R \) are included in Table 4. The values of \( L_R \) can also be used in equation (19) to correct measured crack density, in this case to crack density in that portion of the carbon-bundle carrying approximately full load. Curves corrected in this way, Fig. 53, for the three larger bundle sizes, no longer intersect, and have very similar shapes, but are separated by 0.0005 to 0.001 strain. The strain at the onset of cracking ranges from 0.0150 for 10,000 fibres, to 0.0177 for 630 fibres, compared with 0.00112 for the all-carbon laminate. Expressed as a percentage Hybrid Effect, this ranges from 34\% (10,000 fibres) to 58\% (630 fibres). The Hybrid Effect applies not only to first failure of the carbon-bundle, but also to all subsequent failures. Although only 5 specimens of each of the four bundle sizes were tested, considerable confidence can be placed in the results, as each involved a total of 50 to 200 fractures.

Acoustic emission rate curves also show the Hybrid Effect since the peak rate, which occurs just before the first fracture, is found at higher strains for smaller bundle sizes, reflecting the increase in first failure strain.

2-3-9 The failure strain and mode of failure of spread-tow hybrids

Failure of the spread-tow hybrids differed from that of the other hybrids in that there was no distinct event which paralleled failure of a carbon-ply or bundle, but was, in other respects, very similar. Instead of the spread-tow failing as a lamina with a crack propagating from bundle to bundle, isolated fibres and groups of fibres failed at random throughout the layer, leading to an accumulation of damage. Failures of the larger groups of fibres could be seen as fine white flecks on the bundles of carbon-fibres, Fig. 49, but failures of smaller groups and single fibres could only be seen under the microscope, Fig. 71. With continued straining a regular array of multiple fractures became visible on the larger bundles, with a spacing of about 1 mm (Fig. 49). A similar type of multiple fracture was found in smaller groups and single fibres in the microscopy, but with much finer spacing. The very fine scale of cracking prevented any accurate determination of the strain at which cracking first occurred, indeed, it is doubtful if the concept of a first failure event has much value when the bundle size covers a spectrum from single fibres to groups of 200 or more. Acoustic emission
failure, was a far better indication of the mechanisms and extent of failure, and both are discussed fully in sections 2-3-11 and 2-3-12 respectively.

If the maximum in acoustic emission rate, Fig. 58 (g), corresponds to the maximum rate of failure of the carbon fibre, it represents the strain at which the carbon fibre is most rapidly becoming ineffective in the structure. Taking this strain to be analogous to the onset of carbon-bundle failure in the divided-tow hybrids, gives some indication of the increase in carbon failure strain. The maximum occurs at 0.0189 strain, or the equivalent of roughly 67% Hybrid Effect.

2-3-10 Properties of single fibres

The strength of flawed filaments is length dependent, and can be described by the Weibull distribution.

Later analysis of composite strength requires fibre strength distributions for lengths of the order 100 μm, and it is obviously impractical to measure these directly without specialised equipment. However the Weibull distribution allows extrapolation from data measured at long (∼10 mm) gauge lengths to the shorter lengths of interest. Single fibre tests were performed to determine the two Weibull parameter. Two procedures are available. The mean strength and variance may be measured at a single gauge length, or alternatively the mean strength may be measured at two (or more) gauge lengths. The second approach was adopted with 10 mm and 50 mm gauge lengths. Fifty HTS-carbon fibres, (from the same batch as used for the divided-tow and spread-tow hybrids), were carefully selected from all parts of the tow, and mounted on cards with 10 mm and 50 mm cutouts. Small spots of 'Durofix' secured the fibres at the edges of the cutouts.

Before testing the diameter of each fibre, (both gauge lengths), was measured by laser diffraction with the apparatus shown in Fig. 54.*

*A 2 mW He-Ne laser beam (wavelength = 632.8 nm) was diffracted by the fibre, whose position was adjusted to bring the fourth minima of the diffraction pattern into coincidence with a length calibration on the screen. The separation of fibre and screen was measured with the vernier scale of the optical bench on which all the components were mounted. Fibre diameter is given by

\[ \frac{3 \lambda}{2} \sqrt{1 + 4 \left( \frac{\text{screen to fibre distance}}{\text{separation of minima}} \right)^2} \]

The effective diameter could be measured to within 2%, equivalent to 4% on cross-sectional area, but the true precision of the technique for area measurement was somewhat less because of irregularities (fluting) of the fibre surface.
Testing machine (Instron model TTM, load cell A), then the card was cut, and the fibre pulled at a rate of 0.1 mm/min. Load and crosshead movement were recorded, and values at failure were converted to stress and strain. Stress values were sorted into ascending order by a Fortran programme before plotting as cumulative probability distributions, Fig. 55. Plotting failure stress against failure strain for the 10 mm and 50 mm gauge lengths, Fig. 56, gave moduli of 208 GPa and 290 GPa respectively. Since the two lengths tested came from the same fibre, the difference in apparent modulus indicates that strain values calculated from cross-head movement are in error. Correction for load-cell compliance (measured to be 1.34 \(10^{-3} \text{mm}/\text{g}m\)), reduced the difference in modulus from 82 GPa to \(\sim 60\) GPa. The major part of the remaining error is attributed to deficiencies in the gripping arrangement. In view of this, it was considered preferable to calculate failure strain from stress, and the fibre modulus obtained from the composite modulus.

The Weibull parameters were determined from the \(\log / \log\) plot of mean fibre strength (GPa) against fibre length (mm), Fig. 57. Substitution of the gradient and intercept taken from Fig. 57, with appropriate value of the gamma function, into equation (6) gives

\[
\sigma_0 = 2.17 \text{ GPa} \\
\omega = 6.98
\]

Evaluation of the Weibull expression for cumulative frequency of fibre failure

\[
G_L(J_f) = 1 - \exp(-L_j \omega_j
\]

with these values of \(J_0\) and \(\omega\) for 10 mm and 50 mm gauge lengths generates the solid curves in Fig. 55. At low probability of failure, <0.2, there is significant difference between experiment and the Weibull distribution, but at higher probabilities the agreement is good.

2-3-11 Acoustic emission results

During tensile testing, microscopic damage accumulated by a number of mechanisms before any signs of fracture were outwardly visible. Acoustic emission monitoring was able to distinguish between noise sources associated with the fibre, and with the resin, but direct identification of noise sources was not attempted. There is however substantial evidence from microscopy, that carbon fibre failures, and resin cracking are the major noise sources. Typical A-E rate curves for all-glass and all-carbon pre-preg laminates are shown in Figs. 23, 24, and Fig. 58 presents on a common strain axis, A-E curves typical of all the wet-lay-up
Considering first the all-glass fibre composites, there was a considerably higher rate of A-E from the pre-preg based laminate (code 69 resin), than from the wet-lay-up composite, Fig. 24 and Fig. 58 (b), although both curves have a similar upward trend. No glass-fibre failures have been observed microscopically in either composite, so the A-E is attributed to the numerous resin cracks which have been observed. Their rate of increase approximately follows the A-E rate, see Fig. 59 and section 2-3-12. Both types of glass fibre have similar tensile strengths (in the composites), but the resin failure strains differ by about 0.01, (Code 69 fails at ~0.01, (64), and Epikote 828 system at ~0.02 strain, (63)), which could account for the A-E rate difference between pre-preg and wet-lay-up specimens.

All the composites laminated from pre-preg have similar curves in which the noise from resin cracking predominates. No consistent trends can be discerned for either HTS or HMS based hybrid laminates.

With the wet-lay-up composites a more interesting trend emerges. The level of A-E from the all-carbon HTS composites, Fig. 58 (a), was very much greater than that from the all-glass composites, Fig. 58 (b), and occurred over a range of strain approximately one third smaller. This difference is associated solely with the difference in fibre type as all other factors were the same. This suggests that carbon fibre fracture is a major noise source. The A-E rate from specimens of identical construction was consistent to within ± 50%, (at equal strains), whereas the difference between glass and carbon composites is a factor of about 60 at 0.01 strain.

Up to first failure of the carbon-bundle, the divided-tow hybrids all have similar A-E curves to the all-carbon composite, Fig. 58 (c-f) and (a). After the onset of multiple cracking, indicated by the peak, the emission rate drops to a lower and more irregular level which increases only marginally with strain. The three larger bundle size hybrids inevitably failed ultimately by longitudinal splitting around 0.025 strain, but in the case of 630 fibre bundles, straining could be continued to ultimate strains typical of all-glass composites (~0.03). At these high strains, in specimens without longitudinal splits, the emission rate rose again to a maximum at ultimate failure, following the high-strain behaviour of the all-glass A-E curves. It appears that the total output is simply the sum...
A-E curve peaks are translated to higher strains as the bundle-size is reduced, demonstrating the Hybrid Effect, and the area under the curves (representing the total number of emissions), up to the onset of multiple cracking, decreases roughly in proportion to the number of fibres in the bundle, Fig. 60.

Each 20 mm wide spread-tow hybrid specimen nominally contained 0.6 of a tow, or a total of about 6,000 carbon fibres, and at equal strains, the A-E rate was about 0.6 of that from a whole tow bundle. The two spread-tow hybrid composites with 0.5 mm and 1.0 mm thick glass-plies both gave very similar A-E curves, Fig. 58 (g) and (h), showing there is very little attenuation of carbon fibre noise through the thicknesses of glass-fibre typical of this type of specimen. It is therefore doubtful whether the differences in attenuation arising from different carbon-fibre distributions could significantly influence the A-E rate recorded. Acoustic emission rate curves for spread-tow hybrids have no sharp peak followed by a distinct fall in rate, but have instead a broad hump centred around 0.019 strain. The AE rate then falls by about a half to a minimum at roughly 0.023 strain, before rising again to a maximum at failure (~0.03 strain), reflecting the underlying behaviour of the glass-plies at high strain.

2-3-12 Microscopy of damage features

In this section the microscopic damage found before and after macroscopic failure and the special techniques developed to observe these features without introducing damage during the preparation of specimens is described. Generally, optical microscopy, (including macrophotography) was more useful than scanning electron microscopy as the transparency of glass fibres and resin could be exploited, allowing features below the surface to be observed without the need for sectioning. By focussing successively deeper into a specimen the effective depth of field approached the relief of the microscope objective, which had a minimum value of about 1.5 mm. With practice the eye and brain are able to form an effective 3-dimensional image as the plane of focus is shifted up and down, but this cannot be recorded on film. Many of the micrographs are therefore the result of a compromise between trying to show features in sharp focus, and at different depths, and need to be interpreted with imagination! All optical micrographs were taken on a Zeiss 'Ultraphot' photomicroscope. Scanning electron microscopy, which only examines surfaces was of limited use because most features of interest were internal. Simple examination of a specimen by eye, while turning it in a strong light, revealed many features of microscopic scale, but again such a dynamic image could not be recorded. To
some extent macrophotography was able to record the naked-eye impression, but with lower resolution and depth of field. For example; fine cracks and individual fibres could be clearly seen by eye in spread-tow hybrids, but were not well resolved in macrographs, e.g. Fig. 49.

Some features could only be examined by cutting and polishing sections of composite, and to do this effectively required close adherence to the following preparation routine. For sectioning, small samples of the tensile test pieces were mounted in the desired orientation in epoxy resin (Epikote 828, cured 3 hours at 60°C). The cured block was then sectioned close to the desired plane with a high-speed, fine-grade, water-lubricated diamond saw. Surface damage from cutting was a fibre diameter or less deep, and was removed in later stages of preparation. The cut surface was then ground lightly on 400 and 600 grit, wet carborundum paper, which left numerous scratches and chipped the edges of glass fibres. This damage was removed without developing relief by wet lapping on a cast iron plate with a slurry of 1000 grit silicon carbide, and resulted in a uniform fine pitting. The preparation was completed with two diamond polishing stages, on 6μm grade to remove the pitting left from lapping, followed by a finishing polish on 1μm grade. The time required with each diamond grade depended on the condition of the cloth, so the specimen was periodically examined microscopically to minimise polishing time, (typically 7 mins on each grade). Between each polishing stage specimens were cleaned in an ultrasonically agitated water bath. It is unlikely that any of the damage features described later are introduced by the preparation, as they are not seen in unstrained specimens. Certain details of the diamond polishing technique were essential to achieve a relief-free surface. The diamond compound was carried on a close-woven napless cloth which avoided the brushing action of piled cloths which tends to remove resin from between the harder fibres. Epoxy mounting resin and water lubrication in conjunction with water based medium in the diamond compound were used in preference to polyester mounting resin and petroleum spirit based lubricants which caused gum to build up on the polishing cloth.

The remainder of this section describes damage features, with the relevant techniques, beginning with matrix cracking. Isolated resin cracks, normal to the tensile axis, were seen at comparatively low strain before failure of the carbon-ply in laminated composites (Code 69 resin) Fig. 61. This micrograph of a longitudinal section was taken in reflected polarized light, which shows up strains in the resin on account of its photoelasticity. In this technique a certain proportion of the incident light penetrates the surface, and is reflected by features below the
surface, passing through strained regions before emerging with its plane of polarisation rotated. The analyser was adjusted to some position away from complete cancellation to show up both fibres, and cracks with strain fields. This distinguishes cracks from surface scratches. The reflected polarised light technique increased contrast by removing surface reflections and extends the depth at which resin cracks can be detected as light areas. For this reason it was particularly valuable in locating matrix cracks associated with carbon-fibre fractures, Fig. 71. In addition these cracks could be seen in transverse sections with unpolarised light, by focusing below the surface. Figure 62 is a pair of micrographs showing an annular crack surrounding an isolated carbon fibre. Cured Code 69 glass fibre laminates had a milky translucence of unknown origin, which effectively masked any outward change in appearance as matrix cracking developed. In contrast, the wet lay-up glass-fibre composites were virtually transparent, and with suitable back-lighting cracking could be seen by eye. When light is directed at an angle (~45°) to the fibre direction onto the back of a specimen, cracks normal to the tensile axis reflect light at an equal angle out of the front of the specimen, while the majority passes straight through. Viewed against a dark background the cracks appear as a scintillating lightening often referred to as 'stress whitening', Fig. 63. The cracking first appeared as strings of whiteness following the fibre direction at about 0.015 strain, and developed to the point where the structure of the tows was clearly visible by about 0.02 strain. Thereafter the whitening intensified progressively until failure, when it was fairly uniform. In its early stages the whitening partially disappeared on unloading the specimen, but by about 0.02 strain little change could be detected. Very similar cracking occurred in the spread-tow hybrids, but some of the cracks in the carbon-fibre layer were obviously much larger, and could be seen as distinct bright flecks. Figure 64 shows four of these specimens taken to similar strains to Fig. 63, but photographed against a light background to show up the carbon fibres. This means the whitening is less obvious, but the larger cracks can be clearly seen as bright lines.

Resin cracking was observed within wet-lay-up composites using transmitted light microscopy with a low-power oil-immersion objective, which avoided the need for any form of specimen preparation. Figure 65 is a series of micrographs of all-glass fibre specimens taken to successively higher strains, photographed with the plane of focus about 0.2 mm within the specimen. The surface appearance before testing, and after failure at 0.031 strain, is shown in Fig. 66. (The spots are probably mould release spray.) No cracks are visible in unstrained
the fibres, particularly in resin-rich areas. The cracks clearly do not pass through the fibres, and have a completely different random distribution at the specimen surface where they are not bounded by fibres. From these micrographs, and those taken in polarized light to show strain fields there is no evidence of debonding between fibre and matrix where the cracks meet the fibres. There appears to be a minimum crack spacing which depends on the crack length, which at high strains leads to regular arrays of cracks. The spacing is less for the shorter cracks which appear at higher strain, and at ultimate failure the spread in crack length is about 7:1. The total number of cracks visible in a standard field of view at each strain is plotted in Fig. 59, and is very similar to the normalised integral of the A·E rate curve for the same all-glass composite.

A very similar pattern of cracking was observed in the glass-plies of the spread-tow hybrids, but in addition to the strings of fine cracks the much larger cracks were seen when the central layer of dispersed carbon fibres was brought into focus, Fig. 67 (a) - (d). Because the greater thickness of glass-ply tended to obscure the finer matrix cracking, and because the objective relief was limited, a thinner spread-tow composite was fabricated with 0.5 mm thick glass-plies, and failed in a manner indistinguishable from the thicker composite. The large cracks frequently occurred in lines across the specimen, normal to the tensile axis, but also at isolated positions, Fig. 67, (a) and (b) and at higher magnification the stepped appearance of the crack lines is clearly seen, Fig. 67 (c) and (d). At first it was thought these lines were stray misaligned carbon fibres which gained the stepped appearance as a result of the cylindrical-lens action of the glass fibres. A number of observations suggest this is not so. The primary evidence is that no cracks are seen in unstrained specimens, and that all the lines of cracks are at 45° or more to the tensile axis, whereas misaligned fibres would be expected at all angles. In addition, the glass fibre diameter is several times less than the apparent length of the cracks, and it is hard to see how cylindrical lens action could give rise to such large steps instead of a general broadening. Further convincing evidence comes from micrographs in polarised transmitted light, Fig. 68, which show intense strain fields around the cracks. It is not clear from these micrographs whether these cracks pass through fibres of either type in addition to the matrix.

Before failure of the carbon-ply, no fibre failures were seen in longitudinal sections of laminated hybrids, and after failure of the
the delaminated region, though not necessarily associated with the transverse fracture. Some fibres had two or more fractures spaced by at least 10 fibre diameters, and this has implications for the transfer length. Several examples can be seen in Fig. 28. The volume of material effectively examined for fibre breaks in a longitudinal section is rather small because only cracks in a surface layer, on average one fibre diameter thick, can be detected. Because the distribution of fibre breaks is an important consideration in statistical models of composite failure, a more efficient technique for their location was developed, and is described in the next section. Where the delamination was long enough to permit some separation of the plies, glass-fibres could be seen bridging the failed interface, but proved difficult to photograph with adequate depth of field, Fig. 69.

