METALLOGRAPHIC ASPECTS OF
FATIGUE DAMAGE IN BETA BRASS

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by

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ABSTRACT

An electronic fatigue machine has been constructed and used to fatigue stable and metastable beta brass in reverse bending beneath the lens of a microscope provided with stroboscopic illumination. This enabled the progress of fatigue to be continuously observed during tests.

A study has been made of resulting fatigue damage using X-ray and various metallographic techniques such as taper sectioning and the removal of surface layers by electro-polishing. The fatigue behaviour of stable beta brass is shown to be in some respects unusual and to exhibit certain features previously unreported.

One of the most important observations has been the initiation of fatigue cracks and the formation of extrusions in short slip markings adjacent to grain boundaries. Subsequent removal of fatigue damage by electro-polishing has revealed 'persistent markings' which indicate that these cracks originate in the layers adjoining grain boundaries and not in the grain boundaries. These cracks afterwards propagated into and across grain boundaries. Fatigue cracks have also been seen to be initiated at triple point intersections of grain boundaries.
Intense and 'permanent' slip markings excluding those described occurring adjacent to grain boundaries have not been observed to form. Extrusions were found to be rare and comparatively small. Fatigue debris considered to be an attrition product consisting of oxidised fragments of metal has been observed.

An endeavour has been made to explain these phenomena in terms of the nature of dislocations and their movements in the superlattice. It is thought that cross slip is unlikely to occur and that this inhibits the formation of intense slip markings and extrusions. A disordered grain boundary layer had been proposed to account for the initiation of cracks adjacent to grain boundaries.

A fatigue induced precipitate has been described in metastable beta and an explanation has been given in terms of enhanced diffusion due to the generation of vacancies by moving dislocations during fatigue.
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INTRODUCTION

The study of the basic aspects of the fatigue of metals has been approached in several ways. An approach to the problem has been to investigate changes in surface structure during the fatigue process. Here various metallographic techniques such as optical or electron microscopy and X-ray observations have been applied to determine and correlate these important changes in structure as fatigue progresses. This approach has resulted in increased knowledge of the formation of surface slip markings, and of the initiation and propagation of fatigue cracks under cyclic stress conditions. The results so far obtained have not been completely reconcilable nor have they been completely explained, but the general pattern of fatigue damage is now reasonably well established.

The present study, of the metallographic aspects of fatigue in beta brass, was prompted by the lack of knowledge of the fatigue behaviour of body-centred cubic metals. This excludes alpha iron, the fatigue behaviour of which is fairly well known and which was the subject of pioneer work on fatigue by Humfrey and Ewing (1) in 1903. Beta brass has an ordered body-centred cubic structure of the caesium chloride type, and as far as the author is aware no work has been published on its fatigue behaviour nor, indeed, on the fatigue behaviour of any other superlattice.
1.1. The Structure of Beta Brass CuZn

There exists in the CuZn system a homogeneous region denoted beta, the composition of which is of the order of 50 atomic wt.%Zn i.e. CuZn. This has a disordered b.c.c. structure (i) at high temperatures, and an ordered b.c.c. structure of the caesium chloride type (ii) at low temperatures, (2). The change from ordered to disordered structure takes place in the neighbourhood of 460°C, and is a simple superlattice transformation.

(i) Beta H.T. CuZn, Cu rich boundary 871°C, W type b.c.c. 
\[ a = 2.9907 \text{Å} \]

(ii) Beta quenched L.T. 47.66 Zn At.% CsCl ordered b.c.c. 
\[ a = 2.9539 \text{Å} \]

Experiment has shown that two distinct types of order can exist i.e. long-range and short-range. In long-range order the lattice can be regarded as being composed of two (or more) interpenetrating sub-lattices, one (or some) of which contain most of the atoms of one kind, the other(s) containing most of the atoms of the other kind. Above the critical ordering temperature long-range order breaks down to give short-range order in which small ordered groups of atoms continually form, break down, and reform.
Theory predicts for b.c.c. CuZn superlattice that ordering will be nearly perfect at low temperatures and that the degree of ordering decreases slowly with rising temperature at first but decreases more rapidly as the critical temperature is approached, when ordering should completely disappear (3).

Sykes and Wilkinson (1936) (4) have shown that CuZn does not disorder completely above the critical temperature but changes to short-range order and that in practice the order/disorder change took place within a much narrower range of temperatures than that predicted by theory. Chipman and Warren (5) measured the superlattice lines of quenched and slowly cooled beta brass and concluded that the quenched state had the same amount of long-range order as the equilibrium state. Thus, beta brass orders so rapidly that water quenching will not prevent it to any appreciable extent. Indeed, further experiment has shown that the order/disorder transformation is impossible to suppress by quenching. Imai and Kitajima (1951) (6) found no third element which would prevent the beta phase in the CuZn system from being ordered and of the caesium chloride type structure at low temperatures.

It is difficult to start disorder in a perfectly ordered alloy i.e. to place the first few atoms in 'wrong' positions, since each such move is resisted by all the neighbouring atoms concerned. As the amount
of disorder increases it is easier to create more disorder.

1.2. Domain Boundaries in Beta Brass

Within an ordered domain ordering is complete, but if two domains are out of step i.e. antiphase, the sublattice chosen by Cu in one domain is the same as that chosen by Zn in the other. Strictly speaking, there will be no overall long-range order. Long-range order exists because domain boundaries are unstable at low temperatures. Domain boundaries are unstable because CuCu and ZnZn bonds across them give them a high internal energy. Therefore, at low temperatures we expect domain boundaries to be at a minimum. At high temperatures, in contrast, they will lower the overall free energy, and increase in amount, because they have a high entropy. In CuZn long-range order forms very quickly showing that coalescence of the domain boundaries occurs easily.

1.3. Deformation of Beta Brass

An analysis of dislocation structure in superlattices has been given by Cottrell (1954) (7) and by Flinn (1960) (8). They show that the deformation of a superlattice can be understood in reasonably quantitative detail by considering the properties of pairs of dislocations and their interaction with antiphase boundary material.
In any well annealed crystal of beta brass only a single domain should exist, with antiphase boundary present only in connection with dislocations. As pointed out by Cottrell (7) and Ardley and Cottrell (1953) (9), dislocations of the disordered lattice are only partial dislocations in the superlattice. They must either combine to form larger dislocations with Burgers Vector equal to a primitive translation of the superlattice (Fig. 1a), or occur in pairs connected by a strip of antiphase boundary (Fig. 1b). According to Brown and Herman (1956) (10), at complete order the dislocations should be about 15 atoms apart.

In undeformed beta brass the dislocation pairs will all be in stable (metastable) position i.e. orientations, since any unstable configuration would reform when the superlattice formed (when diffusion and dislocation climb were possible). Many dislocations will be relatively immobile since the motion of any pair not lying in a slip plane, will require the creation of new antiphase boundary, and there will be a corresponding opposing force. However, in each structure, there are certain orientations of singular lines for which stable (metastable) positions are those with both dislocations in the same slip plane. These positions for the CuZn superlattice are those close to either [111] (pure screw) or [112] (pure edge). Portions of dislocation lines with these orientations can act as Frank-Read sources. The loops generated will contain dislocation pairs which are in unstable positions, but if no climb can occur the entire loop will
FIG. 1(a)

FIG. 1(b)

STRIP OF ANTIPHASE BOUNDARY
remain in the slip plane. Screw dislocations in the CuZn superlattice will be confined to the \{110\} planes.

Hence wavy slip lines, and choice of slip, should not be present in the ordered structure as compared with disordered structure. The principle obstacle to slip should come from the effect of anti-phase boundaries, these are only present in beta brass when quenched in contrast to the Cu\_3Au type superlattice. It has been suggested that the age softening of beta brass after quenching is due to the elimination of anti-phase boundaries (11).

Deformation of polycrystalline material requires slip on intersecting slip planes to relieve stress concentrations at grain boundaries. In an ordered material this type of slip results in a high work hardening rate, since domain size is rapidly reduced as new antiphase boundary is produced. Associated with this effect is a tendency towards grain boundary fracture, since the stresses at grain boundaries become relatively large before they are relieved by slip on new systems in adjoining grains. In an ordered alloy the resistance to slip increases with the passage of dislocations, as each one increases the amount of antiphase boundary.

Cold work destroys long-range order, as was observed first by Dehlinger and Graf (1930) (12). Dahl (1936) (13) showed that the mechanical disordering
caused by cold work produces changes in those properties that are affected by long-range ordering. The existence of order has a marked effect on the manner in which an alloy deforms and, therefore, on its mechanical properties, as shown by Sachs and Weerts (1931) (14). Seemann and Glander (1938) (15) suggested that cold work reduced an ordered alloy to a kind of 'gruel' of antiphase domains.

According to Cottrell (7), super dislocations and pairs of dislocations in a slip plane, connected by a strip of antiphase boundary, provide mechanisms which do not change the state of order. Broom (1954) (16) suggests that disorder in an ordered alloy may be created by deformation in several ways. First the passage of odd numbers of single dislocations across a slip plane will give rise to an antiphase boundary. Second if vacancies are generated during deformation and then move away from the slip plane, a 'filament' of disorder will result from the motion of each single vacancy in an originally ordered lattice. However, these processes will be difficult and he considered the most likely cause of disordering is the multiplication of domain boundaries by the passage of dislocations across domains.

The tension of the domain boundary that stretches between the pairs of dislocations and that holds them near together should act to inhibit the generation of dislocations at Frank-Read sources.
It should also interfere with the processes by which obstacles to the movement of dislocations are surmounted, whether this be by climb or by penetration between local obstacles.

The interaction of vacancies with moving pairs of dislocations should create strings of atoms in antiphase positions within an ordered domain, strings which trail out behind dislocation pairs and resist their continued motion.

The difficulty of the passage of dislocations and the difficulty in starting Frank-Read sources operating, coupled with the tendency of dislocations to pile up in the ordered b.c.c. beta, may be the explanation of the inherent brittleness of the material. This might also explain certain aspects of the fatigue behaviour of beta brass.

Taylor (1928) (17), found that beta brass behaves in a manner similar to alpha iron when deformed. Within a certain range of orientations of the crystal axes relative to the direction of deformation, slip did not occur on definite crystallographic planes. In other ranges of orientation slip occurred on a definite crystal plane of the \{110\} type. He also showed that in beta brass the resistance to slipping in one direction on a given slip plane was not the same as the resistance offered to slipping in the opposite direction.
Elam (1936) (18), studied the deformation behaviour of beta brass and concluded that there was no essential difference in the behaviour of alpha iron and beta brass in tension. In most crystals a \{110\} type plane was the nearest plane to the distortion plane, whilst a \{112\} plane was less frequently observed. The \{110\} plane and \{112\} plane are the two planes of greatest atomic density in the lattice. There are eight \{110\} planes to choose from and twelve \{112\} planes. Elam (18), suggested that slip occurred first on a plane of maximum resolved shear stress coincident or near to a \{112\} or \{110\} plane and that probably slip took place on planes which were not of necessity crystallographic but were planes of maximum resolved shear stress. Also the slip direction was not necessarily a crystallographic direction. The vast majority of observations indicated that in general slip was near enough of the type \{110\}[111] and less frequently \{112\}[111]. She also noticed an additional laminated surface feature, which showed no crystallographic agreement with any likely plane, (resulting from deformation) making angles with and apparently distinct from slip bands. These she described as 'broad bands' and suggested that possibly they were planes of distortion analogous to Luder's lines.

Von Goler and Sachs (1928) (19), found that slip in stable beta brass coincided with traces of a \{110\} type plane. Greninger (1938) (20), plastically
deformed stable beta brass (52Cu48Zn) obtaining slip and corrugated relief, which he called 'deformation bands', and considered that like slip these bands must be related to crystal structure as both bands and slip change direction at grain boundaries. The slip bands produced were very much less prominent than in f.c.c. metals and alloys and in general corresponded to \{110\} type planes. The 'broad bands' deviated only 2° - 7° from \{110\}. He concluded that the bands could be the result of composite slip movements having a different direction component from pure slip, leading to considerably more 'lattice rotation'.

Barrett (1938) (21), concluded that banding of the type seen in beta brass was uncommon in other b.c.c. metals and alloys, and observed that low degrees of deformation in alpha iron, tantalum and tungsten did not produce a similar effect.

Kramer and Maddin (1952) (22), in investigations on the delay time for the initiation of slip in metal single crystals found slip only on \{110\} planes in [111] in beta brass single crystals. In general b.c.c. metals may deform by twinning as well as by slip. Accordingly they studied, optically, beta brass deformed at room temperature, -52°C and -190°C but were unable to detect twins, only slip was present.

Ardley and Cottrell (9), showed that the
presence of nitrogen in beta brass can lead to a yield point analogous to that caused by carbon in alpha iron. Jerky flow, that is sudden elongation in the early stages of plastic deformation, produced audible clicks. After more than 0.2% plastic deformation these jerks were observed to be replaced by a sequence of much more frequent and smaller ones. This jerky flow is known to occur in alpha iron, where it is closely related to the yield phenomenon, but has occurred in beta brass where no yield was detected.

They considered three possible explanations for this effect. That possibly jerky flow in beta brass was a more sensitive indication of yield than the yield point itself and that it was caused by nitrogen or some other analogous element which was present in too small a quantity to produce a yield point; that jerky flow was caused by deformation twinning or by a strain transformation (Greninger and Mooradian) (20). This they rejected as an explanation because no twin crystals or new phases could be detected by X-ray, and because strain markings would not reappear on etching after they had been removed by polishing. These last observations, together with the fact that the strain lines coincided with (110), the accepted slip plane for beta brass led them to the conclusion that the markings were almost certainly packets of slip.

They also considered an effect suggested by Fisher, Hart, and Fry (1952) (23), and Mott (1952) (24),
to be a possible explanation of jerky flow. They suggested that the stress needed to start a Frank-Read source working is greater than that needed to keep it working, since the first dislocation loop formed has to start from rest, whereas subsequent loops have a running start. Consequently, once a source has started to create dislocation loops it continues to do so until the stress acting on it is substantially reduced, and an avalanche of dislocations is created, (each avalanche corresponding to one jerk/click).

