FATIGUE OF COPPER - TUNGSTEN FIBRE COMPOSITES

THE PREPARATION AND FATIGUE PROPERTIES OF COPPER - TUNGSTEN FIBRE COMPOSITES

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SCOPE AND CONTENTS: The principles and theory of the tensile behaviour of fibre reinforced composites is reviewed, and a mechanism of fatigue failure is proposed. A technique for preparing copper - tungsten fibre composites was developed. Tensile tests showed the composites to behave normally, and the number of fibres breaking was shown to increase rapidly with stress up to the U.T.S. Fatigue below a small critical volume fraction of fibres was shown to be dominated by the matrix, which under repeated tension failed by cyclic creep. Above this volume fraction the fibres dominated the fatigue failure and cracks were observed to propagate through the matrix; a process not observed in tension. Matrix fatigue cracks were initiated around breaks in the fibres, and could propagate through other breaks, and along fibre - matrix interfaces and could circumnavigate unbroken fibres. The fatigue ratio of high fibre volume fraction samples was poor.

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SECTION 1

Introduction

The strongest materials known today are those in fibre form such as whiskers. Only in the last ten years or so has any effort been made to incorporate such fibres, which may have strengths in excess of 10^6 p.s.i., into matrices to form useable composites. Many very strong fibres are also brittle and therefore useless for most practical purposes, and this gives rise to the concept of associating the fibres with a soft ductile matrix to obtain intermediate properties. If the fibres are of high aspect ratio and are all aligned in the same direction, a very high strength composite can be produced provided there is an appreciable modulus difference between the two components. A stress applied in the axial direction of the fibres will then tend to produce a greater strain in the matrix than in the fibres. The matrix thus exerts a shear stress on the fibres, but since the surface area of the fibres is far greater than the cross - sectional area, a small shear stress gives rise to very high tensile stresses in the fibre. The fibres carry much the greater proportion of the load on the composite, and fail before the matrix under tensile conditions.

The manufacture of composites has been delayed in the past due to difficulties in maintaining the fibre strength during manufacture. These difficulties take various forms but are chiefly associated with interactions with the fibre surface so as to destroy its perfection and hence the strength.

A number of workers have now prepared composites with tensile strengths of several hundred thousand p.s.i.,(8,9,18,21,35.) and long fibres were found to be more effective than very short fibres. The theory of the tensile behaviour of composites is now reasonably well understood, and it is possible to predict the strengths and moduli of the whole spectrum of composites formed between two components, provided the tensile properties of the latter are known. It is apparent that from the point of view of tensile properties, fibre reinforced composites are capable of considerable increases in strength over conventional materials.

The amount of work so far carried out on properties of composites other than tensile is negligible. The field chosen for this work is fatigue, on which there have so far been three experimental papers and no mention of theories or mechanisms. Forsyth, George and Ryder (48) found that fibre meshes incorporated in aluminium sheet delayed fatigue crack propagation at high stresses but not at low stresses. Davis, McCarthy and Schurb (51) found that subjecting a fibre reinforced plastic to tension - compression fatigue caused failure at the fibre - matrix interfaces. Baker and Cratchley (49) subjected silica fibre reinforced aluminium composites to reversed bending, and found that they failed via weak interfaces and breaks in the fibres produced during composite manufacture. Sound fibres were capable of stopping a propagating crack.

The fatigue behaviour which can be expected of composites, therefore, appears to be that fibres can delay crack propagation except where they are broken, and propagation can occur down weak interfaces in tension - compression fatigue.

The purpose of the present work was to develop a manufacturing technique for the chosen composite, and then examine the fatigue behaviour under conditions of repeated tension. All the previous work on composites involved cycles with compressive components, and this introduces the problem of the effect of fibre buckling. The use of repeated tension eliminates the necessity of deducing the part played by buckling fibres in the fatigue process.

SECTION 2

Principles of Fibre Reinforcement

2.1 Strong Fibres

Many substances, when produced and tensile tested in fibres of various small diameters, show a marked strength - size effect as illustrated in Fig. 1. In some cases the effect is not real but is apparently due to production differences. An example of the strengths which can be achieved is 0.9% carbon steel, which can be drawn down to 0.006" diameter to yield a strength of 575,000 p.s.i. (1). It has been estimated that the theoretical fracture strength of solids is about E/10 (2,3), where E is Young's modulus. It has been known for some years that the closest approach to the theoretical strength is achieved by materials in fibre form. Tensile strengths of 1,000,000 p.s.i. and more are frequently encountered, as shown in Table 1.

Strong fibres may be divided into three categories:

1. Whiskers
2. Ceramic fibres
3. Metal fibres.

2.1.1. Whiskers

Whiskers are small single crystal filaments which exhibit the highest known tensile strengths of solids, and show a very marked strength - size effect. (4,5). Whilst their strengths make them the most attractive fibres for composite use, they have severe drawbacks. The first is that attempts at mass production have failed in all but a few cases, though it may only be a matter of time before this is solved. The second and more severe drawback is that a perfect surface must be maintained if high strength is to be retained. Many whiskers form steps on the surface or kink during growth, and these faults provide sufficient stress concentration to seriously reduce the strength. Much of the whisker harvest may have to be rejected due to growth faults. If a metal whisker is mechanically damaged during handling, dislocations will be introduced and a drastic loss of strength occurs. A further disadvantage of metal whiskers is that if incorporation in a metal matrix is attempted, an interaction with the matrix may destroy the necessary surface perfection. All the successful whisker composites so far manufactured are discontinuous and composed of non - metallic whiskers of widely varying dimensions and strengths.

2.1.2. Ceramic Fibres

In this group the glasses and silica are the most promising, being readily available and very easy to manufacture by drawing into continuous fibres. Silica is the strongest of the group, a typical strength of freshly drawn fibre being about 1,500,000 p.s.i. There is no strength - size effect in these materials, but the strength is extremely surface sensitive. A fresh - drawn silica fibre is very strong, but a touch of a finger causes catastrophic strength loss. Bending a 2mm diameter ordinary silica rod causes failure at a small strain, but a fresh - drawn rod of the same diameter can be bent to a remarkable 7.5% strain before failure. (6). If fresh drawn fibres are left in air, there is a considerable loss in strength after a matter of hours. Thus if such fibres are to be used in composites they must be protected immediately after manufacture. Coating of silica with aluminium has been successfully achieved (7) but it causes a drop in strength of about 60%. The coated fibres are not sensitive to handling and can readily be formed into composites.

A further characteristic of silica is that it is prone to static fatigue; subjecting it to a constant load causes eventual failure. There may be a relation between the strength degradation of silica by the atmosphere and static fatigue, but this has not been investigated.

2.1.3 Metal Fibres

These fibres are the weakest of the three types, though many metals can show strengths in excess of 300,000 p.s.i. (1). Most commercial products show a strength - size effect, but there is some doubt if this is real or due to different mechanical and thermal histories. Metal fibres have the advantage of being readily available in continuous lengths with very consistent properties. These fibres are not nearly as sensitive to surface conditions and mechanical damage as metal whiskers, but reaction with a metal matrix in composite production can degrade the strength (8).

2.2 Fibre Composites

It is apparent that fibre materials offer an opportunity for a very large increase in the strength of available materials. To achieve this end considerable numbers of fibres must be bonded together in a composite in such a way that their strength is utilised to the full. Maximum strength, as will be shown later, occurs when the fibres are unidirectional and continuous and the composite is pulled in the fibre axial direction. A number of workers have produced such composites with tensile strengths of several hundred thousand p.s.i. (8, 9, 10)

2.2.1 Choice of Matrix

The choice of matrix material is important, since it may serve all the following purposes:

1. To bind the fibres and in some cases protect them from surface damage which would destroy their strength.

2. Since many fibres are brittle, the matrix should be sufficiently ductile to prevent crack propagation through an entirely brittle path.

3. Transfer of stress to the fibres, for which purpose the modulus of the fibres should be much greater than that of the matrix. Alternatively the matrix must deform plastically at a stress much lower than the U.T.S. of the fibres. Thus at any given strain on the composite, the fibres will be carrying a much greater proportion of the load, as represented in Fig. 2. For any load on the composite the matrix will try to elongate further than the fibres by shearing past them. The value of this shear stress, however, is very small when considering fibres of high aspect ratio.

It is apparent that to form a practical composite the matrix must not destroy the fibre strength in any way, and must be ductile and of low modulus relative to the fibres.

2.2.2 Growth of Fibres in Situ.

The possibility of growing fibres within the matrix is attractive. There are often considerable difficulties attendant on composite manufacture, many of which involve detrimental effects of the matrix on the fibres. Growth in situ eliminates these problems, and may be done by use of eutectic systems. (11, 12, 13). One component will be a soft terminal solid solution which acts as a matrix, and the other component is usually an intermetallic compound. The composition employed may be the eutectic itself or some hypereutectic composition which will give an intermetallic precipitate in a eutectic matrix. Under conditions of unidirectional solidification, fully oriented needles or lamellae of intermetallic are precipitated, and these act as the reinforcing fibres. The value of V_f is fixed if the eutectic composition is used, but is otherwise variable.

2.2.3 Fibre Reinforcement and Conventional Strengthening

All the conventional techniques for strengthening a pure metal, such as cold work, alloying and dispersion hardening, depend on creating obstacles to dislocation movement. In dispersion hardening, for example, there is a fine and uniform distribution of incoherent particles which act as dislocation barriers. A dislocation meeting a particle requires appreciable energy to either cut through it or bend round it to leave a loop. (14, 15). The particle separation for effective dispersion strengthening must not be greater than 2 or 3 microns. (16).

The difference between conventional strengthening and fibre reinforcement is that in the latter case no attempt is made to interfere directly with dislocation motion. The plastic deformation of the matrix is used to transfer the load to the fibres. The fibre spacing is usually very large compared with dispersed particles, and may be greater than the average grain size. The fibre spacing does not affect the yield stress of the matrix but does control the subsequent rate of work hardening. (17).

SECTION 3

Theory of Fibre Reinforced Composites

3.1 Tension

3.1.1 Law of Mixtures

A number of experimental workers (9, 18 - 21) have verified an empirical relation, known as the law of mixtures, for several composite systems. The law gives the U.T.S. of a composite with perfectly aligned continuous fibres -

$$\sigma_c = \sigma_f V_f + \sigma_{M'} (1 - V_f) \qquad 3.1$$

σ_c = UTS of composite

σ_f = UTS of fibres

V_f = Volume fraction of fibres

$\sigma_{M'}$ = Stress in the matrix at the ultimate tensile strain of the fibres. (See Fig. 2)

The law has now been placed on a sound theoretical basis by Hashin and Rosen (23) and Hill (24). The relation between σ_c and V_f should be linear, ranging from the stress in the matrix $\sigma_{M'}$ to the U.T.S. of the fibres σ_f. There are other factors which must be considered, however, at the extreme values of V_f.

3.1.2 Limits of Reinforcement

At low values of V_f there is a point at which the U.T.S. of the work hardened matrix alone exceeds that of the composite. Below this value of V_f we have -

$$\sigma_f V_f + \sigma_{M'}(1 - V_f) < \sigma_M \qquad 3.2$$

Thus for fibre reinforcement to be effective V_f must exceed a critical value defined by the above relation so that -

$$V_{f\ CRIT} = \frac{\sigma_M - \sigma_{M'}}{\sigma_f - \sigma_{M'}} \qquad 3.3$$

If all the fibres in a composite, with a low V_f value, break in one transverse section, the composite fails unless the matrix is able to support the load. At all points at which the matrix is able to take the load we have -

$$\sigma_f V_f + \sigma_{M'}(1 - V_f) < \sigma_M(1 - V_f) \qquad 3.4$$

where σ_M is the U.T.S. of the matrix.

Solving for V_f we have -

$$V_{f\ MIN} = \frac{\sigma_M - \sigma_{M'}}{\sigma_f + (\sigma_M - \sigma_{M'})} \qquad 3.5$$

Since $V_{f\ CRIT}$ is always greater than $V_{f\ MIN}$, it is the former which must be exceeded for effective reinforcement. The relation between V_f, $V_{f\ CRIT}$, $V_{f\ MIN}$ and σ_c is shown in Fig. 3. The relationship between σ_f and $V_{f\ CRIT}$, using σ_M and $\sigma_{M'}$ values taken from the copper used in this work, is shown in Fig.4.

It is apparent that the required reinforcement increases rapidly as the fibres used become weaker.

Kelly and Davies (16) have suggested that at high values of V_f the fibre spacing will become so close that dislocation motion will be interfered with, so that the work hardening rate of the matrix is increased. This will effectively increase $\sigma_{M'}$ to some value $\sigma_{M''}$, which must accordingly be used in the law of mixtures. There will be a transition between the σ_f - V_f plot based on $\sigma_{M'}$, and that based on $\sigma_{M''}$, occuring when the inter - fibre spacing is reduced below about 10µ. This is shown schematically in Fig. 3.

If fibres of circular cross - section are arranged in close packing, a theoretical maximum V_f value of 90.6 % can be obtained. In practice, however, (10, 22) it is found that contact between fibres is deleterious. Efficient load transfer between the fibres is impossible without a continuous matrix, and to achieve this in practice reduces the maximum V_f value to about 80 %. A composite containing brittle fibres put into service where impact resistance is required may have a maximum V_f of only 50 %.

3.1.3. Modified Law of Mixtures

A more accurate modification of the law of mixtures is often proposed -

$$\sigma_c = \alpha\theta\ \sigma_f V_f + \sigma_{M'} (1 - V_f) \qquad 3.6$$

θ = orientation factor

α = strength efficiency factor

The orientation factor (22) varies from unity for perfectly aligned fibres to 0.18 for completely random fibres. It is interesting to note the theoretical predictions of Stowell and Liu, and Kelly and Davies (26, 16), for a composite with unidirectional continuous fibres pulled at various small angles θ to the fibre axis. There are two possible modes of failure. The first is the flow of the matrix parallel to the fibres until the fibres fail, and σ_c is then given by -

$$\sigma_c = \sigma \cos^2 \theta \qquad 3.7$$

σ_c = U.T.S. of the composite when $\theta = 0$

σ = applied stress

θ = angle between fibres and axis of tension.

Alternatively the matrix can fail in shear on a plane parallel to the fibres, and the shear stress to produce failure is then -

$$\tau = \sigma \sin \theta \cos \theta \qquad 3.8$$

Plotting the applied stress σ against θ yields Fig. 5. It can be seen that the first few degrees of misalignment actually cause a slight increase in the U.T.S. of the composite, but thereafter the alternative mode of failure dominates and the U.T.S. falls rapidly.

Since Brenner (25) has stated that random misalignments up to 8^{o} are not severely deleterious, it can be expected that some tolerance is allowable for misalignment in production.

The strength efficiency factor, α, was first proposed by Coleman (27). This factor accounts for the strength of a bundle of fibres being less than the mean strength of the fibres. The coefficient of variation of the fibre strength is given by $\frac{S}{\bar{\sigma}_f}$, where $\bar{\sigma}_f$ is the average U.T.S. of the fibres and S is the standard deviation from the mean. α varies from one when $\frac{S}{\bar{\sigma}_f} = 0$ to 0.5 when $\frac{S}{\bar{\sigma}_f} = 0.5$

It is doubtful if Coleman's analysis is applicable to composites, though α is often included in papers. Coleman specified a tensile test in which all the fibres act independently and without inter - fibre friction. The part played by a fibre ended when it had broken. In the case of a composite with a ductile matrix, the failure of a fibre causes immediate redistribution of the load. If the two broken pieces are sufficiently long, they will continue to provide effective reinforcement. Evidence of this is given in the present work, in which individual fibres were observed to break several times. Further evidence is that McDanels (9) found that the mean strength of the tungsten fibres in his copper - tungsten composites was, in fact, attained.

In this work the value of θ was taken to be one, since only continuous fibre composites were used.

The geometry of the mould in which the fibres were placed is such that the maximum angle a single fibre, leaning against the walls, could make was 7° from the vertical. It is assumed that the fibres within a bundle in the mould would have misalignments much less than 7°, and that the maximum strength would therefore be achieved. On the basis of the arguments given above the strength efficiency factor was assigned a value of one.

3.1.4. Stress Distributions and Critical Fibre Length

If we consider one long straight fibre completely embedded in the matrix of a composite under load, the stress along the majority of its length will be constant at some value near the mean fibre stress. It may be anticipated, however, that this uniform state of stress will not apply to the fibre ends. The stress distribution around the fibre ends is of considerable importance in discontinuous composites, and even for continuous composites, since ends are produced by the breaking of the fibres.

There have been two theoretical estimates of the stress distribution, one by Cox (28) and the other by Dow (29). They independently obtained identical expressions, apart from a slight difference in the value of a constant involving elastic moduli. Neither considered the effect of the shape of the fibre end.

The model used by Dow, shown in Fig. 6., involved a single alumina whisker embedded in an aluminium matrix, the stress - strain curve of the latter being represented by four straight lines.

The U.T.S. of the fibre was taken to be 600,000 p.s.i., together with a modulus of 60,000 k.s.i. The effect of matrix across the end of the fibre was ignored. The stress distribution was calculated for two cases, one in which the matrix was elastic and the other with a plastic matrix. The results are shown in Fig. 7. If the matrix remains elastic, the rise in the fibre stress σ_E is slow since the overall matrix strain is too small to induce a high stress in the fibre. If the matrix flows plastically, then the full tensile load is transfered to the fibre over a very short distance from the end; in this case about 20 fibre diameters.

The shear stresses have only a very small value along most of the interface, but rise very sharply at the fibre ends. Dow has commented that the attainment of very high composite strengths may well depend on the ability to resist the severe shear stresses at the ends.

The complete stress distribution around a long fibre is shown in Fig. 8. The state represented is that where the fibre is loaded slightly below its U.T.S., σ_f. The length of fibre , from the end, required to achieve the full tensile stress σ_f is equal to $\frac{lc}{2}$ and is called the transfer length. The shortest fibre in which σ_f can be achieved and the fibre therefore broken has a length of lc, the critical length. If the fibre is shorter than lc, then σ_f cannot be achieved and the matrix shears past the fibre.

We can equate the tensile load in the fibre at any length x from one end, with the interfacial shear load -

$$\frac{\sigma \pi d^2}{4} = \pi d x \tau \qquad 3.9$$

σ = tensile stress in fibre at distance x from end

d = fibre diameter

τ = interfacial shear stress at x

If we take $\sigma = \sigma_f$ so that $x = \frac{lc}{2}$ we can define a critical aspect ratio for the system -

$$\frac{lc}{d} = \frac{\sigma_f}{2\tau} \qquad 3.10$$

For the system used by Dow, the critical aspect ratio is about 36, but this is quite small when compared with the usual aspect ratios of continuous composites which will be several hundred or more.