Spread-tow hybrids were sectioned at a shallow angle to ensure intersection with the thin, non-uniform carbon-fibre layer. Figure 70 shows such a section of one of the larger groups of carbon fibres. A number of fibre fractures are clearly seen as dark lines across the bright fibres, but the total number of fractures is at least double this. Because the carbon fibres are rising upwards out of the plane of section, (towards the top of Fig. 70), where a fibre has fractured and debonded, nothing remains to hold it in place, and it is lost during polishing. All that remains is the lower portion of the fibre, which can be identified by its square rather than elliptical end. Therefore all square-ended fibres in Fig. 70 should also be counted as having fractured. It could be argued that the fibre fractures are caused by the preparation of the section, but no breaks are seen in unstrained specimens prepared in the same way. Again many fibres have more than one fracture, separated by 10 or more fibre diameters, and such multiple fractures seem to be more numerous than in the laminated hybrids, making the fracture path less distinct.

Polarised reflected light microscopy with a low-power oil-immersion objective was used to examine carbon-fibre fractures by focusing through the glass-plies of spread-tow specimens. Areas of matrix failure at delaminations and debonded fibres showed up bright against a dark background, Fig. 71 (a) - (d). Carbon fibre surfaces are optically active (depending a c-plane orientation) and can also appear lighter in the micrographs. The streaks along the fibres may be due to variations in the graphite crystallite orientation (refs. 2, 3, 4 of Appendix 1) or to fluting of the surface. Figure 71 (a) shows a relatively large group of carbon fibre failures, which are the rather fine and indistinct dark lines across the fibres. The large bright areas are debonding between fibre and matrix. It is interesting to note that
several breaks can be seen in the same fibre, all within the debonded region. A smaller group of fractures is shown in Fig. 71 (b), and a number of isolated carbon-fibre breaks in Fig. 71 (c). Generally the carbon fibres are debonded from the matrix for a distance of about 3 - 5 fibre diameters on either side of a fibre fracture, but this is not always so, and Fig. 71 (d) shows two broken fibres at higher magnification, only one of which is debonded. It appears that debonding occurs in preference to matrix cracking when a single fibre fails, but when the matrix cracks without fibre failure there is no debonding. No attempt was made to count the numbers of different-sized groups of fibre breaks with this technique, but at strains above ~ 0.02 all spread-tow specimens appeared to contain groups of fibre failures with a spectrum of sizes from single breaks to groups of about 30, the majority with debonding either side of the fractures.

2-3-13 The location of carbon fibre fractures

An electrolytic technique for decorating conducting fibres was developed to identify the non-conducting, (and possibly broken) carbon fibres. In this technique a thin (~3 mm) polished transverse section of composite is bonded with conducting cement, polished face outward, onto a metal block. The polished face is then immersed in 2 M $\text{H}_2\text{SO}_4$, and anodically etched to develop colouration on the ends of fibres which are electrically connected to the metal backing block, Fig. 72. The anodic etching is thought to develop a layer of intercalation compound, (probably graphite bisulphate (65)) on the fibre surface, giving rise to interference colours, which incidentally show up the sheath/core structure of the fibre. The etching procedure is described in detail in this context in Appendix 1. Fibres which are not connected to the anode do not develop colouration, and retain their polished lustre, appearing white in the micrographs, Fig. 74. It is assumed that un-etched fibres are discontinuous at some point in the section. Fibres which etch and appear dark in the micrographs are not necessarily unbroken, since, even when a fibre has fractured, it is possible that on unloading, fracture debris could be trapped between the broken ends forming a conductive path. Also the fibres make electrical contacts with their neighbours which provide parallel current paths, effectively shunting any electrical discontinuities in the fibres. The further a fibre fracture is from the surface being etched, the greater is the probability that it is shunted by neighbouring fibres. To estimate the depth to which fibre breaks can be detected, a 5 mm thick block of carbon-fibre composite was mounted with the fibres at 45° to the section plane, Fig. 73, and etched. Fibres up to 5 mm away from one edge of the block had no direct connection to the anode block, but by conduction between fibres the majority etched. Only a narrow
strip, between 5 and 30 fibre diameters wide (average ~ 10 fibre diameters), was effectively isolated from the anode, which indicates fractures can be detected on average up to 0.1 mm from the surface, Fig. 74. For a given area of polished surface, the etching technique examines for fibre breaks a volume of material roughly ten times greater than microscopy of longitudinal sections. It has the additional advantage that surface scratches cannot be misinterpreted as fibre fractures. Even so, the number of breaks found in all-carbon specimens is very small. Figure 75 (a) shows a typical field of view of an etched, unstrained, carbon fibre specimen with no fibre breaks. This should be compared with a similar composite strained to failure, Fig. 75 (b), containing a single fibre break. It is clear from these micrographs why no fibre fractures were seen in longitudinal sections. Surface damage is confined to single isolated breaks. The density of fibre breaks in all-carbon fibre composites caused by straining to failure is estimated from a number of specimens to be about $10^{10}$ m$^{-3}$.

Considerably more fibre-breaks were found after failure in laminated hybrids with thin carbon-plies, often in groups of two or three, Fig. 76 (a). They appear to be at least an order of magnitude more numerous, and in many cases there is associated matrix cracking which can be seen under crossed polars, Fig. 76 (b).

The electrolytic decoration technique was somewhat improved for the spread-tow hybrids by incorporating the cell into the feedback loop of a simple operational amplifier circuit to maintain the current at a constant value set between 10 and 100μA, irrespective of changes in the cell voltage due to polarization (within the limits of 15V supply). This allowed better control of the average charge passed per fibre end, and the process could be accurately timed to give a deep blue shade on the etched fibres which photographed with the much improved contrast seen in Fig. 77. General observations are much the same as for the laminated hybrids, although the 828 epoxy matrix is less susceptible to cracking. Fig. 77 shows selected regions of specimens strained in the range 0.009 to 0.023 which contain the greatest number of (in some cases the only) undecorated fibres, and are not representative of the general appearance, since many areas contain no fractured fibres. Two trends are apparent, the total number of fibre breaks increases with strain, most rapidly in the interval 0.017 to 0.023, and the average size of groups of fibre fractures increases from 1 to about 3 over the whole strain range. Although the total number of fibre ends examined at each strain was about 12,000, the numbers of broken fibres at all but the two highest strains are rather small, see Table 6, and insufficient to draw any conclusions about
correction of $1.43 \times 10^{-3}$ residual thermal strain should be applied to
the strains in Fig. 77 to obtain the true carbon fibre strain.) The total
numbers of fibre breaks have been used to estimate the proportion of
failed fibres in the 100 µm surface layer examined by the technique, Table
6. At the lower strains the sample of failed fibres is so small, (about 1-3),
that the errors are very significant. A proportion of the carbon fibres were
broken before fabrication of the composites, and although there were no
undecorated fibres in the unstrained specimen, (of similar appearance to
Fig. 77 (a)), a possible error of -1 broken fibre has been assumed for the
other specimens. This error margin has been applied to the numbers of
fibre breaks in Table 6 in the calculation of the probability of failure of
a 100 µm length of fibre at each strain. The estimated range of probability
is rather large when the number of broken fibres is small, and extends to
zero for a single fractured fibre. These estimates should therefore be
regarded as upper bounds on the probability of failure.

2-3-14 Thermally induced strains

At some point during the cooling of a hybrid composite after curing,
the resin matrix solidifies and thereafter on further cooling strains are
induced to the differential contractions of fibre and matrix. The effective
"solidification" temperature will generally be lower than either pre-cure or
post-cure temperature so some relaxation would be expected to occur in the
resin as it cools. For the laminated hybrids it was possible to measure
these residual thermal strains directly by using unbalanced laminates which
curved on cooling. This method was not however suitable for the wet-lay-up
composites because it is particularly sensitive to the thickness of the plies,
which in these composites were not uniform. Instead expansion coefficients
were measured separately for grp and cfsp, from which an upper bound for the
thermal strain is estimated.

The two unbalanced laminates contained two pre-pregs each of E-
glass and HMS-or HTS-carbon fibre, and were moulded and cured in the same
way as the other laminates. During the post-cure and subsequent cooling they
were held flat by a weighted plate in order to reproduce the conditions in a
balanced hybrid which would not bend. Their thickness (nominally 0.5 mm)
was a comprise chosen to give a high ratio of length to thickness ($\sim 400$)
and easily measured radius of curvature, while minimising errors introduced
by variations in ply-thickness. After curing, the laminates were cut into
5 strips and the deflections of the centres of each from a plane surface were
measured, and used to calculate the radius of curvature by simple geometry,
assuming the bending to be uniform, and end effects negligible. The bending
Stiffness of each beam was measured in a 3-point bend test, and simple beam theory was used to derive the average tensile compressive stress which would exist in each ply if the laminate were held flat as it would be in a balanced construction. A small correction for volume fraction was applied to these stresses before calculating residual thermal strains based on measured tensile moduli for similar composites. The coefficient of variation is about 10%, and is mostly due to ply thickness variations. The thermal strain mismatch $\Delta \varepsilon_t$ is

$$
\begin{align*}
&1.04 \times 10^{-3} \text{ for HTS carbon/E-glass} \\
&0.93 \times 10^{-3} \text{ for HMS carbon/E-glass}
\end{align*}
$$

The fraction of the mismatch which appears as compression in the carbon-ply in a balanced hybrid is the ratio of glass-ply to total hybrid stiffness $S_g / S_h$ as shown in Fig. 78.

The carbon-ply compression is

$$
\varepsilon_{ct} = \frac{\varepsilon_g \Delta \varepsilon_t}{\varepsilon_h}
$$

and that part of the hybrid effect which can be attributed to thermal strain is listed in Table 4, and indicated by the solid portions of the bars in Fig. 22.

Thermal expansions of the wet-lay-up composites were measured in the fibre direction with a Du Pont thermo-mechanical analyser in the range 30° C - 200° C, Fig 79. Both HMS and HTS carbon composites behave very nearly linearly, but the glass composite shows significant non-linearity near the limit temperatures. Best-fit straight lines to these curves give the expansion coefficients for

<table>
<thead>
<tr>
<th>Material</th>
<th>Coefficient</th>
</tr>
</thead>
<tbody>
<tr>
<td>HMS carbon</td>
<td>-1.00 $\times$ 10^{-6} K^{-1}</td>
</tr>
<tr>
<td>HTS carbon</td>
<td>-0.91 $\times$ 10^{-6} K^{-1}</td>
</tr>
<tr>
<td>E-glass epoxy</td>
<td>10.60 $\times$ 10^{-6} K^{-1}</td>
</tr>
</tbody>
</table>

The freezing-in temperature was not measured, but it cannot exceed the post-cure temperature of 150° C, so in cooling to the test temperature of 25° C the maximum possible temperature difference $\Delta T$ is 125° C. The maximum thermal strain differential calculated from

$$
\Delta \varepsilon_t = (\alpha_c - \alpha_g) \Delta T
$$

(where $\alpha_c$ and $\alpha_g$ are cfrp and grp thermal expansion coefficients) is $1.43 \times 10^{-3}$ for the HTS carbon/E-glass hybrids. Virtually all this strain mismatch is accommodated by compression in the carbon fibre component, since in all the wet-lay-up specimens, the glass fibre contributes more than 95% of the stiffness, making
The expansion coefficient of the composites fabricated with the two types of carbon fibre are only 10% different. Although these values are for a wet-lay-up system, they indicate a maximum strain differential in carbon/carbon laminated hybrids of $13 \times 10^{-6}$, which can be ignored since it is an order of magnitude less than the typical scatter in failure strain.
Chapter Three

DISCUSSION

3-1 INTRODUCTION

This discussion of the experimental results is divided into two parts, considering first the factors which influence the strength of carbon fibre reinforced composite, both as part of a hybrid composite, and in isolation. A statistical model provides the best description of the carbon fibre reinforced component of all the hybrid composites tested, and is able to relate their strengths to that of single fibres. The strengths of all-carbon fibre composites are below those expected from the model, and the variation with the size of specimen follows an opposite trend. This suggests that the hybrid effect arises from a failure to realise the full potential strength of the fibres in all-carbon fibre composites, rather than from an enhancement of their strength in the hybrids.

The second part of the discussion is concerned with the phenomenological aspects of the failures in the hybrid composites, and considers the factors which influence their macroscopic properties after the first-failure in the carbon fibre reinforced component.
The influence of residual thermal strains

The results from the laminated hybrid specimens show a trend of increasing carbon-ply failure strain as the fraction of the cross-section reinforced with carbon fibre, $P_c$, is decreased, Fig. 22 and Table 4. The residual thermal compressive strain in the carbon-ply follows a similar trend, Fig. 78 and Table 4, and immediately suggests itself as an explanation, but for two reasons is found to be inadequate. Firstly, the thermal strains estimated from the bending of the two-ply unbalanced laminates are too small, on average by a factor of 4.5 (coefficient of variation $\sim 20\%$) for the HTS-carbon/glass hybrids with a single carbon pre-preg, and by a factor of 6.8 (c of v $\sim 75\%$) taking all the HTS laminates together. There is a similar but somewhat smaller ($\sim 5\times$) discrepancy in the case of the HMS-carbon/glass hybrids. In the HMS/HTS carbon fibre system the thermal strain is negligible in comparison with the total hybrid effect observed. The maximum thermal compressive strain estimated for the divided-tow hybrids is about 45% higher than that for the laminated hybrids, most probably because it was calculated from expansion coefficients and assumed no relaxation in the resin as the composite was allowed to cool slowly from the post-cure temperature, (i.e. it is an upper bound value). Even so, it can only account for roughly 12% of the 30% or more hybrid effect. It is recognised that estimation of residual strains from thermally induced bending is particularly sensitive to the dimensions and properties of the two plies, and that a less optimistic value can be calculated from the expansion coefficients of wet lay-up grp and cfrp and the post cure temperature for the laminated system. This increases the estimated maximum strain from $1.0 \times 10^{-5}$ to $1.2 \times 10^{-5}$, but this higher strain value is still some 2½ times too small. The thermal expansion of the cfrp and in particular grp, is nonlinear, Fig. 29, but maximum gradients were taken for the expansion coefficients. Although the measurements and estimates of thermal strain have their shortcomings, it is unlikely that they are in error by a factor as great as 2½ to 4½. In the later sections of this discussion where the carbon-fibre strain is required, this has been obtained from the apparent strain by subtracting the anticipated thermal compression calculated from the best experimental measurement for the appropriate fabrication method.

The second reason thermal strains fail to explain the hybrid effect is because they vary only with $P_c$, and not with dispersion. The hybrid effect in the laminated hybrids increases by about 10% for a factor of 4 increase in dispersion at a constant $P_c$ of about 0.3, Fig. 22, and
although this increase is rather small and comparable with the typical scatter of the results (~ 7%), it appears to be a consistent trend. This inadequacy is more obvious in the case of the all-carbon fibre laminates where there is a 33% change in failure strain over a 12 fold range of thickness (or dispersion) in the absence of thermal strains. It is clear that the actual values of hybrid effect (based on the definition earlier) are critically dependent on the datum which is the failure strain of the all-carbon fibre control. This was not constant but showed a significant variation with specimen thickness, so that it might be better to express the differences in failure strain of the carbon fibre in the hybrids in terms of the trends observed as \( P_c \) and dispersion are changed. This is the approach adopted in Fig. 81 when considering statistical aspects of the strength of the CFRP.

3-2-2 Energetic and dynamic considerations

The linear elastic fracture mechanics, and simplified bounding approaches, based on the consideration of the energy changes which occur at fracture initially appeared capable, at least qualitatively, of explaining that part of the hybrid effect which could not be accounted for by residual thermal strains: consider a typical carbon fibre composite specimen which contains a short 'crack' extending right through its thickness. Assuming for the moment that the critical energy release rate for propagation of the crack is a constant, and a function only of the constitution of the composite, the crack would grow when the energy release rate exceeded this value. With a constant tensile loading applied to the specimen the energy release rate would increase as the crack grew, thus remaining above the critical value, and the failure would be catastrophic. Strictly, in the displacement controlled tensile tests performed in this work a constant load was not maintained during failure, but the increase in compliance associated with the formation of a small crack would be minimal in comparison with the overall compliance of a typical specimen, and it is justifiable to simplify the argument by assuming the load applied to the region of the crack is constant. Now if a similar specimen containing a crack was bonded between two glass fibre plies which do not fracture, as in a hybrid specimen, the relaxation of the material around the crack would be restricted. The glass-plies would support some of the load which would otherwise be placed on the unfailed carbon fibre ahead of the crack, and whether or not there is delamination at the carbon/glass-ply interface, the crack opening and the release of strain energy would be decreased. The energy release rate in the hybrid would be less than in the all-carbon specimens, and a higher applied load would be required to propagate a crack of given size. This idea of
Constraint is shown schematically in Fig. 80 for a sandwich construction hybrid with delamination at the interfaces between plies. Thicker glass-plies, and thinner carbon-plies would both increase the constraint, thereby decreasing the energy release rate at given strains and crack sizes, and this qualitatively agrees with the trends observed in the hybrid effect. Higher strains would be required to propagate similar cracks in larger all-carbon composite bodies, which is in agreement with experiment. These trends would be expected if the carbon fibre composite had a fixed intrinsic crack size, but microscopic evidence points to a failure mode in which successive fibre fractures accumulate in groups of increasing size, one of which eventually propagates catastrophically. The control of their strength by other defects is discussed later.

While this reasoning can be extended to smaller cracks which may not be immediately adjacent to the glass-plies, there will be effective constraint only if a significant proportion of the cracks have strain fields which interact with the glass-plies. In the case of single broken fibres the strain field is restricted to other fibres in the immediate vicinity, and the relatively remote glass-plies would be expected to have but little influence. There is no experimental evidence from acoustic emission monitoring or microscopy that the glass-plies influence the density of carbon fibre fractures at any given strain below the first macroscopic failure of the cfrp. It is therefore reasonable to assume, for the moment, that at equal strains the distribution of intrinsic crack sizes will be the same for all the hybrids and all-carbon composites. But if the presence of the glass-plies is to have any significant constraining effect on the cracks, they would have to be large enough to interact with the ply structure. For this to occur they would need to be of similar size to the carbon-ply thickness, and it is unlikely that defects of such a size would not have been detected by naked eye, or under the microscope!