Beta brass is an ordered alloy at room temperature and an ordinary slip dislocation, with a unit Burgers Vector in the lattice of atomic sites, is only a partial dislocation in the superlattice; it is therefore a topographical necessity that this dislocation forms the boundary of a stacking fault in the superlattice i.e. the boundary of an antiphase domain (Flinn) (8).

A unit dislocation thus resembles in the superlattice one of Shockley's half dislocations (1948) (25), and as in the case of half dislocations, it is likely that unit dislocations have to move through the crystals in pairs, in order to leave a perfect superlattice in their wake. Ardley and Cottrell (9) considered that it would obviously be difficult to start a Frank-Read source operating under such conditions, especially in the presence of
interstitial atoms and that this leads to the phenomenon of avalanches of slip dislocations of the right order of magnitude to explain jerks.

Ardley and Cottrell (9) found that slip systems in beta brass were $(110)[111]$ at room temperature; $(112)$ almost exclusively at liquid air temperatures and that at temperatures greater than $250^\circ C$ slip lines were confused in appearance, a few $(110)$ and $(123)$ and most noncrystallographic.

Barrett (1954) (26) found no mechanical twinning in specimens worked in the temperature range -191 to $625^\circ C$. Various types of deformation bands were present, including the type described by Honeycombe (1951/2) (27), as bands of secondary slip and the type known as kink bands. The most common configuration was sets of parallel bands in the form of spikes extending from grain boundaries. Bassi and Hugo (1958/9) (28), have reported the same spiked deformation bands in worked beta brass. They show that only the first stage of deformation occurs by slip, and that essentially this mode of deformation is confined to the surface layers. Subsequent deformation leads to growth of deformation bands which are visible to a considerable depth contrary to slip line behaviour. Etch pits indicated high dislocation concentrations in the interior of these bands which grow from grain boundaries.
The mechanism of deformation band formation in beta brass is considered rather uncertain. Bassi and Hugo (28), attribute them to tilt of part of the lattice, in relation to the matrix. Barrett (26), suggests that they originate from a dislocation or wall of dislocations being halted at an obstacle. From this it can be inferred that grain boundaries in beta brass are an effective barrier to dislocations and that rapid piling up of dislocations causes plastic deformation by glide to cease very quickly. Further deformation is by deformation bands, the lattice tilt necessary being crystallographically limited.

Perryman (1954/5) (29), and Bailey (1952) (30), after work concerning intercrystalline failure of binary beta brass and beta brasses containing 3% aluminum or more concluded that grain boundaries in beta brass must be inherently brittle. Perryman (29), suggested that the nature of the inherent 'brittleness' can be understood from the following considerations:

1. Nature of, and the difference between the atomic arrangement within the grains and immediately in the vicinity of the grain boundaries in polycrystalline materials.

2. Corresponding difference in modes of deformation under stress at the grain boundaries and in the interior of grains.
It is generally considered that at a grain boundary the atomic arrangement is less regular than within the grains, the region being one of imperfect 'transition' lattice structure. Ke (1948) (31) has shown that grain boundaries behave in a truly viscous manner, i.e. the shear strain rate decreases rapidly with temperature. This is in contrast to the mode of deformation of the component crystals, which is commonly slip, a process requiring a resolved shear stress exceeding a critical value on particular planes and is relatively insensitive to temperature.

There is evidence from work on single phase boundary alloy systems (32-36) of so-called 'equilibrium' segregation, whereby, because the transition lattice structure between grains is less regular than the structure within the grains, the concentration of solute and solvent atoms in these regions, where equilibrium is established, may differ from the concentrations within the grains, to an extent depending on the misfit between grains, on prior heat treatment and on solute content.

Beta brasses are notable in comparison with alpha brasses for their high resistance to deformation at temperatures below 500°C. This is presumably connected with the fact that they have an ordered structure at room temperature. On the other hand there is no a priori reason to suppose that the properties of the intergranular transition lattice
in binary beta brasses differ greatly from those of binary alpha brasses.

It is possible that intergranular flow may occur more readily under certain conditions than deformation of grains themselves, (especially as deformation in beta brass is difficult), hence as Zener (1948) (36), points out the relaxation of shear stress at grain boundaries (possibly produced by fatigue) can lead to high stress concentrations within grains and result in formation of intergranular cracks.

1.4. Transformations of Metastable Beta Brass

Transformations of the metastable beta phase in the CuZn system are considered complicated and uncertain.

Johnston (1920) (37), observed markings which he thought were analogous to Luder's lines in deformed, quenched brass (55.6wt%Cu and 44.4wt%Zn).

Clarke's papers (1927) (38), on the heat treatment of Muntz metal contain many photomicrographs of markings in beta brass; these she referred to as Neumann bands and stated that markings present in Muntz metal as quenched were the result of shock imparted during the polishing operation. Van Weert (1929) (39), first suggested that the parallel markings produced in 60/40 CuZn, by quenching from the all beta
region, might be alpha precipitated along certain crystallographic planes of the lattice. He showed that upon reheating to temperatures below 800°C obvious precipitation of alpha within large beta grains first occurred at the markings. Phillips (1929) (40), stated that in his opinion the markings observed by Van Wert (39), were mechanical twins due to quench strains.

Greninger and Mooradian (20), examined (microscopically) quenched beta (60/40 CuZn) and noted the frequent occurrence of long, thin parallel bands or markings, which they thought were not mechanical twins but were manifestations of a lattice transformation. They also observed the formation of markings in quenched beta brass, when immersed in liquid air, and their disappearance on returning to room temperature. Both types of markings were morphologically similar. The markings produced in the metastable beta were roughly parallel to \{155\} beta or \{166\} beta. X-rays showed the lattice transformation product to be distorted f.c.t. a=3.76\AA\ c/a=0.96.

Kaminski and Kurdjumow (1936) (41), found f.c.t. lattice in binary CuZn (Zn 40-44at.\%) as a result of quenching. As the alpha phase boundary was approached an increasing amount of f.c.t. was observed until only f.c.t. was observed, $a=3.755\AA\ a/c=1.047$.

Garwood and Hull (1954/6/8) (42-44), have
demonstrated that an alloy containing 39wt.% zinc, when quenched from above the beta phase boundary can undergo two types of diffusionless transformation. During the quench to room temperature small areas of the beta transform to 'massive alpha' and on further cooling below 0°C the residual beta undergoes the characteristic martensitic transformation.

Phillips (1930) (45), and later Baker (1932) (46), showed that alloys within the composition range 36.8 to 39wt.% zinc gave a uniform alpha structure when quenched from above the beta phase boundary. In this range of composition the single phase alpha field lies below that of the beta, so that the beta to alpha transformation must occur very rapidly. With sufficiently fast cooling rates the transformation is diffusionless in the sense that no atom movements over large distance are required.

'Massive alpha' is not described as martensitic as it occurs in large irregular, (often heavily twinned), masses characterised by straight edged boundaries. This transformation has the characteristics of a nucleation and growth reaction; the interface between the f.c.c. 'massive alpha' and matrix being coherent.

With alloys of higher zinc content the beta phase can be retained at room temperature by quenching. On sub-zero cooling the beta transforms to a lattice
of low and uncertain symmetry. This transformation is reversible and truly martensitic giving rise to pronounced relief effects on flat polished surfaces. The interface between the martensite and matrix is coherent.

According to Garwood and Hull (43), the $M_s$ varies between $-20^\circ C$ and $-178^\circ C$ for zinc contents 38.54 to 40.83wt.% zinc, respectively and according to Hornbogen and Wassermann (1956) (47), varies between $-38^\circ C$ and $-140^\circ C$ for zinc contents 38.8 to 41.1wt.% zinc, respectively.

Hornbogen et al (1957) (48), have reported that metastable beta obtained by quenching high purity 60/40 copper zinc underwent the diffusionless martensitic transformation,

\[
\text{Beta} \xrightarrow{\text{Beta} \rightarrow \text{Alpha}} \text{Alpha}
\]

\[
\begin{array}{cc}
b.c.c. & a = 2.9380^\circ A \\
f.c.t. & a = 3.775^\circ A \\
& c/a = 0.943
\end{array}
\]

(almost ordered alpha brass - CuAu type).

when pulled in tension. The transformation began and finished at 18% and 40% elongation respectively. Elongation greater than 40% caused the alpha to transform further to disordered, supersaturated f.c.c. alpha ($a = 3.7^\circ A$).

Strain induced martensitic transformations of beta brass occurring during cold working have been reported by Barrett and Massalski (1957) (49). The
amount of transformation product increases with increasing zinc content and its structure changes from f.c.c. (plus stacking faults) to c.p.h. (with stacking faults). The f.c.c. structure has parameters that would be expected from an extrapolation of the alpha parameters but with line shifts of the type expected from stacking faults. Alloys of any composition below approximately 50wt.% zinc transform when cold worked at room temperature. The $M_d$ temperature drops similarly to the $M_s$ temperature with rise in zinc content. At 42wt.% zinc an interval of approximately 625°C between the $M_d$ and $M_s$ was found, showing that plastic strain is unusually effective in overcoming the barrier to transformation compared with other low temperature b.c.c. transformations (e.g. in lithium and sodium).

**Summary of transformations in metastable beta brass**

1. Beta (QUENCH) 'massive alpha' + beta
   
   b.c.c. \rightarrow f.c.c. \rightarrow b.c.c.
   
   $< 39\text{wt.}\%$ zinc
   
   Beta sub zero cool beta
   
   b.c.c. \rightarrow Uncertain symmetry
   
   $39\text{wt.}\%$ zinc
   
   Beta
   
   b.c.c. \rightarrow
   
   $> 39\text{wt.}\%$ zinc
   
   (reported f.c.t. in alloy 60Cu39Zn rest Pb/Sn wt.$\%$'s)
22.

(2) C.W.

| Beta | b.c.c. | beta |

Martensitic product
f.c.c. + stacking faults -
c.p.h. + stacking faults
with rise in zinc content.

50wt.%Zn product stable at R.T.
>50wt.%Zn

(3) Deformed in tension

| Beta | b.c.c. | Martensitic product |

f.c.t.

a=3.775\(\text{Å}\)
c/a=0.943

almost ordered alpha
(CuAu type). 40wt.%Zn

1.5. Metallographic Aspects of Fatigue

Plastic deformation due to fatigue produces slip lines in the majority of ductile metals. Gough (1933) (51), Gough and Cox (1929) (52), (1930) (53), and (1931) (54), established that the system of
deformation under fatigue stresses is the same as under unidirectional stresses and that the active slip system in each part of a crystal is the system which has the largest shear stress acting on it. Different systems operate at different places in individual crystals, the transition from one to another often being very sharp. They showed that this is true for both single and polycrystalline materials. The major effect of the grain boundaries is to restrict strain, their effect on stress and slip distribution in individual grains being extremely slight. However, they have shown that anomalous slip can occur in the immediate vicinity of grain boundaries.

There is a marked difference in the appearance of slip lines produced by fatigue compared with those produced by unidirectional strain. In unidirectional strain, slip lines seen under low and moderate magnifications appear generally as quite sharp lines which are, in the main, fairly evenly distributed over each grain. (Except alpha iron and other b.c.c.'s where slip can be ill-defined). With increasing deformation in tensile tests little change occurs to the original slip lines, and new slip lines are formed between the original ones. A characteristic feature of fatigue damage is that whilst some grains suffer heavy deformation others show very little evidence of slip. Slip lines tend to be broader, shorter, frequently wavy and often
there is a comparatively wider spacing between the lines. After a short initial period during which fine slip lines are produced the subsequent changes consist of intensification and broadening of certain of the slip lines to quite broad bands. Little is known about the distribution of slip between obvious slip bands, but Broom and Ham (1957) (55), have shown with the electron microscope that more fine bands are produced here.

If the metal deforms by slip then usually the cracks are initiated in an intense slip band, evidently as an end result of the movements causing the intensification of slip bands. This behaviour was established as early as 1903, in iron, by Humfrey and Ewing (1), and has since been confirmed in many other materials such as aluminium (56-58), copper (57-60) nickel (57-58), silver (54), zinc (52), cadmium (61), and alpha brass (57-58).

If the material deforms by twinning, as in the case of some c.p.h. materials such as zinc and cadmium, fatigue cracks commonly form at the twin-matrix interface, apart from those which form in slip lines in both the matrix and twin. Bullen (1953) (62) found in zinc that during fatiguing twin boundaries moved in jerks and left a black line behind in each place where they had paused. Sometimes cracks formed in these boundaries. A similar effect has been seen by Wadsworth (1954) (61) in cadmium. This mode of
deformation is not known in f.c.c. but grain boundary migration has been observed (63).

Fatigue cracks have also been observed by Gough (51), and Gough and Cox (53), to form on pronounced cleavage planes and twin planes in antimony and bismuth. (Antimony and bismuth both have rhombohedral type structures). Fatigue can lead to intercrystalline failure as shown by Duce (1950) (57) in magnesium, Snowden (1958) (64) in lead, Broom and Nicholson (1960) (65) in aluminium plus 4% copper and Smith (1957) (66) in aluminium.

1.6. Fatigue Crack Initiation

Surface cracks have been seen to be initiated at a very early stage, well within the first 10% of the fatigue life. Cottrell and Hull (1957) (67), have demonstrated using the electron microscope that cracks are initiated in intrusions at approximately 1% of the fatigue life in copper. This has been confirmed by the fact that periodic annealing fails to increase the low stress fatigue lives of materials which do not strain age. This had been shown in steels (68), alpha brass (69), and copper (70-72).

Annealing will certainly cause distorted metals to recover or recrystallise but should not remove slip band cracks. Smith (66), suggests that the argument is not conclusive and puts forward the
possibility that gases present in the fatigue test could diffuse to distorted structures and prevent their recovery. There is also the possibility that heat treatment might modify the structure of surface cracks enabling increased lives to fracture to be obtained. This has been demonstrated in the case of gold (73), where periodic annealing did improve the lives of low stress specimens. The improvement was attributed to filling and rounding of cracks by a process of self-diffusion (73) (74), chemical damage due to the atmosphere being presumably absent, (75).