If the composite has a work hardening matrix, the value of τ is not constant but depends on the plastic strain in the matrix. Kelly and Tyson (18) determined $\frac{lc}{d}$ experimentally and then substituted this value in 3.10, which gave the value of τ as the ultimate shear strength of the matrix. The distribution of strain in the matrix according to Kelly and Tyson is shown in Fig. 9.

3.1.5 Tensile Test Behaviour

A certain amount of tensile testing was carried out in the present work, and all the fatigue tests were carried out in repeated tension, the lower tension limit being zero. Consequently it is of interest to examine the tensile behaviour of composites. The example chosen is identical to the system used in this work, this being strong ductile fibres in a soft ductile matrix. It is assumed that the fibres will have identical yield points and U.T.S. values.

The stress - strain curve for such a material is shown in Fig. 10. Now σ_1 is the stress on the composite at which the matrix yields, and up to this point both components will be elastic and the modulus of elasticity is -

$$E_{c_1} = E_f A_f + E_M (1 - A_f) \qquad 3.11$$

E_{c_1} = Primary modulus of elasticity of the composite

E_f = Young's modulus of fibres

E_M = Young's modulus of matrix

A_f = Area fraction of fibres.

At σ_1 the matrix yields and the plot curves to σ_2. The fibres are still elastic and the modulus at any given point between σ_1 and σ_2 is -

$$E_{c_2} = E_f A_f + \left\{ \frac{\sigma_{M'}}{\epsilon} \right\} (1 - A_f) \qquad 3.12$$

$\sigma_{M'}$ = stress in matrix which corresponds to a composite strain ϵ, where $\epsilon_1 < \epsilon < \epsilon_2$

E_{c_2} = Secondary modulus of elasticity of the composite.

At σ_2 all the fibres will yield, giving rise to a sharp change in slope. All the identical fibres will fail at the breaking strain and catastrophic failure will follow. Real composites may have appreciable fibre strength distributions, in which case σ_2 will be ill - defined.

Arridge (30) has developed a theoretical relation which predicts the type of tensile failure in a continuous fibre composite -

$$x\, f\,(x) = \frac{L}{lc \left\{ 1 - \dfrac{C\, E_f}{3\, E_c} \right\}} \qquad 3.13$$

x = load / fibre

$f\,(x)$ = probability density function for breaking loads

C = density of fibres / unit area

lc = critical length

L = length of composite

E_f = modulus of elasticity of fibres

E_c = modulus of elasticity of composite.

Arridge did not specify whether the primary or secondary modulus should be used. The range of the secondary modulus is usually much greater than that of the primary modulus, so the secondary is probably appropriate here. It is assumed that when a fibre breaks there is equipartition of the load amongst the remainder, this being valid for strong fibres in a ductile matrix such as tungsten in copper. It is also assumed that there are no serious flaws in the fibres.

If $x\,f(x) < 1$, then a serial failure of the fibres occurs. If $x\,f(x) > 1$ then the failure is catastrophic. The latter type of failure occurs as might be expected, in fibres with small standard deviations from the mean strength.

3.2 Load Cycling Behaviour of Composites

There are no theories of the cyclic loading of composites in the literature. The model to be considered is the same as that used in describing the tensile behaviour, Section 3.1.5., and the same stress - strain curve may be used. All the cycles range from zero at the lower limit to some tensile stress σ, and the fatigue behaviour is to be examined for different σ values.

3.2.1 Cycling with $0 < \sigma < \sigma_1$.

The value of σ_1 is usually quite small compared to σ_2, and no tests in the present work were carried out in this range. At stresses $\ll \sigma_1$, the fatigue lives observed will be indefinitely long, and even just below σ_1 they will be very long.

The fibres have a much higher U.T.S. than the matrix, so that it can be expected that for any given stress the matrix will probably be the first to fail. The matrix material in this work was copper, and for pure copper and copper matrix composites with $V_f < V_{f\ CRIT}$ the presence of a mean tensile stress in fatigue means that failure will be induced by cyclic creep (31, 32). This stress - induced type of creep produces failure much more rapidly than static creep, operating at the same stress as the peak of the cycle and the same temperature. In composites with $V_f > V_{f\ CRIT}$ the effect of the fibres will be to retard such creep, so that the life of the specimens will be increased. Eventually a matrix crack will form, and the effect of the fibres on a propagating matrix crack was one of the problems studied in this work. Crack propagation and final failure are discussed later, in this and other chapters.

3.2.2. Cycling with $\sigma_1 < \sigma < \sigma_2$

The first loading cycle will produce a stress - strain curve showing the yield of the matrix at σ_1, as described for tensile testing. Unloading from σ will cause elastic relaxation of both matrix and fibres. Now consideration of Fig. 2. will show that a relatively small strain relaxation completely unloads the matrix but leaves a considerable tensile stress on the fibres. Further unloading causes the fibres to elastically compress the matrix, followed by yielding and plastic flow of the matrix until the net load on the composite is zero. At this point there is a balance between the tensile load on the fibres and the compressive load on the matrix. The unloading curve will show a discontinuity corresponding to the yield in compression of the matrix, as shown in Fig. 11. There will be a permanent set due to matrix deformation.

On reloading for the second cycle, the elastic range will be greater and there will be a further increment in strain which will again not be fully recovered on unloading by plastic compression of the matrix. Further cycling will cause progressively smaller strain increments until the material saturates.

The stress range over which the fibres cycle decreases during the initial cycling period. The peak fibre stress decreases as the work hardening matrix takes more of the load. The lowest fibre stress increases as the matrix develops increasing resistance to plastic compression, leaving increasing residual tensile stresses in the unloaded composite. Conversely the matrix stress range increases, the peak stress rising with work hardening and the lowest stress decreasing into compression. Fig. 12. Thus the matrix is subjected to a tension - compression cycle, but with a mean tensile stress.

The tendency of the matrix to undergo cyclic creep under these stress conditions will be retarded or suppressed by the fibres. Failure will probably commence by a typical low strain fatigue crack in the matrix, rather than a high strain ductile cyclic creep failure.

3.2.3 Cycling with $\sigma > \sigma_2$

The initial loading curve of a composite subjected to $\sigma > \sigma_2$ will be divisible into three distinct parts as described for tensile testing. This means that the yielding of the fibres at σ_2 must be considered. On unloading from σ the curve will only have the two parts mentioned previously. These are elastic relaxation of both components and plastic compression of the matrix.

On reloading in the second cycle, both σ_1 and σ_2 will be raised and in subsequent cycles may become indistinguishable. Both the matrix and the fibres will now be subjected to stresses which are large fractions of their U.T.S. values. In the stress range $\sigma_1 < \sigma < \sigma_2$ the matrix would be subjected to appreciable plastic strains and may initiate fatigue cracks before the fibres do so. The situation is then that of a matrix crack propagating amongst unbroken fibres, and this may not be a serious condition until fibres start to fail by fatigue. When $\sigma > \sigma_2$ the failure of fibres will be quicker, and it can be anticipated that a number of fibres failing in one transverse plane will accelerate failure.

3.2.4 Fatigue Failure of Composites

Consider a fatigue crack propagating through the matrix. In alloys where strength is conferred by particles, there is a continuous path through the matrix which the crack may be able to follow.

In continuous fibre composites, however, the crack soon meets a relatively large fibre, which extends the length of the specimen. It is not likely that the arrival of a small transverse crack at the fibre surface will have any appreciable effect on the overall fibre stress, nor will the crack be able to cut through the fibre. The only choices for the crack are either to go round the fibre in some way or to turn parallel to it.

If the crack circumnavigates the fibre, the facility with which it does so may be affected by the crack front meeting other fibres elsewhere. The crack will then have to bow out between the fibres in a manner rather like a line dislocation being forced against an array of particles. A high density of fibres may be a more effective crack arrester than a low density. The alternative of the crack turning parallel to the fibre becomes more likely the weaker the fibre - matrix interface. A further factor which might favour this behaviour is the crack propagation mechanism observed at high stresses by Laird and Smith (33). They observed that each cycle produced a pair of 'ears' at the crack tip, these being at 60° on either side of the main crack direction. The explanation put forward was that the maximum tensile stresses occur at ± 60° to the plane of the crack. Thus at high stresses applied to composites with weak interfaces there is a very real possibility of turning a crack parallel to the fibre. When the crack has been turned, the forces tending to open up the crack tip will be reduced since it is now parallel to the applied stress, and this mechanism may be an effective crack stopper.

Whatever behaviour the crack shows, it can be seen that the crack will be delayed, if not arrested, by the presence of fibres. The fatigue life of a continuous fibre composite can be expected to be much better than the matrix alone. This will be partly due to the considerable increase in tensile strength and partly because of the crack arresting ability of the fibres.

In the higher stress ranges the effect of the fibres themselves failing by fatigue must be considered. If a fibre fails, the two ends are sources of stress concentration as described earlier. If there is a crack propagating transversely through the matrix, it will be attracted by the stress concentration and will then use the break as a gateway. Unless many fibres fail by fatigue in the same transverse plane, however, the matrix crack will still be held up. It is possible that a matrix crack could spread over an appreciable area, but final failure of the composite under tension - tension conditions must await the failure of the fibres. When an appreciable number of fibres in one transverse plane have failed, the remaining fibres and matrix area will fail by tension overload.

SECTION 4

Previous Experimental Observations on Composites

4.1. Tensile Behaviour

4.1.1. Critical Length and Aspect Ratio

An individual fibre in a continuous fibre composite can undergo more than one break, the weakest point breaking first and then the next weakest and so on. The theoretical limit to this is when all the pieces are $<$ lc for the particular system, in which event they become less effective and the matrix shears past them. A more likely event is that sufficient breaks will accumulate in one transverse plane in the composite for it to fail there, before all the fibres are reduced to sub - critical pieces. Thus for a composite with n fibres of length L, the number of breaks N will be in the range $n < N < \frac{nL}{lc}$

The critical aspect ratio of the Cu - W fibre system was calculated by McDanels et al. (9) to be 16. The minimum aspect ratio used was 75, this being found to be adequate. Kelly and Tyson (15) calculated that the critical aspect ratio for their Cu - W fibre system should be 4 if the matrix is assumed to behave elastically. Subsequent experiments showed that the actual value was between 2 and 4. The value of $\frac{lc}{d}$ for metal matrix composites is much smaller than that for plastic or resin matrix composites, where $\frac{lc}{d}$ can be as much as 1000. The critical aspect ratio can be determined by tests in which the ends of a series of fibres are buried in the matrix at known depths and then pulled.

Dow (26) calculated that stress transfer to the fibres occurs much more rapidly when the matrix is considered to be plastic than when it is elastic. Sutton and Chorné (34) came to the reverse conclusion, but it seems unlikely that stress transfer in an elastic matrix at a given strain, will be more efficient than that in a plastic matrix at the same strain but lower stress. Support for this view is that Sutton and Chorné calculated an $\frac{lc}{d}$ value for the Cu - W system which is twice the experimental value of Kelly and Tyson.

Cratchley (16, 32, 39) showed that the critical aspect ratio in the aluminium - stainless steel fibre system was apparently very large. Aspect ratios of 50 - 400 were used, and specimens with a ratio of 400 were stronger than those at 200. This result was then checked, using specimens with fibre aspect ratio of 50 which were very carefully prepared. It was found that the improved alignment gave strengths approaching the theoretical maximum and much higher than in the earlier experiments. These results serve to emphasise the importance of orientation in certain systems, and in this case the value of θ in equation 3.6 will decrease rapidly with increasing misalignments.

4.1.2 Work Hardening of Composites

The behaviour of composites at both large and small strains was examined by Koppenaal and Parikh (17). They used three different types of metal fibre, which in order of increasing strength were mild steel, molybdenum and stainless steel. All the composites were made with a silver matrix.

Specimens containing different fibres but with equal V_f values were strained 4% and then microexamined. All the mild steel fibres showed slip markings, 5 - 10 % of the molybdenum fibres had slip markings and there were none at all in the stainless steel fibres. Koppenaal (35) showed that the slip in the silver activated the slip in the mild steel fibres. There was a continuity of the silver slip lines into the mild steel, but at a different orientation. In this type of composite it is quite possible that a crack could propagate through the fibres.

In the microstrain region the yield point of the silver was not altered by the presence of fibres. The rate of work hardening after yield, however, showed a marked dependence on V_f. Brown and Lukens (36) showed that in the microstrain region the square root of the plastic strain is directly proportional to the stress for polycrystalline metals. This is also true for these composites, as shown in Fig. 13. Thus the fibres act as slip obstacles, and dislocations pile up at the fibre - matrix interfaces. The available slip length decreases with increase in V_f and larger pile - ups are produced for a given strain. The rate of work hardening increases since a greater stress is required to drive a dislocation into a pile - up. The effect is analogous to that of grain size in polycrystals. Koppenaal and Parikh substituted inter - fibre spacing for grain size in the Petch (37) relation to give -

$$\sigma_Y = \sigma_o + k\,d^{-\frac{1}{2}} \qquad 4.1$$

σ_Y = Yield strength of matrix of composite

σ_o = Grain size independent portion of σ_y

k = Constant

d = Inter - fibre spacing.

Parikh (38) has shown this relation to be valid for the 0.2 % yield strength of silver matrix composites containing mild steel, molybdenum or stainless steel fibres.

McDanels et al. (9) found that the stress - strain curve of their Cu - W fibre system could be split into the three parts described in 3.1.5. The only difference was that there was no sharp yield associated with the fibres, but a smooth transition from elastic to plastic behaviour. This will be due to a distribution in yield points amongst the fibres.

There was a linear relation between both primary and secondary moduli and V_f. The results also showed that the mean fibre strength was achieved for this system, so that the value of a in equation 3.6 is 1.

4.1.3 Tensile Failure

Piehler (21) manufactured silver matrix composites containing either 7 or 19 mild steel fibres, and made so that the fibre distribution was very precise. The 7 - fibre composite, for example, had one central fibre and the remaining 6 in an accurate hexagon around it.

On carrying out tensile tests on fibres extracted from a composite, it was found that the average strain to failure was about 4.5 %. Tensile tests on the composites, however, showed strains of about 7 % before the fibres broke and failure occured. Pichler attributed this to the matrix, which resisted the tendency of the fibres to neck and prolonged the duration of uniform plastic deformation. Eventually a neck formed in the whole composite prior to failure. Examination of necked specimens showed that each fibre had necked within the large neck of the specimen. The outer fibres had not only necked but bent inwards to conform to the neck of the composite. All the samples were stronger than the law of mixtures prediction.

McDanels et al. (9) carried out tensile tests on copper - tungsten fibre composites. They found that failure took place by the following sequence -

1. Random distribution of breaks in the fibres.
2. A localised group of breaks occurs in one transverse plane.
3. The remaining fibres in this plane fail.
4. The matrix fails.

The failure mode was of the serial type predicted by Arridge (30).

4.2 Stress Distribution Around Fibres.

Tyson and Davies (40) examined the stress distribution around a square ended duralumin fibre, in an Araldite epoxy resin matrix, by photoelastic techniques. The results were expressed as the shear stress τ , divided by the applied tensile stress σ, plotted against distance from the fibre end. The experimental values are compared with the theoretical values of Dow and Cox, the latter results being expressed as a single lin since in this case they are closely similar.

Fig. 14. shows that the experimental values correspond closely to theory until the fibre end is approached, where the true stresses are much higher than predicted.

The radial decay of the maximum matrix shear stress was measured one fibre diameter away from the fibre end. The value of $\frac{\tau_{MAX}}{\sigma}$ curved down from 2 at the interface to 0.5 at about one fibre diameter away from the interface.

A further experiment was to estimate the effect of adhesion over the fibre end, a factor which was ignored by Dow in his analysis. The results in Fig. 15. show that without bonding $\frac{\tau}{\sigma}$ declines more slowly down the fibre than is the case with bonding. The maximum value of $\frac{\tau}{\sigma}$, at the fibre end, was about the same in both cases.

Schuster and Scala (41) embedded single alumina whiskers with different end geometries into the centre of the gauge length of a tensile specimen made from a birefringent plastic. The specimen was then pulled on the stage of a polarising microscope, and the stress distributions were calculated by photoelastic methods.

The maximum shear stress around a fibre with a square end was observed at about half a fibre diameter from the fibre end. The value was higher than the Dow prediction but much closer to it than with the results of Tyson and Davies. The value of the shear stress declined radially in similar fashion to the Araldite - duralumin fibre system.

The effect of loading until the whisker broke was to produce a complex stress system at the break, which was not analysed but was a more severely stressed condition than that at the fibre ends. This result can be related to that of Tyson and Davies, in which adhesion across the fibre end was observed to produce lower stresses than a non - bonded end. The significance of these results must be considered with regard to the effect of broken fibres on a propagating matrix crack.

Stress analyses were carried out using whiskers with pointed and finely tapered ends. It was found that the finer the point, the less the stress concentration associated with it. Square ended whiskers gave the worst observed stress concentrations.

There was no appreciable stress concentration associated with whiskers which were oriented at right angles to the applied stress. If this is also true for other systems it may be possible to produce isotropic composites without much loss in strength.

4.3. Effect of Alloying the Matrix

Petrasek and Weeton (8) used the Cu - W fibre system for their work on matrix alloying. The effect of such alloying on the properties of composites is of great interest, since if successful it raises the possibility of tailoring the matrix properties to suit specific requirements in service.

The matrix copper was alloyed with various amounts of aluminium, chromium, niobium, nickel, titanium and zirconium. The strength of the

13 alloys tested was below that of the pure Cu - W composite in all but two cases, and the decreases ranged from mild to severe. The drop in strength usually increased with increase in alloying addition. All the additions changed the interface conditions. Pure copper and tungsten are mutually insoluble, but alloying produced one of three effects -

1. Diffusion penetration accompanied by recrystallisation of the grains at the periphery of the fibre.

2. Formation of a two - phase zone at the interface.

3. Formation of a solid solution without recrystallisation at the interface.

The first effect caused the most severe losses in strength, and was associated with F.C.C. additions. The second and third effects were not serious and were due to B.C.C. additions.

The effect of alloying for this system can be summed up by saying that any modification in matrix properties by these elements will be achieved only with some loss in strength. None of the chosen alloying elements was insoluble in tungsten. The effect of the insoluble elements gold, silver and zinc could be of interest; zinc with regard to strengthening and gold with regard to the effect of ordering on the fatigue of Cu - Au.

Petrasek and Weeton found that the following expression gave the drop in strength on alloying -

$$\Delta \sigma_c = K A_f P (2 d - P) \qquad 4.2$$

$\Delta \sigma_c$ = Difference in strength between composites with alloyed and unalloyed matrices.

K = Constant for particular alloy system.

A_f = Area fraction of fibres.

$\frac{p}{2}$ = Depth of penetration or recrystallisation.

d = Initial fibre diameter.