Although no sub-critical cracks have been found with dimensions that approach the thickness of even a single pre-preg carbon-ply (0.125 mm) it is still possible that such cracks do exist, and it would be instructive to consider how their propagation might be influenced by the proximity of the glass-plies. It was beyond the scope of this work to develop a detailed energetic model for the propagation of a variety of sizes and locations of cracks. Instead a simplified bounding approach was developed for laminates following that of Aveston & Sillwood (7), and this is contained in Appendix 2. Only the assumptions and predictions of the model will be considered here to highlight the major drawbacks of the approach. The total energy change
(i.e. the difference between the conditions associated with the formation of a transverse carbon-ply fracture), is equated to the sum of the energies required to create the transverse and delamination cracks, and the energy dissipated by the sliding which occurs against the frictional stress at the delamination interface. The configuration of the final state is that illustrated in Fig. 25. This assumes that the transverse fracture and delaminations are indivisible parts of a single failure mechanism. But this is in conflict with the model developed earlier, which considered the energy changes for an increment in the length of the delamination crack as a separate independent event, which followed a transverse fracture. The bounding approach over-simplifies failure by considering only beginning and end states, and it is obviously inconsistent to have a model which assumes that a transverse fracture has occurred, and then considers the energetics of delamination, while at the same time adopting an approach which considers the transverse fracture and delamination as inseparable. Although the sequence of events in the development of a carbon-ply failure has not been observed, it seems unlikely that the carbon-ply fracture and delaminations would occur as separate event. Rather, a crack would propagate across the width of the carbon-ply, while the delaminations grew simultaneously along the interface. At an intermediate stage the fracture would appear similar to Fig. 80.

The laser moiré work, which was performed after the analysis in App. 2, shows that delaminations can be propagated at loads below that required to fracture the carbon-ply, (in all but one instance). The exception was the 6-1-6 lamina glass/carbon/glass specimen in which multiple fractures of the carbon-ply occurred before delamination at the end of the carbon-ply. This might simply be because the shear stress concentration at the ply's end is less than at a fracture where there are effectively two ends butted together. From this it is concluded that delamination is energetically favourable in its own right before carbon-ply fracture, so there is justification for dealing with the delamination as a separate phenomenon. This does not necessarily mean that the carbon-ply fracture is possible without some degree of delamination, although this may be the case. The bounding approach for the carbon-ply failure strain which is considered next, has to assume, therefore, either (i) the transverse fracture occurs with no delamination, or (ii) that the full extent of delamination observed after failure is necessary to allow the failure to proceed. This second approach has been adopted in App. 2.

It is recognised that the bounding approach is crude, but it is
not practicable to consider in detail the growth of a minute crack somewhere in the carbon-ply at all stages up to the final state of a complex transverse fracture with delaminations, in order to determine the point at which the propagation is critical, and influenced by the glass plies. It was hoped that the bounding approach would show similar trends to the real behaviour, and in general it does, indicating increased failure strains for thinner carbon-plies and thicker glass-plies. Its major shortcoming is that it predicts failure at virtually zero strain for long lengths of specimen when there are no glass plies! When the carbon-ply is thick and the glass-plies are thin, failure is obviously going to be similar to the all-carbon case, and will be determined by some other failure mechanism which is virtually independent of the glass-plies, representing an additional higher bound.

The model is also particularly sensitive to the values of fracture energy, and makes a major approximation in ignoring shear deformations in the plies around the fracture. These are probably the two factors which are responsible for it predicting failure strains above those observed. Because some other model is required to explain the intrinsic strength of the carbon fibre, the bounding approach does not offer much insight into the mechanism limiting the strength in hybrids, but it is reassuring that numerical evaluation confirms the trends which are intuitively expected. Although the predicted strains are too high, they are of similar magnitude to the actual failure strain, which means the constraining influence of the glass-plies could be an important factor governing the propagation of cracks in the carbon-ply. The analysis can be used with greater confidence for the case of a pre-existing crack in the configuration depicted in Fig. 80 which fits the model precisely. It would then demonstrate the influence of the ply-thicknesses on the load required to propagate the crack. This is of interest where plies of glass fibre are interleaved with carbon fibre plies in a crack-stopping role, because it indicates that as well as acting as barriers to carbon-ply cracks in the through-thickness direction, they could also restrict crack propagation within an individual ply. This discussion is however primarily concerned with the initiation of the crack, rather than the influence of ply-structure on its propagation.

The arguments developed above apply in principle equally well to the divided-tow hybrids since it is still apparently true that there are no such sub-critical cracks large enough to interact with the glass-plies. However, as the size of the carbon fibre reinforced element is reduced further there will come a point where the largest groups of fibre fractures will interact with the glass-plies, which could then have a significant influence. The microscopy of the spread-tow hybrids suggests that, at strains typical for failure in the laminated and divided-tow composites,
the largest individual groups of carbon fibre fractures would be in the range 1-3. Therefore constraint by the glass fibre, and significant enhancement of failure strain would only be expected for bundles of carbon fibre of comparable size. It was not possible, experimentally, to compare the strengths of such small bundles when tested on their own, or incorporated into a glass fibre composite. But it is notionally possible to compare small bundles surrounded by glass fibre, with similar bundles surrounded by other carbon fibres, by comparing the spread-tow hybrids with all-carbon non-hybrid composites, since the latter are effectively just an assemblage of a large number of small bundles. The higher modulus carbon fibres would have a greater constraining influence than lower modulus glass fibres, and the bundle surrounded by carbon fibres would thus be expected to fail at a higher strain. This is not the case! The comparison makes the error of assuming that the fibres surrounding the small bundles are not themselves broken, and a proper analysis becomes the statistical problem discussed later. The statistical analysis assumes that the fibres have a fixed distribution of flaws which determine their strength at any given gauge length, and relates this distribution to the strength distribution for the composite. Extending the idea of constraint to the flaws in individual fibres, it becomes apparent that strong, stiff materials in their immediate environment could significantly increase the stress at which they become critical and cause the fibre to fracture. In practice this is probably a minor consideration in view of other approximations made by the statistical model. Aveston's and Sillwood's (7) analysis of the fracture of single fibres suggests a lower bound failure strain of 0.011 for fibres which debond from the matrix, and 0.004 for the elastic case, without debonding, so the strength of the bond to the matrix could also be important.

So far the behaviour of a crack has only been considered for a situation where it grows relatively slowly in comparison with the speed of sound in the material, in which case the strain distribution is to all intents and purposes identical to that in static loading. In reality the growth of the crack is rapid and catastrophic, and the dynamic loads may be significantly different. Two observations provide evidence for this. When all-carbon composite specimens fail they are often fragmented by two or more fractures perpendicular to their axis, and frequently linked by longitudinal splitting parallel to the fibres. The precise sequence of such a failure is not known, but it is clear that the sudden release of strain energy gives rise to transient loads which exceed the materials' strength in several places. The multiple cracking in the laminated hybrids normally occurred at successively higher loads, but occasionally this pattern was broken and
Fig. 30. The net result of the first crack and its accompanying delaminations was to reduce the load at which continued straining caused the second crack. Thus, any load-induced weakening at the position of the second crack is inferred to be the result of a transient overloading which was of too short a duration to be recorded by the load measuring system. The presence of glass fibre plies introduces an interface which might reflect and refract stress waves by reason of the acoustic impedance mismatch, and reduce the dynamic loading. A hybrid effect could arise as follows: If a small crack suddenly occurred within a carbon fibre reinforced composite it would briefly overstress the surrounding material (by a maximum factor of 1.27 in the 2-D case (46) ) which would be more likely to fail, allowing the crack to propagate. If partially surrounding the crack with glass fibre reduced the temporary overstressing, the probability of continuous crack propagation would be reduced. Hedgepath (46) estimates a maximum of 27% overstress, so even if this was completely removed by the glass-plies, it would only result in a 27% hybrid effect. Allowing for thermal strains this could barely account for the hybrid effects observed, and in any case it is difficult to conceive how glass-plies relatively remote from the small crack could have such a profound effect.

To summarise, in the case of laminated and divided-tow hybrids, any mechanism of strain enhancement which involves the interaction between the strain field of a sub-critical crack in the CFRP and the GRP in a hybrid is unlikely, because the largest cracks observed are too small, and in the main too remote for the interaction to be significant. When the carbon fibres are dispersed as small bundles, as in the spread-tow hybrids, such interactions should be considered, but statistical factors become dominant.

3-2-3 A statistical explanation of the strength of composites

A statistical approach following that of Harlow and Phoenix (39) will be used to relate the strengths of all the composites at first failure of the CFRP, to the strengths of single fibres using a single model. This enables the results as a whole to be compared with the theoretical predictions of their strength, and to account for the differences in strength of the CFRP component in the various hybrid laminates.

The CFRP component of the hybrids, the all-carbon laminate, and the single fibres are considered in the model to be composed of a number of short segments of fibre, combined appropriately in series and/or parallel to model continuous fibres and parallel bundles respectively. The statistical
model discussed in the introduction suggests that the strength of CFRP might be governed by the number of such elements it contains. The volume dependence of the strength of the CFRP phase of the composites is shown in the Weibull probability graph, Fig. 81. In this graph the natural logarithm of the carbon-fibre failure stress (measured in GPa) is plotted against the natural logarithm of the total number of elements contained in the body being considered. The fibre failure stress has been calculated from the failure strain of the composite, (having applied an appropriate correction for the thermally induced compression in the carbon fibre), but the negligible contribution of the resin matrix (~3%) has been ignored. There are two horizontal axes which are linear in \( \ln(-\ln(1-p)) \), where \( p \) is the probability of failure, and in \( \ln(mn) \), where \( m \) and \( n \) are the number of elements in series and parallel respectively. The origin of the \( \ln(mn) \) axis coincides with the \( p = 0.5 \) position so that a curve showing the dependence on the fibre stress can also be interpreted as a curve of median fibre failure stress with respect to the number of elements (mm). The basic element has been chosen to be a 100\( \mu \)m length of a single fibre, so \( m \) is simply the number of such lengths in the gauge length, and \( n \) is the number of fibres. The matrix is ignored, although its behaviour is implicit in the formulation of the load sharing rule. This basic unit is the transfer or ineffective length in the statistical model, and the value of 100\( \mu \)m, which approximately corresponds to the total length of debonding observed at single fibre fractures in the spread-tow hybrids, Fig. 71, was chosen to allow experimental curves to be compared directly with Harlow and Phoenix predictions without the need for translation. The value is somewhat arbitrary, but it is convenient because it also corresponds to the average length of fibre in which breaks can be detected by the electrolytic decoration technique. A better estimate would need to consider the non-uniform load on the fibre within the ineffective length, but this is unwarranted since the statistical model assumes uniform over-stressing within the length. The effect of different ineffective lengths can be seen by translating the curves in Fig. 81 horizontally by a shift factor \( \ln(\text{original ineffective length/new ineffective length}) \).

The way in which the theoretical predictions of strength, and the experimental results for composites and single fibres, have been combined in a single presentation will be described next, before considering the results as a whole.

Single fibre strengths were measured at only two gauge lengths, and their mean strengths are plotted in Fig. 81. In both cases \( n = 1 \), and \( m \) is 100 or 500 for 10 mm and 50 mm gauge lengths respectively. Weibull
behaviour has been assumed in placing a straight line through the points, and this part of the graph is the equivalent of Fig. 57, except that length is expressed in terms of the ineffective length, and is also similar to a 90° counter clockwise rotation of Fig. 12 where fibre load is presented as a non-dimensional quantity $\sigma/\sigma_o$. The load axis is logarithmic and so this scaling can be accomplished by a vertical translation of $-\ln (\sigma_o)$ in Fig. 81, without changing the shape of the curves since

$$\ln \left( \frac{\sigma}{\sigma_o} \right) = \ln (\sigma) - \ln (\sigma_o)$$

As most of the data in Fig. 81 are experimental, stresses have been left in (GPa), and the predictions from Harlow and Phoenix have been translated accordingly.

The gradient of the line for single fibres is a measure of the variability of fibre strength at a given length, and the Weibull modulus $w=7$ is typical of good quality fibres (Moreton (45)), and corresponds to a coefficient of variation of about 20%. There is an error margin of about 5% in the fibre stress, introduced by uncertainty in the measurement of the fibre diameter, but this is not indicated as it is probably negligible in comparison with deviations from Weibull behaviour over two decades of gauge length. Moreton's data for 3 gauge lengths (45) indicate about 10% variation per decade.

The spread-tow results were plotted by a somewhat different procedure using the probabilities of failure given in Table 6, and the probability axis. The horizontal error bars for the two points at lowest stress extend to infinity in the low probability direction because the estimated range of probability includes zero, so that they may be considered to indicate upper bounds on the probability of failure. At higher stresses where there are more fibre breaks the errors are smaller, and here they span a factor of $5\sqrt{e^{1.6}}$, each way to allow for the spread in the depth to which the decoration technique is effective, errors in estimating the sample size from which the probabilities were calculated, and errors arising from the choice of ineffective length. The errors in stress are about 10% comprised of 2% in strain measurement, 8% from other sources such as the modulus value and the discounting of matrix contributions, and are comparatively small (0.1 ln unit).

The divided-tow results were entered on Fig. 81 by taking the mean crack spacing as the effective length of bundle being tested at any
obtained from the curves in Fig. 53 to which a correction based on the most closely spaced 10% of cracks, has been applied to allow for the reduction in carbon bundle stress in the region of each carbon fracture. (The results for the 600 fibre bundles have not been included because of the difficulties in counting the numbers of cracks.) For these hybrids $m$ is the number of $100 \mu m$ lengths in the mean crack spacing, and $n$ is the number of fibres in the bundle.

The mean fibre stress at the first carbon-ply failure is plotted as a single point for each lay-up sequence of the sandwich construction hybrids. Here $m$ is the gauge length expressed in terms of the ineffective length, i.e. $10^3$, and $n$ is the total number of carbon fibres in the hybrid, about 12,000 for each 10 mm wide strip of a single layer of pre-preg. The number of pre-preg laminae in the carbon-ply is also indicated in Fig. 81, and the vertical spread reflects the differences in failure strain between hybrids with different numbers of glass laminae, which remain even after corrections for thermal strains have been applied. The all-carbon laminates are entered in a similar manner.

Finally the composite strength predictions from Harlow and Phoenix (39) for values of $w = 7, 10, 15$ have been transferred from Fig. 12. The median fibre strength and median composite strength should coincide for a single element, i.e. at $\ln (mn) = 0$, and this is the case in Fig. 12. However, the fibre strength plotted in Fig. 81 is the mean strength, which accounts for the small discrepancy in the intercepts on the stress axis. At the median fibre strength the probability of failure is 0.5

\[ 0.5 = 1 - \exp \left( - \frac{\ln (\sigma/e)}{\sigma_e} \right) \]

in the Weibull distribution, whereas the mean strength is given by eq( 7 ). For the value of $w$ of interest here, the difference is rather small because the distribution is reasonably symmetrical, as shown by Fig. 55. Translating the curves predicting composite strength in this way effectively normalises them to the median strength of a single fibre one ineffective length long, and the fact that the load axis uses units of (GPa) is simply a convenience for entering and comparing experimental data in these units.

There are a number of qualifications and points of interest relating to the way in which the data have been plotted in Fig. 81, and these will be discussed after first outlining the more important features of the results which are brought out by this presentation. The curves and
and experimental points can be interpreted in two ways; as showing the probability of failure of a single element at various fibre stresses and in a variety of sizes of composite, or alternatively as curves giving the median strength of an assemblage of elements making up a single fibre or composite body.

The major achievement of this presentation is to bring together on a single line in Fig. 81, all the results for hybrid composites, and the extrapolated strength for single fibres. In view of the very wide range in the individual volumes of cfrp represented by these results, about 5½ decades, the trend (dashed line in Fig. 81) can be accepted with some confidence as convincing evidence of weakest link behaviour in the cfrp phase. In addition the slope of the experimental curve, giving the variation of cfrp strength with volume, agrees exceedingly well with those predicted by the statistical model of Harlow and Phoenix.

As a more general observation the strengths of all the composites are greater than that of the single fibre, and there is a trend towards a greater ratio of composite to fibre strength as the size of the composite is increased. With the exception of the all carbon laminates (and possibly the spread-tow hybrids) the strength of the cfrp is less variable (w ≈ 25) than that of single fibres (w ≈ 7). (The Weibull modulus w = -1/gradient.)

Data for all the hybrids lie between the predictions of composite strength for w = 10-15, within the estimated margins of error. In contrast, the all-carbon laminates show a trend of increasing strength with volume, and are significantly below the trend set by the hybrid results, and the theoretical predications.

The results from the hybrids define a strength trend over approximately 5½ decades of specimen volume which is in fair agreement with theoretical predictions. The all-carbon laminates covering only 1 decade of volume are significantly weaker, and do not fit the statistical model. The variation of the carbon fibre failure strain in hybrids can therefore be explained as a volume-related size effect which follows the statistical model. However, the magnitude of the hybrid effect, as defined earlier, is mainly dependent on the arbitrary choice of an all-carbon fibre composite control, whose strength is determined by other factors.

Although there is good agreement (in the case of the hybrid composites) between the experimental and predicted variation of strength with volume, the actual strengths in the hybrids are somewhat greater than
This is, perhaps, not surprising as a number of approximations and assumptions have been made in both experiment and theory. The remainder of this discussion on strength will examine these more critically, and will also propose an explanation for the apparent weakening of the all-carbon laminates. Suggestions are made, where appropriate, for further experimental work to provide improved information on which to base the statistical model, or to extend the range of structures. This is essentially an open-ended exercise as the statistical approach can be applied in principle to any carbon fibre composite of interest. Significant improvement upon the present data would involve a substantial programme to refine and extend techniques for the fabrication and testing of specialized composites, not necessarily hybrids.