Head (1953) (76), had made a mathematical analysis of rate of fatigue crack propagation and has proposed a definite relationship between length of crack, number of cycles and some function of stress. Using the experimental data due to Moore (1927) (77), De Forest (1936) (78), and Bennett (1946) (68), he found that on extrapolating back to the grain size of fatigued polycrystalline materials one accounts for the greater part of the life.

Thompson et al (60), Modlen (1960) (79), and Harries and Smith (1959) (80) have shown in copper, mild steel and aluminium respectively, that longer fatigue lives can be obtained by periodic removal of the surface layers during the course of fatigue tests.

Slip bands which are going to form cracks can be detected at a very early stage by electro-
polishing to remove surface slip (70), (73), (79), (81). This was first demonstrated by Wadsworth and Thompson (70). They observed that the electro-polishing of a small layer, of the order of 2 microns, from the surface of fatigued copper gave rise to what they called 'persistent markings'. The persistence of some fatigue slip bands after electro-polishing is quite characteristic and differs from the behaviour of slip bands produced by tensile deformation. Here electro-polishing removes the great majority of them, the rest are faint and indistinct, quite unlike the dark bands remaining on fatigued specimens.

An important consideration in connection with 'persistent' slip bands and grain boundaries is whether they represent cracks in the true sense of the word i.e. a definite parting of the lattice, or whether they are regions of highly distorted metal, which behave in a different manner from less distorted material on electro-polishing. The fact that hand polishing (66), shows essentially the same effects confirms that the markings are not the result of spurious electro-polishing effects.

Thompson et al (60), further demonstrated that the 'persistent markings' spread and formed cracks on refatiguing and that they could be opened out to form cracks when subjected to a tensile stress. Where the 'persistent markings' were small a tensile stress did not always clearly open them out but small
amounts of secondary slip frequently occurred at their ends. This they concluded was due to local stress concentrations such as would be present if the markings were true cracks.

They noted this behaviour in both single and polycrystalline copper. In the case of single crystals the 'persistent markings' were rather shorter, first appeared at 18% of the fatigue life, were of the order of 50 microns long and arranged in rows along slip bands. In the polycrystalline material 'persistent markings' formed as early as after 7½% of the fatigue life. They attached importance to the grain boundary observing that it was only after 42% of the life that cracks ceased to be confined to individual grains and that once cracks had entered neighbouring grains they lead to the final failure.

They also showed that with progressive electro-polishing the 'persistent markings' disappeared and concluded that they formed only on the surface and that their length and depth were a function of the number of stress cycles.

'Persistent markings' have since been shown in most materials e.g. aluminium (81), mild steel (79), gold (73), and beta brass (82). Their locations and morphology depend on the material and stresses concerned. In some materials 'persistent markings' (66), (81), are first observed as a series of small dots
which sometimes develop into pits or holes and form an easy crack path or join together to form a more continuous marking. In other materials 'persistent' grain boundaries are common.

There is, however, some unresolved confusion about the occurrence and interpretation of 'persistent' slip bands. Kemsley (1957) (83), applied a similar electro-polishing technique to that used by Thompson et al (60) to copper specimens. He was only able to find 'persistent' slip bands in specimens fatigued at low stresses (lives less than \(10^7\)) and then only if electro-polished in the early stages of the test. It is possible that differences in technique or impurities present can explain the difference in results.

1.7. Etching of Fatigue Damage

Samuels (1955) (84), demonstrated in copper and copper alloys that zones of heavy deformation could be etched up in section. Acid ferric chloride was found to be the most effective etchant. Kemsley (1956) (85), (1956/7) (86), observed that internal fatigue damage in copper could be etched up in a similar manner. He found crystallographic striations were developed by etching but they did not correspond to those types listed by Samuels (84). The striations were shown to be the result of fatiguing and not the sectioning process. It is of interest to note that,
as early as 1924 Thompson and Millington (87) observed
the development of similar striations, by etching, in
a fatigued brass feed-water heater tube. Etching of
electro-polished surfaces of fatigued specimens has
also been shown to develop markings in addition to
'persistent markings' already present (81) (88);
presumably this is a more sensitive test for slip
planes in which cracks are going to be developed.

Forsyth (1951) (89) has used sensitive
polishes and etches on fatigued materials. His early
work done on 3% silver in aluminium solid solution
and later on pure aluminium showed essentially the
same effects. He found that it was possible to etch
up many of the regions which had been slipping. It
appeared that the material in the slip bands had been
heavily deformed and had polygonised into sub grains
of slightly different orientations (in some cases
change in orientations as much as 30°). The poly-
gonisation leading to differential etching was less
marked in specimens which were fatigued at low stresses
and low temperatures.

Similarly Jacquet (1956) (90), fatigued a
67/33 copper zinc alloy and etched up slip bands. He
found that if the brass were annealed at 550°C for one
hour and re-etched the straight traces of the etched
slip bands were replaced by a diffuse polygonised
structure. It seems that in the last two cases i.e.
Kemsley and Jacquet, the metal had possibly not
polygonised during the test but etching had been selecting the regions with the largest internal stresses.

1.8. Extrusions

Slip band extrusion was first observed in a fatigued Al+4%Cu alloy, but has now been found to occur in many other materials (67) (81) (82) (88) (91-93).

It usually takes the form of a thin ribbon or spill of the crystal material projecting as much as 20 microns from the slip band. The thickness and shape of the extrusion varies with different materials and conditions. Extrusions occasionally take the form of whiskers or filaments and have been seen to be exuded in these forms from grain boundaries (64) (65), and small holes in 'persistent' bands (81).

Forsyth (1957) (81), considers that the reverse slip mechanisms by which slip ridges and grooves are formed are the processes by which extrusions are created. Apart from extrusions, crevices or intrusions have been observed. It is thought that they are closely connected with extrusions, and are generated by the same processes. Intrusions have been clearly seen in silver chloride, by Forsyth (81), using transmitted light.
Normal metallographic techniques whilst adequate for resolving surface damage laterally do not easily show the structure in depth. This makes the observation of intrusions in fatigued metals and alloys difficult. Wood and Segall (1957) (94) have shown that for this purpose sufficient magnification can be obtained by taper sectioning a specimen at a very small angle to the surface and then viewing the polished section under high optical magnification. Taper angles of 2° - 3° were used to obtain an increased magnification (approx. x 20) of the surface disturbance in section. Giving an overall magnification, in one direction, of 10,000X-20,000X.

Intrusions closely associated with extrusions, in copper and alpha iron, have been shown using a taper sectioning technique, (95).

Cottrell and Hull (67) have demonstrated that extrusions and intrusions are formed in copper at a very low percentage of the fatigue life (1% at 90°K) and at temperatures as low as 20°K. They are of the opinion that cracks originate from intrusions at a very early stage. They suggest a purely mechanical process of cyclic slip, at a free surface, is responsible and have proposed a model based on this. The formation of extrusions and intrusions on different planes is explained as the result of the Bauschinger effect and sequential slip movements on two intersecting planes. It is interesting to note that
Forsyth (81) observed extrusions and intrusions on the same plane in fatigued silver chloride.

Mott (1958) (96) has suggested an alternative model which, making no special assumptions about substructure, predicts crack formation along slip bands under any conditions of cyclic stress. Unlike the Cottrell-Hull (67) model, cracks should be formed underneath extrusions. Mott's model depends on the existence of a cavity about which cyclic cross slip can occur leading to an extrusion. He considers that dislocation interaction of the type proposed by Fujita (1954) (97) will initiate suitable cavities if none already exist. Fujita's mechanism requires slip on two parallel planes within a few atomic distances of one another, such a process can occur, repeatedly in a broad slip band. Mott's (96) mechanism can be applied equally to single or polycrystalline material because the initiation of the required cavities does not require dislocation pile-up at grain boundaries. Extrusions are not considered essential to crack initiation but just to be a by product of surface reverse slip movements.

1.9. The Effect of Grain Boundaries on Fatigue Behaviour

It is to be expected that grain boundaries and hence the grain size have an important influence on fatigue properties. In general it appears that smaller grain sizes are associated with higher fatigue
strengths. This has been shown to be true for a variety of materials (57) (98-102). There appears to have been no attempt to investigate separately the effect of grain size on crack initiation and propagation. Since in each grain the plane of the crack is strongly influenced by crystallographic orientation, and since neighbouring grains are in general randomly orientated with respect to one another, the passage of a crack from one grain to the next is likely to be difficult. Thompson et al (60) have observed in fatigued polycrystalline copper that cracks initiated as early as after 7½% of the fatigue life were confined to individual grains until after 42% of the fatigue life and that those cracks which propagated through grain boundaries led to the final failure. Crack propagation might be expected, therefore, to be more rapid in single crystals and in large grain sized materials.

Gough, Cox and Sopwith (1934) (103) first studied the influence of the intercrystalline boundary on fatigue characteristics. They found in aluminium that the effect of the boundaries on stress distribution was negligible. The close approach of general systems of slip to the boundaries and the small amounts of anomalous slip in that region, indicated that even locally the boundary has a very limited field of influence. They concluded that the major effects of grain boundaries were to restrict strain and strengthen crystals against fatigue damage.
Kemsley (1956/7) (104) showed that fatigued copper exhibited intercrystalline failure at high stresses. He observed marked surface rumpling, resulting in the formation of grooves adjacent to grain boundaries which he suggested acted as stress concentrators and were the cause of the intercrystalline failure. He further suggested that an increase in grain size would increase high stress fatigue life by minimising intercrystalline failure. At low stresses an increase in grain size results in longer slip bands and, therefore, longer resultant cracks. He suggests that this is the reason why, at low stresses, an increase in grain size results in a lower fatigue life, at low stresses intercrystalline failure is absent.

1.10. Effect of Stress Levels on Mode of Fatigue Failure

It has been suggested by many workers (85) (86) (105-108) that there are two extremes of fatigue behaviour, both fundamentally different.

At high strain amplitudes hardening is most pronounced and progressive and it is evident that the resulting fatigue fracture is closely related to static fracture, whereas at low strain amplitudes specimens survive an initial strain hardening and fail after many more cycles.
Kemsley's work (85) (86) (107), supports this view, in that it has shown definite metallographic differences between the effects of high and low fatigue stresses. At low stresses the damage was localised and characteristic of fatigue, at higher stresses it became less localised and more similar to static deformation. He further demonstrated that the anomalous annealing behaviour, also characteristic of fatigued materials, progressively disappeared at high stresses. Increased bending of the lattice (109), with higher fatigue stresses, as shown by X-ray would seem to further support this view.

Porter and Levy (1960/1) (108) have shown a marked discontinuity in the SN curve for copper, which they termed the Intermodal Transition Point (I.M.T.P.). This discontinuity they confirmed corresponded to a change in the mode of fatigue failure. They showed that at stresses below the I.M.T.P., cracks were initiated in slip bands and that crack paths were markedly dependant on the slip pattern. At high stresses above the discontinuity cracks were initiated at L or Z shaped nuclei and propagated in a more random manner; intercrystalline failure being a common occurrence. Both forms of damage were observed at stresses in the region of the SN curve discontinuity.

1.11. The Surface
It would seem that the fatigue strength of the metal away from the surface must be considerably
greater than at the free surface. There are further unique aspects of the surface, which may be important, in causing fatigue failure to begin here. The surface is the only part of the specimen not fully supported by adjoining metal. Since the slip systems in neighbouring grains are unrelated the presence of surrounding grains makes slip in the bulk more difficult. Nearer the surface, however, the effect is less pronounced and slip is easier. The amount of plastic deformation at a free surface is likely to be greater, and that due to the action of a Frank-Read source should begin near the surface at a lower stress than in the interior. Dislocation sources at a free surface are usually of the single ended type and are more easily activated than double ended types.

1.12. Corrosion

In addition to freedom from mechanical constraint the surface is unique in another respect in that it is the only part of the specimen exposed to chemical attack due to the atmosphere.

That fatigue damage only occurs at the surface is considerably supported by the effect of the surrounding atmosphere on fatigue life. Corrosion and fatigue operating together have a much greater effect than either have when acting alone or consecutively. In the normal fatigue test, carried out in air, no obvious corrosion usually occurs. There
is, however, evidence that sufficient corrosion does occur in these tests to affect the fatigue life appreciably.

The most extensive and thorough work on this subject is that by Wadsworth and Hutchings (75), who fatigued OFHC copper, super-pure aluminium, and fine gold in reverse bending in air and in vacuum. In tests on copper in various gaseous environments, it was clearly established that the partial pressure of dry oxygen had a controlling influence on life, and that although water vapour alone had no effect it did enhance the effect of oxygen. For aluminium, water vapour alone was as detrimental as water vapour with air; dry air had a slightly lesser effect. Thompson et al (60) showed that when oxygen was excluded from a copper specimen the initial stages of crack formation occurred at the same rate as in air but cracks took longer to spread i.e. oxygen speeded up propagation but did not seem to influence crack initiation.

In earlier experiments Gough and Sopwith (1932) (110) found significant improvements in fatigue life when mild steel, copper and alpha brass were fatigued in vacuum. Earlier still in 1930 Haigh and Jones (111) found that the lives of fatigued lead specimens could be improved by a factor of 10 if air was partially excluded by a film of oil, grease or water.
When interpreting the results of fatigue tests it must be borne in mind that probably it has been a corrosion fatigue test. From the metallographic point of view when tests have been conducted in air corrosion has no real effect, except possibly in the case of some types of extrusion and fretting debris (65).

1.13. General Effects of Fatigue Stressing

It is known that cyclic stressing leading eventually to a fatigue failure is accompanied by work hardening and that the damage produced by fatiguing is of a different nature from that produced by unidirectional strain. In addition to hardening, fatigue has been shown to cause softening in materials previously hardened by static methods and also in age-hardened alloys. The annealing behaviour of materials which have been fatigued to fracture is similar to that of irradiated materials but differs from that of statically deformed materials.

1.13.1. Fatigue hardening

Many workers (96) (112-114) have shown that the fatigue hardening process is similar to hardening produced in unidirectional stressing. New dislocations are formed, which impede each others motion and which are responsible for most of the hardening and stored energy. Their arrangement is not the same as
after unidirectional stressing, it is thought that positive and negative dislocations are more uniformly distributed in the cyclic case; certainly fewer X-ray asterisms or centres for recrystallization are formed.