4.4. Fatigue Behaviour

Forsyth et al. (48) put steel wire mesh in volume fractions up to 10% into aluminium alloy sheet. The aim was not to achieve a higher U.T.S. but to examine the effect of the meshes on fatigue crack propagation. Slots were machined in the samples to provide a source of stress concentration, and crack propagation tests were made. The meshes caused a considerable increase in fatigue life. Under low stress conditions in which the sample spent most of its life initiating a crack, the improvement in total life was small. The reinforced samples could withstand the effect of a longer crack before becoming unstable. The factors which affected the rate of crack propagation were fibre orientation, strength and volume fraction.

Davis, McCarthy and Schurb (51) carried out fatigue tests on fibre reinforced plastics, and concluded that the fatigue resistance depends on fibre orientation, the bonding resin and design. Their results agreed with those of Forsyth et al. in that fibres aligned parallel to the axis of tension did not give the best results. An angle between fibres and axis of tension of 5^{o} was better than both 0^{o} and 10^{o}. Specimens with 0^{o} orientation split in the fibre irection. Splitting was overcome either by using a 5^{o}orientation or using 85% of the fibres at 0° and 15% of them at 90°. The effect of fibre buckling in compression may have aggravated the splitting problem, though this was not commented upon in the paper.

Baker and Cratchley (49) fatigued silica fibre reinforced aluminium in reversed bending. The silica fibres were not affected by cyclic loading but were broken in some places during the manufacturing process. Deformation zones ran out from the fibre breaks and developed into cracks, but these cracks could not propagate through an unbroken fibre. The specimens with strong interfacial bonding showed a jagged fracture made up of transverse and axial components. The latter ran for several fibre diameters in the axial direction. Specimens with weak interfacial bonding failed by splitting open at these interfaces. No quantitative data were given and the effect of the compression half cycle in fibre buckling was not commented upon.

Salkind (56) has fatigued unidirectionally solidified eutectics. The lamellar $CuAl_2$-Al eutectic showed superior fatigue properties to the whisker eutectic Ni_3Al-Al.

SECTION 5

Experimental Procedure

5.1 Choice of System

The requirements of the components of a composite system have been examined in previous chapters. The choices of fibre are whiskers, silica, metal wires or fibres grown in situ.

Whiskers were rejected due to the difficulties in manufacture and quality control. Manufacturing problems can be avoided by the purchase of alumina whiskers, which are obtainable in reasonable quantities, but the cost is extremely high and the problems of incorporation into a matrix remain.

Silica was rejected due to protection problems. The strength can only be maintained if the fresh - drawn surface is protected immediately, and this has been successfully achieved only in the case of aluminium (7). The resulting loss in strength makes the coated fibres little better than the best of the metal fibres, and with a poorer elastic modulus.

The growth of fibres in situ is very attractive since it eliminates all the considerable difficulties attendant on incorporation of fibres into a matrix. The technique has certain drawbacks, however, as found by Davies (42, 43). Fibres extracted from such composites were found to exhibit whisker strengths if the surfaces were perfect, but a considerable proportion had growth steps on the surface which drastically reduced the strength. There is little control that can be exercised either over such steps or over the fibre aspect ratio. A lesser problem in some cases is gravity segregation of the fibres to top or bottom of the melt.

Metal fibres can be obtained cheaply in long continuous quantities, with good surface condition and very consistent cross - section and mechanical properties. These qualities alone make them attractive for fatigue purposes, since scatter in results is usually considerable, so that it is best to eliminate as many variables as possible.

The copper - tungsten fibre system was chosen for this work, since the tensile properties have already been investigated (9). Copper wets tungsten, but they are mutually insoluble so a good bond should be achieved without introducing the variable of alloying effects. Tungsten has a much lower strength than whiskers or silica, but the problems of handling and composite manufacture are not nearly as formidable.

The copper used was of 99.99 % purity, and the tungsten wire was type 218 of the General Electric Company, and was 0.010" in diameter.

5.2 Composite Manufacture

5.2.1 Materials Preparation

Two sizes of specimen were required, one for the Instron tensile machine grips and a larger size for the Baldwin fatigue machine. The large specimens were 2.5" long and 0.266" in diameter. They were necessary since the minimum load the fatigue machine could apply was 200 lbs. The manufacturing procedure was the same for both types. The preparation procedure for the small specimens, which were 1.5" long and 0.187" in diameter, is described below.

The tungsten wire was cut into lengths of about one foot, and a known number of them packed into one end of an 18" long, 0.5" diameter copper tube. The tube was then swaged down to 0.25" diameter, a process which did not deform the wires but pressed them into a more closely packed arrangement. The resulting rod of copper and tungsten was then machine sawed into 1.5" lengths. This procedure saves a considerable amount of hand - cutting of wires and gives bundles of exactly the same length. Machine sawing was necessary since hand sawing of a bundle of tungsten fibres is extremely slow. The specimens at this stage are contaminated due to swage lubricant, sawing and handling, so the copper tube was dissolved away in nitric acid. The fibre bundles, which were not attacked during this treatment, were then separated into loose piles and cleaned in alcohol and acetone. They were then ready for placing in the graphite mould.

The copper was obtained either as rod or shot, and was cleaned in nitric acid, alcohol and acetone before placing in the mould.

The moulds were machined from 2" diameter graphite rod, and two designs were developed. The large specimens were made in a mould with a single central hole the size of the specimen. The mould had a machined shoulder at the top, onto which a hollow graphite cylinder was pressed. The fibres were placed into the central hole and the cylinder pressed on. The copper was dropped into the cylindrical cap. After the specimen preparation the cap was removed to reveal a solid cylinder of copper, which was gripped and pulled to remove the attached specimen. When the technique of manufacture had been developed, a similar mould was used to produce seven specimens simultaneously. Six holes were drilled in a hexagon around the central one. Some difficulty was experienced with specimen removal from this type.

The small specimens were made up in a split mould shown in Fig.16. This design made specimen removal very easy. The specimens were separated from the solidified block of melt by hand sawing. This was facilitated considerably if the fibres hadbeen pressed down into the mould, so that the top part of the specimen adjacent to the block was fibre free copper.

All the freshly machined moulds were assembled and placed in the apparatus, described later, and heated to about 1300°C in a nitrogen atmosphere. This had the effect of driving off a considerable amount of moisture and volatile matter. This procedure was repeated with used moulds if they had not been in use for some time. After this preheating treatment, the moulds were ready for use.

5.2.2 Manufacturing Technique

A considerable time was spent developing a good manufacturing technique for the composites. Melting was first attempted in an evacuated and sealed silica tube placed in a vertical tube furnace, but this frequently caused failure of the silica. A graphite mould was then tried, this being placed in a silica tube clamped in the coil of a Tocco high frequency generator. Nitrogen was passed through the tube before, during and after the heating cycle. The specimens produced showed gross porosity and the fibres were frequently oxidised. In an attempt to eliminate these problems, the copper was charged in such a way as to leave the fibres open to the atmosphere and permit complete flushing with nitrogen. A 0.025" hole was drilled in the bottom of the mould to assist gas flow and escape from the descending copper. At the end of each run the input power was reduced and the mould lowered gradually out of the coil, this method having previously been successful in reducing porosity (57). These modifications eliminated oxidation and reduced porosity. A backing pump and manometer were introduced, and the column was evacuated to 3 mm of mercury after flushing and melting down under nitrogen. Evacuation caused considerable outgassing which lasted for five minutes or so, after which the melt was directionally solidified. Porosity was again reduced so that pure copper and specimens with low fibre content could be satisfactorily made. Specimens with high V_f values still contained some pores.

The final modifications included the construction of Perspex end pieces to the tube, between which O - rings were clamped. A better backing pump reduced the vacuum to 0.2 mm. When using vacuum the hole underneath the mould merely acted as a leak source, so this was eliminated. A diagram of the column and ancillary equipment is shown in Fig.17 , and Fig. 18 . is a photograph of the whole assembly. The specimens of high V_f now produced were usually pore - free, though some porosity was observed in a few specimens.

The finally developed procedure, after preparation and assembly of the materials and the mould, was as follows -

1. Column sealed and evacuated

2. Previously warmed - up pyrometer assembly calibrated and aligned optically.

3. Generator switched on and mould heated slowly to the melting point of copper.

4. Pyrometer reading taken at the instant the molten copper appeared in view.

5. Melt held at 1100 - 1200°c for one hour at 0.2 mm mercury pressure.

6. Generator input power reduced and mould lowered out of coil for directional solidification.

Temperature control was effected with a Lactronics Coloratio two - colour pyrometer, sighted through a Bausch and Lomb total reflecting 45° prism set on top of the column. Errors due to stray light were reduced by enclosing the light path between pyrometer and prism with a cardboard tube, and surrounding the prism with dull black paper. This did not give black body conditions since the light path also ran down the silica tube, but since the pyrometer was calibrated by the melting of pure copper and only used within 100°C of this reading, the technique was adequate.

The Perspex end pieces, the glue used to attach them, and the prism needed protecting from the heat. The prisms are delicate and moderate heating causes craze cracking. The topmost Perspex plate held a small fireclay ring which fitted inside the silica tube. A glass tube ran through the centre of the plate and fireclay. On the top of the glass tube an optical flat was glued and the prism rested on the flat. The fireclay acted as a heat shield and the whole area was cooled with an air blast. The lower end of the column was cooled with an air blast.

The air blasts, vacuum pump and generator cooling water were kept on during cooling of the specimen.

5.3 Examination and Testing

5.3.1 Tensile Testing

Tensile tests on the as - received tungsten wire were carried out in an Instron machine. The ends of the wires were pushed into 0.125" diameter copper tubes, which were then filled from the other end with silver solder. The tubes were then bent into a hook shape and suspended from special adaptors in the Instron. This technique resulted in more gauge length failures than achieved by mechanical clamping. The wires extracted from the composites were short, and this presented problems of gripping. The method used was to clamp the wires between two flats, pressure being applied by the screws of a collet. The collets fitted into a tension - compression cage in the Instron, which was used in the high sensitivity range with a 40 lbs full scale deflection.

The tensile tests on the pure copper and composite specimens were carried out in the Instron. The specimens were brazed into mild steel end pieces. The brazing formed a smooth radius at the end of the gauge length, and proved strong enough to withstand the U.T.S. of the composites. The specimens were cleaned up with emery and polished down to a 0000 grade emery finish in a lathe. Electropolishing the copper in an orthophosphoric acid solution was tried at first, but this resulted in the outer fibres being exposed completely before a good surface was obtained. The fine emery finish could be achieved without exposing fibres. Most specimens appeared to have a thin outer rim of fibre - free copper.

The large specimens were mounted in self - centreing adaptors. The small specimens, which were used for the more accurate work, were mounted in collets. The mounting was checked for accurate axial alignment in a jig. The collets were then screwed into a tension - compression cage. Load - extension curves were recorded automatically, and were then replotted with a correction for cage extension. Load cycling was also carried out using this cage.

5.3.2 Fatigue Testing

The fatigue testin was carried out in a Baldwin Lima - Hamilton S F 1 U machine, using a repeated tension assembly. The large specimen was loaded into the assembly collets and a load applied with an automatic preload device. The dynamic load was set by means of an eccentric, so that on starting up the load cycle was symmetric about the preload value. The lowest load value was always zero for all tests. If left alone the whole cycle would shift towards compression due to specimen extension. This was avoided by use of automatic preload maintenance, which maintained the same load limits whilst allowing for specimen extension. The rate of cycling was 1800 c.p.m.

The load cycle was made wholly tensile to eliminate the variable of fibre buckling in compression. The specimens were prevented from fracturing completely by very close adjustment of the shut - down micro - switches. This adjustment was made five minutes or so after starting up. Restricting the amount of elongation before shut - down usually had the effect of stopping the main fatigue crack when it had covered more than 50 % of the cross - sectional area. This condition, or complete failure, was taken to be the end of the test for the purpose of quantitative results.

The extracted wires were fatigued in repeated tension in a machine at De Havilland of Canada, Malton. The machine was designed at the Institute of Aerophysics of the University of Toronto, and has been described and evaluated by Carr (53).

An electromagnetic shaker applies a vibrating load to a hinged heavy section beam, which applies a direct tensile pull to a thin steel transfer plate. The plate pulls on a heavy cage which contains a collet to hold $\frac{3}{4}$" diameter specimens. A similar cage acts as a fixed head. The wires were held in small steel clamps. The clamps were bolted to $\frac{3}{4}$" diameter clamp holders which were held in the collets. There was an aluminium spacer spar between one clamp and clamp holder. A strain gauge compensated for aluminium was mounted on either side of the spar. These gauges were connected with two dummy gauges mounted on identical material, to form a Wheatstone bridge. The spar gauges compensated for any bending effect, and the dummy gauges were placed close to the spar and acted as temperature compensators.

The spar was made small in order to give a good strain reading with the 10 or 20 lb loads employed. An identical piece of aluminium was pulled in the Instron, using an extensometer and the X - Y recorder. The material was elastic up to the maximum load applied, this being 200 lbs. The strain gauge bridge was connected to the Brush strain recorder at De Havillands and the system calibrated by hanging weights up to 30 lbs on the aluminium spar.

An extracted fibre was mounted in the machine, and the mean load applied. This preload was not automatically maintained during the test, but the strain recorder showed no appreciable extension. This observation is in agreement with a test carried out in the Instron, in which a fibre was load cycled for several days and 125,000 cycles with no noticeable change in the extension indicators. The alternating load amplitude on the De Havilland machine was controlled by the oscillator and was independent of the mean. The cut - off after specimen failure was effected by a galvanometer acting as a strain amplitude monitor, this activating a relay which shut off all power when the amplitude deviated beyond certain limits. The stability of the system was very good, the load amplitude remaining constant within about 1%.

The rate of cycling was usually 60 c.p.s., resonance being avoided since it merely caused instability of the load amplitude and bending vibration in the strain gauge spar. Tests were carried out at various load ranges, the minimum load used being 0 and the maximum 20 lbs.

5.3.3 Metallographic Examination

To obtain a specimen for metallographic examination it was sometimes necessary to remove the specimen from the grips. The consequent heating recrystallised the work hardened areas near the cracks and this could be observed on some of the micrographs. Longitudinal and transverse sections were taken, and successive polishing of the former allowed the path of the crack to be followed. Sectioning was carried out by hand sawing at the ends of the gauge length, care being taken not to load the cracked area. In some cases the crack was filled with braze before sectioning. The etchant used on some of the specimens was acidified ferric chloride.

Photomicrographs were taken on Reichert and Vickers microscopes.

SECTION 6

Results

6.1 Tension

Tensile tests were carried out on the as - received fibres and on the fibres extracted from composites. The results are shown in Table 2, the mean strain at the U.T.S. of the extracted fibres being 1.0%. If the ductile fibres were removed from the list, the mean strain falls to about 0.75% strain. This figure is in closer agreement with that arrived at by inspection of the histogram in Fig. 19, in which most of the breaks occured around 0.7%. The least squares value of the extracted fibre U.T.S. with 95% confidence limits was-

$$274{,}228 \pm 71{,}200 \text{ p.s.i.}$$

Approximate stress - strain curves for a brittle fibre and copper prepared as for the composites is shown in Fig. 2. A stress - strain curve for one of the Cu-23%W composites is shown in Fig. 19. A histogram of the fibre breaks heard during the latter test is shown below the curve, and it can be seen that the number of fibre breaks for a fixed strain interval increases to a maximum at the U.T.S.

The relation between σ_f, σ_M, $\sigma_{M'}$, and V_f is shown in Fig. 20. The law of mixtures line is a least squares plot evolved by an IBM 7040 computer, which was fed with $\sigma_{M'}$ values and the extracted fibre strengths. The $\sigma_{M'}$ and σ_M values are together with all the tensile results on composites in Table 3. The intersection of the least squares line with σ_M horizontal gives the value of $V_{f\ CRIT}$ as 5% (see Fig. 4.)

Fig. 4 shows the variation of σ_f with $V_{f\ CRIT}$, this curve being calculated using the σ_M and σ_M' values from copper prepared as for the composites used in this work. The dot on this curve corresponds to the mean extracted fibre U.T.S., and gives a $V_{f\ CRIT}$ value of 5%.

The fractures of the specimens with $V_f < V_{f\ CRIT}$ were of the necked, ductile type. Specimens with $V_f > V_{f\ CRIT}$ showed jagged but more or less transverse fractures with little evidence of ductility. The copper of the Cu-7.4%W samples showed more evidence of deformation than in the case of the Cu-25%W samples.

6.2. Fatigue

On commencing a fatigue test, the specimen underwent permanent tensile strain in the first 100 cycles. In the case of copper and the low V_f samples, this straining was primary cyclic creep and was quite considerable. The amount of such strain for a given stress, and the attendant number of audible fibre breaks, decreased with increasing V_f so that the Cu-25%W samples did not usually undergo noticeable strain.

A Cu-1.5%W sample was subjected to about 100 cycles at 1[illegible],900 p.s.i. The sample was then removed and the copper dissolved away in nitric acid. Initially present were ten 65mm long fibres, but after cycling there were 54 definite pieces and some small debris. The lengths of the pieces were in the range 27mm to 1mm. Thus within 100 cycles about 5 breaks per fibre were produced. A typical life for such a sample at this stress level was [illegible]0,000 cycles. Such breaks were purely tensile behaviour and none were heard after the initial cycles.

The numerical fatigue data are presented in Table 4, and also as an S – N plot in Fig. 21. In order to make a quantitative analysis of these results the following assumptions were made–

1 The pure copper and Cu-1.5%W samples had identical fatigue strengths.

2 The S – N plot for a fixed V_f was linear in the range of lives observed for any one degree of reinforcement.

All the data except that for unbroken specimens or those with pores was then processed in an IBM 7040 computer to give least squares lines with 95% confidence limits. The computed equations were then used to give full cycle*fatigue strengths at 10^6 cycles, which were divided by the corresponding mean U.T.S. values to give the fatigue ratios shown in Table 6. The numerical results are presented in the as – computed form, but the limitations on accuracy are obvious. The assumptions can be questioned, the number of results is not great and the criterion of fatigue failure was loose. This means that only gross effects can be considered valid. Much more work would be needed to establish the validity of the fatigue ratio difference between the copper and Cu-7.4%W , but the difference between copper and Cu-23%W is indeed gross.

A number of extracted fibres was fatigued in repeated tension at high fractions of the U.T.S. The results are shown in Table 5. The full cycle fatigue strength at 10^6 cycles will be at least 200,000 p.s.i.

* Full cycle means that the whole stress range is quoted.

The extension of the fibres during the test was negligibly small. There was no noticeable change in the extension limits during the load cycling of a fibre in the Instron.

6.3 Metallography

The first step taken in examining the specimens was to etch the surface in acidified ferric chloride. The grain size thus revealed was very large, there being a few long dendritic grains each extending half an inch or more. A typical transverse section of a Cu-23%W sample is shown in Fig. 22.