The presentation of the experimental results in the probability graph, Fig. 81, has three aims: to link the behaviour of all the hybrids in order to demonstrate the trend towards lower strength with increased volume, to show that the all-carbon composites deviate from this trend, and to relate the strength of cfrp in the hybrids to the fibre strength. The degree of confidence placed in the strength trend in the hybrids depends on a number of factors. The strengths of the cfrp in laminated hybrids with a constant carbon-ply thickness, but different thicknesses of glass-plies, follow a consistent upward trend as the ratio of glass to carbon fibre is increased. The differences are small and comparable with the typical scatter about the mean for a given geometry (≈7%), but they rank in correct order in all but two cases. Three explanations are considered likely. The corrections for thermal stress could be too small, in which case applying a greater correction would draw the points together, and lower them generally. The error estimates could be too small, and the ranking fortuitous, i.e. no trend. Or the trend could be real, and a result of a more subtle interaction between the glass and carbon-plies, possibly involving fibre alignment, (discussed later).

The curves for the divided-tow hybrids embody a very large number of individual fractures of the carbon bundle, (≈100) and show fairly smooth behaviour, Fig. 51. These results have been corrected to allow for the length which becomes un-stressed at each fracture. This correction is based on the average spacing of the 10% most closely spaced cracks and is somewhat arbitrary, but it is effective in separating the curves. This procedure may cause the increase in curvature at the high stress ends of the curves as plotted in Fig. 81 because the correction becomes increasingly sensitive to the effective length relieved of load as the number of cracks increases. Although it is not very obvious in Fig. 81, the slope of the
relatively straight lower portions of these curves increases with the bundle size, with \( w = 21, 28, 35 \) for 1620, 3620, and 10,000 fibre bundles respectively, contrary to the trend of decreasing slope predicted by Harlow and Phoenix which indicates decreasing variability as the volume of the composite in increased. This could also be a result of the correction applied, or it could reflect a true increase of variability in the larger bundle size, caused for example, by some parts of the tow being slacker and deviating sinusoidally from the specimen axis.

The considerable uncertainty in the estimates for the probability of failure in the spread-tow hybrids could affect the position of these results in relation to the other hybrids. This is unfortunate because they cover the portion of the experimental curve between that of the divided-tow results, and the predicted strength of the single fibres, (which is itself subject to considerable error). The geometry of the carbon fibre layer in the spread-tow hybrids is somewhat different from that assumed in the Harlow and Phoenix model because it is fairly continuous in one direction, albeit with substantial variations in the volume fraction of fibre, while it ranges in thickness from 0 to a maximum of about 10 fibre diameters in thickness. The convergence limit for \( W(r) \) is reached for bundles of about 5 or 9 fibres when the Weibull modulus for the fibres is 10 and 5 respectively, so it is clear that there are substantial areas where the layer is essentially 2-dimensional in terms of the statistical model. Although Harlow and Phoenix do not specifically predict the expected sizes of groups of broken fibres, the number of fibres at which \( W(r) \) effectively converges does give some indication. When the thickness of a layer approaches the size of groups of fibre breaks deviations from the 3-D model might be expected. Empirically thinning and widening of the layer can be seen to reduce the number of overstressed carbon fibres associated with each fibre fracture without changing the total number of fibres involved, and this would reduce the 'weakest link' characteristics of the composites cdf for failure. Put another way, distributing the fibres as a thin layer reduces the opportunity for fracture to propagate from one area to another, and the layer will behave as a number of smaller bundles of fibres. In practice the resin-rich areas and misalignment of the fibres also tend to separate them into small bundles, Fig. 49. Notwithstanding these qualifications, the technique does offer the opportunity of investigating a wider range of probability of failure in hybrids of the same basic geometry, provided that more material can be examined to give a more reliable estimate of the lower probabilities. Somewhere between a 100 and 1000 fold increase in the sample of fibres would be required to provide useful data at similar probabilities of failure as for the laminated hybrids, and this could be accomplished by incorporating a greater number of
similarly spread glass-fibre tows. This type of hybrid would also be of
caliberae interest because the limited debonding possible would give
rise to a failure mode differing considerably from that of laminated
hybrids with a relatively coarse ply-structure.

It is rather optimistic to hope that extrapolation of single fibre
strengths to gauge lengths two orders of magnitude smaller will give a true
estimate of strength at the 100\mu m ineffective length, but there was
insufficient time in the schedule of other experimental work to develop
apparatus capable of measuring fibre strengths directly at these very short
gauge lengths. Even if the single fibre strengths do follow a Weibull
distribution, which other workers (Moreton (45), Metcalfe and Schnitz (43),
and Hitchon and Phillips (52)) suggest is unlikely, it is questionable whether
the strengths measured in isolation are representative of those realised
in a composite where the fibres are closely packed together. It is possible
that simply pulling a fibre from the tow causes surface damage. This
procedure also favours the extraction of stronger fibres. The proportion of
fibres broken while being pulled from the tow and mounted on cards for testing
was only about 10\%, and most of those were the result of clumsy manipulation
rather than intrinsic weaknesses in the fibre. Only 2\% of correctly mounted
fibres were accidentally broken during testing, and again this only marginally
favoured the stronger fibres, if at all. Further degradation of their strength
probably arose from 'crinkling' of the glue spots holding the fibre to the
card which caused them to be locally misaligned with the tensile axis. When
the fibres failed the whole gauge length was usually lost, and it was not
possible to tell if the gripping regions were limiting strength. (Moreton
(45) has tested fibres submerged in damping fluid and is satisfied with the
method of fixing fibres to the card.)

The comparison of predictions for the strength of composites with
the single fibre strengths in Fig. 81 assumes that their cdf is not altered
by incorporating them in a closely-packed array coupled by the resin matrix,
while the fracture mechanics arguments suggest that the stress at which
fibre flaws become critical will depend on their immediate environment.
Moreton (45) has shown that many fibre fractures are initiated at surface
flaws, (typically \(\sim l_{\mu m}\)), and these would be more susceptible to constraint
than internal flaws. Although the experiment has not been performed, it
would be quite feasible to incorporate single carbon fibres into thin glass
fibre laminates in a similar manner to the divided-tow hybrids, and to strain
these close to the grp failure strain (\(\sim 0.03\)), counting the number of
fractures at each stage microscopically. Unfortunately the grp would fail

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before the projected strength of the 100 µm gauge length carbon fibres was
reached, but Fig. 81 suggests that the minimum spacing of carbon fibre
fractures attainable would be about 7-12 ineffective lengths (0.7 - 1.2 mm),
with a maximum equal to the gauge length, 100 mm. This would cover roughly
two decades of effective gauge length, and would include the conventional
single fibre results spanning only a factor of 5 range of gauge length. If
other aspects of the exercise attempting to relate fibre strength to
composite strength were improved, it would be worthwhile to develop the
techniques for fabrication and microscopy of such embedded single fibres.

The dashed curve drawn through the experimental points for the
hybrids in Fig. 81 is of similar shape to the family of predictions from
Harlow and Phoenix (39), and by interpolation would coincide with one of
these curves for w ≈ 12. The predicted strength curve for w = 7 (based on
the single fibre tests) lies substantially below these experimental results.
Over the range of (mn) for which there are experimental results the strength
curves are fairly straight, (indicating approximate Weibull behaviour), and
of similar gradient, so virtually any predicted curve in the range w = 7-15
would fit within the error margins with an appropriate vertical shift. A
relatively small change in the slope of the Weibull straight line through the
single fibre results could give such a shift. Alternatively the W = 7 curve
could be made to fit by a horizontal shift of 5 ln units, which is a 150
fold reduction in the number of elements (mn). This can only be accomplished
by a change in the ineffective length, which would then be 1.5 mm, since the
number of fibres is fixed. Such a large ineffective length is not in accord
with the observations of debonding at isolated fibre fractures, but is not
implausible in the case of larger groups of fibre fractures. The statistical
model, however, assumes a constant ineffective length, and this is one reason
why the predictions themselves could be inappropriate for the actual failure
mode observed. The model is also unrealistic in its assumption of a uniform
overstress in the ineffective length, in the adoption of a very localised
load sharing rule, and is approximate in that convergence is assumed for ~ 9
fibres, and because weak link behaviour is assumed in scaling G (σ) back to
single fibre size to obtain W (σ) for the various bundle sizes. The analysis
is based on a 3-D model, but the laminated and spread-tow hybrids are to some
extent 2-Dimensional in nature. Harlow and Phoenix have not yet published
results for a more wide ranging load sharing rule, but suggest that the main
effect would be to increase the number of fibres required for convergence.

Some of these points could be resolved by additional experimentation.
The two most productive areas would be a more accurate determination of single
fibre strengths, when tested in isolation, and also incorporated into a glass fibre composite, particularly at short gauge lengths, and the investigation of a wider range of bundle sizes. In theory the bundles could be tested in isolation, simply as a test of the statistical model of composite strength, but incorporating the bundles into hybrid specimens has a number of advantages. Not least is the relative ease of fabrication, handling, and testing, but hybridization also allows a large number of fractures to be observed in a single specimen. Other advantages are the reduction of surface effects, ease of strain measurement, and the containment of the fractures enabling the fracture mechanism to be deduced. There are interesting possibilities for measuring the lower probabilities of failure by monitoring the changes in electrical resistance in the fibre direction during straining. A reduction in the scatter and the extension of the experimental results to larger and smaller bundle sizes would be a check on the approximate 'weak link' behaviour which assumes that changes in volume arising from changes in the number of fibres or in the gauge length are equivalent. It is also conceivable to design an experiment to measure the ineffective length. If the strength of the bond between the carbon fibres and the matrix is varied, for example by suitable combinations of resin system and level of fibre surface treatment, the length of debonding at carbon fibre fractures would be altered, and this would change the number of elements (mn) for a given size of composite body. The scatter in the present data would require a difference in (mn) of about two ln units, or a factor of \( e^2 = 7.4 \) difference in effective length for such an effect to be seen. It would be rather difficult to alter the debonded length by such a large factor, but if the scatter could be reduced a smaller difference might be acceptable.

To summarise, the already good agreement between experiment and the Harlow and Phoenix strength predictions could be further improved by consideration of the following factors:
(i) Inaccurate extrapolation of single fibre strengths to short gauge lengths where they do not fit a Weibull distribution.
(ii) A difference in the variability of fibre strength between fibres tested in isolation, and when surrounded by other fibres in a composite.
(iii) Over simplification of the load redistribution at fibre fractures.
(iv) The 2-Dimensional nature of the hybrids.
(v) The failure mode where the ineffective length increases with the number of adjacent fractured fibres in a group, and which therefore does not fit the assumptions of the simple model.

These qualifications do not invalidate the conclusions that
(i) strength differences between hybrids are simply the result of differences in the volume of their CFRP components, and (ii) that the all-carbon laminates do not realise the full potential strength of the fibre, but are limited in strength by some other mechanism. Possible reasons for this weakening are discussed next.

The all-carbon laminate strengths plotted in Fig. 81 fit the straight line drawn through them with a correlation coefficient of 0.91, but it is unlikely that this trend can persist much beyond the decade range of size studied, because ultimately as the specimen is reduced in size it becomes a single element, whose strength is given by the projected intercept of the single fibre results on the strength axis. Similarly there must be an upper limit to strength as the specimen volume is increased, and it is hard to see how this can be greater than the carbon fibre strengths in the hybrids, or those reported for larger structures, (which are similar to that of the 12-laminae carbon laminate). Indeed this laminate was chosen as the control because its strength appeared to be 'typical' of manufacturers' and other workers' results. In the absence of thermal strains, strength is almost certainly limited by flaws, and there are indications that a specific weakening mechanism operates over the range of thicknesses tested.

First we may discount the effect of specimen bending through misalignment in the testing machine as being the source of stress inhomogeneities as large as the size effects observed as the curvatures required would be so large as to be obvious. Apart from the volume differences which do not explain the size effect on a statistical model, the effect could be related to differences in thickness, surface area, cross-sectional area, or the aspect ratio of the cross-section. Any explanation based on the probability of there being a flaw of given severity on the surface or within the test piece would predict a trend of opposite direction to that observed.

The most probable explanation involves an intrinsic defect of constant size whose weakening effect depends on the thickness of the laminate. This defect could be a volume of misaligned fibre which would enhance the the general stress level in the surrounding material where fracture would then be more probable, rather than initiating the fracture itself. In the following analysis of this situation fracture is still assumed to initiate and develop in this overstressed region by the accumulation of fibre fractures in groups, one of which ultimately propagates catastrophically through the section. The composite as a whole is therefore weakened according to the over-stress in the region of the defect. In order to recognise the effectively 2-Dimensional nature of relatively wide strip specimens cut from thin laminates,
width equal to the thickness of the laminate \( t \), Fig. 82. The defect is of width \( t \), and extends a depth \( s \) through the thickness. The laminate as a whole can be considered as a number of such sections coupled together by the matrix. The material in the immediate vicinity of the defect will experience an overstress which is relatively independent of the rest of the laminate. A very approximate estimate of the strain concentration in the unflawed part of the section can be made by redistributing the entire load which would normally be carried by the defect, equally over the remaining area of the square. The strain concentration is then

\[
K = \frac{t}{t-s}
\]

If ultimate failure results from the failure of the overstressed fibres in this square, the failure strain at thickness \( t \) is

\[
\varepsilon_{\mu}(t) = \varepsilon_\omega K^{-1} = \varepsilon_\omega \left(1 - \frac{s}{t}\right)
\]

where \( \varepsilon_\omega \) is the failure strain of an unflawed laminate. \( \varepsilon_\omega \) will in fact be the same as the hybrid's failure strain given by the statistical model, and will change with specimen size, but for the purposes of this approximate analysis it is assumed constant because it changes much less rapidly than does the all-carbon composite failure strain, Fig. 81. A graph of \( \varepsilon_{\omega} \times t^{-1} \) should be linear with slope \(-\varepsilon_\omega s\), and intercept \( \varepsilon_\omega \), and this is approximately so (correlation coefficient 0.93) as shown in Fig. 83. The intrinsic strain limit \( \varepsilon_\omega \) is 0.113, which is about 10% below that of the carbon fibre in the laminated hybrids. The estimated depth of the defect \( s \) is 43\( \mu m \), or about 5 fibre diameters. This is slightly over one third of the thickness of a pre-preg lamina. However, only single fibre fractures have been found in the all-carbon laminates even after stressing to failure, so the defect is unlikely to be an accumulation of fibre fractures. If, instead of a crack, the strain concentration was caused by a region of lower longitudinal modulus, the defect would need to be proportionally larger. The pre-pregs as supplied contained slack tows and portions of tows which adopted a sinusoidal aspect when rolled into the sheet, and also areas which contained sworls of loose fibre, both of which gave rise to substantial regions of local mis-orientation in the moulded composite, such as those in Figs. 20 and 28. The typical mis-orientation in one of these areas is about 15°, and its longitudinal modulus would be reduced by up to 60% according to calculations by Bishop (67) for a similar fibre and resin combination, Fig. 84. In this case the depth of the defect would be proportionally greater, about 110\( \mu m \), or very nearly the thickness of a pre-preg, and this is quite plausible. This very approximate analysis is intended only to demonstrate that a constant size of
defect, considerably larger than the fibres, but of the same order of thickness as the pre-preg, could account for the variation of strength with thickness. The weakening is probably the result of misaligned bundles of fibre which are a feature of the pre-preg material, although it is rather surprising that they apparently have such a profound effect, because by their nature the transition between aligned and misaligned zones is gradual. There is some support for this mechanism of weakening in the work of Hughes, Morley and Jackson (53,54) who found comparable strength differences between impregnated tows fabricated in the usual manner, and those with specifically aligned fibres moulded in shrink-fit tubing.
3-3-1 Failure at the microscopic level

As increasing strain is applied to any of the composites the first microscopic failures are isolated fractures of the carbon fibres. In the 828 epoxy system debonding is seen in unloaded specimens between the fractured fibre and the matrix, Fig. 71, and this almost certainly occurs at the time of fracture. Similar observations were not possible in Code 69 resin on account of its cloudiness. The load at which the fracture occurs, the configuration of other fibres surrounding the fracture, the condition of the fibre matrix interface, and residual strains in the matrix are all factors which could determine whether or not the fibre debonds. However, it is not known which if any of these is the dominant factor. Where the fibre has not debonded there is presumably an intense plastic or elastic strain field over a similar volume of material, but there is no indication of how the load distributions in the fibre might differ in the bonded and debonded cases. It is possible that initially bonded fractures debonded at a later stage of failure. As the strain is further increased, fibre-fractures continue to occur in isolation, but also in the regions over-stressed by other fibre fractures, and groups accumulate, Fig. 71. In all-carbon laminates only single fractures have been observed using the electrolytic decoration technique, and it may be deduced that the maximum size of a stable group is rather small, but in the case of the spread-tow hybrids the carbon fibres have fewer carbon fibre neighbours and the constraint resulting from the glass fibres allows larger groups of carbon fibre breaks to occur before catastrophic propagation ensues. No glass fibre fractures have been observed in any of the hybrids, and this is the essential feature which allows the progressive multiple cracking failure modes. Where there is a group of carbon fibre fractures in the spread-tow hybrids there is always debonding from the matrix, Fig. 71, and it is more extensive for larger groups. It is not clear whether the individual debonds at each fracture grow as more fractures add to the group or whether the debonded area observed is a sheath surrounding the whole group, with smaller debonds around each fractured fibre within the group. As a group of fractures increases in size the ratio of the debonded surface area to the volume it encloses diminishes, and the debonding of an outer sheath would be expected to become progressively larger. There appears to be a limit to the debonding of about 50\,\mu m to either side of individual carbon fibre fractures, and this length also manifests itself as the typical short range fluctuation in the fracture path seen in a longitudinal section through a macroscopic carbon-ply fracture, Fig. 28, and also as the pull-out length on a fracture surface. In the spread-tow hybrids there is no clear distinction
of fibre bundles, but in the case of the divided-tow hybrids the bundle size is very much larger than the critical group of fractures, and the distinction is obvious, and the sheath of debonding surrounding the whole bundle, Fig. 48, is much larger (≈ 10x) the ineffective length. This is presumably because the carbon fibre bundle fails essentially as a single ligament, releasing considerably more strain energy than a single fibre. Also, the shear failure occurs at an interface between grp and cfrp which is somewhat weaker (≈ 0.7x) than the cfrp/cfrp interface as measured in short beam interlaminar shear tests, which could make a marginal contribution to increased debond length. Most of the observations of carbon fibre fractures were made in the spread-tow system, but the behaviour in the other composites is expected to be very similar.