In 1953 Bullen et al (59) showed that the hardening caused by fatigue at low stresses did not cause severe bending of the lattice as did hardening by static methods. This was later confirmed by Wood (1955) (113) and Broom and Ham (55). Clarebrough et al (1957) (109) have also shown that as the stress level used in fatigue hardening was increased bending of the lattice was introduced, eventually approaching that of metals deformed by conventional work.

Kemsley (1956) (115) observed that, in the case of fatigued copper, as the stress level was increased the localised damage characteristic of fatigue became less localised and exhibited similar features to those produced by unidirectional strain. Wood and Segal (1957) (106) have shown that the rate of increase of strain hardening which occurs during the early cycles decreased during subsequent cycles, indicating that the plastic strain in cyclic stress may be non-hardening. Further, when hardening thus ceased deformation continued to proceed by the process of slip, indicating that slip movements in cyclic deformation may also be non-hardening.
Dislocations responsible for hardening in cyclic stressing seem to have a higher than normal density of jogs. Broom and Ham (55) suggest that this may be due to the repeated cutting of other dislocations or to vacancy condensation on them if the temperature is high enough. They also consider that the possibility exists of hardening by point defects which are known to be produced by moving dislocations, during plastic deformation processes (116). Louat (1954) (117) on the other hand has suggested that fatigue may give rise to the formation of special types of dislocation networks.

1.13.2. Fatigue softening

If a metal specimen, work hardened in tension to a known yield stress, is subjected to a given number of cycles of alternating stress whose semi-amplitude is less than the yield stress then a subsequent tensile stress will usually show that the yield stress has been reduced. A back reflection X-ray photograph will indicate an accompanying reduction in the gross lattice disorientation associated with work hardening.

This process of fatigue softening has been demonstrated by Polakowski and Palchoudhuri (1954) (118) and they have given an explanation in terms of the Bauschinger effect. If a specimen had been pre-strained in tension its yield stress in compression would almost certainly be low and smaller than the
maximum compressive stress applied during fatigue. Some dislocation redistribution might then occur and result in softening. Another type of explanation given by Broom and Ham (55) depends on the generation of vacancies by moving dislocations. Vacant lattice sites would be required to move to dislocations in such numbers as to produce dislocation climb and so encourage local re-arrangement akin to those taking place in polygonisation. Increased vacancy concentration may also give rise to increased diffusion rates thereby leading to over ageing effects, (92) (119).

1.13.3. **Annealing of fatigued metal**

Kemsley (115) showed that copper fractured by fatigue at low stresses resisted annealing as compared with copper hardened an equal amount by static means. Siede and Metcalfe (1959) (120), and McCammon and Rosenberg (1956) (121) demonstrated that the multistage annealing behaviour of irradiation hardened copper and that the irradiation softening of work hardened material shown by Makin and Minter (1956) (122), bore a strong resemblance to the effects caused by fatigue. Clarebrough et al (1955) 114), (1957) (109), showed that the energy stored in metal when fatigued is released at different temperatures and different rates from that which is stored in specimens strained in tension. They showed that fatigue hardening could be annealed
out at 100°C in copper whereas work hardening to the same degree, as the result of tensile strain, could not be annealed out at that temperature. It is, therefore, apparent that the nature of hardening introduced by fatiguing is different from that, say, introduced by tensile straining. It is not known why fatigue hardening in copper can be annealed out at 100°C. It is highly improbable that at 100°C dislocations can themselves migrate; on the other hand it does seem probable that the dislocation configuration, produced by fatigue, is such that some of the 'pile-ups' involved can become unlocked at this temperature and hence produce the observed softening.

1.14. **X-ray Techniques Applied to Fatigue**

It is well known that progressive static deformation of polycrystalline metal reduces initially annealed grains to disorientated mosaics. This effect is shown by the broadening out into arcs and blurring of originally sharp X-ray deflection spots with progressive static deformation. The effect implies first a rumpling of the grains as a result of the deformation and then a relaxation of the elastic strains, at points of high lattice curvature, by local slip movements due to the process of polygonisation.

The massive specimens usually employed in fatigue work demand the use of a back reflection technique such as the Laue back reflection method.
Wood and Segal (58) demonstrated that in copper, brass (to a lesser extent in aluminium and nickel) fatigued in torsion, the amount of blurring of X-ray deflections increased with the number of cycles initially. Eventually the blurring due to the fatigue process settled down to a steady condition. The degree of diffuseness was greater at greater strain amplitudes. The resulting blurring of spots was shown to be negligible in comparison with that produced by a similar unidirectional strain. This has also been demonstrated by other workers (59) (109) (117) (123-126).

It must be noted, however, that various workers have shown that back reflection X-ray photographs are often inconsistent at neighbouring points in the same specimen. In the case of fatigued copper, Broom and Ham (55), concluded that there were no typical back reflection X-ray photographs. They found that sometimes spots stayed sharp and sometimes became blurred tending to form Debye-Scherrer rings. Spurious effects have also been recorded by Davies (1954) (127) due to a slight variation in the mean stress of 2%.

If a metal which has been cold worked by a unidirectional straining process is subjected to a fatigue test, the initially very diffuse X-ray reflections tend to become sharper. This effect was first observed on rolled copper and silver in 1931.
by Dehlinger (128), and has since been recorded by a number of workers e.g. Clarebroigh et al (109). It is clearly related to the work softening process brought about by fatigue.

1.15. Effects of Fatigue on Diffusion

A number of instance have been reported in which metallurgical changes have been observed during a fatigue test which would not have occurred in the same time at the same temperature in the absence of the fatigue stress (119) (126-136). The changes are similar to those produced by heating and appear to be caused by an increased diffusion rate. A most obvious example is the apparent over-ageing of aluminium alloy specimens investigated by Hansock and Murray (129-133). Holden (1954) (119) showed that silica particles were precipitated in a copper-0.05% silicon alloy in an oxidising atmosphere at 325°C, during fatigue, whereas no precipitation occurred at this temperature if the specimens were unstressed or under a static stress equal to the peak fatigue test. A temperature of 750 - 1,000°C was needed to cause precipitation in the absence of stress.

All these phenomena may be explained by an increased rate of diffusion near the active slip bands. A number of explanations have been put forward to account for the enhanced diffusion, (137). The simplest assumes that the active slip bands become
considerably hotter, during the course of fatigue, than the rest of the specimen. This is not considered to be true as the mean temperature of slip bands cannot be much above the temperature of the specimen as a whole.

While the mean temperature of slip bands is low in cool specimens it might be possible for it to reach high instantaneous values occasionally in restricted areas; there being two chief sources of short pulses of heat i.e. dislocation movement and annihilation and formation of vacancies or interstitials, (116) (137-189). Various workers have calculated the amount of heat produced in these pulses and show that they are in general too small to have any effect in enhancing diffusion appreciably.

Dislocations, vacancies and interstitials which are produced during fatigue can all directly assist diffusion. Diffusion down dislocations is considerably more rapid than through the bulk of the metal, (140). That dislocations can in fact lead to enhanced ageing has been demonstrated by Gayler (1946) (141), who showed that the rate of ageing of aluminium 4% copper was greatly increased by previous cold work and that visible precipitates first occurred along slip planes.

It is generally considered that the enhanced diffusion observed during fatigue tests is due to
dislocations, vacancies and interstitials, produced in concentrations above their normal equilibrium values rather than by actual heating, (137).

1.16. Dislocation Theories of Fracture

It is well known that dislocations play an important part in the fracture of metals, possibly by the reinforcement of associated stress fields when obstacles to dislocations cause them to 'pile-up'. Various workers (142-148), have considered such piled up groups of dislocations and are of the opinion that the resulting stresses will be great enough to produce fracture.

Mott (1953) (149), and Stroh (1954) (142), have made a precise model of this and calculate that the stress associated with a thousand dislocations is enough to initiate a crack. Stroh (142), (150-152), considers edge dislocations to be the more important type in this respect, since they are surrounded by regions of tensile and compressive stresses, and if a large number of such dislocations arranged themselves with their stress fields reinforcing one another they will jointly be able to produce a stress large enough to cause fracture.

The fact that grain boundaries in polycrystalline materials can be efficient barriers to dislocations has been demonstrated by Gilman (1957) (156)
using zinc bicrystals and by Barrett (26) in beta brass. Further experimental evidence which lends support to the Stroh mechanism of crack formation has been offered by Stokes, Johnston and Li (1958) (157). They observed in compressed magnesium oxide crystals minute cracks formed in the region of a kink band generated by nonhomogeneous plastic strain. There is little doubt that the cracks they observed were formed as a result of piling up of dislocations, associated with one slip system only.

Mott (1951) (158), has considered the development of kink bands within a crystal. Initially dislocations of opposite sign pile up against a common barrier. This barrier may be the effect of their own stress field interaction or due to imperfections created previously as a result of dislocation interaction on intersecting slip planes. Once a localised lattice curvature has started, its influence spreads to neighbouring planes causing further dislocation pile up. The end result is a kinked region across which dislocation pile ups of opposite sign face one another. Further deformation forces more dislocations into this region which becomes more severely kinked and thereby more effective as a barrier. Applying Stroh's (142) analysis of stresses due to edge dislocations, it can be shown that pile ups of opposite sign, distributed around a kink band, will assist the applied shear stress in nucleating cracks. Fewer dislocations per slip band are required at a
kink for crack initiation than at a grain boundary.

Jacquet (1954) (145) has produced evidence of lines of etch pits (assumed to be sites of single dislocations) terminating at grain boundaries, which he offered as evidence of dislocation pile up. Bilby and Entwhistle (1956) (146), analysed this evidence in terms of Stroh's model and found good agreement.

Crowe (1954) (153), Korchendorfer (1954) (154), and Fujita (97), have independently proposed somewhat different mechanisms to form a small flat void by annihilation of dislocations; their mechanisms do not require a grain boundary. Fujita (97) (147) states that if two edge dislocations of opposite sign exist on planes approximately 10^0A or less apart, they attract each other so strongly that a crack is opened up and other dislocations moving down the same planes move into the crack and widen it. He suggests that this mechanism might apply to the case of fatigue in which many dislocation sequences will also occur and flat voids be created.

Mott's (96) theory for the formation of extrusions and intrusions is based on the formation, in the first stage, of just such a void.

Fisher (1955) (155) also suggests the possibility of the formation of a very thin cavity by
cutting of screw dislocations. Another possible mechanism is that of vacancy condensation; vacancies excessively produced in fatigue may tend to coagulate and form small flat cavities. However, Clarebrough et al (109) suggest that no abnormal concentrations of free vacancies are produced by fatigue.

1.17. Theories of Fatigue

There has been no general agreement regarding the movements leading up to the formation of the fatigue crack. The early suggestion, by Ewing and Humfrey (1903) (1), that cyclic movements intensify the slip lines by forming debris as a result of friction on the slip planes, has not been supported by electron microscopy, which shows no sign of attrition. The later theories that the movements produced an amorphous layer, also have not proved acceptable. More recent suggestions, notably those due to Gough and Hanson (1923) (159), and Orowan (1939) (160), are that local deformation consists of cumulative slipping, accompanied by strain hardening, and that cracking occurs when the capacity of the metal for further slip is exhausted. This view considers that the same processes take place in the local regions as take place on a more widespread scale in progressive static deformation and that fracture occurs for the same reason. The large amount of work carried out in the last ten years or so, has shown that this is only partly true.
Various other suggestions have now been made based on the clearly established fact that in ductile materials cracks are initiated, in the very early stages of the fatigue life, at the surface in an intense slip band (in an initial hardening stage). Some require the coagulation of vacancies, generated by moving dislocations (116), in slip planes and others the formation of oxide films or absorption of gases on surface slip steps, which may be drawn into the crystal by the relative to-and-fro movements in slip bands. Wadsworth (61), suggests that 'persistent bands' as shown by electro-polishing after fatiguing, are due to an increased oxygen concentration in the bands. This, he suggests, may be facilitated by enhanced diffusion due to vacancies and moving edge dislocations in the fatigue slip bands.

These explanations must be considered in the light of the outstanding evidence presented by McCammon and Rosenberg (161), that fatigue occurs at the very lowest temperatures. They investigated the fatigue properties and ultimate tensile strengths of copper, silver, gold, aluminium, magnesium, zinc and iron at 4.2, 20, 90 and 293°K. Except for zinc and iron which exhibited brittle fracture at low temperatures, the fatigue characteristics improved very considerably as the temperature was reduced. They showed that at low temperatures if fracture is to be produced a higher fatiguing stress is necessary both for the formation and spread of the crack.
Since fatigue is possible at 4.2°K it seems highly unlikely that it is needed to postulate any corrosion mechanism as an essential to crack initiation. Nor does absorption of gas molecules or atoms in metal, as slip planes rub together, seem to be essential. Diffusion at 4.2°K is very unlikely and it does not seem possible that at this temperature vacancies produced by dislocation movement could agglomerate to form a crack. All these processes could operate at higher temperatures but they are clearly not essential to the phenomenon.

Other explanations involve the accumulation of lattice defects in a slip band, leading to the destruction of the original lattice and subsequent recrystallisation or polygonisation (137). Volume changes associated with this process are supposed to destroy the coherence of the crystal across slip planes and hence initiate cracks. Dislocation interactions of the Stroh and Fujita types leading to fracture are also considered to be possible.

There is a great deal of experimental evidence that fatigue cracks can be and are usually produced by slip alone. Based on this idea Cottrell and Hull (67) have offered a purely geometrical explanation. They have suggested a mechanism involving sequential slip movements on intersecting planes leading to an intrusion and extrusion, cracks being initiated in the intrusion.
Mott (96) has put forward a dislocation mechanism involving cross slip, by which an internal cavity, produced by dislocation interaction, can grow and an extrusion be formed on the surface with a void beneath it. Like the Cottrell-Hull mechanism it gives preference to the free surface. This process can occur at all temperatures, no diffusion being required, only cross slip. This, then, would not apply to materials in which cross slip was either very unlikely or impossible e.g. cph materials.