During the first few cycles some fibres fail by tension, and others fail later due to fatigue. The effect of fibre breaks is shown in Fig. 23. This particular break was double, and shows a small piece of tungsten in between the breaks. Such pieces doubtless account for the debris from the dissolved sample, (Section 6.2), and are unlikely to have been produced by two breaks separated in time. The cracks running from the breaks are typical, and are not found under tensile test conditions. In a tensile test the fibres break until there are insufficient in a particular plane to support the load, upon which the matrix necks down in between the fibre breaks and fails in ductile fashion. This type of failure was observed in copper - tungsten by Kelly (50), McDanels et al. (9) and in the present work. In no case was there any evidence of cracks running from the breaks as in Fig. 23. The cracks in this case are fatigue damage, and they were observed around the majority of fibre breaks.

There was an indisputable relation between fibre breaks and the propagation of the main transverse fatigue crack, Fig. 24 being one example of many. This sample shows a main transverse crack which has split into two at the tip. Tracing the crack back a little leads to two smaller ear cracks similar to those at the tip. The tributary crack at the top of the micrograph also has ears at the tip. There is another tributary running along the interface with the lower fibre.

Fig. 25 shows a propagating transverse crack meeting an unbroken fibre. The crack turned parallel to the fibre and ran along it until it either broke the fibre at a weak point or passed through an existing break. The main crack is well developed and definitely runs down the interface.

Fig. 26 shows the crack tip running from a main transverse crack which has been turned parallel to the fibres. The tips run appreciable distances through the matrix in this particular section, but a slightly different section may show these cracks to meet the interface.

The fibre orientation was not perfect, so that the micrographs of longitudinal sections often show that the fibres have been taper sectioned and appear elliptical. If the major axis of an ellipse is extended from the fibre into the matrix, the matrix area around the extension will be very close to the fibre surface. Fig. 27 shows the tip of such an ellipse. A main transverse crack has been turned parallel to the fibre, and Fig. 27 shows the crack tip behaviour. The crack has passed from one side of the fibre to the other by running a thin tributary around the circumference. Further evidence of this is in Fig. 28, which shows a sharper taper section. The crack in this case met the fibre at a small angle, and ran down the interface and around the fibre. The crack follows a reasonably smooth curve in the matrix until it crosses the fibre axis, where it shows a distinct kink towards the transverse direction. This may be associated with the crack leaving the interface and running freely in the matrix. A further piece of evidence of the ability of a matrix crack to circumnavigate unbroken fibres was provided by a specimen in which a well developed crack had been stopped when the fatigue machine micro-switches tripped out. The jaws of the crack were spanned by unbroken fibres.

SECTION 7

Discussion

7.1. Tension

The extracted fibres showed a considerable drop in strength from the as - received state, and also showed considerable scatter in strength. These effects are due to differing degrees of recovery and recrystallisation during the manufacture of the composites, since the temperature employed was close to the usual recrystallisation range of work hardened tungsten.

The composite copper was in the as - cast state and had very large soft grains, this accounting for the low U.T.S. The material showed the expected large strain to failure and necked ductile fracture.

The composites behaved normally in tension despite a wide variation in fibre strengths, and obeyed the law of mixtures. The composite U.T.S. values were well within the 95% confidence limits of the least squares line in Fig. 20. The figure of 5% obtained for the $V_{f\ CRIT}$ value in this system is close to the 6% figure obtained by McDanels et al. (9).

Fig. 19 shows the number of audible fibre breaks on tensile testing a Cu-23%W sample. The number of breaks counted in this way is not far wrong, the only errors being underestimates of the occasional multiple break. It is interesting to note that the number of breaks up to the U.T.S. is just two short of the total number of fibres in the specimen, though this may not mean that every fibre breaks once but that some are unbroken and a few have broken twice. The histogram also gives the number of fibre breaks in this particular specimen which would occur in the first half cycle of fatigue. There were only two breaks up to 18,700 p.s.i. and only four up to 22,400 p.s.i. The breaks were too rapid to count during an actual fatigue test, but it sounded as if more breaks than this occurred during the first few cycles. Subsequent cycles to the first under slow tensile loading conditions did not cause further fibre breaks. The extra breaks in rapid fatigue testing may have been due to the different characteristics of the other specimens or erratic loading conditions as the fatigue machine built up to operating speed and load. The fibre breaks in Fig. 19 beyond the U.T.S. are mostly secondary breaks, and plastic strain in the matrix beyond the average of 0.7% is needed to achieve the breaking stress of a fibre for the second time.

Copper has been reported as insoluble in tungsten (59), but molten copper wets tungsten satisfactorily. An interface such as this may be expected to behave in a manner similar to that reported by Herzberg (46) for the Cu - Cr system. Herzberg prepared tensile specimens from a unidirectionally solidified Cu - Cr eutectic, which was essentially 1.5% by volume of pure chromium whiskers in a copper matrix. Pulling a specimen with the whiskers parallel to the applied stress caused the whiskers to fracture, but pulling with the whiskers transverse to the stress caused failure of the interface. The interface could withstand the small shear stress required to build up the fracture stress of the whiskers, but could not withstand a direct tensile load.

All the workers with the copper - tungsten fibre system used copper of only 99.99% purity and various atmospheres over the melt (9),(18), (50). The effect of small amounts of impurity can have a considerable effect on bonding (45, 61). Nicholas and Poole (61) carried out sessile drop tests using molten copper on tungsten, the system being under vacuum. The copper was of the same purity as used in this work. The contact angle decreased as temperature was increased from 1100°C to 1450°C , and the poorer wetting at the higher temperatures was thought to be due to the formation of an oxide at the interface. The bond strength of this system was typically 3600 p.s.i. The effect of impurities was very strong, since a similar series of tests carried out using copper of spectroscopical purity gave excellent wetting and a consistent bond strength of about 78,000 p.s.i. Tests were carried out showing that the bond strength necessary to achieve the law of mixtures tensile strength is lower than might be expected. Tungsten wires were pushed into thin copper tubes and then tensile tested, so that the only bond was friction. The strength of a bundle of such tubes with wires inside so that V_f = 50% was calculated to be 217,000 p.s.i. as compared with 280,000 p.s.i. for chemically bonded composites. The strength of the interface achieved with the purity of copper used in this and other work (9, 18) appears to be low in direct tension, but more than adequate to transmit the shear stresses necessary to break the fibres under tensile stresses in their axial direction.

7.2 Fatigue

The extracted fibres were extremely strong under conditions of repeated tension fatigue. No room temperature fatigue tests of any description were reported in the literature for tungsten, but Bredfick and Fritch (54) carried out high temperature repeated tension fatigue tests on 0.250" diameter tungsten rods. The lowest temperature employed was 1670°C , and an extrapolation of their results to 10^6 cycles gave a fatigue ratio of 0.56. The closest comparison apart from this are the results of Bruckart and Hyler (55) , who claim their fatigue tests to be the first ever on molybdenum. The rotating bending fatigue strength at 10^6 cycles was extremely good, the fatigue ratio being about 0.75. The previous work thus suggests that tungsten might be good in fatigue , and the present work suggests that the fatigue ratio based on the repeated tension fatigue strength at 10^6 cycles is at least 0.75.

The pure copper prepared as for the composites failed by cyclic creep , the strain to failure being very large and of the same order as in a tensile test. The fractures were necked and of the ductile type expected in cyclic creep. This type of failure is typical for these conditions of loading (31, 32) .

The mode of fatigue failure of the composites was dependent on V_f . Introducing 1.5% by volume of fibres had no noticeable effect on either tensile or fatigue strengths. The few fibres undoubtedly provided reinforcement and were broken many times as indicated by the 100 cycle test in which the breaks were counted. The small number of fibres took little load however, and both tensile and fatigue behaviour were dominated by the matrix. The breaks in the fibres did not reduce the strength of the composite, since the matrix took a high fraction of the load and failed before fatigue cracks had time to develop at the breaks. The breaks were likely to be pure tension rather than fatigue, caused by the considerable amount of plastic strain during cyclic creep.

The mode of failure of the specimens with V_f values around 23% was totally different. In this case the fibres took most of the load. The final fracture appeared brittle in that there was little strain to failure, no necking and a jagged but more or less transverse fracture path. The fibres undoubtedly dominated this type of fatigue failure. The number of audible fibre breaks during the initial fatigue cycles was small compared with the 1.5% V_f sample. The 100 cycle test on one of the latter resulted in 54 breaks, whilst the breaks in a 23% V_f sample were little more than the 2 or 4 indicated by the tensile test. The bulk of the fibres must therefore fail by fatigue, and it is likely that the fatigue strengths will show similar scatter to the tensile strengths, which is considerable. In the event of a fibre failing, there will be a redistribution and increase in load amongst the others. A simple quantitative idea of the effect of successive fibre breaks is as follows; there are in fact about 170 fibres in the cross - section, and these may carry 800 lbs load. Considering one transverse plane in isolation, the failure of a fibre effectively removes it from the plane so that 800 lbs is now carried by 169 fibres. The mean fibre load increases in this way slowly at first but rapidly later, so that when less than 100 fibres are left each break adds 1000 p.s.i. to the mean fibre stress. The mean stress exceeds the mean U.T.S. when about 40 fibres remain unbroken. This mechanism explains how the crack propagation will become increasingly rapid towards the end of the test, but it cannot explain the poor fatigue ratio of the Cu-23%W samples. It appears that the addition of increasing volume fractions of fibres causes reductions in fatigue ratio. The fatigue strength of the Cu-23%W specimens was some 30% better than the matrix , but this contrasts with a 300% increase in tensile strength.

A first analysis of this problem can be made by applying a law of mixtures of fatigue, so that -

$$\sigma_c^f = \sigma_f^f V_f + \sigma_M^f (1 - V_f)$$

where σ_c^f = Repeated tension fatigue strength of composite at 10^6 cycles.

σ_f^f = Repeated tension fatigue strength of extracted fibres at 10^6 cycles.

σ_M^f = Repeated tension fatigue strength of matrix at 10^6 cycles.

Substituting the fatigue strengths obtained for the composites we obtain at 7.4% V_f reinforcement

$$16{,}654 = 0.074\,\sigma_f^f + 0.926\,\sigma_M^f$$

at 23% V_f reinforcement

$$20{,}241 = 0.235\,\sigma_f^f + 0.765\,\sigma_M^f$$

Solving for σ_M^f we have

$$\sigma_M^f = 14,400 \text{ p.s.i.}$$

and this figure is only 400 p.s.i. different from the direct experimental value for σ_M^f, and a long way inside the 95% confidence limits. Solving for σ_f^f we obtain

$$\sigma_f^f = 39,700 \text{ p.s.i.}$$

which is far below the experimental value. The tests on extracted fibres showed a fatigue strength of at least 200,000 p.s.i., and the problem is thus to establish why the calculated effective strength of the fibres is so low.

The metallographic results show that the regions of high stress concentration around breaks in the fibres are favourable sites for fatigue crack initiation in the matrix. Since the copper is subjected to a fairly high plastic strain amplitude the rate of crack propagation will also be high (60). Now the maximum stress in the fibres is typically 70,000 p.s.i. or so, a stress which should be easily withstood for 10^6 cycles. The only way in which these fibres can be made to fail under these conditions is to be subjected to a localised stress concentration. The source of stress concentration can only come from the propagating matrix fatigue cracks, which must exert a stress concentration factor in the fibres of at least 3 to cause a fibre to fail in 10^6 cycles. If the whole composite fails in 10^6 cycles a higher stress concentration factor must be present, since the cracks must have time to propagate and will not influence all the fibres simultaneously. If the tensile stress in the fibres is raised by a factor of about 4, the localised stress is then sufficient to break the fibre in tension.

To achieve such stress concentration factors in the copper means that the matrix must work harden in fatigue until it is almost completely elastic over the applied stress range. Cook and Gordon (47) examined the stress distributions around an elliptical transverse crack propagating slowly through an elastic medium. The theoretical stress concentration at the crack tip magnified the applied stress 200 times if the ratio of the ellipse axes was 100 : 1. The copper is initially ductile, but cyclic stressing permits a high degree of fatigue hardening, particularly in the region of a crack tip. When the crack is running through the matrix the stress ahead of the tip can be relieved by plastic deformation. When the crack is delayed such as on encountering a fibre, the material around the crack tip work hardens more and more until it is nearly elastic. Very high stress concentrations are possible in elastic material, and it can be appreciated that there will be a strong tendency for the matrix to try and shear past the fibre, and hence induce high tensile stresses in the fibre.

If we postulate that the whole matrix is fatigue hardened until it is completely elastic, then a quantitative check that this is so can be made with the relation

$$E_c = E_f V_f + E_M (1 - V_f)$$

The modulus figures obtained by McDanels et al. (9) for their tungsten, which was the same type as in the present work and obtained from the same maker, was 58.8×10^6 p.s.i. The figure obtained for copper, which was of the same purity and in the same state, was 17×10^6 p.s.i. Thus when $V_f = 0.23$, $E_c = 26.6 \times 10^6$ p.s.i. Then the strain in the composite ϵ_c, at the peak fatigue stress for a life of 10^6 cycles is given by

$$\epsilon_{c_1} = \frac{20,241}{26.6 \times 10^6}$$

$$= 7.6 \times 10^{-4}$$

$$= \epsilon_{M_1}$$

where ϵ_{M_1} is the strain in the matrix.

Therefore

$$\sigma_{M_1} = E_M \epsilon_{M_1}$$

$$= 17 \times 10^6 \times 7.6 \times 10^{-4}$$

$$= 12,900 \text{ p.s.i.}$$

When $V_f = 0.074$, $E_c = 20.1 \times 10^6$ p.s.i.

The composite strain at the peak fatigue stress is

$$\epsilon_{c_2} = \frac{16,654}{20.1 \times 10^6}$$

$$= 8.29 \times 10^{-4}$$

$$= \epsilon_{M_2}$$

Therefore

$$\sigma_{M_2} = 8.29 \times 10^{-4} \times 17 \times 10^6$$

$$= 14,100 \text{ p.s.i.}$$

Thus the peak fatigue stress for a matrix life of 10^6 cycles averages out at 13,500 p.s.i. This result is well within the errors in the experimental figure of 14,812 p.s.i., and the result calculated from the fatigue law of mixtures, 14,400 p.s.i. The closeness of the experimental result to the figure for a totally elastic matrix indicates that the copper does fatigue harden until it is almost elastic.

Further confirmation of this view was provided by load cycling tests on a Cu-23%W composite in the Instron, in which the mechanical hysteresis after 2 cycles was $< 10^{-4}$.

The hysteresis measurement also gives an estimate of the residual stresses in the components. A residual strain in the fibres of 10^{-4} gives an elastic tensile stress of

$$\epsilon E = 58.8 \times 10^{6} \times 10^{-4} = 5880 \text{ p.s.i.}$$

In the unloaded state the residual tensile load on the fibres is balanced by the residual compressive load in the matrix, so that

$$\sigma_f^u V_f = \sigma_M^u (1 - V_f)$$

where

σ_f^u = mean residual tensile stress in fibres

σ_M^u = residual compressive stress in matrix.

Therefore

$$\sigma_M^u = \sigma_f^u \frac{V_f}{1-V_f}$$

$$= 5880. \frac{0.23}{0.77}$$

$$= 1{,}760 \text{ p.s.i.}$$

Thus the compressive stress on the matrix is small and the matrix is fatigued effectively in repeated tension, as assumed in the law of mixtures calculation.

A further question to be answered is why the interfaces can open up under fatigue conditions. It is also necessary to explain what happens to such a longitudinal crack once it has formed. Cook and Gordon (47) considered the effect of a transverse crack slowly propagating under steady tension towards a longitudinal weak interface. There is a considerable direct tensile stress exerted on the interface, this stress acting in a direction parallel to the crack propagation direction. As the crack approaches the interface, this stress builds up to a maximum a short distance ahead of the crack tip. If the interface could not withstand this maximum, it would open up the same distance ahead of the crack tip. The main crack then propagates into the interfacial crack and is deflected longitudinally. Similar stresses will be experienced by a fibre - matrix interface, should a matrix fatigue crack approach. The stresses will be cyclic , however, and there is the additional shear stress at the interface due to the difference in moduli. The fibre on one side of the interface is cycling in repeated tension, and the matrix on the other side of the interface is cycling in tension - compression with a tensile mean. It appears that this complex stress system eventually causes either a transverse break in the fibre, as discussed previously, or a longitudinal split in the interface.

The fracture surface of the Cu-23%W sample was similar to that obtained by Baker and Cratchley (49) for strongly bonded silica fibre reinforced aluminium. The fracture was jagged but was more or less transverse. Weakly bonded specimens failed by wholesale longitudinal splitting of the interfaces. This was also the case with the fibre - reinforced plastics of Davis, McCarthy and Schurb (51), and the steel wire reinforced aluminium of Williams and O'Brien (62). All these workers used tests involving a half cycle of compression, and it likely that compression as well as weak bonding aggravates the problem of splitting. The matrices have been observed to fail completely by splitting, whilst the bulk of the fibres remains intact. The tests carried out in the present work induced a small amount of matrix compression and no fibre compression at all, so that the difficult problem of analysing the effect of fibre buckling is eliminated. The interfacial failures observed in this work are due entirely to the effect of propagating matrix fatigue cracks and cannot be produced by fibre bending. The interfaces in this system were strong enough to resist any tendency to produce wholesale splitting under the applied stress conditions.

Once a crack has formed at the interface , there is still a transverse stress with an appreciable value extending some length along the fibre, and this can assist propagation. During the unloading part of the fatigue cycle the unbroken fibre contracts and compresses the matrix, and the crack then tends to buckle open and extend further. It is possible that when an interfacial crack opens up near a fibre break as in Fig. 25, propagation towards the break is assisted by the severe stress system around the break. It is also possible that the crack propagates down the interface until the severe triaxial stress system at the tip causes failure at a weak point in the fibre. It is not possible to distinguish the breaks produced under initial tensile loading and those produced by the matrix crack under fatigue conditions, except where a tensile break is in isolation as in Fig. 23.

The whole picture of the fatigue failure of Cu-23% W composites can now be outlined. Failure commences with the tensile failure of a few fibres during the first few cycles. The next event is the easy initiation of fatigue cracks at the breaks , which are randomly distributed about the specimen. The propagation of matrix cracks is quite rapid at the high strain amplitude, and they interact with fibres in the following ways;

1 Pass through an already existing fibre break.

2 Cause the accelerated failure of the fibre by localised stress concentration, and then pass through the break.

3 Turn parallel to the fibre, cause failure of a weak point by local stress concentration and then pass through the break.

4 Turn parallel to the fibre and then run down and around it by tearing open the interface.

5 Continue propagating transversely and engulf the broken fibre.

Mechanisms (1), (2) and (3) are not distinguishable by use of metallography, and (2) and (3) essential to the explanation of premature failure of the composites.

Nicholas and Poole (61) have shown that the interfacial bond strength is not great under direct tension, so that (3) and (4) may not be able to occur if a very strong bond was produced.