There is however a considerable difference in the amount of cracking in the two resin systems, being greater in Code 69 Figs. 61, 62, than in the 828 epoxy Figs. 65, 66. There is no evidence that the matrix cracks significantly influence the longitudinal strength, and this is not expected since the matrix represents less than half the cross-sectional area, and is roughly an order of magnitude less stiff. For this reason there is no detectable modulus change accompanying the cracking. The main interest in the fine matrix cracking is its contribution to the acoustic emission, but even when this is comparatively small it is not possible with the present equipment to distinguish between noise emanating from fibre fractures, debonding and delamination, or movement at debonded interfaces. It is unlikely that acoustic emission monitoring could easily be developed into a reliable quantitative technique for measuring the number of fibre fractures. In the 828 epoxy system it is useful as a semi-quantitative indication of general activity associated with the carbon fibres. It is interesting to note that the matrix cracks in Fig. 65 are a particularly good example of the multiple cracking phenomenon described and analysed in refs. (18-26, 63). They illustrate near-random crack positions at low strain, which become more uniform as the strain is increased. There are many instances where several cracks can be found roughly twice as closely spaced as the otherwise fairly regular array. The shorter matrix cracks are found at higher strain in accordance with the ideas of constrained cracking where a region of matrix is bounded by stiff fibres. The shorter cracks are more closely spaced demonstrating the effect of the shorter transfer length. Although these fine matrix cracks appear to have little influence on the tensile strength of the composite, they could have undesirable effects on the matrix dominated properties such as shear modulus and strength, and they are also a capillary pathway for the ingress
of moisture or other media to the fibre/matrix interface, acting in parallel with the diffusion path. It has not been established whether the larger matrix cracks in the carbon layer of the spread-tow hybrids pass through the fibres because they can only be seen clearly in transmitted light which does not show up fibre fractures. They appear to be most common in areas where there are relatively few carbon fibres, and are probably associated with large concentrations of resin. A more even distribution of fibres would remove these resin rich areas in which the cracks first appear, but a more effective way to eliminate them would be to use a matrix with higher failure strain (Garrett and Bailey (24)).

3-3-2 Macroscopic aspects of failure

The delamination which accompanies fracture of the cfrp is the major feature governing the load/extension behaviour. The simple ideas developed in Chapter 1 to predict whether or not the first failure of the cfrp element will precipitate catastrophic failure of the hybrid as a whole cannot be applied to most of the hybrids in this study because of the load drop which accompanies the delamination in displacement controlled loading. For those hybrids in which the proportion of the hybrid reinforced with carbon fibre, $P_c$ is greater than $P_{c\text{crit}}$ (the critical value above which failure is expected to be catastrophic), the length of delamination is generally large, often extending the entire gauge length, and consequently the load drop is large and pre-empts catastrophic failure. The value of $P_{c\text{crit}}$ given by eq. (2) is 24% or 35% for HTS- and HMS-carbon fibre/glass fibre hybrids respectively. Only the HMS/HTS-carbon laminated hybrid failed catastrophically, but in this case $P_c$ was 5.5% and well below the predicted $P_{c\text{crit}} = 22\%$, so in this system the localised stress concentration at the fracture in the HMS ply dominates the effects of the general increase in net section stress. At the microscopic level this means that the cdf's for the failure strain of both types of carbon fibre overlap sufficiently for the crack to propagate through both without interruption, whereas the glass fibre cdf is high enough for it to be an effective barrier to propagation.

The number of carbon-ply cracks and the extent of the delamination at each determine the effective stiffness of a hybrid and its ultimate strain at failure. There is no evidence that the fractures are in any way detrimental to the properties of the glass-plies which are expected to fail at a similar ultimate strain to all-glass composites. The ultimate strength of the glass-plies is limited by a mechanism involving shear failure in the resin, and for this reason has not been investigated in detail. To predict the load/strain behaviour of a glass/carbon hybrid it is necessary to consider both the probab-
The energy based model for the length of a delamination crack agrees reasonably well with experiment over a rather limited range of glass-ply thickness in laminated hybrids containing a single carbon fibre pre-peg lamina, and the assumptions of load transfer between plies by a low shear stress over most of the delaminated interface, with a higher shear stress at the crack tip are generally substantiated by the laser moiré study of the fractures, Figs. 37-41, 43-46. Friction is probably responsible for load transfer in the lower stress region, although this could also include bridging of the crack by unbroken fibres, and mechanical interference between the fracture surfaces. The estimate of this frictional stress from the energetic model ($\sim 0.6$ MPa) is of the same order as the upper limit estimated from the laser moiré work ($\sim 1.5$ MPa). Neither technique is sufficiently sensitive to permit a reliable measurement of the stress. If the energetic model is correct, and the length of the delamination in laminated hybrids is inversely proportional to the frictional shear stress, it is of considerable importance for the load/extension behaviour. Unfortunately the delamination length is extremely variable, either as a result of severe variations in the interface strength and geometrical factors, or because of its extreme sensitivity to differences in the load transfer rate, and it has not been possible to verify the general applicability of the energetic model for other than laminated hybrids. In view of the wide range of delamination lengths encountered, from 1 mm to $>100$ mm, it is unlikely that the frictional stress would be constant for all geometries or lengths of delamination, or even in all regions of a single delamination. A lower average frictional stress would be expected in the laminated hybrids when the delamination is large because the glass-plies are then more easily separated reducing the normal force at the interface.

The laser moiré study of the hybrids containing a half-length carbon-ply demonstrated the existence of a 2-3 mm zone of much higher shear stress behind the delamination crack-tip, and a typical value of 40-60 MPa at the failed interface can be inferred from the surface strain gradient, making allowance for diffusion of the strain gradient through the thickness of the glass-plies. The fact that the zone is behind the crack-tip, and the evidence of residual strains on unloading suggest that the load transfer mechanism is not elastic. The alternatives are friction, plasticity and bridging fibres. The lighter appearance of the delaminated area implies some kind of interface failure, although this could take the form of micro-cracks in a field of
There is some evidence of bridging fibres from the micrograph in Fig. 69, and the fragments adhering to the fracture surfaces in Fig. 27. It is probable that all these mechanisms contribute to the load transfer.

The most notable characteristic of this second series of laser moiré specimens was the transition from controlled growth of the delamination, to stick-slip behaviour when its length exceeded 2-3 mm, the length of the high shear stress zone at the crack-tip. This is presumably caused by a breakdown of the mechanism responsible for the high rate of load transfer when the delamination reaches a critical length, and this is quite possible if the mechanism involves plasticity or bridging fibres. If the load transfer mechanism did not break down, but maintained some maximum shear stress, the delamination would continue to propagate in a controlled manner in a very similar manner to that described by Wright (55) for a butt-strap joint in cfrp which he has analysed by finite element methods. The subsequent stick-slip propagation implies strain rate sensitivity, and it has not been determined how well the behaviour of the laser moiré specimens represents the rapid growth of the delaminations when they accompany a carbon-ply fracture.

With the current level of understanding of the delamination it is not possible to accurately predict the form of the load/extension curve beyond the onset of multiple fracture of the cfrp, but if the characteristic delamination length is known, a reasonable estimate of the modulus of laminated hybrids, (and by inference also that of mixed-tow hybrids) could be made by assuming that no load is carried in the carbon-ply within the delaminations when they are longer than about 3 mm. Hybrids in which carbon and glass fibres are more intimately mixed are of potentially greater interest because the damage associated with individual carbon fibre fractures in less severe and more evenly distributed, and because there is also the possibility of significantly improving the utilization of the carbon fibre's strength by reducing their bundle size.

A study of hybrids fabricated with different proportions of spread-tows of both glass and carbon fibres in an epoxy resin matrix has demonstrated a change of failure mode from the progressive behaviour typical of all-glass composites, to a 'brittle' mode more typical of all-carbon fibre composites, as the proportion of carbon fibre is increased. (Wood (68)). The transition to catastrophic failure is expected at \( P_{\text{crit}} \) because there is only a very small load drop accompanying the delaminations and debonding at fractures of
hybrids extended the simplified 3-layer sandwich construction to a structure of greater practical interest, and their behaviour is very similar to that of the simpler hybrids of similar dispersion and fibre ratio.
The least well understood aspect of this work is the dependence of the strength of the all-carbon laminates on specimen size. Further work is required to establish the mechanism of weakening in the specimens laminated from pre-preg, but this system is too inflexible and variable to permit a systematic investigation of fibre alignment and specimen thickness. The alternative is a wet lay-up system with which tows of carbon fibre would be fabricated into laminates which are subsequently cut into test coupons, or single bundles of desired size could be fabricated indivually. The main objectives would be to determine the factors influencing tensile strength, and to test further the general applicability of statistical models of strength to all-carbon composites fabricated in different ways. If these fabrication procedures do introduce imperfect areas, this effectively alters the population of flaws, which in this work has been assumed to contain only defects in single fibres. It is possible that similar statistical arguments could be applied to these larger flaws. A better understanding of all-carbon composites is required before the value of hybridization with other fibres can be properly appreciated.

HMS/HTS-carbon fibre hybrids would be of considerable interest if the failure strain of the HMS fibre could be enhanced to that of the HTS fibre by reducing the size of the bundles of HMS fibre. Model hybrids containing small bundles of HMS fibre of known size in HTS fibre composites could be used to investigate the critical size for propagation of a group of fractures directly.

Hybrids with a uniform intimate mixing of fibres would be a fruitful area for future study now that the principles of their behaviour are known, but improved methods of dispersing the fibres would be desirable. Intermediate levels of dispersion can be achieved more simply by combining tows of different fibres and different weights. In less well dispersed glass/carbon fibre hybrids the delaminations between volumes reinforced with each fibre is a major feature of their failure, and it would be interesting to investigate other fibre and matrix systems to determine to what extent it is a function of the fibre/matrix adhesion, or relative moduli and strain to failure of the fibres. This delamination has only been reported for carbon/glass hybrids, whereas catastrophic failure has been reported for carbon/carbon by Edwards et al (17) and for carbon/Kevlar by Zweben (10), but only for a limited number of geometries.
significantly to the apparent enhancement of the failure strain of the crimp in carbon/glass hybrids, but improved techniques are required to establish how these strains are influenced by the cure cycle and the shrinkage and creep properties of the matrix. This would reduce the uncertainties in comparing the carbon fibre strength with statistical predictions, and would also have a practical benefit in indicating suitable fabrication procedures to optimise the favourable effect of the thermal strains.
CONCLUSIONS

(a) This work has been concerned with the tensile failure of a range of unidirectional composites in which the reinforcing glass and carbon fibres are mixed to give a variety of structures. The results confirm earlier reports of enhancement of the first failure strain of the carbon fibre in hybrid composites with respect to similar composites containing only carbon fibres.

(b) Part of this Hybrid effect has been shown to arise from differential thermal contractions which result in a residual compressive strain in the carbon fibre when the hybrid composites are cooled after curing of the resin matrix, but these strains account for only a small proportion of the effect.

(c) The strengths of the composites have been rationalised by presenting them in the form of a graph of the strength against the discrete volumes of cfrp, (on logarithmic axes). This presentation is essentially a Weibull probability graph, and enables the strength of specimens with different cfrp volumes to be compared on the basis of a statistical model which predicts the strength of the composite in terms of the strength distribution in single fibres. In the case of the hybrid composites there is good agreement with the numerical predictions from such a model presented by Harlow and Phoenix, showing that volume is the parameter of overriding importance limiting the strength realised in the cfrp component. This volume dependence has 'weakest link' characteristics, whereby the strength is limited by the catastrophic propagation of a small group of fibre fractures which has exceeded some critical size. In contrast the strengths of the all-carbon fibre composites are lower, and increase with volume instead of decreasing as expected from statistical considerations. This behaviour is thought to be caused by fibre misalignment, but is not fully understood.

Three factors therefore contribute to the hybrid effect:-
(i) Residual strains of thermal origin
(ii) The variation of the first failure strain of the cfrp component of a hybrid with its volume.
(iii) Thickness related variations in the defect-limited strength of the all-carbon fibre composites with which the hybrids are compared.

(d) Failure in all the carbon/glass fibre hybrids involved fracture of the cfrp, with localised delamination or debonding failure of the interface
between the grp and cfrp, but without fracture of the grp. When the extent of this delamination was less than the length tested further similar failures occurred as the load was increased, each contributing to a progressive loss of stiffness. The size of the delamination is extremely sensitive to the geometry, and has only been predicted successfully for a limited range of laminated hybrids.

(e) Laser moiré strain measurements in the failure zone indicate high shear stress in a small region of the failed interface near the delamination crack tip where it approximates to the interlaminar shear strength, and a very much lower 'frictional' stress elsewhere in the delamination which is approximately 1% of the interlaminar shear strength. The shape of the load/extension curve can be explained in terms of the redistribution of load in the glass fibre surrounding the fractures in the cfrp.

(f) The single HTS-carbon/HMS-carbon fibre hybrid tested failed catastrophically at an intermediate strain between those expected for each of the constituent fibres, and demonstrated that the lower elongation fibre has a weakening influence on the higher elongation fibre. The failure is catastrophic because there is too small a difference in the failure strains of the two fibres to obtain "hybrid toughening" effects.

(g) Most of the hybrids tested contained only a single structural element of cfrp, but the behaviour of hybrids laminated from thin (≈100μm) layers has been shown to remain similar as the number of plies is increased, while maintaining the same dispersion and ratio of fibres.

(h) The number of fibre fractures in unstrained composites is negligible, but as load is applied carbon fibres fail initially at isolated positions, and then accumulate in small groups on account of the over-stressing of the adjacent fibres. These groups have a maximum size in the range 2-5, beyond which there is catastrophic propagation of a crack through all the carbon fibres in a bundle.

(i) Regions reinforced with glass fibre act as barriers to the propagation of cracks between bundles of carbon fibre.

(j) Techniques have been developed to examine these fractures microscopically, and have revealed extensive matrix micro-cracking, but this does not have any effect on the primary tensile properties.
Overall the work has been successful in its aim of clarifying the nature of the so called hybrid effect. The failure mode of the carbon/glass composites, in which multiple fractures of the CFRP are contained, has allowed the more fundamental aspects of composite strength to be investigated, and there are good prospects for the continuation of research along these lines. The results demonstrate there is a considerable advantage in dispersing the low-elongation fibre as a number of small bundles, but further research is required to make this a practical proposition beyond the laboratory scale.
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<td>54</td>
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(c) Isolated " " "
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<th>Melting point °C</th>
<th>Density x 10^3 kg m^-3</th>
<th>Strength (GPa)</th>
<th>Modulus (GPa)</th>
<th>Specific strength x 10^6 m^-2</th>
<th>Specific modulus x 10^9 m^-2</th>
<th>Thermal expansion coeff x 10^-6°C^-1</th>
<th>Diameter (μm)</th>
<th>Reti of cost to that of E-glass fibre</th>
<th>Date</th>
<th>Source of data (major)</th>
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<td>28</td>
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<td>1</td>
<td>'78</td>
<td>(a,c)</td>
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Data sources: (a) Bib.1, (b) Courtaulds Ltd. Data sheet FC 12, (c) Bib.4, (d) Ref.16.
# Mechanical Properties of Hybrid Composites Described in the Literature

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<th>Fibre type, lay-up sequence, Rule of Mixtures predictions</th>
<th>Fibre volume fraction</th>
<th>Modulus (GPa)</th>
<th>Failure strain, specified. (GPa)</th>
<th>Failure strain, case stress</th>
<th>Fracture reinforced with low-elongation fibre</th>
<th>Hybrid composite reinforced with low-elongation fibre, %</th>
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<td>All type I carbon, All type II carbon, (1 type I, 1 type II), R M prediction</td>
<td>0.6 205</td>
<td>0.50</td>
<td>0.0024</td>
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<td>0.0042</td>
<td>0.5</td>
<td>75</td>
<td>2/2 &amp; 5/5 hybrids are laminated.</td>
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<tr>
<td>All IWS carbon, A-S carbon (1 IWS, 1 AS)</td>
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<td>137</td>
<td>0.66</td>
<td>0.0068</td>
<td>0.5</td>
<td>10</td>
<td>Most composites tested in flexure, but failure criterion for which stress is specified is not given.</td>
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<td>117</td>
<td>0.60</td>
<td>0.0056</td>
<td>0.5</td>
<td>168</td>
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<td>0.50</td>
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<td>0.5</td>
<td>168</td>
<td>Failure mode is not given.</td>
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## Bunsell & Harris, (Ref. 6)
- All glass 0.6 142 0.40 0.0026
- All hybrids laminated from pre-preg. Epoxy matrix.

## Zwehen, (Ref. 10)
- All kevlar - - - - 0.180
- All carbon - - - - 0.014
- Thermal contraction differences not considered.

## Aveston & Sillwood, (Ref. 7)
- All glass - - - - 0.0050
- Veil is 10⁶ filament tow spread to 300 mm width, and combined with 204 filament glass twp in RT curing epoxy. Coeff. var. 18%.

## Edwards, Parratt, Potter, (Ref. 17)
- All IWS carbon 0.5 97 1.16 0.012
- Discontinuous aligned fibres in epoxy resin matrix. Tensile test.

## Phillips, (Ref. 2)
- All glass 0.6 40 0.73 0.020
- Unidirectional woven hybrid cloth with tows of glass and carbon alternating in the ratio given. Vinylester matrix. Thermal strains not considered.

