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The Fatigue of Fibre-Reinforced Aluminium

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In order to make use of the available strength of strong fibres, the metal matrix of a composite will have to undergo plastic deformation. The effect that this will have on the fatigue behaviour and damping capacity is discussed, with reference to aluminium reinforced with silica fibres and with stainless steel wires.

1. General Introduction

To make use of the available strength of the fibres in a fibre-reinforced metal component, some plastic yielding of the metal matrix must be expected. For instance in a composite with fibres having a modulus of 35×10^3 kg/mm^2 (5×10^7 psi) and a usable strength of 280 kg/mm^2 (4×10^5 psi), to utilise only half of the strength available from the fibres a strain of 4×10^{-3} must arise in the composite. Whilst elastic strains of this magnitude could conceivably be attained by a suitable metal matrix at ambient or modest temperatures [at a stress of approximately 84 kg/mm^2 (1.2×10^5 psi) for a metal such as nickel] there are a number of reasons why this is undesirable.

In general the fibre and the matrix will have large differences in the coefficients of thermal expansion and it is desirable that these are able to be accommodated by yielding of the matrix, otherwise large internal stresses and distortion would occur in the composite; exactly the same considerations apply to other mechanical incompatibilities such as differences in Poisson's ratios. Although fibre-matrix interface failure, or delamination, has been suggested to be the best method of obtaining toughness in fibre-reinforced materials [1–3] the ability of the matrix to plastically deform and so blunt, and absorb energy from, a crack in a fractured fibre is a very important alternative or second line of defence. Conversely, the stress at the tip of a propagating crack in the matrix is determined by the yield strength of the matrix [4] and this, in the absence of weak interfaces, will influence the tendency to failure of the fibres. Where the fibres have similar moduli, preferential fibre loading can only occur by plastic yielding of the matrix [3]. Finally, it is not possible to avoid plastic deformation in the matrix when a composite component is subjected to creep conditions.

Thus, if we ignore for a moment the problem of matrix fatigue, it is desirable to have a matrix of high yield stress only when the fibres are relatively weak, highly dense, or when it is not possible to obtain sufficiently high transverse properties in the component by controlling the orientation of the fibres.

Plastic deformation of the metal matrix leads to non-linear stress/strain behaviour in the composite [3, 5]. For instance Baker and Cratchley [3] have shown that the stress/strain behaviour of the composite aluminium reinforced with silica fibres [both components in this composite have a modulus of 7×10^3 kg/mm^2 (10^7 psi)], after prestressing, consists of a closed hysteresis loop reproducible at least for hundreds of stress cycles. Although this behaviour is somewhat more complicated than the behaviour normally encountered in engineering materials, it does have the advantage of giving a high damping capacity. The main disadvantage is that this behaviour will eventually result in matrix fatigue damage.

In this paper the fatigue behaviour of pure aluminium reinforced with silica and with stainless steel fibres will be considered in the light of some recent results.

2. Fatigue Behaviour

2.1. Testing Methods and the Failure Criteria

In composites reinforced with continuous fibres the way in which fatigue damage will be observed depends very much on the mode of fatigue-testing. Many investigators have used tension/tension cycling [6–9] to avoid the complication of compression loading. Clearly under these conditions, complete fracture of the composite is the most likely criterion of failure. Thus this test measures the fatigue properties of the fibres in the presence of the matrix. If, however, the mode of fatigue-testing involves compression, for instance in reversed or rotating bending or in a straight tension/compression test, loss of compressive properties will be the most likely criterion of failure, provided that the fibres are not easily fractured by the matrix cracks. In this instance we are measuring the ability of the composite to perform as a composite rather than as a bundle of fibres; this is thus strongly dependent on the fatigue behaviour of the matrix. There may be other failure criteria, for instance the appearance of a significant crack in the matrix [10]. This measures the properties of the matrix in the presence of the fibres and, although a difficult criterion to use for practical reasons, extremely important in a metal-matrix, filament-wound pressure vessel where gas tightness is required. It is thus not possible to compare the results of different modes of testing directly. In the work described in this paper the fatigue testing was carried out under conditions of reversed plain bending; for details of the composite fabrication techniques used and the fatigue testing methods used, see [11–13]. Although in many respects reversed bending is not the most ideal method of fatigue testing these materials*, it does at least measure the ability of the fatigue-damaged material to behave as a composite.