A cross slip mechanism has also been put forward to account for the continuance of slip band formation after the cessation of the initial hardening (162,163). It is suggested that screw dislocations can move with neither multiplication or hardening by cross slipping back and forth between closely spaced planes and so generate ridges and crevices on the free surface. Mott, (163) suggests that this is the mechanism by which slip lines broaden and intensify, when no further hardening is observed, become 'persistent' and finally develop cracks.

Wood and Segal (58,94,164,165,) are of the opinion that the mode of fatigue fracture depends on the strain amplitude and that at high strain amplitudes progressive hardening leads to a fracture similar to that produced by static deformation. At small amplitudes, they suggest, internal stresses which otherwise would be stored in the lattice find an outlet by supplying the energy required to form
notch like surface contours in which cracks initiate. Hence, internal stresses are relaxed, little strain hardening occurs and, in intense slip bands, totals of plastic strain without fracture are permitted which, when summed irrespective of sign, are many orders higher than occur in static deformation.
APPARATUS

2.1 Fatigue Machine

The type of fatigue machine constructed was influenced by several factors; the most important being the intended metallographic nature of the investigation and the strip form of the 'as received' beta brass.

No suitable fatigue machine was readily available for this work and rather than purchase a necessarily expensive machine one was constructed. In order to achieve good resolution at high optical magnifications a flat as opposed to a curved surface is desirable. Fatiguing of flat specimens is commonly carried out in two ways, i.e. in reverse bending or in push-pull.

A reverse bending machine was decided upon, in preference to a push-pull machine, because of the simplicity of design and low cost of construction. Specimens are of the cantilever type. The non-uniform stress distribution, inherent in such a system, was recognised. This was not considered to be serious as a fairly uniform stress distribution can be obtained, over a limited area, by suitable tapering of specimens. The distribution of stress amongst individual grains in
a polycrystalline material is complex. No attempt was made, therefore, to determine more than an approximate value of the stress.

A main requirement of the machine was that it should be possible to observe fatigue damage, whilst it was occurring during the test. This necessitated fatiguing of specimens beneath the lens of a microscope, illuminated by a stroboscopic light source.

Before constructing the machine a survey was made of the literature concerned with machines of a similar nature.

Forsyth (166), has constructed a fatigue machine that facilitated continuous microscopic examination of specimens. Cantilever specimens were driven at resonance by an audio frequency oscillator. The drive was in the form of a moving coil attached to the specimen at its free end. Movement was detected as a transmitted vibration in the body of the machine by a crystal pick-up. Dolan (167), has produced a similar machine.

Forsyth recognised that it would have been a great improvement to have used the pick-up signal, amplified, to maintain vibrations at the resonance peak. Hanstock and Murray (129), have maintained torsional vibrations at resonance in this manner during a study of variation of damping capacity.
The fatigue machine and ancillary electronic equipment, see overleaf.
during fatigue, but their pick-up device was not suitable for reverse bending specimens.

A fatigue machine of different construction which utilised some of the principles from the apparatus discussed plus major modifications was developed.

2.2. Apparatus Development

The initial apparatus consisted of a power amplifier fed with a signal from a beat frequency oscillator, the power being dissipated in an exciter coil creating an alternating magnetic field. The magnetic field was used to cause cantilever specimens, fitted with high silicon iron pole pieces, to vibrate at resonance in reverse bending. Resonant vibrations were used because maximum amplitudes of vibration could be obtained with a minimum of power. A stroboscopic light source, triggered by the amplifier output signal, was used to illuminate specimens. The exciter coil was designed to load the amplifier correctly and to provide an efficient magnetic circuit. A schematic diagram of the apparatus is shown in Fig. 2.

A suitable wave form and power factor was achieved for frequencies in the range 60 c/s - 10,000 c/s, by means of a condenser bank in series with the exciter coil. To check wave forms and measure frequencies an oscilloscope was used as an integral part of the apparatus.
FIG. 2

SCHEMATIC DIAGRAM OF THE APPARATUS
This system was inherently unstable. Specimens vibrating at resonance would only remain so far short periods of time. The main reason responsible for this instability, apart from fluctuations and drift in output frequency, was the change in the natural frequency of specimens brought about by fatigue damage.

This change in resonant frequency necessitated continual manual adjustment of the oscillator. It was not possible, therefore, to maintain a constant load on the specimens, to obtain efficient stroboscopic illumination of the specimens, and to observe the specimens without distraction. Also specimens could not be left vibrating unattended. To overcome difficulties with the oscillator and to maintain resonant vibrations some form of pick-up was required; the function of the pick-up being to detect the resonant frequency of specimens and provide a signal of this frequency to the amplifier in place of that supplied by the oscillator. It was essential that the pick-up should neither physically interfere with free vibrations nor be interfered with by the alternating magnetic field set up in the exciter coil.

A phototransistor (Mullard ocp 71) type pick-up was decided upon after various other types had been experimented with. The advantages of the phototransistor as a pick-up were, that no physical contact with either the machine or the specimens was
required, it could be located remote from the specimens and was not interfered with by the alternating magnetic field.

To the tip of the specimen was fixed a black flag which, when the specimens vibrated, interfered cyclically with the illumination of the phototransistor, Fig. 3. The phototransistor, flag and light source formed an oscillator the signal of which was the resonant frequency of the specimen.

The beat frequency oscillator and pick-up signals were displayed on an oscilloscope, whilst specimens were driven at resonance by the beat frequency oscillator. Observation and theoretical reasoning showed the vibration frequency to be twice the exciter frequency and that specimens were fatigued such that a mean tensile stress existed in the surface remote from the coil, Fig. 4. This frequency doubling effect was due to specimens being attracted to the exciter coil on each half cycle of the alternating magnetic field.

To cause the specimens to vibrate at the exciter frequency, which was necessary if the pick-up was to be used, the exciter coil had to be polarised. This was achieved by means of the permanent magnet (4,000 Guass). Maintenance of resonant frequency gave stable vibrations and enabled accurate triggering of the stroboscopic light source which, in turn, made
SCHEMATIC DIAGRAM OF PICKUP.

Fig. 3

Fig. 4
possible continuous microscopic examination of specimen surfaces during the fatigue test.

The exciter coil, pick-up device and specimen clamp were mounted on a microscope stage, as shown in Figs. 5(i) and (ii). The normal direct illumination of the microscope was replaced by the stroboscopic light source. Means for adjusting the exciter coil and pick-up positions were provided.

2.3. Mode of Operation

The beat frequency oscillator was used only to produce the initial vibrations until the alignment of the phototransistor and its light source with respect to the specimen was correct. This was determined by reference to the oscilloscope where driving and pick-up signals were displayed.

The amplitude of vibration was controlled by the amplifier gain and the gap between the exciter coil and specimen. Once vibrations at resonance were established specimens continued to do so until fatigue damage reached an advanced stage when cracks made further vibrations impossible.

2.4. Stroboscopic Unit

Stroboscopic illumination was provided by a neon filled discharge tube. The steady specimen
image produced allowed magnifications up to X 500 to be used. Difficulty in strobing specimens exactly plane to the objective lens and the orange colour of the neon light made resolution at magnifications greater than X 100 poor. The most useful magnification was found to be X 100. Using this magnification a large surface area, in which most of the fatigue damage occurred, could be observed by swivelling the objective lens turret.

The strobe unit was constructed according to the circuit diagram shown in Fig. 6. Trigger signals of the correct frequency, phase and voltage were supplied by either the amplifier output or by the pick-up device.

2.5. The Pick-up Device

The phototransistor used in this device was essentially a conventional junction transistor whose inherent photoelectric properties had been exploited. The 'chopped' nature of its illumination when operated as a pick-up device made possible the use of the circuit shown in Fig. 6, in which fluctuations in output due to temperature variations were minimised.

The phototransistor was illuminated by a 3W bulb at a distance of 4 cm. To produce the optimum pick-up signal the relative positions of the bulb and phototransistor were adjusted so that the
Circuit diagram of apparatus.
(For list of components see Appendix I).
phototransistor was fully illuminated at one extreme of the specimen's amplitude and fully blacked out at the other.
3.1. Specimen Design

Fatigue specimens were made according to the dimensions shown in Fig. 7. These dimensions were chosen as being the most suitable, after various prototype specimens had been fatigued. The resonant frequency of such specimens, prior to fatiguing, varied between 80 and 140 cycles per second. During fatiguing frequencies tended to rise initially then fall with the onset of fatigue cracking.

3.2. Stress Distribution

Reverse bending gave a uniaxial stress system, surface layers experiencing alternating tensile and compressive stresses ranging from zero on the axis to a maximum on the surface. The simplest arrangement possible was used with specimens clamped at one end and loaded at the other i.e. cantilever specimens. This gave rise to a longitudinal stress gradient, with a specimen of uniform cross section, the maximum stresses occurring at the point of clamping.

The stress diagram, neglecting the mass of the specimen is as shown in Fig. 8.
TYPICAL LARGE GRAIN CONSIDERED EFFECTIVELY AS A 'SINGLE CRYSTAL' SPECIMEN.

FATIGUE SPECIMEN

FIG. 7
Stress
0
1
2
3
4
5
Kg/mm²

Length of specimen

Stress diagram for a plain specimen

Stress diagram for a notched specimen

FIG. 8.

Specimen

Free end of specimen

5.7 Kg/mm²

FIG. 9.

Both stress diagrams are calculated for a 1mm deflection of the specimen tip.
To prevent failures occurring at the point of clamping circular notches were made in the specimens. The stress diagram for notched specimens is as shown in Fig. 9. Hence, most fatigue damage occurred in the area between the notches where continuous observations were made.

3.3. Stress Control

Most fatigue experiments are run under conditions which ensure that the peak values of the applied stress are constant (brevity: constant stress). More exactly, the applied load is kept constant; the accuracy to which this implies constant stress depends on the variation during fatigue of the cyclic stress/strain curve for the different parts of the specimen. Alternatively, the amplitude of strain may be kept constant; in reverse bend tests this implies that one end of the specimen is clamped and the other end moved through a fixed amplitude. Sometimes the results of such observations are presented in terms of constant strain amplitude, but are more often converted to stress by multiplying by the elastic modulus. In some work the strain amplitude is adjusted several times in the test in an attempt to maintain a constant stress despite changes in the properties of the specimen.

In this investigation the power input to the specimen was constant, both stress and strain then changed to an extent which depended on the variation
of the damping capacity due to fatigue damage. These variations were manifested in changes in resonant frequencies of specimens and amplitudes of vibrations.

The stress values quoted were the initial values of stress in the first few cycles as calculated from classical theory of bending beams. A 2,500 lb Avery tensile testing machine was used to establish a stress/strain curve Fig.10 for the beta brass from which a value of the elastic modulus was calculated.

3.4. Specimen Preparation

3.4.1. Polycrystalline beta brass

Beta brass (52Cuwt.%48Znwt.%), as received from the British Non-Ferrous Metals Research Association, was in the form of strip 0.28 cm thick. Strips 18 x 1.5 x 0.28 cm were cut and by a process of cold roll and anneal were reduced in thickness to 1.2 mm.

Only 10% cold reduction was possible at each rolling stage before re-annealing became necessary. The rapid cold working characteristics of the material led to serious edge cracking if greater reductions were attempted. Even with reductions of 10% edge cracking occurred and had to be kept in check by filing off the edges after each reduction. Annealing was carried out in an air circulating furnace at 500°C for 15 minutes.
STRESS/STRAIN CURVE FOR BETA-BRASS

![Stress/Strain Curve Diagram]

- 11.4 Kg/mm² - 2 mm deflection of specimen tip
- 2.85 Kg/mm² - 1/2 mm deflection

FIG. 10
Some oxidation and dezincification occurred during the many annealings but were insufficient for any alpha solid solution to be microscopically detected. The final 1.2 mm thick strip had an average grain size of 2 mm and a hardness of 107 V.P.N. (in general) grains occupied the whole thickness of the strip.

Blanks were cut from the strip, clamped between case hardened steel formers, and filed to the final specimen shape, as shown in Fig. 7. After further annealing to remove the effects of shaping, specimens were carefully ground, using wet silicon carbide papers to grade 0000, to a thickness of 1 mm, prior to final polishing. This achieved the final specimen thickness and removed any dezincified surface layers.

3.4.2. Single crystals of beta brass

It was decided that it would be useful to fatigue single crystal as well as polycrystalline material. Kuper et al (1956) (168), have produced single crystals of beta brass by passing an ingot in a graphite mould through a Bridgeman furnace at the rate of 1 in./hr. Specimens were cut from the single crystal ingot with an electrolytic cutter. The apparatus required for this method was not conveniently available hence a simpler method of strain anneal was adopted.
Strips of the 1.2 mm thick material were strained to give permanent extensions from 0 - 30% and subsequently annealed at temperatures ranging from 625°C to 850°C, just below the liquidus temperature for times varying from 1 to 14 hours.

A 2,500 lb Avery tensile testing machine was used for the straining. Strips were annealed in sealed refractory tubes, and were stacked one on another to minimise oxidation. Temperatures greater than 830°C and times longer than 2 hours led to excessive dezincification and pitting of the surface. The largest crystals were formed with 3.75% permanent extension and 2 hours at 830°C. The crystals produced occupied the whole thickness of the strip and were large enough to enable specimens to be cut in which the notched area was completely within a single grain, Fig. 7. Hence, the specimens were effectively single crystals, in that most or all of the fatigue damage occurred in one grain.

After cutting to shape the specimens were annealed for 24 hours at 200°C. No recrystallisation occurred and the large grains remained. Wet grinding to the final specimen thickness of 1 mm was carried out as in the case of the polycrystalline material.

3.5. Polishing of Specimens
3.5.1. **Hand Polishing**

Specimens were polished in four stages, namely 6μm diamond, 1μm diamond, ½μm diamond and finally on microid alumina. This proved to be very laborious and time consuming, bearing in mind the large number of specimens to be polished. This method was not persevered with as scratch and strain free surfaces could not easily be obtained.

3.5.2. **Chemical polishing**

Various mixtures of nitric, orthophosphoric and acetic acids (169), were used at temperatures between 50 and 80°C. Reasonable results were achieved using this method following the use of 0000 grade silicon carbide paper. It was difficult to avoid some pitting and etching. Some grains polished well, whilst others pitted badly. The technique gave, in general, worse results the larger the grain size; often the pitted grain was the large grain at the notch of the specimen.