## Other Comments
- COE, var. 15-17%.
- 3-layered sandwich. COE, var. 15-17%.
- Failure mode of all hybrids is brittle, COE var. 10%.
- Fibres intimately and randomly mixed before incorporation into pre-preg.
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<td>Polyester/epoxy compatible finish. Silenka Ltd.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>HMS-carbon fibre:</td>
<td></td>
<td>✓</td>
<td></td>
</tr>
<tr>
<td>High modulus, surface treated. Courtaulds Ltd.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(Pre-preg supplied by F&amp;H Ltd.)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>HTS-carbon fibre:</td>
<td></td>
<td>✓</td>
<td></td>
</tr>
<tr>
<td>High tensile, surface treated. Courtaulds Ltd.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(Pre-preg supplied by F&amp;H Ltd.)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>HTS-carbon fibre:</td>
<td></td>
<td>✓</td>
<td>✓</td>
</tr>
<tr>
<td>Unsized tow, surface treated. Courtaulds Ltd.</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

130
### Laminate Properties

<table>
<thead>
<tr>
<th>Lay-up sequence</th>
<th>Dispersion (m²)</th>
<th>Carbon fraction (Yc)</th>
<th>Modulus (GPa)</th>
<th>Failure stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Hybrid Effect/Size Effect</th>
<th>Thermal strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>3</td>
<td>3.67</td>
<td>0.00</td>
<td>45</td>
<td>1.20</td>
<td>0.07</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>12</td>
<td>3.67</td>
<td>0.00</td>
<td>44</td>
<td>1.21</td>
<td>0.08</td>
<td>-</td>
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<td>-</td>
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</table>

**E-Glass composites**

<table>
<thead>
<tr>
<th></th>
<th>No. of laminae</th>
<th>Fraction</th>
<th>Stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Hybrid Effect</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
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<td></td>
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</table>

**High Tensile carbon: All-carbon composites and hybrids**

<table>
<thead>
<tr>
<th></th>
<th>No. of laminae</th>
<th>Fraction</th>
<th>Stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Hybrid Effect</th>
</tr>
</thead>
<tbody>
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</tbody>
</table>

**High Modulus carbon: All-carbon composites and hybrids**

<table>
<thead>
<tr>
<th></th>
<th>No. of laminae</th>
<th>Fraction</th>
<th>Stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Hybrid Effect</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
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<td></td>
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<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
</tr>
</tbody>
</table>

**High Modulus carbon / High Tensile carbon hybrid**

<table>
<thead>
<tr>
<th></th>
<th>No. of laminae</th>
<th>Fraction</th>
<th>Stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Hybrid Effect</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
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<td></td>
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</tbody>
</table>

### Properties of Wet Lay-up Composites

<table>
<thead>
<tr>
<th>Lay-up</th>
<th>Modulus (GPa)</th>
<th>Failure stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Closest spacing of fractures (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>All HTS-carbon (1mm)</td>
<td>135</td>
<td>1.52</td>
<td>0.0112</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>All E-glass (2mm)</td>
<td>35</td>
<td>&gt;1.2</td>
<td>&gt;0.05</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

**Divided-tow Hybrids**

<table>
<thead>
<tr>
<th>Fibres</th>
<th>Modulus (GPa)</th>
<th>Failure stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Closest spacing of fractures (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>10,000</td>
<td>3.620</td>
<td>-</td>
<td>-</td>
<td>14.5</td>
<td>7.6</td>
</tr>
<tr>
<td>1,620</td>
<td>3.620</td>
<td>-</td>
<td>-</td>
<td>0.71</td>
<td>3.5</td>
</tr>
<tr>
<td>650</td>
<td>3.620</td>
<td>-</td>
<td>-</td>
<td>~0.25</td>
<td>3.5</td>
</tr>
</tbody>
</table>

**Spread-tow Hybrids**

<table>
<thead>
<tr>
<th>Glass-plies</th>
<th>Modulus (GPa)</th>
<th>Failure stress (GPa)</th>
<th>Failure strain</th>
<th>Debond length (mm)</th>
<th>Closest spacing of fractures (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1mm</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>&lt;1</td>
</tr>
<tr>
<td>0.5mm</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>&lt;1</td>
</tr>
<tr>
<td>Specimen</td>
<td>Number of 0.125 mm pre-preg laminae in each ply</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>----------</td>
<td>-----------------------------------------------</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>Glass</td>
<td>Carbon</td>
<td>Glass</td>
<td></td>
<td></td>
</tr>
<tr>
<td>A</td>
<td>3</td>
<td>1</td>
<td>3</td>
<td></td>
<td></td>
</tr>
<tr>
<td>B</td>
<td>4</td>
<td>1</td>
<td>4</td>
<td></td>
<td></td>
</tr>
<tr>
<td>C</td>
<td>6</td>
<td>1</td>
<td>6</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Predicted $\varepsilon_{max}$</th>
<th>Observed $\varepsilon_{max}$</th>
<th>Shear stress (MPa)</th>
<th>Frictional shear stress $T_f$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>1.55</td>
<td>1.77</td>
<td>21.1</td>
<td>&lt;1.4</td>
</tr>
<tr>
<td>B</td>
<td>1.41</td>
<td>1.41</td>
<td>17.1</td>
<td>&lt;0.45</td>
</tr>
<tr>
<td>C</td>
<td>1.27</td>
<td>1.32</td>
<td>21.9</td>
<td>&lt;1.5</td>
</tr>
</tbody>
</table>

**Series Two Laser Moire Specimens**

(With half-length carbon-ply)

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Number of 0.125 mm pre-preg laminae in each ply</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Glass</td>
</tr>
<tr>
<td>D,E</td>
<td>3</td>
</tr>
<tr>
<td>F</td>
<td>6</td>
</tr>
<tr>
<td>G</td>
<td>3</td>
</tr>
<tr>
<td>H</td>
<td>6</td>
</tr>
</tbody>
</table>

**TABLE 6**

Probability of Fibre Failure in Spread-tow Hybrids.

<table>
<thead>
<tr>
<th>Number of undecorated fibres.</th>
<th>Strain in carbon fibre (thermally corrected)</th>
<th>Fibre stress (GPa)</th>
<th>Range of probability of fibre failure, allowing one less broken fibre. $x 10^{-6}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>0.007</td>
<td>2.17</td>
<td>0 - 0</td>
</tr>
<tr>
<td>1</td>
<td>0.009</td>
<td>2.65</td>
<td>0 - 83</td>
</tr>
<tr>
<td>1</td>
<td>0.012</td>
<td>3.37</td>
<td>0 - 83</td>
</tr>
<tr>
<td>3</td>
<td>0.015</td>
<td>4.09</td>
<td>167 - 250</td>
</tr>
<tr>
<td>12</td>
<td>0.019</td>
<td>4.81</td>
<td>916 - 1000</td>
</tr>
<tr>
<td>90</td>
<td>0.022</td>
<td>5.53</td>
<td>7410 - 7500</td>
</tr>
</tbody>
</table>

Based on sample of approx 12,000 fibres.
Fig. 1 Possible structures for unidirectional hybrids with different levels of dispersion.

Fig. 2 Test coupon designs for:
(a) Spread-tow hybrids
(b) Divided-tow hybrids
(c) Laminated hybrids.
Fig. 3  Rule of mixtures behaviour of the Young's Modulus of a hybrid.

Fig. 4  Property vs. composition diagrams for,

(a) failure stress
(b) failure strain
(c) load in the glass and carbon fibre reinforced components.

contd...
Load Strain

R.O.M.

Percentage Carbon

0% 100%

(b)

Constant strain

R.O.M.

Extra load on glass

Carbon load

D 0% 100%

(c)

135
Fig. 5 The build-up in load in a fractured member sandwiched between unfractured members for the case of frictional load transfer (dashed) and elastic load transfer, (solid).

Fig. 6 The load distribution in the cracked member at three stages of multiple cracking, increasing in strain from (a) to (c).
Fig. 7 Schematic load/extension curve for a hybrid with successive fractures of the low-elongation component.

Fig. 8 The effect of residual compressive strain in the cfrp of a hybrid on its load/strain behaviour.
Fig. 9  The variation of intrinsic and critical defect sizes with applied stress or strain.

Fig. 10  Load/extension curve for a single fracture occurring at constant load.
Fig. 11 A composite conceived as a chain of slices, each of which is a bundle of fibres.

Fig. 12 Convergence of predictions for composite strength for (a) $n = 1-9$ fibres, $w = \rho = 5$
(b) $n = 1-9$ fibres, $w = \rho = 10$
(c) convergence limit for $w = 3-50$

From Harlow & Phoenix, ref. (39). contd...
Convergence of the transformed probability distribution for bundle strength as bundle size increases. Local load sharing (LLS) is assumed.

Conjectured limit for the transformed distribution of bundle strength as bundle size \( n \) increases (actual results for \( n = 9 \)).
Composite: pre-preg or tows wound on frame (not shown)

Fig. 13 Open-ended trough mould, with shims used to determine the thickness of the moulded composites.

TOW SPREADING APPARATUS

Fig. 14 Divergent-flow apparatus for spreading a tow of carbon fibre.
Fig. 15 The divergent-flow nozzle used to spread a tow of carbon fibre.

Fig. 16 The definition of dispersion for 3-ply, and multiple-ply laminated hybrids.
Fig. 17 The range of laminated hybrids tested, based on:

(a) HTS-carbon / E-glass fibre
(b) HMS-carbon / E-glass fibre

Solid circles represent sandwich constructions, and open circles represent multiple-layer laminates.
Fig. 18 Acoustic-emission pulses showing 'ring-down'.

Fig. 19 The system used to monitor acoustic emission.
Fig. 20  Tensile failure of an all-carbon fibre composite.

Fig. 21  Tensile failure of an all-glass fibre composite.
Fig. 22 The variation of the carbon-ply first failure strain with the fraction of carbon fibre, and the dispersion, for hybrids based on, (a) HTS-carbon, and (b) SMS-carbon fibre. Solid bars indicate the contribution of thermal strains to the increase in failure strain.
Fig. 23  Acoustic emission/strain trace for an all-carbon fibre composite. (12 ply, HTS-carbon)

Fig. 24  Acoustic emission/strain trace for an all-glass fibre composite. (12 ply laminate)
Schematic diagram of a transverse fracture of the carbon-ply in a laminated hybrid, showing the associated delaminations.

Longitudinal section through a fracture similar to Fig. 25. The arrows indicate the delaminations which extend further out of the field of view.
Fig. 27  Scanning electron micrograph of a glass-ply after fracture and delamination of the carbon-ply. Short lengths of carbon fibre remain bonded tracing the fracture path.

Fig. 28  Section of a carbon-ply failure in the plane of lamination.
Fig. 29  Multiple-cracking (arrowed) and accompanying delaminations (lighter areas) in laminated hybrids.

Fig. 30  (On next page)

Load/strain traces for three-ply laminated hybrids. The letters denote successive fractures of the carbon-ply as the specimens underwent multiple-cracking.

The lay-up sequences are:

(i) 6 glass, 7 HMS-carbon, 6 glass pre-pregs
(ii) 8 " 3 " 8 " "
(iii) 9 " 1 " 9 " "
(iv) 9 " 1 HTS-carbon 9 " "

150
Fig. 31
Model for the separation of the plies at a delaminated interface.

Fig. 32 Model for the loads in the carbon and glass-ply plies to one side of a carbon-PLY fracture.
\[ \beta (\times 10^{-3}) \cdot \frac{1}{\sqrt{Pd}} \]

**Fig. 33** Variation of the length of delamination with the stiffness parameter \( \beta \) for hybrids with a single carbon fibre lamina.

\[ \beta (\times 10^{-3}) \cdot \frac{1}{\sqrt{Pd}} \]

**Fig. 34** Variation of the first failure strain of the carbon-ply in hybrids containing a single lamina of carbon fibre.
Fig. 35 The principles of the laser moire technique for the measurement of surface strains. The reflected beams have been separated for clarity.
Fig. 36 Laser moiré fringes for specimen A at failure of the carbon-ply, and two lower loads.
Fig. 37 Laser moire fringes for specimen 3 at failure of the carbon-ply and two lower loads.
Fig. 36 Laser moire fringes for specimen C at failure of the carbon-ply, and two lower loads.
Fig. 39 Strain traces for specimen A from the moire fringes in Fig. 36.

Fig. 40 Strain traces for specimen B from the moire fringes in Fig. 37.
Fig. 41  Strain traces for specimen C from the moire fringes in Fig. 38.

Fig. 42  Laser moire specimen containing a half-length carbon-ply between two glass-plies.
Laser moire fringes for Series Two specimens with half-length carbon-plies. In all cases the carbon-ply extends to the right from its end left of centre, and appears darker. The tensile axis is horizontal. Delamination, which initiates at the end of the carbon-ply appears lighter in the photographs without fringes.

Fig. 43  Moire fringes for specimen D.
(3 glass-1 carbon-3 glass laminae)

<table>
<thead>
<tr>
<th>Frame Load (KN)</th>
<th>Comment</th>
</tr>
</thead>
<tbody>
<tr>
<td>(a) 6.6</td>
<td>Delamination initiates.</td>
</tr>
<tr>
<td>(b) 6.6</td>
<td></td>
</tr>
<tr>
<td>(c) 8.0</td>
<td>Controlled growth of delamination.</td>
</tr>
<tr>
<td>(d) 8.0</td>
<td></td>
</tr>
<tr>
<td>(e) 8.4</td>
<td></td>
</tr>
<tr>
<td>(f) 8.4</td>
<td></td>
</tr>
<tr>
<td>(g) 9.0</td>
<td></td>
</tr>
<tr>
<td>(h) 9.0</td>
<td></td>
</tr>
<tr>
<td>(i) 9.3</td>
<td>Unstable propagation of delamination.</td>
</tr>
<tr>
<td>(j) 9.3</td>
<td></td>
</tr>
<tr>
<td>(k) 9.3</td>
<td></td>
</tr>
<tr>
<td>(l) 9.3</td>
<td></td>
</tr>
<tr>
<td>(m) 0.3</td>
<td>Unloaded.</td>
</tr>
</tbody>
</table>

160
Fig. 44 Moire fringes for specimen E.
(3 glass-1 carbon-3 glass laminae)

<table>
<thead>
<tr>
<th>Frame Load (KN)</th>
<th>Comment</th>
</tr>
</thead>
<tbody>
<tr>
<td>(a) 3.9</td>
<td>No damage</td>
</tr>
<tr>
<td>(b) 8.0</td>
<td>&quot;</td>
</tr>
<tr>
<td>(c) 9.3</td>
<td>Controlled growth of delamination.</td>
</tr>
<tr>
<td>(d) 9.3</td>
<td>&quot;</td>
</tr>
<tr>
<td>(e) 9.3</td>
<td>&quot;</td>
</tr>
<tr>
<td>(f) 3.9</td>
<td>Unloading</td>
</tr>
<tr>
<td>(g) 1.2</td>
<td>&quot;</td>
</tr>
</tbody>
</table>
Fig. 45  Moire fringes for specimen G.
(3 glass-2 carbon-3 glass laminae)

<table>
<thead>
<tr>
<th>Frame Load (KN)</th>
<th>Comment</th>
</tr>
</thead>
<tbody>
<tr>
<td>(a) 8.2</td>
<td>No damage</td>
</tr>
<tr>
<td>(b) 9.0</td>
<td>Initiation of delamination.</td>
</tr>
<tr>
<td>(c) 9.0</td>
<td>&quot;</td>
</tr>
<tr>
<td>(d) 2.0</td>
<td>Unloaded.</td>
</tr>
</tbody>
</table>
Fig. 46  Moire fringes for specimen H.
(6 glass-2 carbon-6 glass laminae)

<table>
<thead>
<tr>
<th>Frame Load (KN)</th>
<th>Comment</th>
</tr>
</thead>
<tbody>
<tr>
<td>(a) 5.4</td>
<td>No damage</td>
</tr>
<tr>
<td>(b) 5.4</td>
<td>&quot;</td>
</tr>
<tr>
<td>(c) 17.0</td>
<td>Controlled growth of delamination.</td>
</tr>
<tr>
<td>(d) 17.0</td>
<td>&quot;</td>
</tr>
<tr>
<td>(e) 17.0</td>
<td>Unloading</td>
</tr>
<tr>
<td>(f) 8.2</td>
<td>&quot;</td>
</tr>
<tr>
<td>(g) 3.8</td>
<td>&quot;</td>
</tr>
</tbody>
</table>
Fig. 47 Essential features of a failure of a carbon fibre bundle in a divided-tow hybrid.

Fig. 48 Multiple fractures of the carbon fibre bundles in divided-tow hybrids containing:

(a) 10,000 fibres
(b) 3,620 fibres
(c) 1,620 fibres
(d) 630 fibres

(Micrographs on next page)
Fig. 49  
Carbon fibre fractures in a spread-tow hybrid strained to 0.02. The lighter areas are debonding at the glass/carbon fibre interface.

---

Fig. 50  
Variation of the characteristic debond length in divided-tow hybrids.
Fig. 51 Density of carbon fibre bundle fractures vs. strain for the divided-tow hybrids.

Fig. 52 Curves from Fig. 51 corrected to allow for the debonded zone at each fracture.
Fig. 53 Curves from Fig. 51 corrected to allow for the length of carbon fibre bundle carrying reduced load at each macroscopic fracture.

Fig. 54 Diffraction method of measuring the diameter of carbon fibres.
Fig. 55  Cumulative distribution for failure of single HTS-carbon fibres, with fitted Weibull curves.

Fig. 56  Failure stress vs. nominal strain for HTS-carbon fibres tested individually.
Fig. 57  Mean failure stress of single HTS-carbon fibres as a function of gauge length, (logarithmic axes)

Fig. 58  (On next page)

Acoustic emission and load/strain curves for the wet lay-up composites.

(a) all-carbon fibre
(b) all-glass fibre
(c) 10,000 carbon fibre bundles
(d) 3,620 " " "
(e) 1,620 " " "
(f) 630 " " "
(g) spread-tow, 1 mm glass-plies
(h) " " 0.5 mm " "
Load 5,000 AECounts \( \frac{s^{-1}}{20 \text{ KN}} \)

Acoustic emission rate

(a) 

(b) 

(c) 

(d) 

(e) 

(f) 

(g) 

(h) 

STRAIN

0

0,02

\( x \times 10 \) scale change.

\( x \times 10 \) scale change.

176
Fig. 59
(a) Correlation between total acoustic emission and number of matrix cracks.

(b) The same data presented against strain.