Mechanism (4) has been observed and can be readily understood, since if the crack starts to run round the fibre it will be increasingly assisted by a component of the applied stress.

Mechanism (5) has only been observed once, and this was in a specimen with a low V_f value. The advancing crack was held up at the fibre, but was able to propagate past it on either side since the inter-fibre spacing was large. The cracks at either side then met beyond the fibre, thus engulfing it without turning parallel.

Where two or more fibres failed in positions close to one another the breaks were eventually joined by matrix cracks to form what can be termed a damage area. It is conceivable that such damage areas could cover the whole cross - sectional area if accumulated in one plane, but spread over the whole gauge length in small areas would not be immediately fatal. A damage area will grow slowly at first, but will spread more rapidly as it grows since the associated stress concentration increases with size and this will influence an increasing number of undamaged fibres. It has been observed that cracks can run around fibres, and down fibre - matrix interfaces at least for short distances. Thus a matrix crack can join two fibre breaks separated longitudinally to form a damage area. Damage areas continue to grow and may join up until the remaining cross - section cannot support the load and fails in tension.

The Cu-7.4%W samples showed behaviour intermediate between that of the copper and the Cu-23%W specimens. The initial strain and number of fibre breaks was greater than that of Cu-23%W, though the strain was much less than the copper. They were stronger in tension and slightly better in fatigue than the copper. The samples had a reinforcement level just above $V_{f\ CRIT}$ and the fibres took slightly more than half the load. The fracture was similar to that of the Cu-23%W but showed signs of slightly better ductility.

The ability of the fibres to delay crack propagation in other systems has been noted previously (48, 49). In the present work the fatigue life of Cu-23%W was several hundred times better than that of the pure copper at the same stress, though this is partly due to the trivial reason that the fibres take a high proportion of the load. It has been seen , however, that a propagating crack can be turned parallel to an unbroken fibre, and this must exert a delaying influence. The presence of fibres in greater quantities than $V_{f\ CRIT}$ prevents the copper from undergoing cyclic creep.

SECTION 8

Conclusions

1 A satisfactory technique of preparing copper - tungsten composites was developed.

2 The composites showed normal tensile behaviour, the results being in agreement with those of previous workers and the theory of composites.

3 The number of fibres failing in a fixed strain interval increases to a maximum as the tensile stress rises to the U.T.S. of the composite.

4 The value of $V_{f\ CRIT}$ determined for tensile testing is still significant under repeated tension fatigue conditions.

5 The fatigue failure with $V_f < V_{f\ CRIT}$ is dominated by the matrix behaviour, which fails by cyclic creep under repeated tension.

6 The fatigue failure with $V_f > V_{f\ CRIT}$ is dominated by the fibres, and the sequence of events is

(i) Initial tensile strain and breaking of fibres in tension.

(ii) Initiation and propagation of fatigue cracks in the highly stressed matrix around the fibre breaks.

(iii) Failure of fibres by fatigue, accelerated by high stress concentrations associated with matrix cracks.

(iv) Joining of fibre breaks by matrix cracks to form damage areas.

(v) Growth and eventual joining of damage areas so that the remainder of the cross - section fails in tension.

7 The fatigue ratio of the Cu-23%W samples was poor. This was due to

(i) The easy initiation of matrix fatigue cracks in the highly stressed areas near fibre breaks.

(ii) The accelerated fatigue or tensile failure of the fibres produced by the stress concentration of the matrix cracks.

(iii) The ability of running matrix fatigue cracks to join fibre breaks.

8 The extracted tungsten fibres had excellent repeated tension fatigue properties.

9 The theoretical behaviour of composites in fatigue was largely as predicted in Section 3, but the theory did not take into account the stress concentrations in the matrix which proved damaging to the fibres.

10 The fatigue life of composites with $V_f > V_{f\ CRIT}$ was much better than that of the matrix alone, and the improvement increased with V_f.

11 The relatively weak bond at the interface opened additional channels for matrix crack propagation, but strong bonds will not give a great deal of improvement in properties.

SECTION 9

Future work

Fatigue cracks were observed to propagate through the matrix and down the interfaces. There are several likely methods of counteracting these processes. The first is by using an age - hardening alloy as a matrix. The difficulty here would be to obtain an alloy which wets tungsten and yet does not destroy the strength. The Cu-Be system is a possibility. The complete Be-W equilibrium diagram is not known, but there are similarities with Be-Mo in some regions and it seems likely that the solubility of Be in W will be quite small. Cu-Be alloys are well known for good fatigue strength, and are likely to resist matrix crack propagation much better than pure copper. Another possibility is to use an ordering alloy such as Cu_3Au as the matrix. Gold and copper wet tungsten and do not alloy with it. Work on other alloys in the ordered and disordered states (52) shows that the ordered alloys are appreciably stronger in fatigue. The propagation of cracks down the fibre - matrix interface may be counteracted by producing a stronger bond with high purity materials and very clean manufacturing conditions. Work on alloyed matrices in the Cu-W system by Petrasek and Weeton (8) showed that the strength reduction in the fibres was severe in all but a few cases. It may be possible to tolerate a reduction in fibre strength in order to improve the fatigue life.

Discontinuous composites are likely to be worse in fatigue than continuous composites, due to the presence of fibre ends.

Reducing the associated stress concentrations by using pointed ends (41) is unlikely to be much help, since fibres will still fail possibly by initial tension and certainly by fatigue.

Fatigue tests should be carried out under conditions which do not cause the initial tensile breaks in the fibres. The role of such breaks could be further checked by prestraining to break a greater number of fibres and then fatigueing strained and unstrained samples at the same stress. It is likely that the fatigue life but not the fatigue strength is improved by eliminating tensile breaks.

An excellent matrix material to use would be one which does not work harden at the temperature of testing. At room temperature only materials such as lead would be available but at moderate temperatures a wider range of choice would be afforded. A matrix such as this could not support severe stress concentrations.

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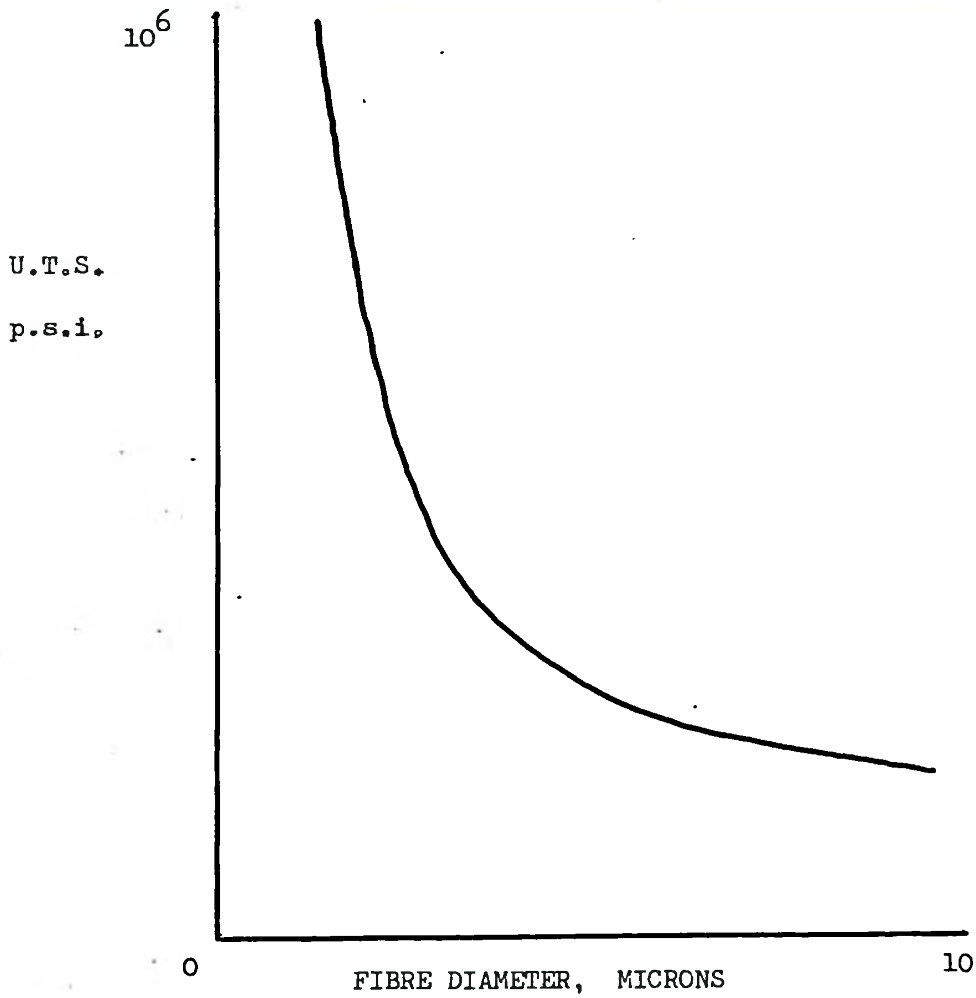


FIG, 1

Typical strength - size effect.

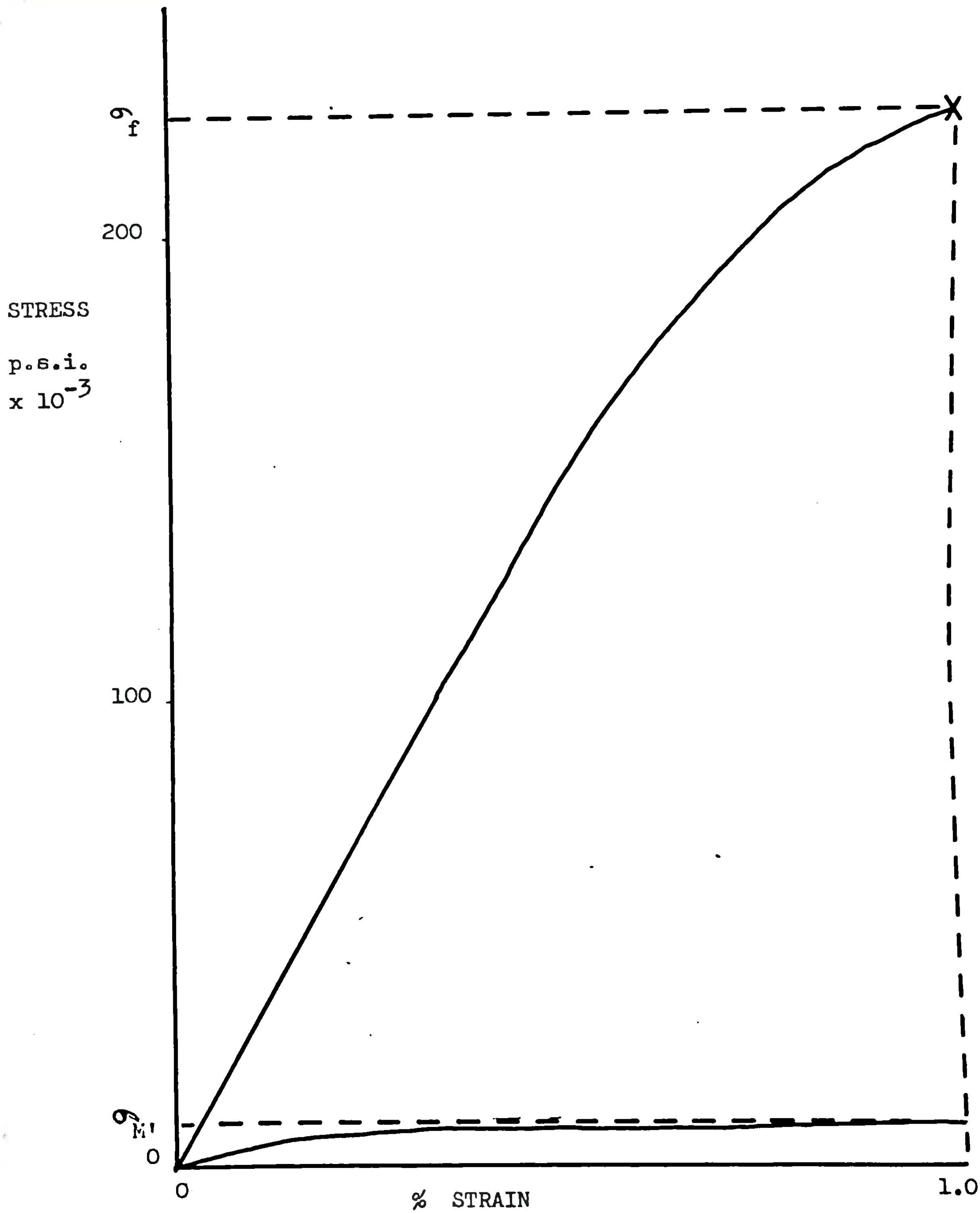


FIG. 2

Approximate stress - strain curves for a brittle tungsten fibre and copper as used in composites.

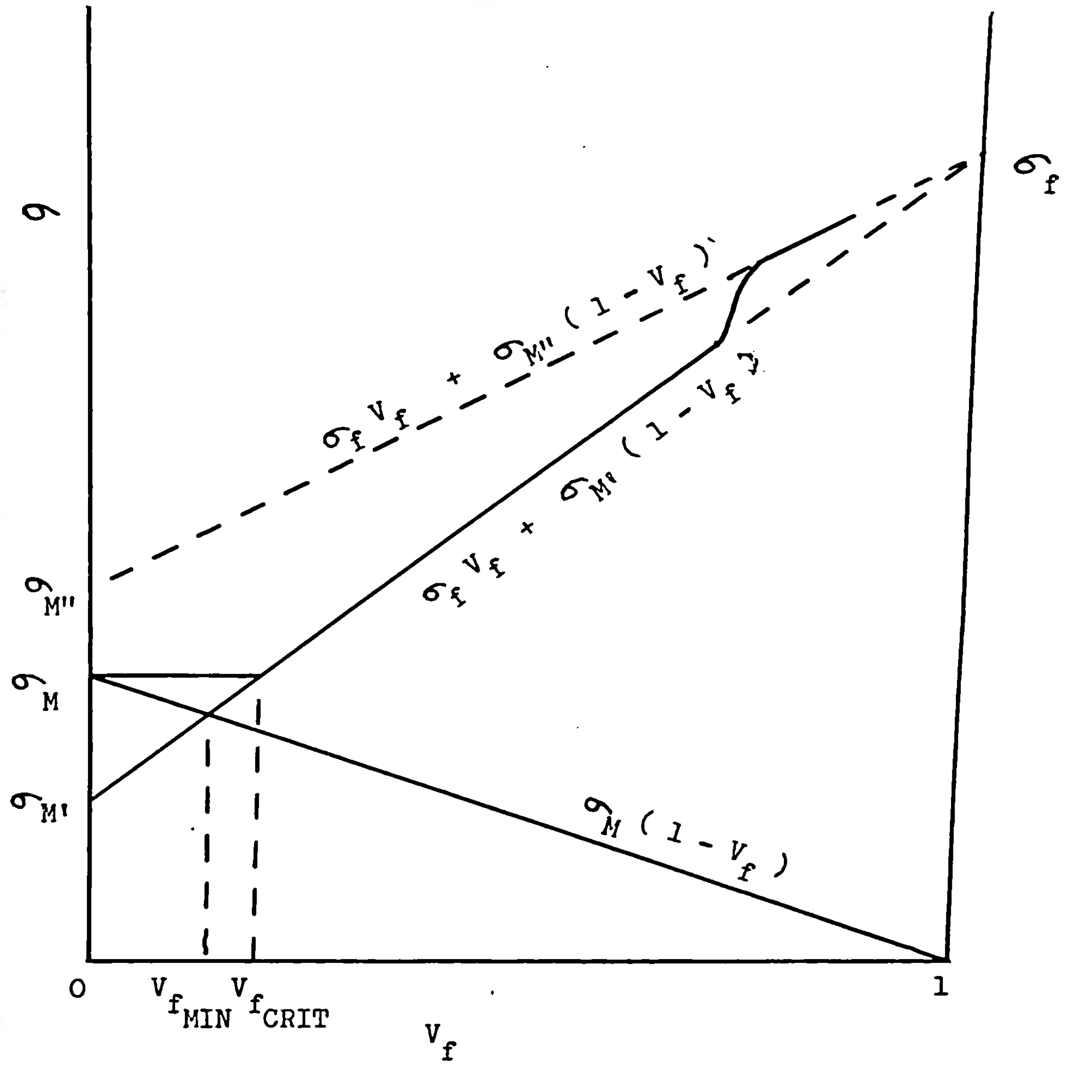


FIG. 3

Theoretical stress - V_f relationships

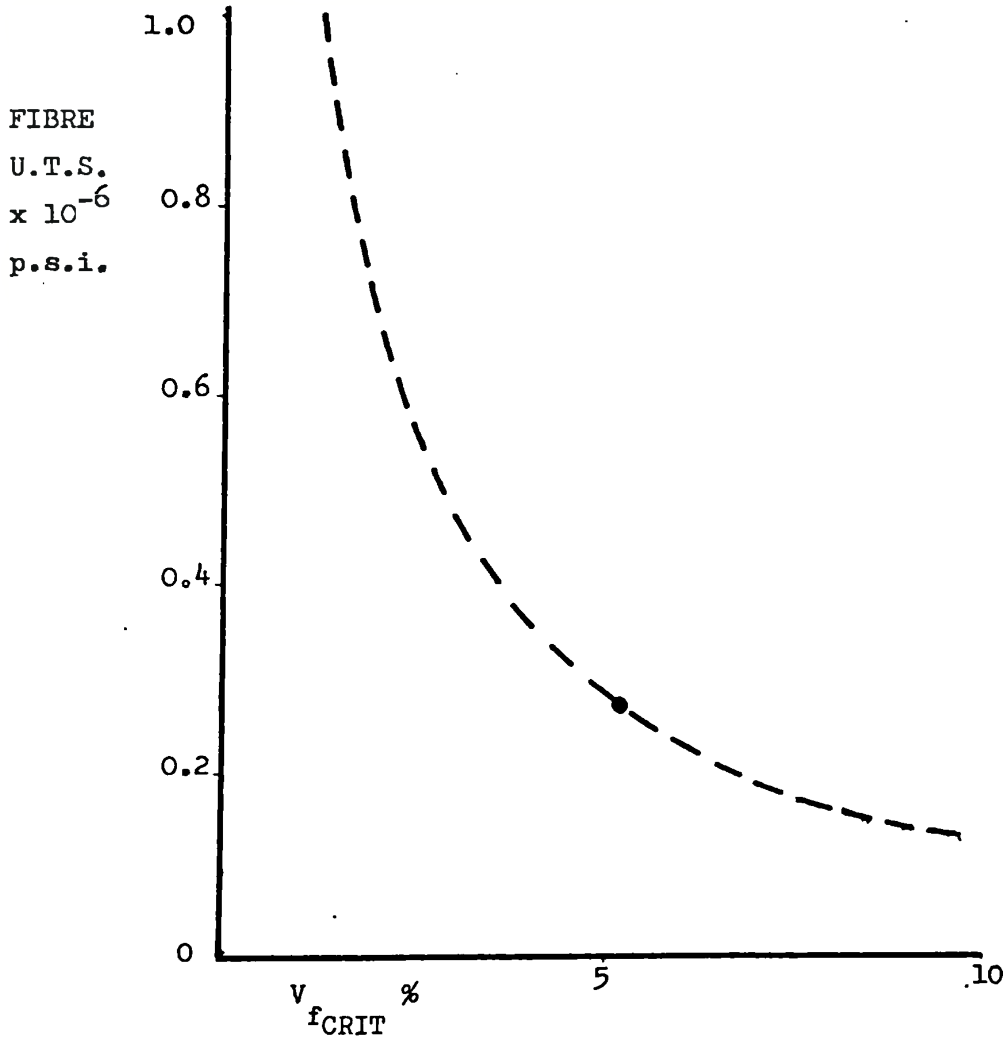


FIG. 4

Variation of mean fibre U.T.S. with $V_{f\ CRIT}$ using copper figures from this work.