Chemical polishing was also not persevered with and was only used on early prototype specimens used for apparatus development.

3.5.3. **Electro-polishing**

Electro-polishing was carried out in a
Specimens were roughly prepolished on a diamond and suspended horizontally, by means of a lacquered copper wire, 1 cm below a large flat copper cathode. This method gave usually an excellent polish on the top surface facing the cathode and on the edges of the specimen. The other surface, though reasonably polished, was unsuitable for metallographic examination. Attempts to achieve good polishes on both sides of the specimens, by suspending them vertically surrounded by a tubular cathode, were not successful. Edge effects and longitudinal flow lines could not be eliminated.

The effects of thick oxide films, formed during electro-polishing, on surface damage have been observed by Wilms (1949/50) (170), Forsyth (1951/2) (171), and Louat (117). Such oxide films were shown to restrict surface strain and to suppress fine slip markings. Louat (117), has shown photomicrographs of anomalous fatigue damage, due to oxide films, produced by an unsatisfactory electro-polishing technique.

Anomalous fatigue damage similar to that
shown by Louat (117), was initially obtained in beta brass, see Figs. 11 and 12. Steps were taken to eliminate this effect, which was found to be due to long polishing times. The polishing time was cut down by prepolishing on 1μ diamond, as described.

3.6. Preparation of Metastable Beta Specimens

60/40 CuZn specimens were cut from 1.4 mm, thick strip obtained from material supplied by Delta Metals Limited.

To obtain metastable beta, specimens were quenched from the all beta region Fig. 13. The heat treatments were carried out in a vertical tube furnace Fig. 14. The top of the furnace was sealed by a removable refractory plug, through which passed a chromel/alumel thermocouple and iron wires on which the specimens were suspended. The bottom of the furnace was sealed by a refractory trap door.

Quenching of the specimens into a bath below the furnace was carried out by the simultaneous cutting of the suspension wires and opening of the furnace trap door. Specimens were held at 850°C for times ranging from 1 - 10 minutes before quenching.

Two minutes at 850°C were found to be sufficient for complete re-solution of the alpha
FIG. 11. X500. Oblique Light.

FIG. 12. X1000. Oblique Light.
COPPER-ZINC

G. V. Raynor, Institute of Metals (Annotated Equilibrium Diagram Series, No. 3), 1944.

FIG. 13.
FIG. 14

HEAT TREATMENT FURNACE

WIRE SUPPORTING SPECIMEN

REFRACTORY PLUG

THERMOCOUPLE

SPECIMEN

HINGED TRAP DOOR

QUENCH BATH 14 cm. BELOW
solid solution. Longer than two minutes led to excessive dezincification and oxidation, though this could be controlled by sandwiching the specimens between other pieces of brass.

The specimens, after quenching, were ground to a taper section using silicon carbide papers to grade 0000, and finally electro-polished by the technique previously described. This permitted examination of the structures existing at varying depths below the surface.

Structures in specimens held for two minutes at 850°C and quenched into brine at 4°C, ranged from all retained beta to all alpha at the dezincified surface. In between these extremes both 'massive alpha' (Figs. 15, 16) and martensite (Figs. 17, 18) were observed (43). Also what appeared to be a martensitic deformation product caused by a scribe mark, in grains containing the 'as quenched' martensite and in adjacent grains, was observed. Much of the martensite was striated (Fig. 18).

It is thought that the combination of dezincification and the very drastic quench, which the material near the surface received, had suppressed the reaction at high temperatures (which produces 'massive alpha') and allowed the martensitic beta to form just above room temperature, (172). As the martensite, which forms on sub zero cooling (172),
is normally free from subsidiary markings, the observation of striations appears to confirm the argument that quenching strains have assisted the formation of martensite in the dezincified surface layers.

All metastable beta fatigue specimens were obtained by quenching material held at 840°C for 2 minutes, into brine at 4°C. Surface layers were ground off to achieve the final 1 mm specimen thickness and to remove the dezincified layers containing phases other than metastable beta.
4.1. Experimental Procedure

The surfaces of both stable and metastable beta brass specimens were studied before, during and after fatiguing. Continuous observations allowed a general impression to be formed of the progression of fatigue damage from the first slip markings to the final failure. Tests were interrupted when damage was seen to have reached an interesting stage, to enable surfaces to be studied at higher optical magnifications under more ideal conditions. Bench microscopes, a Vickers projection microscope and a Bausch and Lomb projection microscope were used for this purpose. Further work was then carried out to investigate more fully individual stages in the fatigue process.

The initial frequency of vibration of individual specimens, of the same dimensions and material, varied between the limits 80 and 140 cycles per second. This was attributed to differences in size and orientation of grains. The resonant frequency could be varied by altering the distance between the point of clamping and the free end of the specimen. No differences in fatigue damage were observed, within the range of frequencies used in this investigation, which could be attributed to frequency effects.
Specimens were fatigued, at stresses within the range $2.8 - 4.4\text{Kgm/mm}^2$ which corresponded to amplitudes of vibration of 1 mm and 4 mm, respectively. These values of stress were the maximum tensile stresses experienced at the surface, in the centre of the notched area of specimens. It should be noted that a stress gradient of the form shown in Fig. 9, existed along the length of specimen surface.

A variety of techniques were used to study fatigue damage, which included the removal of damaged surfaces by electro-polishing to detect 'persistence' of slip, subjection of specimens to tensile stresses before and after fatigue, taper sectioning and X-ray.

The X-ray work consisted mainly of Laue back reflection pictures in order to orientate grains and hence index slip planes accommodating fatigue damage. Phenomenological changes and the structure and cell dimensions of a fatigue induced precipitate, in metastable beta, were also determined by X-ray.

4.1.1. Taper sectioning

Taper sections were used to increase lateral magnifications, to overcome the limited depth of focus associated with high optical magnifications and to show more clearly the nature of the surface damage in relation to the interior of the specimen.
The following method of taper sectioning was used. Tapered blanks were made from colourless transparent mounting material in a standard mounting press using a block with the desired taper angle. Surfaces of specimens were coated with Araldite which was allowed to harden. Care was taken to ensure that the Araldite adhered properly to the surface and that bubbles were absent. Specimens were then mounted as shown in Fig. 19, with the surface of interest in contact with the tapered blank.

Taper sections were obtained by grinding on wet silicon carbide papers followed by conventional hand polishing techniques. The use of transparent Araldite, in preference to a metal plating, had the advantage that the surface corresponding to the tapered section could be seen. Accurate values of the taper angle were calculated from interference fringes produced in the Araldite. The magnification being increased by a factor of times \( \frac{1}{\sin \theta} \), where \( \theta \) is the taper angle, see Fig. 20.

An advantage of this method was that any desired point in the surface could be taper sectioned. Points of interest were located by means of a series of micro hardness indentations. The disadvantages of this technique was that it was laborious especially where a section at a point of interest was required. Great care was necessary in polishing to maintain a good edge on the Araldite, so as to
METHOD OF MOUNTING A SPECIMEN TO BE TAPER SECTIONED

FIG. 19

TAPER ANGLE $\theta$

$\begin{align*}
a &= \text{true height of surface disturbance} \\
b &= \text{apparent height in section}
\end{align*}$

$b = a \csc \theta$

FIG. 20
preserve surface damage.

4.2. Fatigue Damage in Stable Beta Brass

4.2.1. General damage - slip

Fatigue damage, in beta brass, was observed to follow a different general pattern from that exhibited by either typical face-centred cubic materials (e.g. copper) or body centred cubic alpha iron. The main differences lay in the development of slip after the initial stages of fatigue, and the mode of crack initiation. The damage was also unlike that due to unidirectional strain; there being much less distortion and rumpling of the surface.

Within the first few thousand cycles of stress slip bands were observed to have been formed in the majority of grains. In some grains, however, relatively little slip was observed. This is characteristic of fatigue damage. After the initial stages no new slip bands appeared to form, those slip bands already present broadened and became more developed. The spacing of slip bands was seen to be related to the fatigue stress, the higher the stress level the closer the spacing of slip bands.

At low stresses, with increasing number of cycles of stress, slip bands broadened considerably
forming smooth surface contours that were sinusoidal in section. At high stresses slip bands were sharper in appearance, and frequently due to the more rapid failure of specimens did not develop to the same extent as those at low stresses.

Slip bands in general were long and straight but wavy shorter slip markings were also observed. The shorter wavy markings being formed more frequently at higher stresses, in the later stages of the fatigue life. Different slip systems operated in different parts of individual grains. The transition from one to another in some cases being very sharp and in other cases there was an area where slip gradually bent from one system to the other. Curvature at the extremities of slip bands frequently occurred. In some cases it appeared as if slip in one grain was bending to align itself with slip in another grain, see Figs. 21, 22.

Grain boundaries, apart from simply restricting strain, appeared in places to exert a large influence on slip. In some instances slip stopped short of the grain boundary leaving a zone where either no slip could be detected optically or unusual short wavy slip markings occurred. Yet, in other instances the grain boundary appeared to have no effect at all, slip extended from one grain to another as if it did not exist, see Figs. 23, 24, 25.

In face centred cubic materials and in alpha
iron some slip bands intensify as fatigue proceeds. Fatigue cracks are initiated in such intensified bands. This type of behaviour was not observed in beta brass. Slip bands did intensify during the process of fatigue but no one band intensified more than another.

Thin layers of the order of 2 microns or less were removed from the surfaces of fatigued specimens by electro-polishing. No 'persistent' slip markings, of the type reported by Thompson et al (60), were observed. Subsequent etching with acid ferric chloride also failed to reveal any slip markings in contrast to those demonstrated by Kemsley (85) (86) in fatigued copper, Jacquet (190) in fatigued brass and Forsyth (89) in fatigued ½% silver in aluminium solid solution.

The following observations refer to the single crystal specimens and to the centre of grains in polycrystalline specimens and exclude effects seen at grain boundaries and immediately adjacent areas, where fatigue cracks were initiated.

Fatigue damage in the larger crystals, that have been considered to be effectively single crystals was similar to that in polycrystalline specimens. Generally only one or two slip systems operated. Slip was more regular with no curvature, except in some cases where fatigue cracks had propagated.
Taper sections of various fatigued specimens showed surface slip to form smooth regular contours, the average height of slip bands, being of the order of 1000 – 2000 Å. No extrusions or intrusions were detected.

4.2.2. Fatigue crack initiation and propagation

Fatigue cracks have been observed to be initiated at the following sites:–

(a) The edge of specimens.
(b) Grain boundaries.
(c) The intersection of three grain boundaries.
(d) Deformation bands and severe slip damage.

(a) Crack initiation at the edge of specimens

This was the only mode of fatigue crack initiation observed in single crystal specimens. Cracks initiated at the edge of specimens were invariably in the notched area, at grain boundaries or in slip lines, see Figs. 26,27. The cracks propagated at about right angles to the applied stress, but tended to follow slip markings and in some cases grain boundaries. Where there was no convenient slip direction or grain boundary a random path was followed, see Fig. 28. Forked cracks and cracks zig-zagging from main to secondary slip systems were common, see Fig. 29. Where cracks did not follow slip markings
FIG. 29. X500.

FIG. 30. X250.

FIG. 31. X500.

FIG. 32. X500.
there was much disturbance of the surface on either side of them, as shown in Figs. 30, 31, 32. Cracks initiated at the edge of specimens usually linked up other cracks which had formed away from the edge, in the centre of specimens. In a great many specimens it was this crack that led to the final failure. Slip bands in which cracks initiated at the edge of specimens were not intense relative to surrounding slip bands.

(b) Crack initiation at the grain boundaries

This form of crack initiation was the most common and was considered to be the most important observation. Cracks were initiated in short intense slip markings, which formed adjacent to grain boundaries, and in a number of cases at an angle to the main slip system. These cracks were afterwards propagated into and across the grain boundaries and frequently became linked together, see Figs. 33, 34, 35, 36, 37. Away from the grain boundaries they often did not follow operative slip systems. Exudations were commonly associated with this phenomenon and took the form of either spills of metal protruding from such cracks or what appeared to be an attrition product, see Fig. 38.

Although the effect always followed the same general pattern, as outlined, the details varied in individual cases. For instance, in some cases the
effect occurred on both sides of the grain boundary, as shown in Fig. 39. More commonly it occurred only on one side of the grain boundary. Microscopic examination showed that cracks were initiated at the root of intense slip markings and then propagated into or across the grain boundaries. Splitting of the grain boundaries was frequently observed to be either the result of cracks initiated adjacent to it or the result of a propagating crack initiated elsewhere.

A few boundaries were observed where the adjacent metal darkened and became greyish blue in colour, in contrast to the normal yellow appearance of the brass. Observation of such a darkened area at magnifications of the order of X 1000 revealed fine micro cracks, which at a later stage developed sufficiently to become visible at low magnifications, see Fig. 40.

At some grain boundaries, there was observed to be a strip of the order of 200 \( \mu \text{m} \) in width, between the ends of the intense slip markings and the grain boundary, itself. Cracks propagating from one grain to another changed direction on entering and leaving this strip, see Fig. 41.

Intense slip markings adjacent to both sides of a grain boundary were removed by electro-polishing, (39). 'Persistent' slip markings typical of fatigue
FIG. 37. X 750.

FIG. 38. X 680.

FIG. 39. X 1200.

FIG. 40. X 1000.
damage, were revealed, see Figs. 42 and 43. These markings were then opened out to form cracks by subjecting the specimens to a tensile stress. It can be seen in Figs. 42 and 43 that 'persistent markings' i.e. incipient cracks, existed on either side of the boundary and had in some instances become linked together across the boundary. It was also noted that in places where cracks met at the boundary, the boundary, itself, was 'persistent'. Cracks were seen to have been initiated at points in slip markings, which were still faintly visible, away from the grain boundary and frequently did not exist wholly in the slip system in which they had initiated.