Fig. 60
Correlation between acoustic emission and the number of carbon fibres in divided-tow hybrids.
Fig. 61  Longitudinal section of a matrix crack.  
( Code 69 epoxy resin )

Fig. 62  Disc or annular matrix crack around a carbon fibre. (a) focus at surface, (b) below surface.
Fig. 63  Matrix cracking in all-glass fibre wet lay-up composites.

Fig. 64  Matrix cracking in spread-tow hybrids.
Fig. 65 Matrix cracking within all-glass fibre composite
(a) 0.0 (b) 0.02 (c) 0.022 (d) 0.031 strain.

100 μm.
Fig. 66  Surface resin cracks in all-glass fibre composites
(a) Unstrained, (b) strained to 0.023

100 \mu m

Fig. 67  Large matrix cracks in the carbon fibre layer
of a spread-tow hybrid strained to 0.02

(a) & (b) separate cracks
(c) & (d) cracks in lines
(b) & (d) are at higher magnification.

(On next page)
Fig. 68 Matrix cracks similar to Fig. 67 in polarised light showing strain fields as lighter areas.

Fig. 69 Glass fibres bridging the delaminated interface between carbon and glass-plies.
Fig. 70  Section of a carbon fibre layer in a spread-tow hybrid strained to 0.02

The carbon layer of a spread-tow hybrid strained to 0.02, taken in polarised reflected light to show up the debonding between fibre and matrix as lighter areas.

(a) Large group of carbon fibre fractures.
(b) Intermediate "
(c) Isolated carbon fibre fractures.
(d) A pair of adjacent fibre fractures, only one of which is debonded.
Fig. 72 The method of electrolytically etching sections of composite to decorate conducting fibres.

Fig. 73 Method of determining the depth to which the electrolytic decoration can detect non-conducting sections of fibres.
Fig. 74 Sections of carbon fibre composite etched as shown in figure 73. Dark fibres are conducting.

Fig. 75 Electrolytically decorated sections of carbon fibre composite, (a) unstrained with no fibre fractures, (b) strained to failure, 0.013, showing a single non-conducting (bright) fibre end.
Fig. 76  Groups of carbon fibre fractures in a laminated hybrid, (a) focus at surface, (b) focus below surface in polarised light to show matrix cracks.

Fig. 77  Etched sections of spread-tow hybrids strained to; (a) 0.009  (b) 0.011  (contd. over...)

188
(c) 0.014 strain

(d) 0.017 strain

(e) 0.020 strain

(f) 0.023 strain
Fig. 78 Model for the calculation of thermal strains.

(a) Contractions in each ply.
(b) Proportion of strain mismatch appearing as compression in the carbon-ply.
Fig. 79 Thermal expansions for composites fabricated by wet lay-up.

Fig. 80 The principle of 'constraint' in a carbon-ply sandwiched between two glass-plies.
Fig. 81 Weibull probability graph, similar to Fig. 12, showing the variation of the strength of the carbon fibre in composites and single fibre tests, with the number (mn) of hypothetical 'elements' in a composite, or single fibre.

The curves predicting the strength of composites are taken from Harlow & Phoenix, (ref. 39'), and are based on a chain-of-bundles statistical model.
Fig. 82  A section of a composite containing a defective area, (shaded).

Fig. 83  Variation of the failure strain of all-HTS carbon fibre laminated composites with the reciprocal of the thickness.
Fig. 84 Effective longitudinal modulus of HTS-carbon fibre reinforced epoxy as a function of the fibre orientation, calculated by Bishop (67).
Carbon fibre structure by electrolytic etching

Studies of polycrylonitrile (PAN)-based carbon fibres have demonstrated a concentric three-zone skin–sheath–core structure originating in the oxidation of the precursor fibre before pyrolysis heat-treatment. I describe here an electrolytic oxidation treatment that is effective in showing up different zones in carbon fibre sections.

PAN fibres, 1.5 denier were oxidised in air at 230 °C for either 30 min or 2 h, and heat-treated at 1,000 °C, 1,400 °C or 2,500 °C giving in total six fibre types. Single tows of each fibre type were mounted in epoxy resin blocks and 3-mm thick disks were cut normal to the fibre axis. The disks were polished on one side and sputter-coated with gold on the other to make electrical connection to all fibre ends. For electrolysis the polished face was made anode in a 2 M sulphuric acid bath with carbon cathode. Longitudinal fibre sections were also prepared in a similar manner. In these, conduction through the disk is by fibre-to-fibre contact, and it was necessary to use a high-volume fraction composite of commercial type 2 PAN-based fibre rather than single tows in order to achieve sufficient conductivity. The electrolytic treatment is basically anodic oxidation. Average current density at the fibre ends was limited to 140 A m⁻² for convenient treatment times, though neither acid concentration nor current density were found to be critical. Specimens were removed after intervals of 10–60 s, and rinsed in distilled water before microscopy.

In the initial stages of the treatment, fibre surfaces developed colours which later dulled as oxidative removal of carbon formed relief. The colours are believed to arise from interference in a thin surface layer with different optical properties, formed by intercalation of sulphuric acid into the fibre. As the optical path difference inferred from the interference colour increases linearly at a rate of 0.18 nm per C m⁻², the advancement of any particular area in the interference series is an indicator of its extent of reaction in forming an intercalation compound. The colours extend only to second order blue, which is maintained after relief is developed. Specimens were examined soon after etching as the colours fade over a period of several weeks at room temperature.

For all sections cut normal to the fibre axis the sheath etched more rapidly than the core, and increasingly so for higher heat-treatment temperatures (HTT). In every case etching in the sheath decreases outwards, but there are differences in this gradation between HTT's. A dark ring of very etched material, approximately 1 µm wide, immediately surrounds the core in 2,500 °C HTT fibres, while the remainder of the sheath is uniformly etched at a lower rate. Fibres heat-treated at 1,000 °C and 1,400 °C show a more even outward decrease in etching of the sheath which is shown clearly by the (coloured) concentric rings in Fig. 2b, and the smooth contours in Fig. 1d, e, i, j. The core etched uniformly for all heat treatments and is seen as a plateau in fibres showing relief. In longitudinal sections both sheath and core etched uniformly, but the core more rapidly than the sheath, a reversal of the order in transverse sections (Fig. 2a).

Watt and Johnson have observed oxidation zones in thin sections of the PAN fibres from which the carbon fibres examined here were made. The thickness of the darker oxidised zone corresponding to the sheath increases with the square root of oxidation time (Fig. 1a, f) and after heat treatment the structure is apparent in polarised light (Fig. 1g). Direct comparison of the zone dimensions in oxidised and carbonised fibres is not possible because of the shrinkage from 13 to 8 µm diameter, but the relative sheath thicknesses correspond. In polarised light a Maltese cross is seen in both sheath and skin, whereas the core is optically isotropic. The contrast is lower in 1,000 °C and 1,400 °C HTT fibres but the pattern remains the same. These patterns have been interpreted in terms of the c-plane orientation about the fibre axis, which for the 8 µm PAN-based fibres is thought to be concentric in the skin, radial in the sheath and random in the core. This model agrees well with the structure seen on fracture surfaces and in plasma-oxidised fibre sections, where radial striations may be seen in the sheath, and circumferential rings in the skin. Orientation is less apparent in these surfaces below 2,000 °C HTT, and it is noted that only in Fig. 1h (2,500 °C HTT) are there dark radial streaks in the sheath.

It is becoming evident that the sheath/core distinction is not simply one of c-plane orientation. Electron microscopy of crushed fibre fragments, and electron diffraction studies of taper-thinned fibres show decreasing layer–plane misorientation and
decreasing crystallite stacking height from the fibre centre outwards. A uniform outward decrease is associated with the lower heat-treatment temperatures, but for 2,500 °C HTT fibre the transition takes place in a narrow 1-μm wide band; patterns which are mirrored in the etched fibre surfaces (Fig. 1 c–e, h–j).

The thickness of the intercalated layer might be expected to vary with the inclination of the c planes to the surface, rising to a maximum when they are normal. If there are preferred circumferential and radial crystallite orientations any non-diametral longitudinal section will show some variation of the mean c plane to surface angle within zones, which might be reflected in variation of the etching rate across the zones. In fact, the zones etch uniformly (Fig. 2a), showing that the technique is insensitive to such orientation as may exist. Differences in crystallite misorientation in sheath and core will give rise to differences in plane edge density in transverse sections, but these do not seem able to account for the contrast in etch rates.

In longitudinal sections intercalation will initially take place in crystallites presenting c-plane edges to the surface, and the depth of penetration will be limited by the crystallite width. A higher etching rate will then be associated with wider crystallites. The core is known to have greater crystallite thickness than the sheath, and it is possible that the effective crystallite width is also greater, which would account for its higher rate of etching.

Because of the high axial c-plane alignment all crystallites in transverse sections will be suitably oriented for intercalation. But, the rate of intercalation is expected to be greater in the less misoriented sheath where there is greater separation of lattice distortions hindering penetration. It is probable also that axial curvature and folding of layer planes produces channels in which intercalation is facilitated, and a higher density of these may accompany the smaller sheath crystallite thickness.

These ideas come some way to explaining the reversal of sheath and core etch rates in longitudinal and transverse sections. Although the etch patterns can only be correlated with what is already known of fibre fine structure, the technique is worth pursuing as a simple and sensitive monitor of morphological variations arising during the processing of PAN fibre to carbon fibre. As such it is an aid in relating strength properties to structure, and for this application a particular advantage of current controlled etching is the fact that insulating resin matrices are not attacked.

I thank W. Watt and W. Johnson of the Royal Aircraft Establishment, Farnborough for useful discussions and for providing carbon fibres and the photographs of PAN fibres.

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APPENDIX TWO

FAILURE STRAIN ENHANCEMENT

IN CARBON/GLASS FIBRE HYBRID COMPOSITES

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1. INTRODUCTION

Hybrid fibre-composites, consisting of two, or more, fibre types embedded in a common resin matrix have recently attracted considerable attention, see Phillips [1], since they offer the possibility of achieving a balance of mechanical properties that cannot be realised in composites of one fibre type. Amongst the possibilities claimed for hybrid composites are the more cost-effective utilisation of expensive reinforcing fibres such as carbon, and a modification of the mode of failure from the often sudden and catastrophic sequence observed in many all-carbon composites, to a more gradual and controlled mode where one fibre continues to carry stress after failure has initiated in the other.

Typically a high modulus fibre with a low strain to failure is combined with one of lower modulus but higher failure strain, e.g. carbon with glass. Bunsell and Harris [2] tested three layer glass-carbon hybrids and reported that the failure strain for high-modulus carbon was increased by 80% in the glass-carbon sandwich compared with that for an all-carbon-fibre composite. However, their all-carbon composite showed an unusually low strain at failure of only 0.3%. They were able to account for some 10% of the observed increase in the hybrid by the residual compressive stress induced in the carbon layer, due to the thermal expansion mismatch between the glass and carbon-fibres, when the material was cooled down from its moulding temperature. Aveston and Sillwood [3] made hybrids by incorporating separated carbon-fibre tows with glass so that the individual carbon-fibres were highly dispersed with respect to the glass. They found up to 100% increase in the carbon-fibre failure strain. Their results are, however, hardly comparable with those of Bunsell and Harris since the latter used a thick (0.4mm) carbon ply thickness. Zweben [4], dispersed tows of carbon-fibre (Thornel 300) with polyaramid (Kevlar® 49) but found a failure strain enhancement of only 4% in the carbon.

In the present work we have attempted to explore the effects of the state of dispersion and the fibre ratio on the failure strain enhancement, and have studied the modes of failure observed in simple sandwich laminate hybrid composites.

Published in:

NOMENCLATURE

Symbols

CT  High tensile carbon fibre composite
CM  High modulus carbon fibre composite
EG  "E" - glass fibre composite
P  Proportion of hybrid thickness represented by each reinforcing component (subscripted)
t  Thickness of a hybrid component (subscripted)
D  Dispersion
c  Strain of hybrid or hybrid component (subscripted)
T  Temperature
a  Composite thermal expansion coefficient (subscripted)
S  Stiffness of hybrid or hybrid component (subscripted)
YP  Debonding fracture energy
YX  Transverse fracture energy
τ  Frictional force per unit area at debonded interface
x  Distance (along specimen axis) from carbon-ply failure
β  
\[ \beta = \left( \frac{1}{\frac{1}{S} + \frac{1}{S_C}} \right)^{-1} \]
Lb  Debond length to one side of carbon-ply fracture
W  Energy term

Subscripts

c  Carbon-ply
g  Glass-ply
h  Hybrid
u  Ultimate tensile
r  Smallest representative repeat (thickness)
t  Thermal

2. EXPERIMENTAL

2.1 Materials

All the composites were fabricated from unidirectional fibre "pre-preg" sheets. The fibre types were surface treated high-modulus carbon (CM), surface treated high-strength carbon (CT), and E-glass (EG). The impregnating resin was a proprietary epoxide system (Fothergill and Harvey Ltd., Code 69). The pre-preg sheets were designed to give a lamina thickness of 0.125mm at 0.6 volume fraction. The nominal mechanical properties of the fibres as derived from composite tests are given in Table I.

<table>
<thead>
<tr>
<th>Fibre Type</th>
<th>Tensile Modulus GPa</th>
<th>Tensile Strength GPa</th>
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<td>EG</td>
<td>72</td>
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2.2 Fabrication

Laminates were prepared by laying up the appropriate sequence of
100 x 200mm sheets of pre-preg (fibres parallel to 200mm dimension), in an open-ended trough mould, and pressing between the heated platens of a hand-operated screw press. The pre-preg pack was pre-heated in the press under minimal pressure for 10 minutes at 130°C to melt the resin. It was then removed to a vacuum oven and degassed for 15 minutes at 130°C when it was returned to the press. The platen temperature was then raised to 175°C over a period of 10 minutes whilst the press was gently screwed down onto metal stops which established the final laminate thickness. In this way the excess resin and any remaining voids were expelled through the open ends of the trough mould and good quality laminates produced. The cure was continued at 175°C for 60 minutes after which the mould was removed from the press, the laminate extracted and finally post-cured for a further 3 hours at 175°C.

Parallel sided tensile test pieces 200mm x 10mm wide were cut from the laminates with a diamond saw and aluminium alloy and tabs bonded on with an epoxy resin adhesive. Strain gauges were positioned on the central portion of the gauge length as shown in Figure 1.  

2.3 Testing

At least five test pieces of each lay-up were tested in an Instron testing machine at a strain rate of 1.5 $10^{-6}$ s$^{-1}$. The tests were monitored with an acoustic emission detector which measured count rate. Both load-strain and acoustic rate were autographically recorded, a typical trace being shown in Figure 2. After fracture the general failure pattern was noted and specimens were taken from typical regions for optical and scanning electron microscopy.

Fig. 1. Construction of standard specimen.

Fig. 2. Load-strain, and acoustic emission rate-strain traces for a hybrid showing successive fractures of the carbon-ply at a, b, and c.
2.4 Range of Composites Tested

The aim of the work was to investigate the effect of hybrid ratio and the state of dispersion of the two fibre types. This has been done within the restrictions imposed by the choice of a pre-preg system which allows only for a through thickness variation in dispersion of the laminate, and variations in quanta of one pre-preg layer.

Most of the work was done on glass-carbon-glass sandwich laminates using both CM and CT carbon, but some multiple-layer laminates were also prepared. Details of laminate geometry together with basic mechanical test data are given in Table II.

When comparing the properties of hybrid composites two fundamental constitutional parameters should be considered, the hybrid ratio and the state of dispersion.

The hybrid ratio is expressed here as the proportions of the total laminate thickness represented by each reinforcing component. (e.g. for a 1:1:1 glass-carbon-glass hybrid the carbon fraction, $P_C = 0.33$, and the glass fraction, $P_g = 0.67$.) We have defined dispersion as the reciprocal of the thickness (in metres) of the smallest representative repeat unit of the laminate:

$$D = \frac{1}{t_r}$$

In the case of the simple sandwich laminates $t_r$ is the total laminate thickness.

Figure 3 is a two-dimensional representation of the range of laminates studied. The left-hand vertical axes show dispersion, and the number of pre-preg laminae in the thickness respectively, whilst the horizontal axis is the carbon fraction, $P_C$. The two sets of hyperbolae illustrate the restrictions that all laminates be constructed from integral numbers of laminae, and that for a balanced three-layer construction there must be an even number of glass laminae. Laminate geometries satisfying these conditions are found at the intersections of the hyperbolae, and those fabricated are indicated. The open circles represent multiple-layer laminates.

3. MECHANICAL PROPERTIES

Mechanical testing has shown that laminate geometry has a profound effect on the properties of the hybrids. The most notable is an increase in carbon layer failure strain of up to 35% (CT) or 45% (CM), and this is termed the "hybrid effect". Further important findings are a size effect in all-carbon composites, and the observation of a multiple fracture failure mode, in which progressive failure occurs in the carbon component of the hybrid.

3.1 Observations Preceding Carbon-Ply Failure

The first portion of the load-strain curve is virtually linear and the observed elastic modulus is in close agreement with that predicted by the parallel rule of mixtures. Although no external changes are noted in this region, the acoustic emission rate count becomes significant at strains of about 0.005 and then increases progressively up to the first carbon ply failure, Figure 2. Microstructural examination of specimens loaded to just below the first carbon-ply
Fig. 3. The range of dispersion and carbon fraction covered by composites based on (a) high tensile carbon, (b) high modulus carbon. Open circles represent multiple-layer laminates. The hyperbolae indicate the number of laminae in the component plies.
Fig. 4. Longitudinal section of a matrix crack limited in extent by glass fibres. Scale bar 10μm

Fig. 5. Matrix crack surrounding a carbon fibre at a resin-rich glass/carbon-ply interface. (a) Surface reflection (b) Focus below surface showing crack arrest near adjacent fibres. Scale bar 10μm
failure strain typically shows few fibre breaks, but extensive matrix cracking. The matrix cracks are normal to the fibre axes (this was also the tension axis), and are clearly seen to have been arrested at the fibres, Figures 4 and 5. These matrix cracks appear to be more common in resin rich regions.