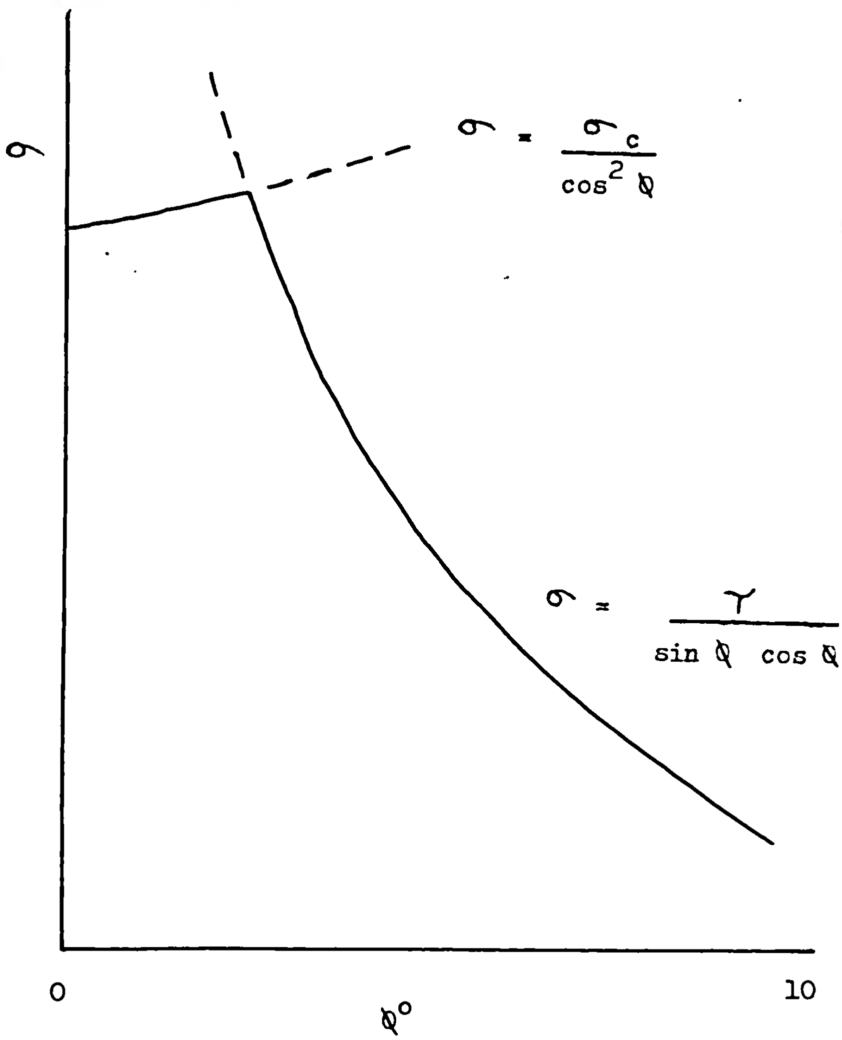


FIG. 5

Theoretical relation between breaking stress on composite and angle between applied stress and fibres.

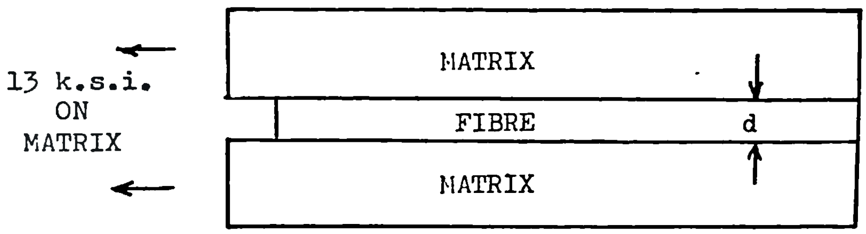


FIG. 6

Dow model

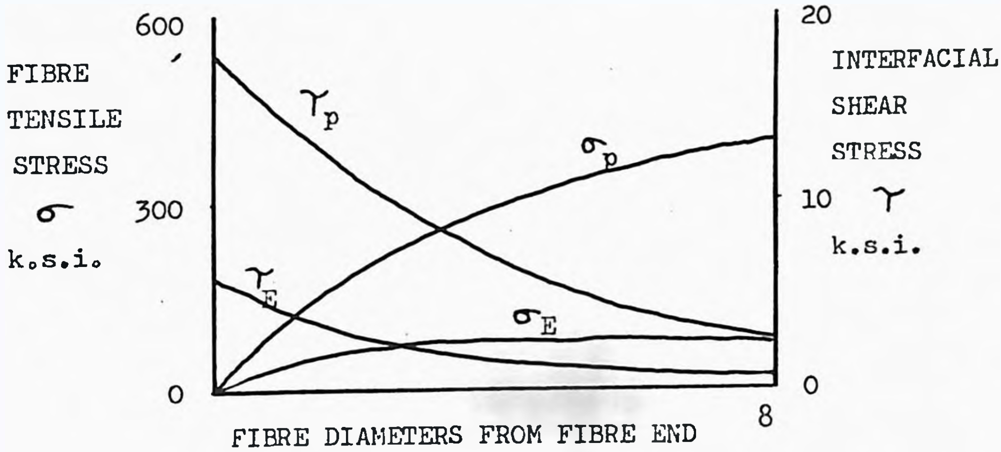


FIG. 7

Stress distributions on Dow model.

σ_p = Tensile stress in fibre with plastic matrix

σ_E = Tensile stress in fibre with elastic matrix

τ_p = Shear stress at interface with plastic matrix

τ_E = Shear stress at interface with elastic matrix.

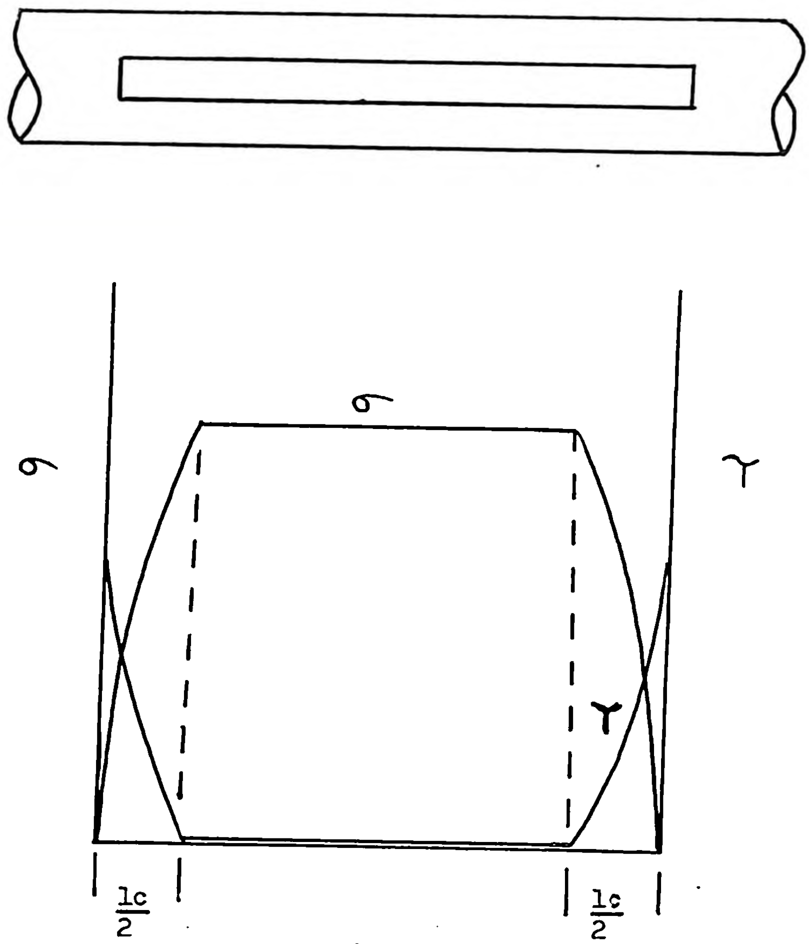


FIG. 8

Stress distributions around a fibre.

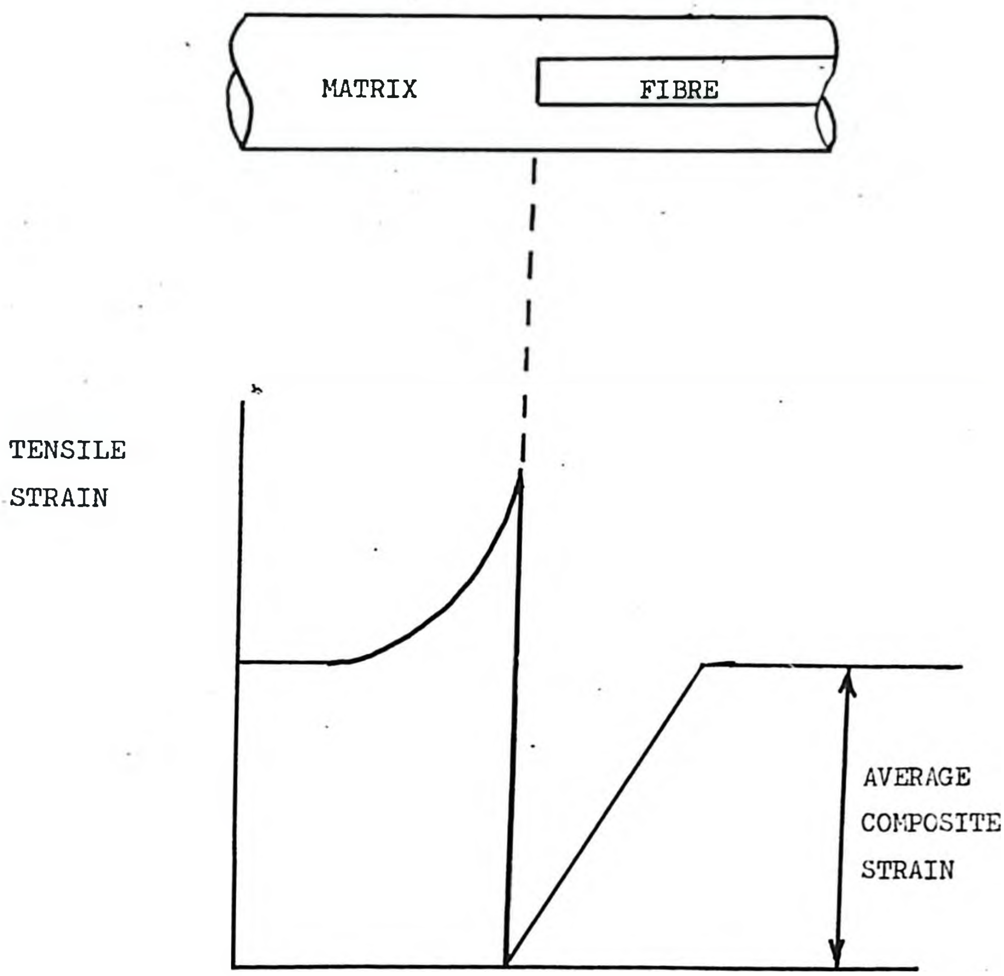


FIG. 9

Strain distribution at end of fibre.

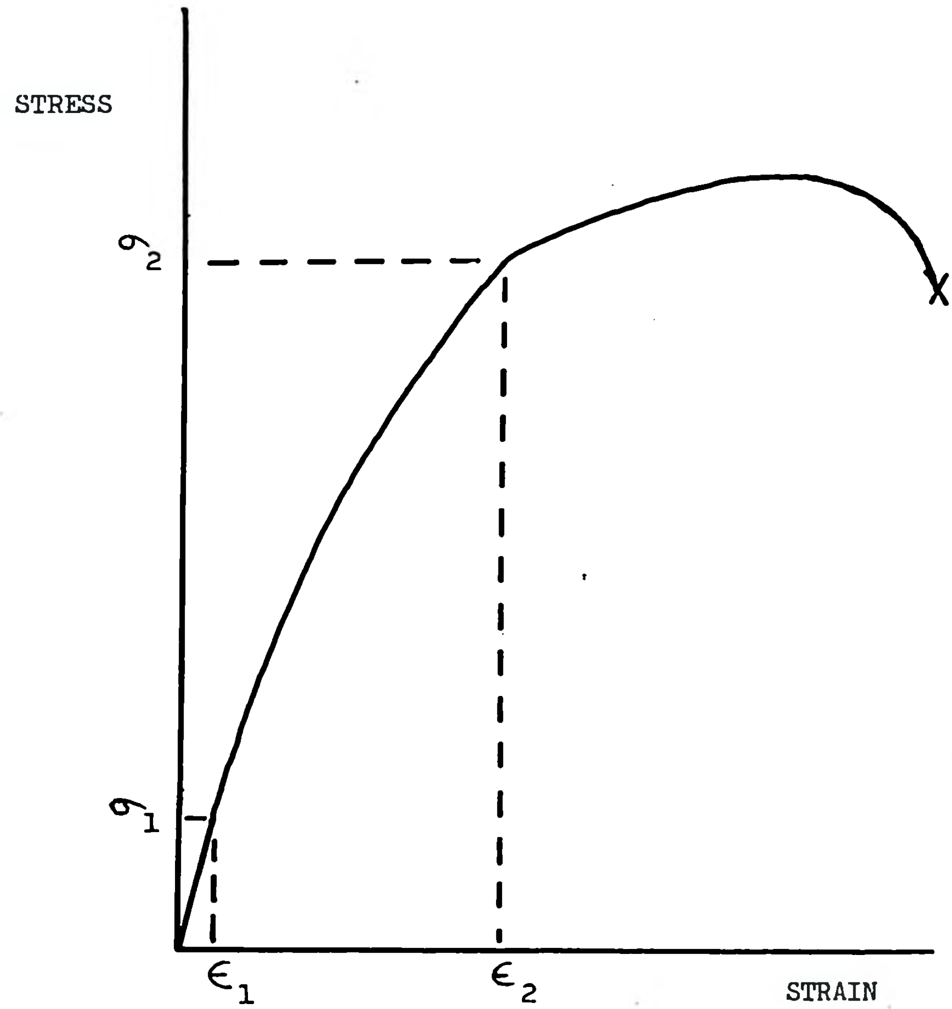


FIG. 10

Composite stress - strain curve.

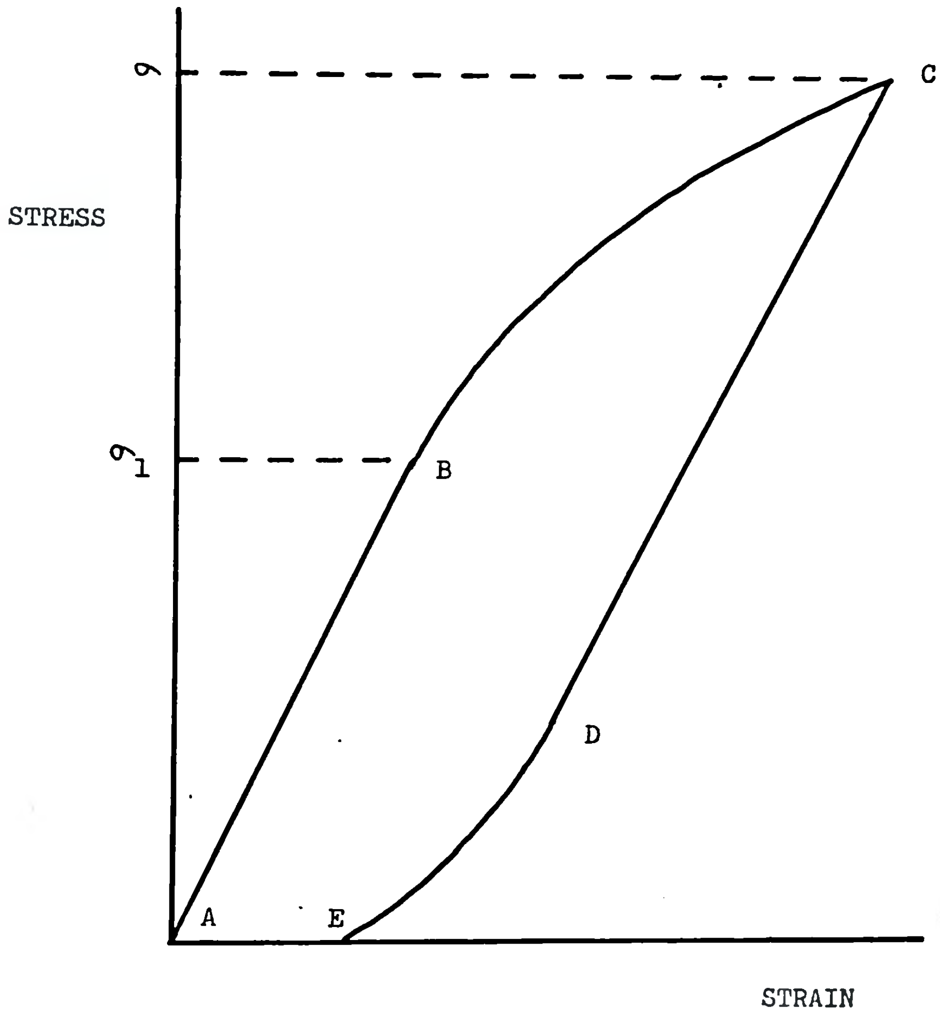


FIG. 11

Composite behaviour during one load cycle.

A - B Elastic extension of fibres and matrix.

B - C Elastic extension of fibres, and plastic extension of matrix.

C - D Elastic relaxation of fibres, elastic relaxation and compression of matrix.

D - E Elastic relaxation of fibres, plastic flow of matrix.