3. Reversed Bending Fatigue of Reinforced Pure Aluminium

Two fibre reinforcements have been used: (i) silica fibres, continuous, 5×10^{-2} mm (2×10^{-3} in.) diameter [11], (ii) stainless steel, continuous [12] and discontinuous [13], 5×10^{-2} mm (2×10^{-3} in.) diameter.

3.1. Previous Observations

At low stress amplitudes, failure in the aluminium/silica system [11] occurred mainly by a process of gradual fragmentation of the matrix owing to fatigue crack propagation in the matrix. This resulted in a gradual decay in flexural stiffness, followed finally by a more rapid decay when the specimen was considered to have failed. Fatigue cracks, at least at relatively low stress amplitudes, could not penetrate the fibres and were deflected along and around the fibre. The process of failure can [be most simply visualised as one of the gradual reversion of the maximum stress surface regions to a bundle of fibres. Thus a large degree of tensile strength was retained even after failure. This situation was complicated by the presence of fibres damaged by the composite manufacture process; matrix fatigue cracks could propagate directly through these regions.

At high stress amplitudes, one or two gross fatigue cracks were produced in the matrix and these apparently propagated directly through the fibres; however, it was not possible to ascertain whether or not this effect was associated with predamaged fibres.

In most respects the continuous fibre aluminium/stainless steel system performed in an almost identical fashion, with the most important exception that direct crack propagation at the fibre-matrix interface occurred. It was found that this effect could be minimised by producing a limited amount of reaction between the fibre and the matrix during composite manufacture [12] which resulted in the formation of an intermetallic compound at the interface. This treatment gave the optimum fatigue properties, which was largely attributed to improved mechanical keying of the fibres into the matrix and the more difficult interfacial crack propagation produced by the irregular nature of the intermetallic compound at the interface, rather than an improvement in bond strength. At higher stress amplitudes, fibres were fractured by gross matrix fatigue cracks; this was particularly noticeable in reacted specimens owing to the stress-concentrating effect of the intermetallic compound when cracked.

*For instance, the use of elastic stress analysis in a composite in which the matrix deforms plastically must lead to some error, although comparative bending and tensile measurements on the same specimen have suggested that this error is small [14]. The fatigue specimens were produced by file-shaping the hot-pressed material [11, 12] (as against directly forming the shape). This results in a tendency to splitting between the fibres and prevents failure in the region of theoretical maximum stress; this problem is found to a large extent in all methods of fatigue-testing these materials.

In composites having discontinuous fibres [13] it was found that in addition to the processes of failure in specimens reinforced with continuous fibres, the presence of fibre ends causes additional complications. Fatigue damage occurs at fibre ends in the surface and sub-surface regions due to the processes of load transfer. In addition, the region of high strain at fibre ends facilitates the direct propagation of fatigue cracks through the matrix.

The fatigue mechanism in these composites is therefore complex. However, in many respects matrix fatigue damage is basic to the final failure. This damage is caused by the cyclic plastic deformation imposed on the matrix during fatigue stressing of the composite. In the work just described it was found that the theoretical plastic-strain-range endured by the matrix was a very useful criterion of the damage imposed on the composite by the external stress and allowed a good fit to the fatigue results. For a given matrix and stress, the theoretical plastic-strain-range is dependent on the fibre-volume-fraction, modulus and aspect ratio. However, the fatigue properties of the composites would be expected to be superior to those of the unreinforced matrix *at a given plastic-strain-range*. This is because the fibres have the additional effects of (i) impeding the propagation of matrix fatigue cracks, and (ii) supporting the fatigue-damaged matrix. This was found to be the case in the work just described [11–13] except at very high values of plastic strain, where the properties of the composite approached those of the unreinforced matrix (of course on the basis of stress the composite is always superior).