Figs. 44, 45, 46, show a typical grain boundary effect where cracks have been initiated. Fig. 44 shows the grain boundary after surface slip markings have been removed by electro-polishing. The results of removing further layers from the surface are shown in Figs. 45 and 46. It can be seen how the grain boundary, which initially had been completely 'persistent' became 'persistent' only in places where it was connected with 'persistent' slip markings. Eventually, the majority of the boundary ceased to be 'persistent'. The markings in the grains became shallow grooves and in two instances were reduced to points that appeared to be on one side of the boundary. Further polishing removed the entire effect.

Figs. 47, 48, 49, 50 show another typical
example of crack initiation at a grain boundary. In this example the grain boundary was not 'persistent' after polishing. It could be seen clearly how on further electro-polishing cracks which had initially straddled the grain boundary contracted to shallow grooves on one side, only, of the boundary. It should be noted that cracks did not lie in the operative slip system but were at an angle to it and were frequently curved.

Some grain boundaries, which darkened during the process of fatigue, were also found to be 'persistent' on electro-polishing, see Figs. 51, 52, 53, 54. On further electro-polishing the effects outlined above were again observed.

Frequently cracks initiated at grain boundaries, became linked together as shown in Fig. 55.

Intense short slip markings adjacent to a grain boundary were taper sectioned as shown in Fig. 56. Short cracks extending from the root of some slip markings were observed and in two cases extended into and along the grain boundary. These cracks were difficult to observe and required very careful polishing to reveal them. Due to lack of contrast they were difficult to show in a photomicrograph. Etching in acid ferric chloride showed them more clearly and also etched up the grain boundary see Fig. 57. Measurements taken from Fig. 57, showed the smaller
cracks to be less than 10 microns in depth.

The grain boundary shown in Fig. 39, was taper sectioned as shown in Fig. 58. This section demonstrated more clearly how only short lengths of the grain boundary had parted, see Fig. 59.

(c) Crack initiation at the intersection of three grain boundaries

This form of crack initiation was commonly associated with the grain boundary effect described in Section (b), (see Fig. 60). Cracks propagated from the point of intersection either into adjoining grains, along grain boundaries or both, see Fig. 61. Continuous observation during fatigue tests, showed that sometimes cracks formed in grain boundaries between two such three point intersections and were followed shortly by the initiation of cracks at one or both of the three point intersections.

(d) Crack initiation at deformation bands and severe slip damage

Cracks initiated in severe localised slip damage were less common and only occurred at the higher stresses used, see Figs. 62, 63. This type of crack initiation was usually associated with grain boundary cracks as described in section (b).
SECTION LINE

INTENSE SHORT SLIP MARKINGS

FIG. 56

INTENSE SHORT SLIP MARKINGS

SECTION LINE

FIG. 58

CRACK

FIG. 60
FIG. 59. X500
X TAPER MAG.4.

FIG. 61. X250.

FIG. 62. X 90.

FIG. 63. X 800.
Figs. 64, 65, 66, show where a narrow zone of severe slip damage had linked cracks initiated at two opposite grain boundaries in the same grain. The example shown was the only clear case observed where two grain boundary effects had been so linked together. Usually the effect was only obviously associated with one grain boundary.

At higher stresses cracks were formed also in triangular shape deformation markings at the grain boundaries, which were similar in form to those produced by tensile strain, see Figs. 67 and 68.

On a few occasions cracks were continuously seen propagating along what appeared to be a kink band of the type described by Mott (158). Short cracks were formed across the band in advance of the main crack which zig-zagged from one crack segment to another, down the length of the band, see Fig. 69.

4.2.3. Extrusions

Large amounts of black fatigue debris, in the form of spiral, spill and whisker shaped exudations, have been continuously seen issuing from cracks during fatigue, see Fig. 70. At high magnifications some had a definite metallic lustre whereas others appeared as greenish, translucent, frondes. The limited depth of focus, associated with high optical magnifications and interference fringes, made
FIG. 68. X 120.

FIG. 69. X 800.

FIG. 70. X 250.

FIG. 71. X 1850.
their exact nature difficult to determine.

Several taper sections were made in an endeavour to obtain a cross section of a crack and associated fatigue debris. It was found that in all cases the debris floated away from the crack into the Araldite used in the taper sectioning technique. Particles of debris so sectioned appeared to be metallic and were difficult to show in a convincing photomicrograph, see Fig. 71.

It was considered that these exudations were not extrusions in the usually accepted sense i.e. were not surface protrusions brought about by reverse slip movements, but were an attrition product. They appeared to consist of either oxidised metallic fragments or just oxide particles. They collected in cracks, in which they had formed, and also were scattered about the surface by the vibrations of the specimen.

True extrusions, typical of fatigue and of the same form as those found in fatigued copper and alpha iron, though on a smaller scale, were also observed. These extrusions were less common and were associated with cracks formed in intense slip markings at grain boundaries and in other areas of localised damage, see Figs. 72, 73, 74.

In one instance a large spill of metal of
the order of 5,000 microns in length was seen protruding from a crack. This was not thought to be an extrusion in the accepted sense but to be a large fragment of metal which had broken away from the crack wall, see Fig. 75.

4.2.4. X-ray results

Laue back reflection photographs were used to determine orientations, of a number of large crystals, relative to fiducial marks on the edge of specimens. After fatiguing, measurements of slip traces were made in two surfaces at right angles. Slip traces were indexed and observations concerning slip planes are as shown in Fig. 76. It was found that if slip traces were examined at an early stage in the fatigue life they appeared straight and smooth when viewed from any direction. At later stages in the fatigue life deformation traces became wavy and difficult to determine. The majority of traces corresponded quite closely to (123) planes, a few to (110) planes and a large number appeared to be irrational.

Back reflection Laue photographs were taken of several representative areas of fatigue damage. They showed that in the majority of cases little or no blurring occurred of the X-ray spots which remained quite sharp.
Slip traces from 18 investigations are as shown corresponding closely to (123), (110) and (221). Another 10 slip traces investigated proved to be irrational.

FIG. 76
Fig. 77 shows a typical back reflection X-ray photograph of an area of severe slip damage (corresponds to area shown in Fig. 64). It can be seen that the spots are still relatively sharp but have tended to form triplets.

Fig. 78 shows an X-ray photograph taken of the area shown in Fig. 40, where grain boundary cracks have been initiated. Here spots have become somewhat diffuse.

Crystal planes accommodating cracks initiated adjacent to grain boundaries were also determined by X-ray. Grains containing cracks were orientated relative to a reference line TT as shown in Fig. 79. The angles between surface traces of cracks and the reference line were measured. To determine planes accommodating cracks it was necessary to know how they were positioned in the solid. This was found by means of a taper section along the reference line TT. The angle between cracks and the surface in a plane normal to the surface and parallel to the reference line was calculated, as shown in Fig. 80.

Figs. 65 and 65A show cracks which were investigated in this manner and a corresponding taper section. Slip traces containing cracks were found to be rational crystal planes whereas the cracks, themselves, did not lie in rational crystal planes.
FIG. 65A. x1230
A taper section of the crack marked X in Fig. 65, sectioned along the line AA.
FIDUCIAL LINE

FIG. 79

FIDUCIAL LIBE • B

CRACKS

TAN 1

FIG. 80

TOPT = SECTIONED SURFACE

THETA TAPER ANGLE

SECTIONED SURFACE

APPARENT CRACK ANGLE
AS MEASURED IN
SECTIONED SURFACE

ANGLE CRACK MAKES
WITH PLANE NORMAL TO
SURFACE PARALLEL TO
LINE TT

\[
\tan \phi = \tan \gamma \csc \theta
\]
4.3. Fatigue Damage in Metastable Beta Brass

The first visible damage was the appearance of many small cracks, which were initiated at grain boundaries. These cracks frequently joined together to form a 'herring bone' pattern astride grain boundaries, see Figs. 81, 82. Cracks propagated across grains and along grain boundaries; where they were transcrystalline they were quite straight, see Fig. 83. The crack leading to the final failure was usually initiated at the edge of the specimen from where it propagated across the specimen to the other edge, linking other cracks together, see Fig. 84.
Slip markings were very much less prominent than in fatigued stable beta brass. They appeared to have no connection with crack initiation and had little or no influence on crack propagation, see Fig. 83. The formation of cracks in the early stages of the fatigue life appeared to prevent the development of slip markings.

Fatigue induced precipitates were observed to form either in slip markings or adjacent to grain boundaries, see Figs. 85, 86, 87. Alpha solid solution and metastable beta appeared yellowish when electro-polished whereas 'massive alpha', martensite and these precipitates appeared pink in colour.

FIG. 88
A glancing angle X-ray photograph taken of fatigue induced precipitates in metastable beta shown in Fig. 86.
A glancing angle X-ray photograph indicated that these precipitates were face-centred cubic in type and that their cell dimensions were approximately 3.65°A (at 21°C), see Fig. 88.

Removal of fatigue damage by electro-polishing showed that precipitates existed only in the surface layers and that they often outlined cracks which went much deeper. No 'persistent markings' were detected.

Specimens were tempered for 1½ hours at 200°C before and after fatiguing. Fatigued specimens, after tempering, showed preferential precipitation of alpha solid solution, in areas where damage had been greatest, and in general an increase in the amount of precipitation, see Figs. 89, 90, 91.
FIG. 90. X110
Etched in acid ferric chloride.

FIG. 91. X110. Dark field illumination. (as Fig. 90). Etched in acid ferric chloride.
5.1. **Discussion of the Fatigue Damage in Stable Beta Brass**

The general pattern of fatigue damage in stable beta brass is unusual and exhibits features which have not been reported previously, either in beta brass or in other materials.

Grain boundaries appear to exert a large influence on the fatigue behaviour of the material. Yet, in some instances, their effect apparently was negligible and slip bands extended from one grain into another; presumably this occurred where only a small difference in orientation existed between the two grains. Frequently slip bands stopped short of the grain boundary leaving a narrow zone in which unusual wavy slip markings were observed.

The most interesting observation was the formation of cracks in short intense slip markings adjacent to some grain boundaries. The results of removing small layers of fatigue damage by electropolishing and of taper sectioning such grain boundary areas suggests that these cracks are of the form shown in Figs. 92 and 93, and are described below.
Note grain boundary ceases to be 'persistent'.

(a)

(b)

(c)

FIG. 92
Fig. 92 refers to the example shown in Figs. 44, 45, 46. The surface of the fatigued specimen is represented by the plane ABCD, the grain boundary by the plane WXYZ. Fatigue cracks are shown in slip markings and in the grain boundary. A plan view of the surface corresponding to Fig. 44 is shown in Fig. 92a. Planes A'B'C'D' and A'B"C"D" represent the surface after small layers have been removed by electro-polishing. Plan views of these planes are shown in Figs. 92b and 92c and correspond to Figs. 45 and 46 respectively.

Similarly, Fig. 93 refers to the grain boundary effect shown in Figs. 47, 48, 49, 50. Plan views of the surface, after each electro-polishing stage, corresponding to Figs. 48, 49, 50 respectively are shown in Figs. 93a, 93b and 93c.

It is clear that fatigue cracks are formed in the roots of intense slip markings at points away from the grain boundaries and not in the grain boundaries, in which they are eventually propagated. The taper section shown in Fig. 57, provides further evidence that cracks are not initiated in the grain boundary but are the result of cracks formed in slip markings adjacent to it. In fact no cracks have been initiated in grain boundaries with the exception of those formed at the intersection of three grain boundaries, which are discussed later. All grain boundary cracks observed have been the
result of propagating fatigue cracks.

Fatigue debris and what appeared to be extrusions have been observed to be associated with the cracks. The cracks are formed next to the extrusions which are small being less than ten microns in height. However, taper sectioning has revealed no obvious extrusions, as are seen in fatigued copper and alpha iron. The fatigue debris is commonly observed protruding from the cracks themselves and in many instances has lost all adherence to the surface. It is thought to be an attrition product consisting of thin laths of metal which appear to be oxidised.

These phenomena may be explained in several ways all of which take into account the ordered body-centred cubic structure of beta brass.

Deformation of polycrystalline beta brass requires slip on intersecting planes to relieve stress concentrations at grain boundaries. This kind of slip should result in a high work hardening rate due to reduction in the ordered domain size as new antiphase boundary is produced. It is suggested, that possibly large stresses are developed by this mechanism at grain boundaries, and that these stresses are responsible for the formation of cracks in the vicinity of grain boundaries. Where the difference in orientation between two grains is small such a condition should
not arise because stress concentrations at one side of the grain boundary can be relieved easily by slip in the adjoining grain. If this is so, then at high angle boundaries high stresses are relieved by fracture rather than by intersecting slip or by slip in adjoining grains. At very low angle boundaries slip would be expected to spread from one grain to another. This has, in fact, been observed. This effect will obviously be dependant on the distribution of fatigue stress which, in turn, will be a function of the overall orientation of grains in the specimen.

If this mechanism is accepted then intercrystalline failure, rather than failure in the grain adjacent to the grain boundary, would be expected. This has not been observed.

It has been proposed, by Bassi (28), that grain boundaries in beta brass are effective barriers to dislocations. This coupled with the difficulty of passage and tendency of dislocations to pile-up in the material and the difficulty of causing Frank-Read sources to operate may well lead to dislocation pile-up at grain boundaries. Such dislocation pile-ups may lead to the initiation of brittle cracks at the grain boundaries in a manner as has been suggested by Stroh (142) and others (143) (149). This may be the explanation of those cracks seen to be formed adjacent to the grain boundaries.
Polycrystalline metal is made up of a large number of small interlocking crystals or grains. Each grain within the mass is joined to its neighbours at all points on its surface by a grain boundary. The packing of atoms in the boundary is almost as compact as that within the grains themselves so that there is no appreciable difference between the density of a polycrystal and a corresponding single crystal. For many years the 'Amorphous Cement' (173) theory was accepted in which grains were considered to be separated by a layer roughly 100 atoms wide, consisting of an irregular arrangement of atoms resembling the structure of a liquid rather than a crystal. It was believed that in a grain boundary the atoms were less tightly held than those in the crystals, resulting in an asymmetrical atomic arrangement. Crystal boundaries would thus have no slip planes and be difficult to deform at low temperatures. The view most widely accepted at the present is the transition lattice theory, proposed by Hargreaves and Hills (1929) (174), in which the boundary is considered as a narrow transition region, about 2 atoms thick, across which the atoms change over from the set of sites of one crystal to that of another.