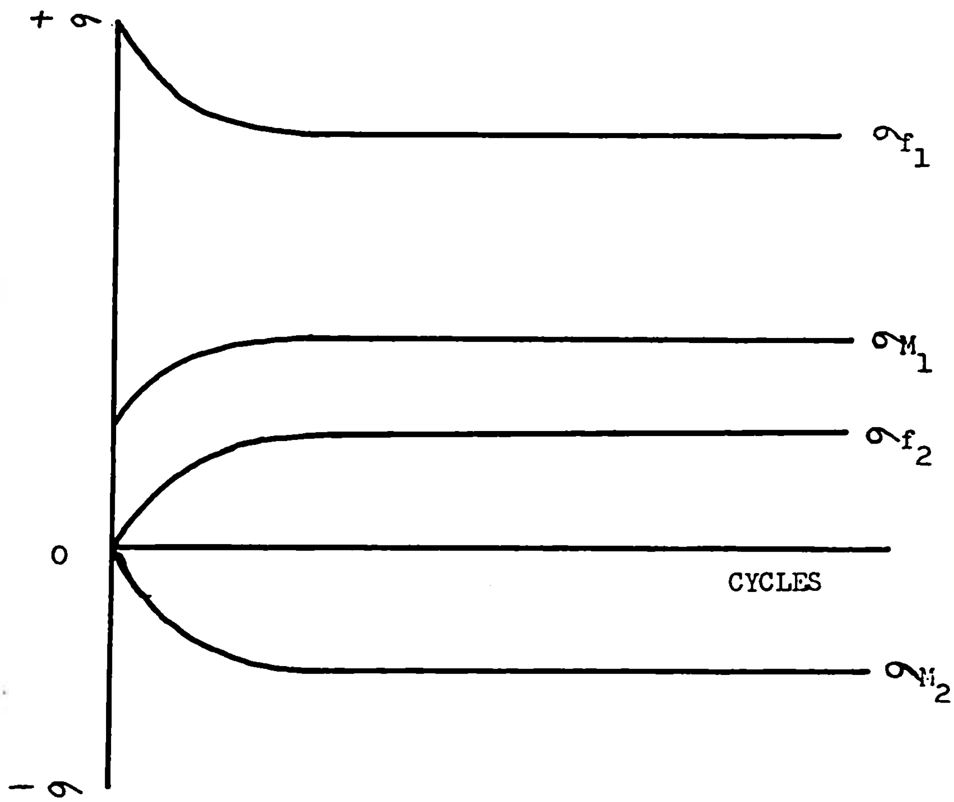


FIG. 12

Stress changes during initial load cycling after first half cycle.

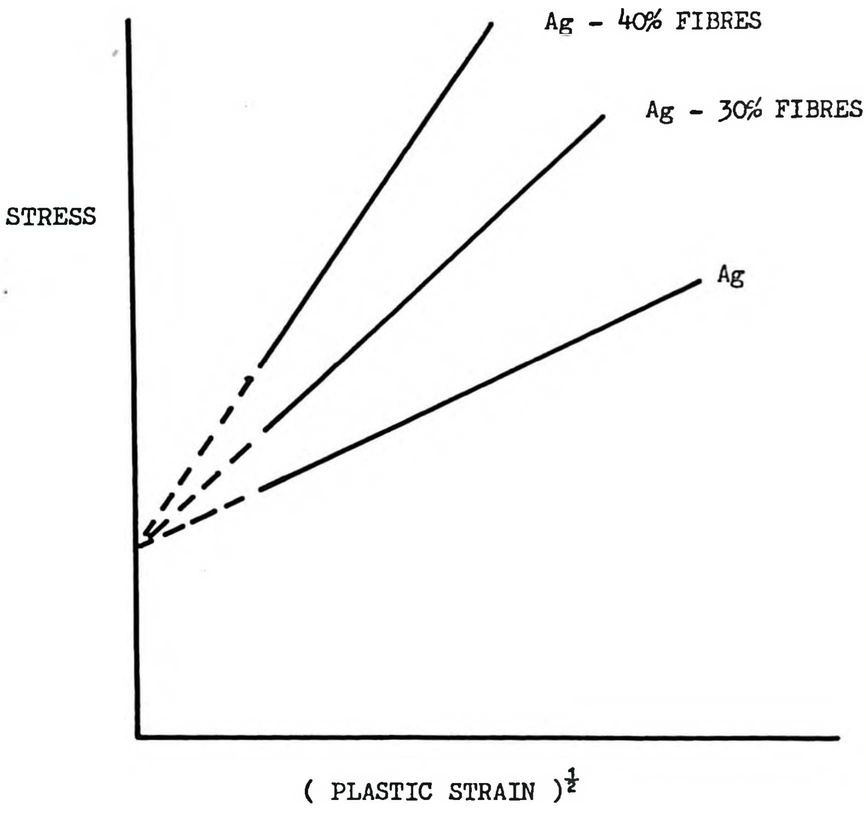
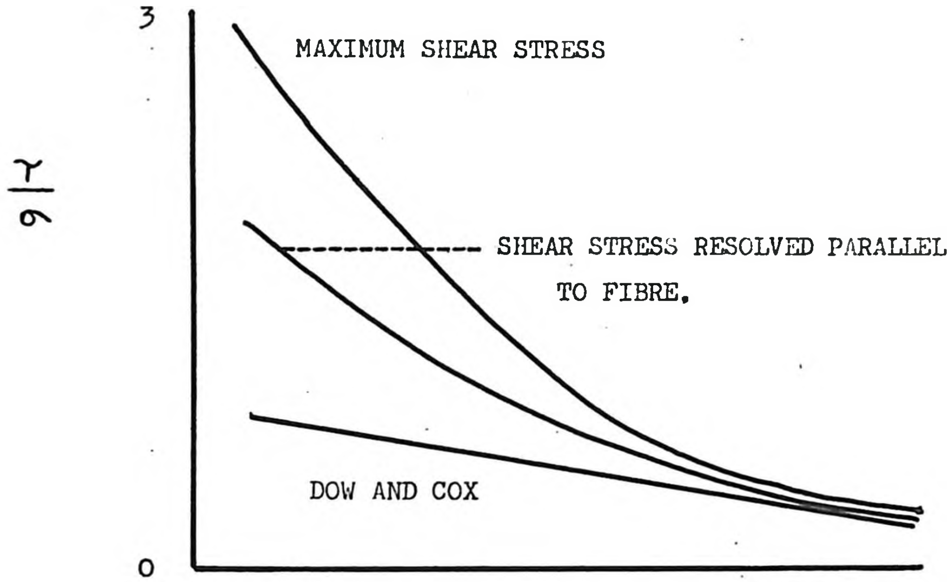


FIG. 13

Work hardening of silver matrix composites.

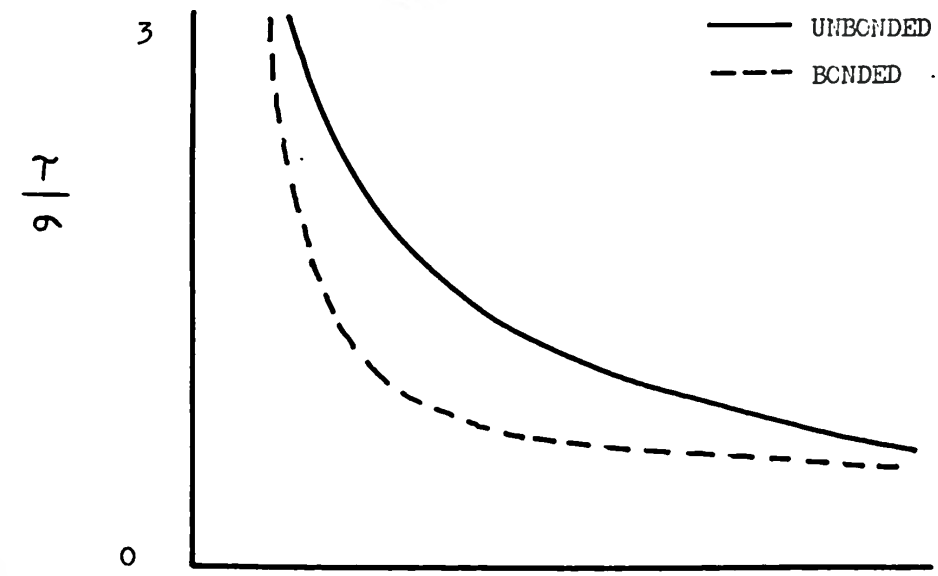
(Koppenaal and Parikh)



DISTANCE FROM FIBRE END

FIG. 14

Experimental stress distribution at fibre end. (Tyson and Davies).



DISTANCE FROM FIBRE END.

FIG, 15

Experimental stress distribution with and without end bonding. (Tyson and Davies).

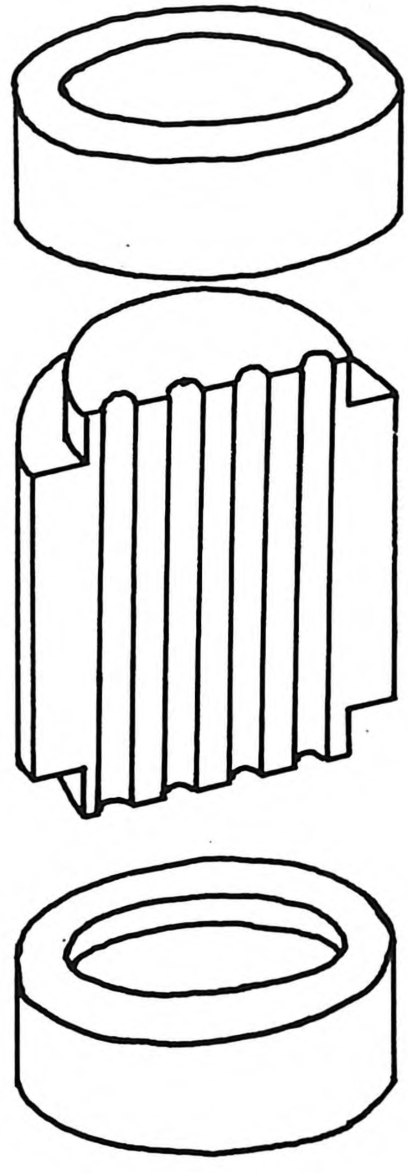


FIG. 16

Graphite split mould.

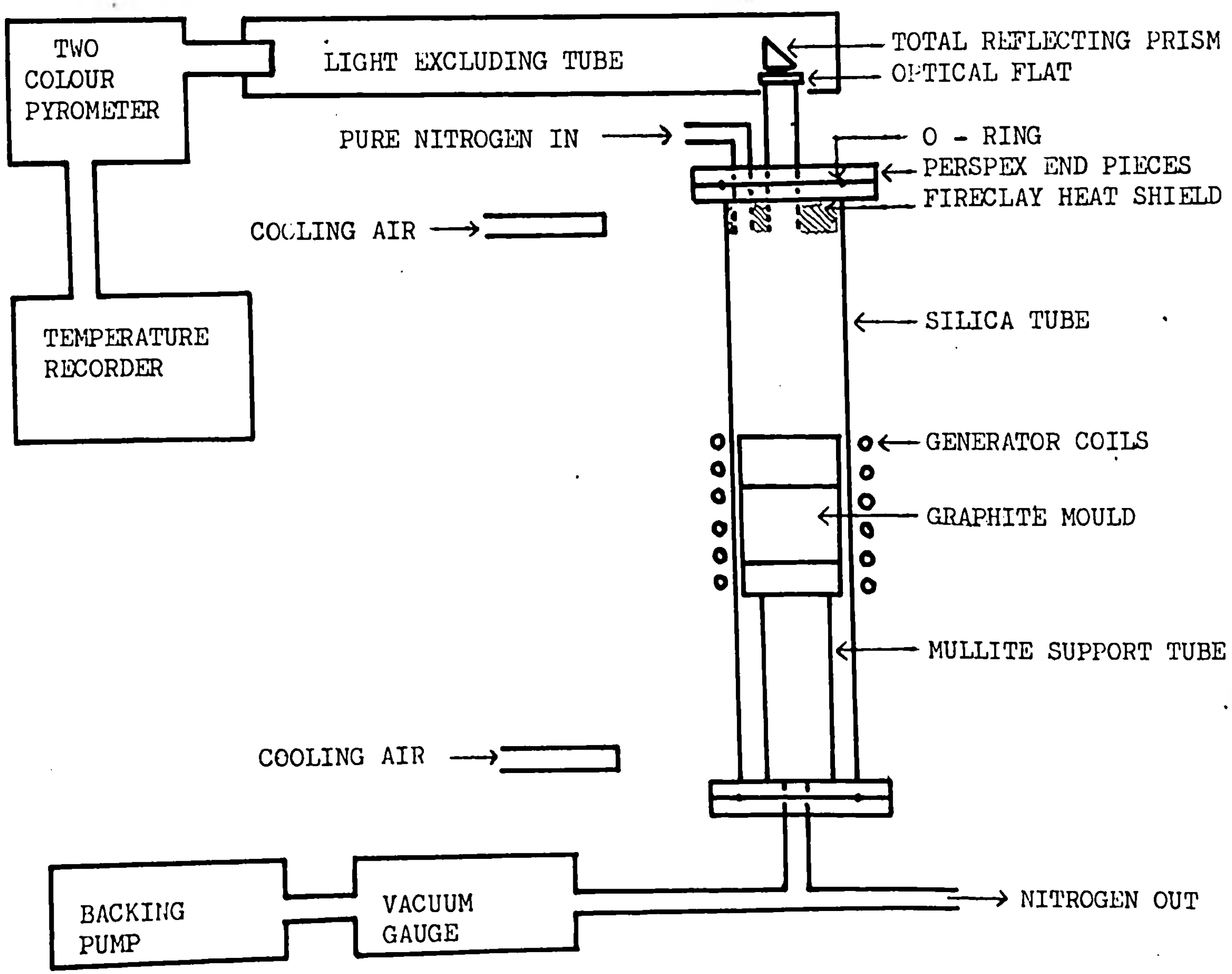


FIG. 17.
Diagram of apparatus

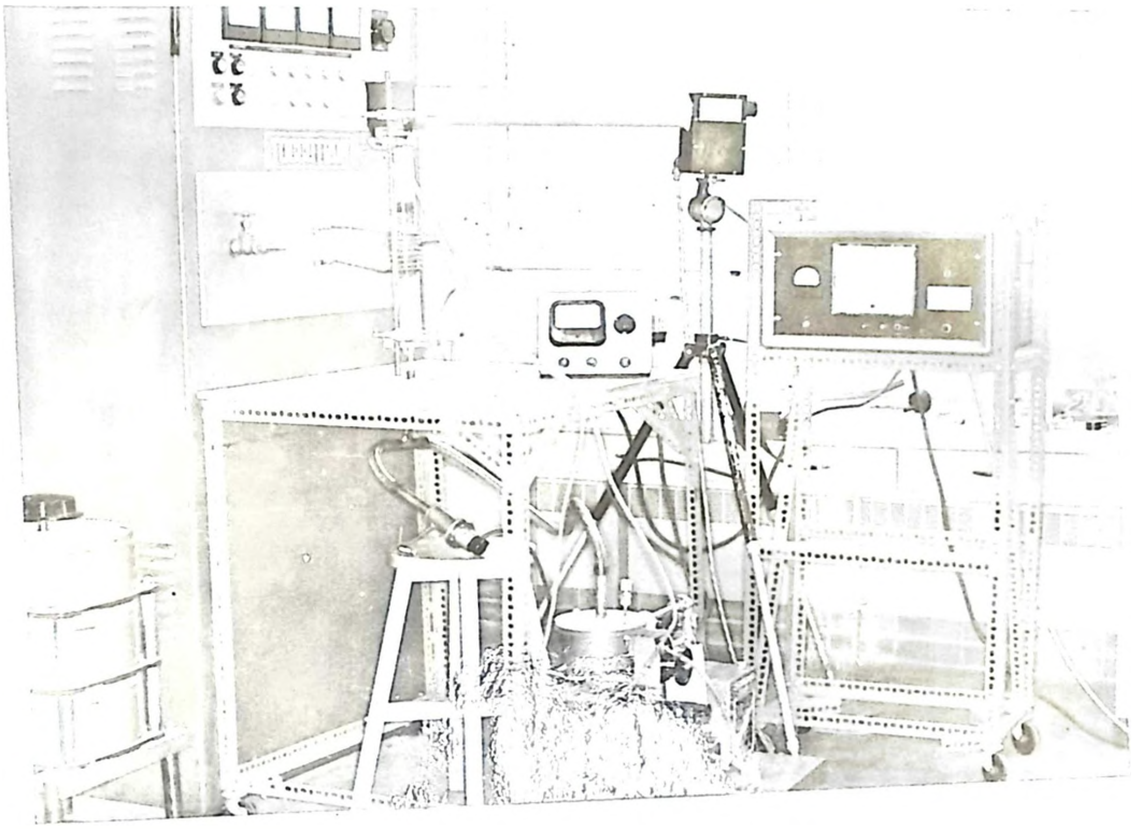


FIG. 18

Whole assembly of apparatus.

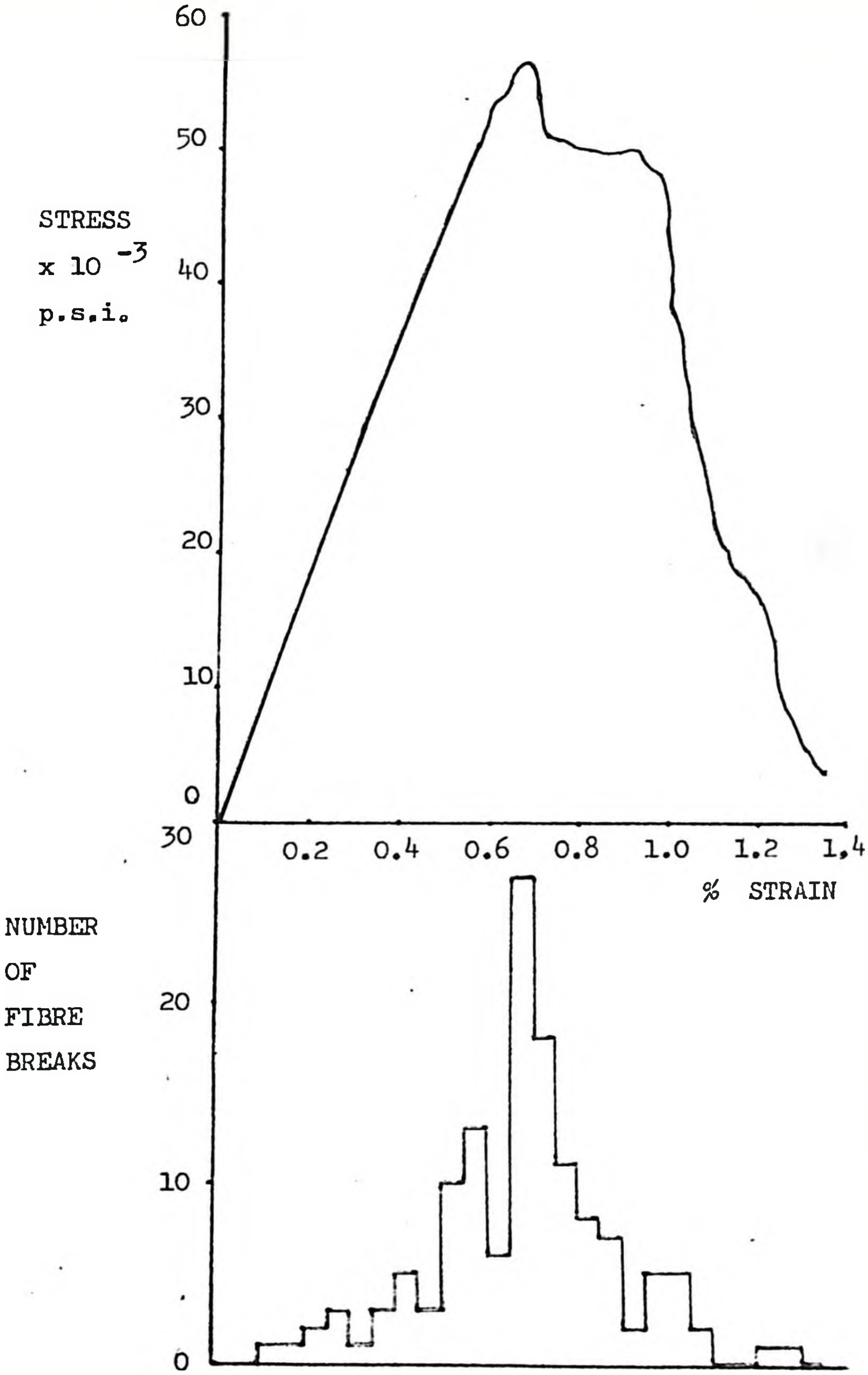


FIG. 19

Approximate stress - strain curve for Cu 23% W sample, with histogram of associated fibre breaks.