For the *unreinforced* matrix, the fatigue life at plastic strains above about 0.1% may be predicted by empirical relationships, such as that proposed by Coffin and Tavernelli [15].

$$N_f^{\frac{1}{2}} \Delta\epsilon_p = C \qquad (1)$$

where N_f is the number of cycles to failure, $\Delta\epsilon_p$ the plastic-strain-range, and C a constant approximately equal to half the true fracture strain. This is a very valuable relationship for comparing the behaviour of the reinforced and unreinforced matrix.

More recent work on the effect of fibre-volume-fraction and fibre modulus on the fatigue properties has confirmed the usefulness of this criterion.

3.2. Effect of Fibre-Volume-Fraction Variation

For this investigation two series of aluminium/silica specimens were produced, having approximately 45% and 25% volume-fractions of 5×10^{-2} mm (2×10^{-3} in.) diameter fibres. Special precautions were taken in composite manufacture to avoid fibre damage in the 45% volume-fraction series [16]. In fig. 1 the results are plotted on the basis of stress and in fig. 2 on the basis of the plastic-strain-range criterion. Thus it is shown that the use of this later criterion allows a good superposition for the results for the two different volume-fractions at lives of over 2×10^5 cycles. The reason for the higher fatigue properties of the low-volume-fraction specimen at lives less than 2×10^5 cycles is not known. It would appear from this result that the numbers of fibres present in the range studied do not greatly influence the fatigue properties. However, on the basis of the Coffin relationship [15], in all but the most highly stressed specimens a substantial increase in life over that predicted for the unreinforced matrix was obtained. There must therefore be a critical fibre content or fibre spacing for good fatigue resistance in terms of the numbers of fibres present to interact with the propagating fatigue crack; for very small fibres this content in terms of volume-fraction may be very small. This situation may not hold for a stiffer or more rapidly work-hardening matrix since in this case fatigue cracks can apparently propagate directly through the fibres [7], particularly if the the fibres are weak. If it is assumed that the crack is slowed down on its path through the fibres, in this case, the fatigue properties would be expected to be strongly dependent on fibre number.

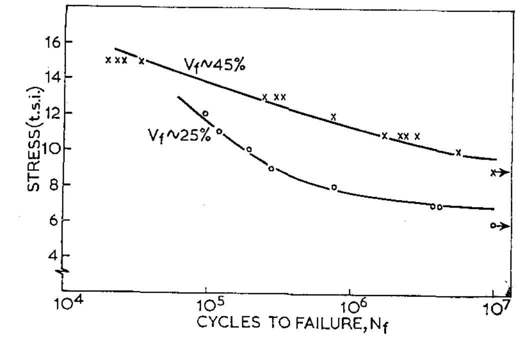


Figure 1 Stress versus log cycles to failure for aluminium/silica composites having 45% and 25% volume-fraction of 5×10^2 mm (2×10^{-3} in.) diameter fibres. *Note* tsi units are used in this diagram. 1.0 tsi = 1.0 ton/in.^{2} = 1.57 kg/mm^2.

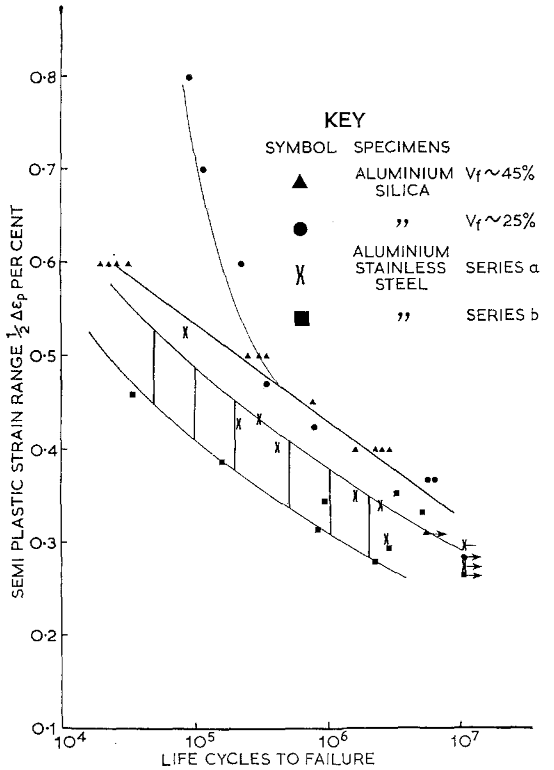


Figure 2 Semi-plastic-strain-range, calculated from the idealised model (see section 4) versus log cycles to failure for aluminium/silica and aluminium/stainless steel composites with 5×10^{-2} mm (2×10^{-3} in.) diameter fibres.