It is suggested that the large effect grain boundaries appear to have on the fatigue behaviour of beta brass and indeed, on its deformation characteristics in general, can be accounted for by a wide transition layer of the order of
100 atoms existing on either side of a 2 atom wide boundary of the type proposed by Hargreaves and Hills.

This layer is suggested to be essentially disordered in contrast to the interior of the grains. It is not considered to be amorphous but to consist of a range of crystal structures e.g. starting from the grain boundary, (i) the beginning of a definite crystal structure, (ii) a disordered body-centred cubic lattice of the tungsten type, (iii) short range order of the caesium chloride type, which gradually merges with the long range order existing within the grains. The degree of disorder and the width of such a band is probably dependant on the difference in orientation between the two grains on either side of the boundary. The greater this difference, the more distinct the layer.

If this disordered zone does exist it can be expected to accommodate dislocation movement more readily and to have a shorter fatigue life than the interior of the grains. This is consistent with fatigue damage observed. It is probable that larger amounts of plastic strain occur in the grain boundary layer because of the easier motion of dislocations in the disordered lattice and the large stresses exerted on it by the deformation of the interior superlattice. It is to be expected that the fatigue behaviour of the layer adjacent to the grain boundary should be similar to that of a disordered body-
centred cubic material such as alpha iron, that it should exhibit, on a smaller scale, intense slip markings, 'persistent markings', extrusions and intrusions. These in fact have been observed. The layer observed in Fig. 41 might well be considered as evidence of the proposed transition layer.

It is possible that high fatigue stresses in the grain boundary layer can lead to dislocation pile-up and the initiation of brittle cracks and hence damage closely related to that produced by static deformation. This would explain the point initiation of cracks at a short distance from the grain boundary as demonstrated in Figs. 42, 43. It should be pointed out that all 'persistent markings' investigated were shown to be true cracks when examined in hand polished taper sections.

The explanation of the discolouration at the grain boundary where fine micro cracks had been initiated, see Fig. 40, is not understood but it is thought that it is probably the result of oxidation. Why this oxidation should occur is not known.

It is considered that the initiation of cracks at the intersection of three grain boundaries is due to the development of triaxial stresses at the triple points (175), though Rhines (176), has pointed
out that true hydrostatic conditions can only occur at quadruple points. Large tensile stresses normal to the grain boundary surface may reach or exceed the cohesive strength of the grain boundary and lead to the cracks demonstrated in Fig. 61. Similar behaviour has been reported in materials subjected to creep, where it appears that a critical stress must be exceeded before triple point cracking can occur. This stress is said to be dependant on the material constants and to be inversely proportional to the square root of the grain size, (177).

An important observation has been the absence, in the grains, of intense slip bands, of 'persistent markings' after electro-polishing and of crack initiation in slip bands. It was also noted that true extrusions in the grains were uncommon and that they were small in comparison with those found in fatigued copper and alpha iron.

These observations may be explained in terms of the nature of dislocations and their interactions, in the superlattice, under the influence of fatigue stresses. The extended dislocations that exist in beta brass are such that cross slip is very unlikely to occur (7 - 10). According to Mott (96) and others (162) cross slip plays an essential role in the formation of intense slip bands and extrusions. He suggests that slip band formation continues after the cessation
of the initial hardening stage by cross slipping of dislocations back and forth between closely spaced planes. This he considers is the mechanism by which slip lines broaden and intensify, become 'persistent' and finally develop cracks. He further suggests that cross slip allows an internal cavity, produced by dislocation interaction, to grow and that an extrusion is formed on the surface with a void beneath it. This process can occur at any temperature, requires no diffusion, only cross slip.

Slip bands in stable beta brass have been observed to develop evenly across grains, no one band developing more than another. This is suggested to be due to the absence of cross slip. Hence, large surface contours develop because there is little tendency for cracks to form in one slip band rather than another. It is expected that with sufficient reversals of stress, cracks would initiate in slip bands in the grains. This stage is rarely reached because cracks initiated adjacent to grain boundaries, at triple points and at the edge of specimens, limit deformation elsewhere. Even in the single crystal specimens intense slip markings did not develop, neither did extrusions but fatigue debris was common. This provides further evidence that intense slip markings and 'persistent markings' are not normally formed, in the grains, in fatigued beta brass.

The comparative absence of extrusions is also attributed to the lack of cross slip. The for-
mation of extrusions by the Cottrell-Hull (67) mechanism is also considered unlikely. Sequential slip movements on intersecting planes in the superlattice would lead to the production of antiphase boundary, with consequent hardening and is expected to be difficult.

The nature of dislocations in a superlattice of the caesium chloride type, i.e. the ease of dislocation pile-up, lack of cross slip, difficulty in operation of Frank-Read sources, the interaction of vacancies produced by fatigue creating strings of atoms in antiphase positions within ordered domains which trail out behind dislocation pairs and resist their continued motion, might be expected to lead to work hardening during fatigue. However, this is not evident from the X-ray results. The absence of appreciable work hardening may result from the absorption of the greater part of the plastic strain by the grain boundary layer, hence their early failure, or internal stresses may be relaxed by the formation of the large surface contours in a manner suggested by Wood and Segall, (58) (94) (164) (165).

Where cracks are initiated in severe slip damage, in the grains, at higher fatigue stresses, they are linked in the majority of cases, to cracks formed adjacent to the grain boundaries. This effect is demonstrated in Fig. 64. It is suggested that
these cracks are due to the stress concentrating effects of the cracks adjacent to the grain boundaries. It is also possible that partial disordering may have been caused by fatigue stressing and that this produced the normal fatigue behaviour of disordered or partially disordered body-centred cubic material. Alternatively, the higher fatigue stresses might have resulted in some cold working of the material which has led to crack formation via a Stroh (142) type dislocation pile-up or to the formation of a Mott (158) type kink.

The predominant slip planes, (123), indicated in the X-ray results are, according to Ardley and Cottrell (9), the slip planes operative at elevated temperatures in unidirectional strain. At elevated temperatures there is an increase in vacancy concentration or partial disordering or both. By analogy, it is also possible that fatigue stressing produces in beta brass, an increased vacancy concentration or partial disordering or both.

To summarise the results, two outstanding features are apparent;

(i) crack initiation and extrusions occur in intense slip markings at grain boundaries;

(ii) in general, the only fatigue damage observed
within the grains in polycrystalline and in single crystal specimens, is well developed slip (excluding propagating cracks, associated fatigue debris and rare extrusions).

These two features are attributed to the existence of a disordered grain boundary layer and the superlattice structure of beta brass; lack of cross slip within the ordered interior of grains, prolongs the fatigue life of the interior, the disordered outer layer fatigues normally and due to the nature of the deformation occurring within the superlattice this outside layer may be subjected to increased plastic strain. There is thus a competing process existing between the disordered grain boundary layer and the ordered interior which usually results in failure occurring within the grain boundary layer. Lack of cross slip inhibits intense slip markings and extrusions, occurring within grains.

5.2. **Discussion of the Fatigue Damage in Metastable Beta Brass**

It is clear that the low fatigue life of metastable beta brass is due to the early initiation of cracks at grain boundaries and their rapid propagation. These cracks are similar in some respects to those observed in fatigued stable beta brass and are possibly produced by a similar mechanism. They may also be due to brittleness or weakness of the
grain boundary. Bailey (30) and Perryman (29), have suggested that the grain boundaries in stable beta brass are inherently brittle. It is thought that alpha solid solution, which has not been retained in solution by quenching from the all beta phase field, has exaggerated this effect. However, it should be noticed that crack initiation was observed at grain boundaries where no alpha solid solution was detected by optical microscopy. This is demonstrated in Fig. 81.

Excluding grain boundary cracks the first stage in the fatigue process is considered to be the motion of dislocations in a few relatively widely spaced planes. The generation of vacancies by moving dislocations, as described by Seitz (116), is thought to cause local acceleration of diffusion and hence the precipitation shown in Figs. 85, 86, 87. Subsequent dislocation movement in these softened planes should be easier and should lead to a concentration of deformation and crack initiation before other regions suffer to any extent. However, crack initiation in planes containing fatigue induced precipitates has not been commonly observed. This may be explained by the catastrophic failure, due to cracks initiated at grain boundaries, limiting deformation elsewhere. This effect is demonstrated in Fig. 86.

It is also possible that the precipitated phase is a strain induced transformation product such
as 'massive alpha' or martensite as has been described by Garwood and Hull (43) and others (45) (46) (49).
The indefinite morphology of the precipitate, its face-centred cubic structure and cell dimensions suggest that it is probably alpha solid solution and not 'massive alpha' or martensite.

The effect of fatigue deformation on the dispersion of subsequent precipitation in aluminium alloys has been reported by Stubbington (178). He attributed the virtual removal of the long incubation time, observed in normal aged material, to the large number of nuclei present in fatigued material. Thomas and Whelan (179) have shown that dislocation loops formed by the collapse of vacancy discs also act as precipitation nuclei and recent observations have shown many such dislocation loops in fatigued metal.

It would appear, then, that the increased amount of precipitation observed in tempered metastable beta brass is due to an increase in nucleation sites. These might be dislocation loops formed by the collapse of vacancy discs produced by moving dislocations.
5. 3. Conclusion. Stable Beta Brass

1. According to Mott (96), cross slip is essential to the formation of intense slip bands and the production of extrusions. However, owing to the extended nature of dislocations in beta brass, cross slip is considered to be unlikely. This may explain the absence of intense slip markings, 'persistent markings' and the comparative rarity of extrusions, in the grains, of fatigued beta brass.

2. The absence of crack formation in slip bands is thought to be due to failure at or adjacent to grain boundaries limiting deformation elsewhere.

3. Extrusions are comparatively rare and where they do occur they are small. The absence of extrusions may be explained by the difficulty of cross slip, if Mott's (96), mechanism for their formation is accepted or by the difficulty of sequential slip movements on intersecting planes if the Cottrell-Hull (67), mechanism is accepted.

4. Fatigue debris is considered to be an attrition product brought about by fretting of the crack walls. It is thought to consist of oxidised fragments of the material.

5. The most important observation has been the initiation of fatigue cracks adjacent to grain boun-
daries in intense slip markings. These may be explained by the difficulty of intersecting slip in a superlattice causing high stresses to be developed at the boundaries, which are relaxed by fracture. This is not considered to be the most likely explanation as intercrystalline failure rather than cracks in the grains may be expected to result from such a mechanism. The existence of a disordered or partially disordered grain boundary layer is considered to provide a better explanation.

6. X-ray results indicate no appreciable hardening during fatigue. Predominant (123) slip planes and irrational slip planes observed may be explained in terms of an increase in vacancy concentration or partial disordering or both produced by fatiguing.

5.4. Conclusion. Metastable Beta Brass

1. Cracks initiated at grain boundaries are responsible for the relatively low fatigue life of the material. Rapid propagation of these cracks, in what appears to be a brittle manner, limits deformation elsewhere.

2. The fatigue induced precipitate observed is considered to be alpha solid solution. The precipitation is attributed to an acceleration of diffusion due to vacancies produced by the motion of dislocations under fatigue stresses.
3. The increased precipitation found in fatigued material, after tempering, is thought to be due to an increase in nucleating sites generated by fatigue, particularly in heavily damaged areas.
APPENDIX I
List of components as shown in Fig. 6, page 67.

Resistors

<table>
<thead>
<tr>
<th>Resistor</th>
<th>Value</th>
<th>Power</th>
</tr>
</thead>
<tbody>
<tr>
<td>R1</td>
<td>180kΩ</td>
<td>1/8W</td>
</tr>
<tr>
<td>R2</td>
<td>10kΩ</td>
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</tr>
<tr>
<td>R3</td>
<td>22kΩ</td>
<td>1/8W</td>
</tr>
<tr>
<td>R4</td>
<td>2.2kΩ</td>
<td>1/8W</td>
</tr>
<tr>
<td>R5</td>
<td>100 Ω</td>
<td>1/8W</td>
</tr>
<tr>
<td>R6</td>
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</tr>
<tr>
<td>R7</td>
<td>30kΩ</td>
<td>1/8W</td>
</tr>
<tr>
<td>R8</td>
<td>10kΩ</td>
<td>1/8W</td>
</tr>
<tr>
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<td>R16</td>
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<td>R27</td>
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<td>R28</td>
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Capacitors

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Capacitors

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Valves

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List of Components Continued

Sl-4  Switches

TI  Phototransistor (Mullard OCP 71).

II-3  Chokes – approximately 18H.
II  Tuned to frequency of vibration.

Output and other transformers - Mains 350-0-350.
Guide to Photomicrographs.

All photomicrographs shown are as electropolished and unetched, excluding the following:--

Fig. 15. Etched in 10% ammonium persulphate.
Fig. 16. " " " " "
Fig. 17. " " " " "
Fig. 86. " " " " "
Fig. 89. Etched in acid ferric chloride.
Fig. 90. " " " " "
Fig. 91. " " " " "
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Initiation of Fatigue Cracks in Beta-Brass

Fatigue cracks have been seen to be initiated in slip bands, twin boundaries and grain boundaries, in a number of investigations on the fatigue behaviour of metals.

During recent work on the fatigue behaviour of beta-brass a different mode of fatigue crack initiation has been observed, which has not been reported previously in the literature. Certainly, it does not occur in the case of fatigued copper.

Beta-brass was chosen because of the lack of knowledge of the fatigued behaviour of body-centred cubic metals. Polycrystalline material, with an average grain-size of 2 mm., has been fatigued in reverse plane bending at a frequency of 8,400 c./min. Specimens in the form of cantilevers were mechanically polished with care and finally electro-polished prior to fatiguing.

One of the main observations has been the initiation of fatigue cracks and the formation of extrusions in short-slip markings, which form adjacent to grain boundaries, and in the majority of cases at an angle.
to the main slip. These cracks are afterwards propagated into and across the grain boundaries and frequently become linked together.

The most unusual features of this phenomenon are that these cracks form at an angle to the main slip but not in the main slip, and that they form adjacent to the grain boundaries but not in the grain boundaries (Figs. 1, 2 and 3).
Further work leading to a model for the mechanism is in progress.

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M. G. Bader

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