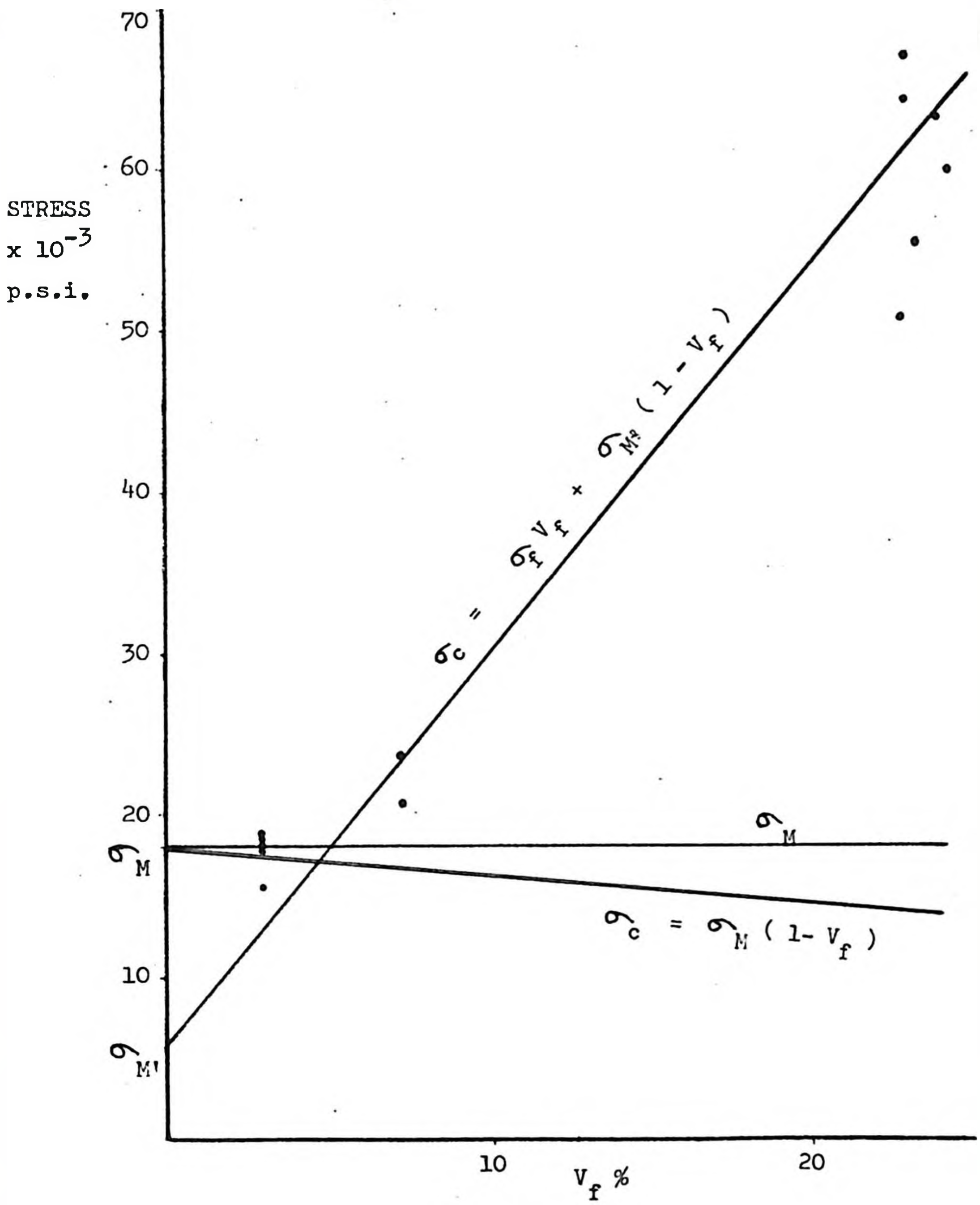


FIG. 20

Experimental variation between stress and V_f.

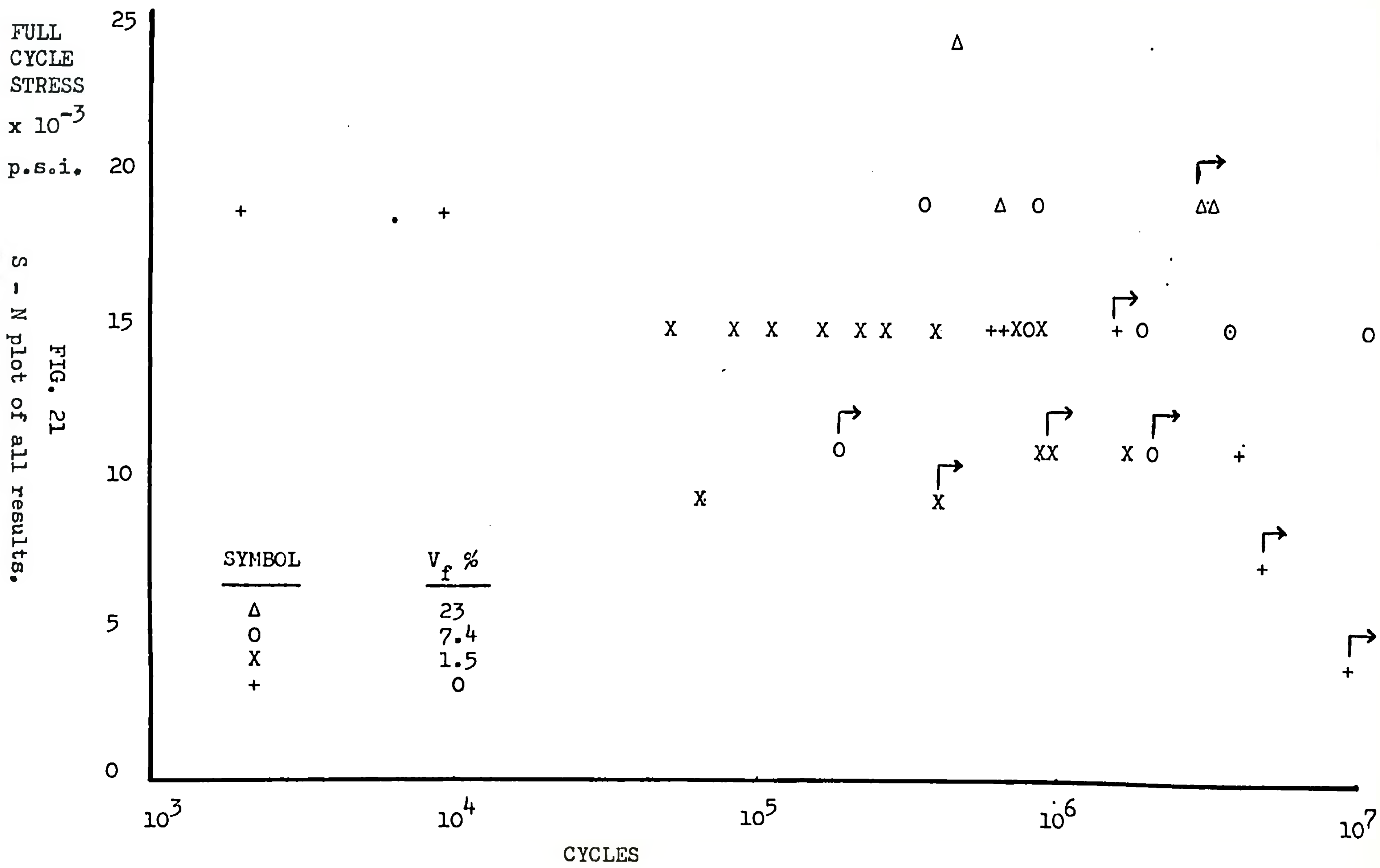


FIG. 21
S - N plot of all results.

x 12

FIG. 22(a)

Transverse section of Cu - 23% W.

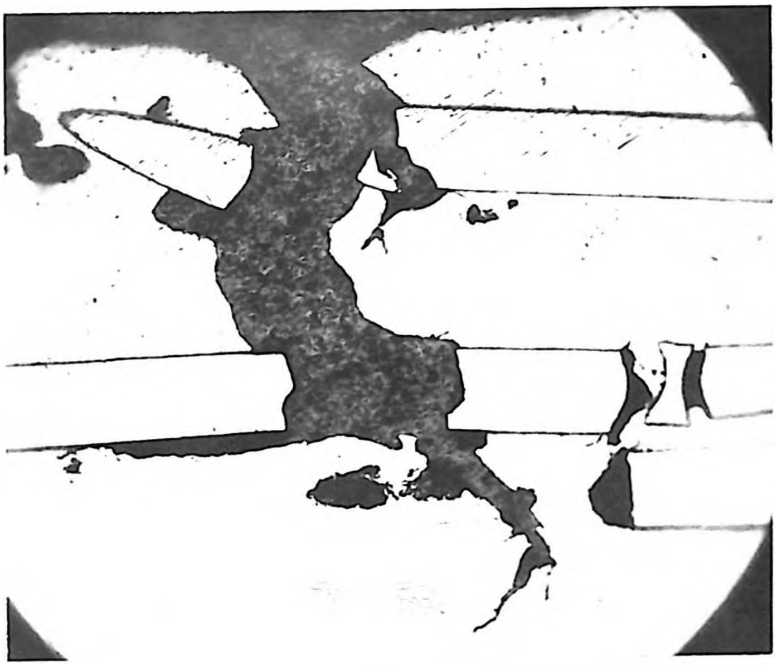
Fig. 22(b) X 60

Typical transverse section of Cu-23%W, etched in ferric chloride to show a grain boundary.

x 156

FIG. 23

Double fibre break showing propagating cracks



x 47

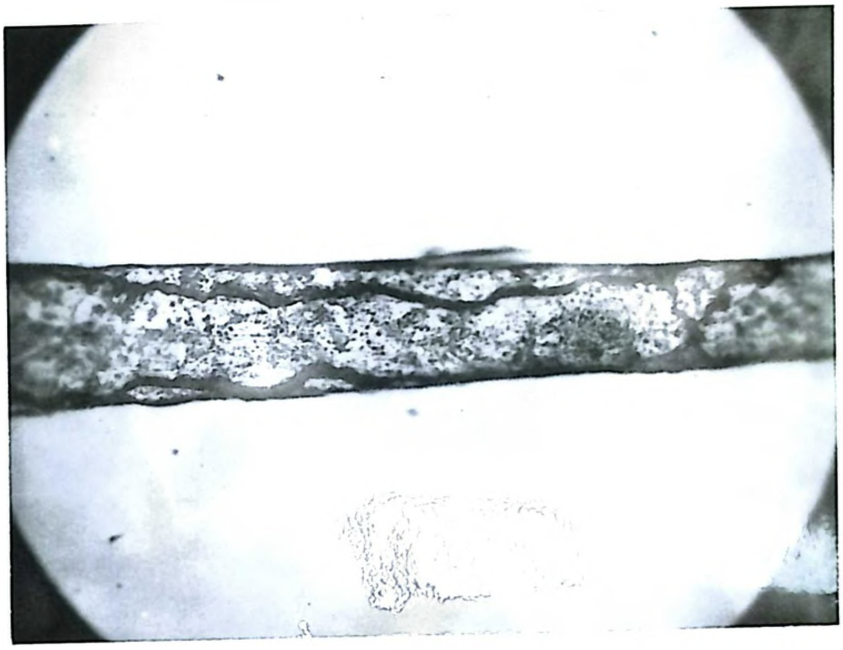
FIG. 24

Fatigue crack running through two fibre breaks.

x 21

FIG. 25

Propagation of a fatigue crack along an interface to a break.



x 340

FIG. 26

Fatigue crack tips running parallel to fibres.

x 60

FIG. 27

Tributary fatigue cracks crossing axis of a taper sectioned fibre.

TABLE 1

Strengths and elastic moduli of some strong fibres.

Material (Fibres)	U.T.S. lb in^{-2} x 10^{-6}	E lb in^{-2} x 10^{-6}
Graphite whisker	3.0	98
Al_2O_3 whisker	2.2	76
Fe whisker	1.8	28
Si_3N_4 whisker	2.0	55
SiC whisker	3.0	100
Asbestos	0.85	27
Mica	0.45	33
Soda Glass	0.40	9.8
Silica	1.5	10.5
Boron Glass	0.35	55
.9C Steel	0.575	30
β Ti	0.320	17
18-8 Stainless	0.347	29
Mo	0.300	53
W	0.420	50

TABLE 2

Strengths of 0.010" diameter Tungsten wires.

Condition		U.T.S (σ_f) p.s.i. x 10^{-3}	Failure
As received		335	Brittle
"		365	"
"		373	"
"		386	"
"		382	"
"		386	"
"		382	"
"		381	"
"		377	"
"		381	"
Extracted from composites		293	
"	"	280	
"	"	268	
"	"	245	
"	"	286	
"	"	293	
"	"	255	
"	"	242	
"	"	285	Ductile
"	"	295	"

TABLE 2 (cont'd)

Condition	U.T.S (σ_f) p.s.i. x 10^{-3}	% Strain at U.T.S.	Failure
Extracted from composites	227	0.8	Brittle
	197	0.3	"
"	290	2.1	Ductile
"	209	0.6	Brittle
"	284	2.3	Ductile
"	278	1.5	Brittle
"	189	0.5	"
"	210	0.9	"
"	248	1.4	"
"	246	0.8	"
"	192	0.5	"
"	226	0.9	"
"	206	1.0	"
"	223	0.5	"
"	265	1.3	"
"	229	0.7	"
"	240	1.2	"
"	277	1.2	"
"	242	1.0	"
"	246	0.9	"
"	264	1.1	"
"	191	0.6	"
"	282	1.7	Ductile

TABLE 3

Strengths of Composites

Number of Wires	V_f	U.T.S. p.s.i.	σ_{M}' measured
0	0	17,500	
0	0	17,700	
0	0	20,100	5,100
0	0	17,400	4,950
0	0	20,500	6,400
20	2.9	18,300	
20	2.9	18,800	
20	2.9	18,550	
20	2.9	15,500	
20	2.9	17,900	
50	7.4	20,650	
50	7.4	23,600	
157	23.0	67,300	
77	23.0	60,200	
77	23.0	63,400	
77	23.0	50,800	
77	23.0	64,600	

TABLE 4

Fatigue of Composites

V_f	Max stress p.s.i.	Cycles to failure $\times 10^{-3}$	Remarks
0	3740	9,613	Unbroken
0	7480	4,981	"
0	11,200	4,295	Broken
0	14,900	1,587	"
0	14,900	605	"
0	14,900	631	"
0	18,700	9	"
0	18,700	2	"
1.5	9,350	68	Pores
1.5	9,350	390	Unbroken
1.5	11,200	834	Pores
1.5	11,200	1,655	Unbroken
1.5	11,200	950	"
1.5	14,900	108	Broken
1.5	14,900	51	Pores
1.5	14,900	442	Broken
1.5	14,900	891	"
1.5	14,900	712	"
1.5	14,900	409	"
1.5	14,900	161	"
1.5	14,900	225	"

TABLE 4 (cont'd)

V_f	Max stress p.s.i.	Cycles to failure x 10^{-3}	Remarks
1.5	14,900	85	Pores
1.5	14,900	267	Broken
7.4	11,200	180	Unbroken
7.4	11,200	2,222	Unbroken
7.4	14,900	3,667	Broken
7.4	14,900	728	Broken
7.4	14,900	1,837	Broken
7.4	14,900	10,112	Broken
7.4	18,700	817	Broken
7.4	18,700	355	Broken
23.0	18,700	3,165	Unbroken
23.0	18,700	616	Broken
23.0	18,700	3,204	Broken
23.0	22,400	380	Broken

TABLE 5

Fatigue tests on Extracted Fibres

	Fatigue Machine	Minimum Load lbs	Minimum Stress p.s.i. $\times 10^{-3}$	Maximum Load lbs	Maximum Stress p.s.i. $\times 10^{-3}$	Cycles $\times 10^{-6}$	Comments
1	UTIA	1	12.7	7	89	1.000	Unbroken
2	UTIA	0	0	10	127	3.850	Unbroken
3	UTIA	1	12.7	13	166	1.000	Unbroken
4	UTIA	1	12.7	15	191	3.400	Unbroken
5	UTIA	1	12.7	18	230	1.000	Unbroken
6	UTIA	2	25.4	20	255	0.223	Broken
7	UTIA	2	25.4	20	255	2.200	Broken
8	UTIA	2	25.4	20	255	0.387	Broken
9	Instron	0.7	9	11.6	148	0,125	Unbroken

TABLE 6

Fatigue Strengths of Composites.

V_f %	Mean U.T.S. p.s.i.	Full Cycle Fatigue Strengths at 10^6 with 95% confidence limits. p.s.i.	Fatigue Ratio
0 - 1.5	18,600	14,812 $\pm$ 3,005	0.80
7.4	22,225	16,654 $\pm$ 3,723	0.75
23	60,300	20,241 $\pm$ 3,051	0.34