In an attempt to identify broken fibres, we have developed an electrolytic etching technique [11], which is able to distinguish between conducting and non-conducting fibres running through a thin (~0.1mm) transverse section of the laminate. There is a colour change on conducting fibres, whereas unetched (broken) fibres remain bright, and are seen to be grouped, and frequently associated with matrix cracking, which appears bright under crossed polars (Figure 6). Up to fracture the number of isolated fibre breaks is small, about $10^{10}$ m$^{-3}$, and they have not been observed in longitudinal sections. It is probable that both fibre fractures and matrix cracks contribute to the build-up in acoustic emission observed, but it has not been possible to attribute their relative contributions.

3.2 Carbon-Ply Failure

On loading the hybrid laminate the near-linear load strain characteristic is maintained until a major failure occurs in the carbon-ply (Figure 2). At this point there is a transverse fracture across the carbon-ply with consequent debonding along the carbon-glass interface. The load-strain trace shows a load drop at this point but on further extension, the load rises, though at reduced slope. The exact shape of the curve is determined by the machine stiffness and by the position of the strain gauge relative to the position of the carbon-ply fracture. A high and irregular rate of acoustic emission is observed during the sequence of carbon ply failure and associated events.

![Fig. 6. Etched cross section of failed hybrid showing: (a) Grouping of broken carbon fibres (bright unetched small fibres) (b) Association of matrix cracks (bright areas) with carbon fibre breaks in the same field. (Polarized light). Scale bar 100µm](image)
Fig. 7. The hybrid and size effects plotted against carbon fraction and dispersion (logarithmic axis) for laminates based on (a) high tensile carbon, (b) high modulus carbon. The contribution of thermal strains to the increased failure strain is indicated by solid bar.
<table>
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<th>Lay-up sequence</th>
<th>Dispersion ( \times 10^3 )</th>
<th>Carbon fraction ( F_c )</th>
<th>Modulus ( (\text{G Pa}) )</th>
<th>Failure stress ( (\text{G Pa}) )</th>
<th>Failure strain</th>
<th>Lebond length ( L_b ) (mm)</th>
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<td>119</td>
<td>0.98</td>
<td>0.0073</td>
<td>-</td>
<td>+4</td>
</tr>
</tbody>
</table>

**TABLE II**

**LAMINATE PROPERTIES**
The strain at which the first carbon-ply failure occurred was found to be dependent on laminate geometry. The details are given in Table II, and in Figure 7. The hybrid effect (relative to a standard 1.5mm thick all-carbon laminate) is shown in a 3-dimensional representation against carbon-fraction and dispersion. It will be seen that the effect increases strongly as the carbon fraction is decreased and also increases as the dispersion is increased. Note also the reduced failure strain for the thinner all-carbon laminates. Part of the hybrid effect may be explained by the thermal stresses induced in the laminate (section 4.1) and this is also indicated on the diagrams. The effect is more pronounced in the case of the EG-CM hybrids.

Figure 8 shows the general characteristics of the region around the carbon-ply fracture. The main transverse fracture is typically of cruciform shape and debonding occurs between the carbon and glass plies outwards from the line where the transverse fracture intersects the glass-carbon interface. Note that the region between the two arms of the fracture is not debonded. This is clear from the photomicrographs of Figures 9 and 10. Figure 11 is a section cut in the plane of lamination and shows the irregular branched form of the fracture. We interpret this evidence to indicate that the fracture was initiated within the carbon-ply and that the cruciform cracks then grew out across the ply to the glass-carbon interface, which then debonded under the influence of the induced shear stresses. There is also a suggestion, Figure 11, that the fracture path propagated in one direction across the width of the test piece as evidenced by the direction of the branching.
3.3 Multiple Cracking

The measured range of debond lengths is rather large, from several millimetres to greater than 100mm (i.e., greater than the gauge length). In this latter case continued loading of the completely debonded specimen after the carbon ply failure results only in failure of the remaining glass plies at their ultimate strength, and the carbon has no further influence on the load carrying capacity of the laminate. When the debond length is less than the gauge length, further straining after the initial failure causes further fractures at random positions in the remaining bonded portions of the specimen. Repeated carbon ply failure at approximately constant load is shown in Figure 2, and is seen to be accompanied by high acoustic output. (In a multiply-cracked specimen the strain is not uniform along the gauge length, which accounts for the mismatch in the traces in Figure 2, as the A.E. trace was related to cross head displacement and the load trace to strain gauge reading.) Ultimately most of the hybrid becomes debonded, and the strength and stiffness approach that of the glass plies alone. As noted previously, there is a load drop associated with each carbon-ply fracture. On further straining the debonded zone has been observed to propagate, ultimately reaching a value which appears to be a characteristic of the particular hybrid geometry.

The practical significance of these observations is that the initial fracture in the carbon-ply does not propagate across the glass plies but is contained by the debonding at the carbon-glass interface. This debonded zone is also restricted and this implies that load can be progressively diffused back into carbon-ply away from the fracture so that the carbon-ply, away from the fracture zone, remains capable of bearing load. This will have a significant effect on the overall work of fracture, although this aspect has not been investigated in the present work.
4. DISCUSSION

4.1 Thermal Contribution to Hybrid Effect

Carbon fibre has a small negative longitudinal thermal expansion coefficient, whereas glass fibre has a much larger positive coefficient, so on cooling from the resin-cure temperature the carbon component of a hybrid is placed in compression, and the glass in tension. The hybrid strain at carbon failure is expected to increase by an amount equal to the compressive strain in the carbon.

Figure 12 shows the thermally induced loads and strains in a hybrid after cooling (away from the regions of end effects) and is the model for thermal strain calculation. The thermal strain mismatch \( \Delta e_t \), between carbon and glass plies, after cooling through a temperature difference \( \Delta T \) is \( \Delta e_t = (a_c - a_g) \Delta T \), where \( a_c \) and \( a_g \) are longitudinal expansion coefficients of the carbon and glass components. The fraction of this mismatch which appears as a strain in the carbon ply is \( S_g \), where \( S_g \) and \( S_h \) are the glass component and hybrid stiffnesses respectively. The carbon-ply compression is then \( \varepsilon_{ct} = \frac{S_g \Delta e_t}{S_h} \).

The fraction \( S_g = \varepsilon_{ct} \) is a function only of material constants and the hybrid ratio, and is independent of dispersion as seen in Figure 13 where it is plotted against \( P_c \). The strain mismatch has also been measured experimentally by observation of the thermally induced bending of unbalanced laminates. This is \( 1.0 \times 10^{-3} \) for the CT - EG system, and \( 0.93 \times 10^{-3} \) for the CM - EG system.
The proportion of the strain increase attributable to thermal contraction, shown in Figure 7, is only a small fraction of the total, and the thermal effect is therefore not a complete explanation of the observed phenomena.

4.2 Failure Theories

Failure in unidirectional composites may be considered in terms of either the statistical failure of fibres leading to some critical weakening of the composite [5-8], or by the application of linear elastic fracture mechanics, [9,10], assuming the material to be macroscopically homogeneous. The fracture mechanics approach is best suited to composites containing macroscopic flaws of known size, and much larger than the fibres, while the statistical approach is applied to materials containing no flaws beyond those intrinsic to the constituents. Applying either alone to hybrids presents a problem, in that they initially contain no macroscopic flaws, and would therefore require a statistical treatment of crack growth. However, the energy made available for fracture depends on the debond length (which itself varies widely with hybrid geometry) so that the macroscopic energetics of failure must also be considered.

Ideally we would wish to unify both treatments and consider the energetics of crack propagation as flaws develop during straining. In practice our approach has been determined by the lack of theoretical or experimental flaw size data so that we have been unable to adopt a crack propagation criterion for failure. Instead we consider the energy balance as a function of strain, before and after a failure in the observed mode, and arrive at a minimum strain criterion for such a failure. The observed failure strain is expected to be higher than this minimum value, because not only must the net energy balance be favourable, but also at all stages of failure, from the very first fibre break through to the ultimate propagation of a macroscopic crack, the energy conditions must be favourable. We are not at present able to discuss a full failure model incorporating the statistical aspects of flaw development, but the simpler "before and after" treatment adopted indicates that the glass plies restrict the relaxation of the carbon ply, which necessitates higher hybrid strains before sufficient energy is available for fracture. This behaviour we have termed constraint.

Consider a hybrid laminate containing a failure of the form depicted in Figure 8. At the transverse fracture the load in the carbon-ply is zero, but to either side of this position there is transfer of load by friction from the glass to carbon across the debonded interface. This results in a progressive build up of stress in the carbon-ply away from the fracture. Detailed consideration of the through-thickness displacements in the debond zone indicates a small separation at the interface, but this anticipated separation is much smaller than the observed roughness and does not counteract the frictional force, which we have assumed to be constant. The linear increase in carbon-ply load, and concomitant decrease in glass-ply load combine to a constant loading on any hybrid cross-section, Figure 14. The shear stress supported by friction is considered to be low compared with that across a bonded interface and this is shown in Figure 14 as a sharp step. The elastic load transfer rate in this bonded zone approaches to the resin shear strength, and this is shown to be much higher than the frictional stress.
We now consider the energy change for an infinitesimal increase in the debond length. The carbon-ply load at any point, x distant from the transverse fracture, is $2\tau x$, where $\tau$ is the frictional force per unit area acting on each side. The sum of carbon and glass loads on any section of the debonded region equals the total hybrid load

$$2\tau x + \varepsilon_S S_g = \varepsilon_h S_h$$

(where stiffness is defined as the product of the modulus and thickness of the respective components, $S = E l$). From which:

$$\varepsilon_g = \varepsilon_h S_h - \frac{2\tau x}{S_g} \quad \frac{S_h}{S_g} \quad \ldots \ldots (1)$$

The stiffnesses of all three plies act in parallel.

$$S_h = S_g + S_c \quad \ldots \ldots (2)$$

Because the debond extension is infinitesimal there is no load drop. The change in the carbon-ply strain energy over length $dx$ is,

$$\frac{1}{2} \varepsilon_h^2 S_c \ dx - \frac{1}{2} \varepsilon_c^2 S_c \ dx \quad \ldots \ldots (a)$$

The change in glass-ply strain energy over length $dx$ is,

$$\frac{1}{2} \varepsilon_h^2 S_g \ dx - \frac{1}{2} \varepsilon_g^2 S_g \ dx \quad \ldots \ldots (b)$$
The work of debonding is,

\[- 2\gamma_D \, dx\]  \hspace{1cm} (c)

where \(\gamma_D\) is the energy to debond unit area of interface.

As the debond extends \(dx\) there is a relative displacement of glass and carbon plies of \((\epsilon_g - \epsilon_c)\) \(dx\) against the total frictional force transferred to the carbon-ply, \(2\tau\). The work of friction is,

\[- 2\tau (\epsilon_g - \epsilon_c) \, dx\]  \hspace{1cm} (d)

For the debond extension \(dx\) the hybrid extends \((\epsilon_g - \epsilon_h)\) \(dx\) and the load does work,

\[(\epsilon_g - \epsilon_h) \epsilon_h S_h \, dx\]  \hspace{1cm} (e)

For static equilibrium the total energy change is zero. Therefore,

\[(a) + (b) + (c) + (d) + (e) = 0\]

and substituting (1) and (2) for \(\epsilon_g\) and \(S_h\) throughout yields,

\[\frac{2\tau x^2}{\beta^2} = \frac{2 \tau \epsilon_h S_c x}{\epsilon^2} + \frac{\epsilon_h^2 S_c^2}{2\beta^2} - 2\gamma_D = 0\]

where \(\beta = \left(\frac{1}{S_c} + \frac{1}{S_b}\right)^{-1}\)

The appropriate solution for the stable debond length \(L_D\) (Figure 8) is given by:

\[L_D = \frac{\left(\epsilon_h S_c - 2\beta \gamma_D^4\right)^{1/2}}{2\tau}\]  \hspace{1cm} \ldots(3)

The physical significance of this equation can be seen from Figure 14.

The six measurable values of \(L_D\) provided by the CT - EG series of hybrids are fitted to expression (3) to obtain values for \(\tau\) and \(\gamma_D\). This was done by substitution of the \(L_D\) values in Table II together with the appropriate values of hybrid strain at the failure of the carbon-ply. These results are shown in Figures 15 and 16, and the values for the interface friction \(\tau\) and debond energy \(\gamma_D\) are 0.59 MPa and $5.0 \text{KJ/m}^2$ respectively. The value for \(\tau\) is comparatively low being only 1% of the resin shear strength (or interlaminar shear strength). This justifies the initial assumptions on frictional load transfer.

In order to estimate the strain at which the carbon-ply in the hybrid will fail we consider the energy balance before and after failure. For a small \(L_D\) the load drop at failure is negligible and the fracture is considered to proceed at constant load. As the load, and hence the strain energy of the bonded portion of hybrid, remain constant, only the volume of material contained within the debonded portion need be considered, Figure 8.
Fig. 15. Variation of the failure strain of a single carbon lamina hybrids with the stiffness parameter $\beta$.

Fig. 16. Variation of debond length with the stiffness parameter $\beta$ for single carbon lamina hybrids.
The energy terms associated with tensile deformations are,

- $W_1$: Carbon-ply strain energy before failure
- $W_2$: " " " " after "
- $W_3$: Glass-ply " " before "
- $W_4$: " " " " after "
- $W_5$: Frictional work at debonded interface
- $W_6$: Transverse fracture energy
- $W_7$: Debonding fracture energy
- $W_8$: Work done by loading system.

The critical condition is that the energy change at failure should be zero, i.e.

$$ (W_1 + W_3 + W_8) - (W_2 + W_4 + W_5 + W_6 + W_7) = 0 \quad \text{.....(4)} $$

These energy terms are evaluated below for a hybrid of unit width.

The carbon strain energy density is $\frac{1}{2} \varepsilon_c^2 S_c$.

$$ W_1 = \varepsilon_c^2 S_c \ell_D $$

$$ W_2 = 2\int_0^{\ell_D} \frac{1}{2} \varepsilon_c^2 S_c \, dx $$

The glass strain energy density is $\frac{1}{2} \varepsilon_g^2 S_g$.

$$ W_3 = \varepsilon_g^2 S_g \ell_D $$

$$ W_4 = 2\int_0^{\ell_D} \frac{1}{2} \varepsilon_g^2 S_g \, dx $$

The relative movement of plies against the carbon ply load of $2\tau x$ does frictional work,

$$ W_5 = 2\int_0^{\ell_D} 2\tau x (\varepsilon_g - \varepsilon_c) \, dx $$

The transverse fracture energy is $\gamma_T$, so,

$$ W_6 = \gamma_T \ell_c $$

$$ W_7 = 4 \gamma_D \ell_D $$

The hybrid load $\varepsilon_h (S_g + S_c)$ moves through a distance equal to the glass-ply extension at failure, doing work,

$$ W_8 = \varepsilon_h (S_g + S_c) 2 \left( \int_0^{\ell_D} \varepsilon_g \, dx - (\varepsilon_h \ell_D) \right) $$

Combining $W_1 - W_8$ in (4) while substituting $\varepsilon_c = \frac{2\tau x}{S_c}$ and (1) for $\varepsilon_g$ gives,

$$ \varepsilon_h \left( \frac{3}{6 \tau^2 b^2} S_c \right) - \varepsilon_h \left( \frac{2\gamma_D S_c}{\tau} \right) + \left( \frac{8}{3} \frac{\gamma_D^2}{\tau} \frac{b}{t} \right) - \gamma_T \ell_c = 0 \quad \text{.....(5)} $$
\( \gamma_f \) has been measured at 82KJm\(^{-2}\) using the Tottersall-Tappin test configuration \([9,10]\) and the higher positive root of (5) is plotted in Figure 17. The lower positive root of (5) corresponds to a negative debond length in (3) and has no physical meaning. Equation (5) predicts a failure-strain enhancement with increasing dispersion and decreasing carbon ratio and the experimental results for the 1-carbon lamina series show a similar trend, although the extent of the effect is much less than that predicted. The results for the thicker carbon-plies series show virtually no hybrid effect. These experimental results indicate that the constraint is only effective when the carbon-plies are very thin. The 3-carbon laminae hybrids all fail at virtually the same strain so we conclude that failure here is controlled by the intrinsic strength of the carbon. The 1-carbon lamina series shows a significant enhancement from a lower (the size effect) intrinsic failure strain of 0.0077, to 0.0155 in the hybrid.

The extent of the deviation between theory and experiment is in part explained by the crudeness of the model, which neglects shear displacements (these would be most significant for small debond lengths). Another source of error lies in the value used for \( \gamma_f \). This was measured using a slow bend test which allowed full fibre pull out whereas in the hybrids there was a fast fracture with limited pull out.

5. CONCLUSIONS

(i) The present work has confirmed the existence of a hybrid effect in glass-carbon sandwich laminates. The failure strain of the carbon (brittle) layer may be enhanced to a degree dependent on laminate geometry.

Fig. 17. Carbon-ply failure strain for a range of one-, and three-carbon lamina hybrids, as found experimentally, and as predicted by the constraint model of hybrid failure. Experimental curves show only 1-carbon lamina hybrids are subject to significant constraint.
(ii) The extent of the hybrid effect is greater when the proportion of carbon is low and the dispersion is high. The carbon-ply failure strain in the hybrid can be as much as twice that of an all-carbon ply of similar thickness.

(iii) The hybrid effect is only significant when the absolute thickness of the carbon-ply is small. Virtually no effect was observed when the carbon ply thickness exceeded three laminae (0.4mm).

(iv) There is clear evidence of a size effect in all-carbon fibre laminates. Thinner laminates fail at lower strains than thicker ones over the range explored (0.13 - 2.0mm).

(v) The failure mode observed in the hybrid composites investigated was transverse fracture of the carbon ply with limited debonding along the glass-carbon ply interfaces. The extent of debonding has been shown to decrease as the dispersion is increased and the carbon proportion reduced. The analysis suggests that debonding could be inhibited if the carbon-ply could be made sufficiently thin.

(vi) Thermal stresses introduced during fabrication account for only a minor part of the observed failure strain enhancement. The bulk of the effect is attributed to a constraint mechanism but there is only moderate agreement between the experimental results and the simple model proposed.

REFERENCES

[1165]