3.3. The Effect of Fibre Modulus Variation

If the plastic-strain-range criterion holds and if there are no other differences between the two composites, the fatigue results for silica [$E_f = 7 \times 10^3$ kg/mm² (10^7 psi)] and for stainless steel [$E_f = 19.6 \times 10^3$ kg/mm² (2.8×10^7 psi)] reinforcement on the basis of this criterion should be identical. The results of this comparison are shown in fig. 2. (The aluminium/stainless steel composites have a range of fibre-volume-fraction of 20 to 33% and again demonstrate the effectiveness of this criterion.) It is seen that some separation exists between the two sets of results; however, this separation is very much less than was obtained on the basis of fatigue strength. For example composites with 25% fibre-volume-fraction gave fatigue strengths at 10^7 cycles of 20.4 kg/mm² (13 tsi) with stainless steel fibres and 10.2 kg/mm² (6.5 tsi) with silica fibres. It can be concluded from this result, and from observation that the aluminium/silica composites never suffered from interfacial failure, that the separation between the two sets of results was due to the superior bond strength of the aluminium/silica composites, rather than a failure in the applicability of the criterion. These conclusions suggest that, at least for a ductile matrix such as aluminium, the fatigue properties could be predicted for any fibres (provided they are resistant to matrix fatigue cracks) and for any fibre concentration.

4. Matrix Fatigue: Theoretical Considerations

Since the plastic-strain-range criterion appears to be so useful for assessing the fatigue behaviour of fibre-reinforced aluminium it is worth looking in more detail into the methods used to derive this criterion.

4.1. Simple Stress/Strain Model

The simple model for tension/compression stress/strain behaviour advanced by Baker and Cratchley [3] is used (fig. 3). It is assumed that a cyclic stress/strain curve for the matrix can be drawn as shown in fig. 4 and amplified in fig. 5. This is a curve of the peak stress in the matrix in any hysteresis loop versus the strain in the composite (two representative hysteresis loops

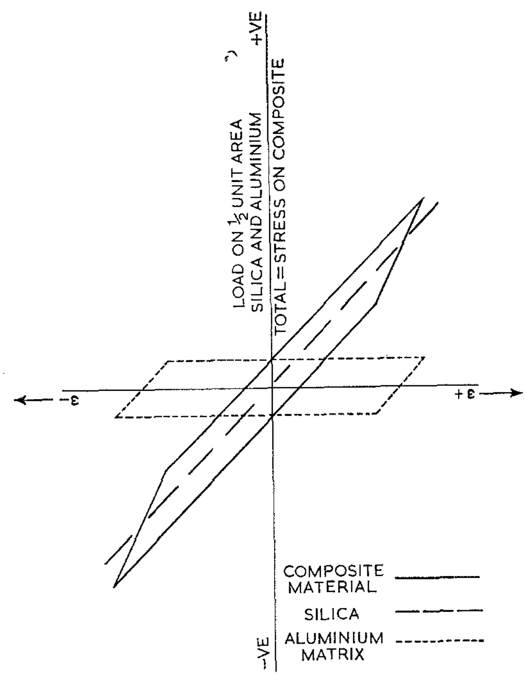


Figure 3 Schematic representation of the stress/strain behaviour for a full tension/compression cycle.

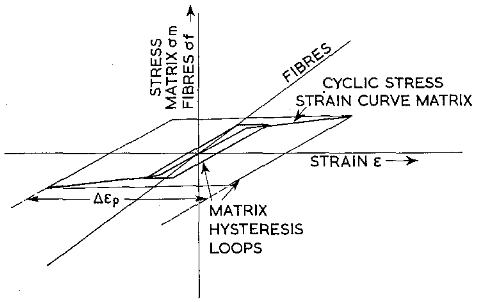


Figure 4 Schematic representation of matrix hysteresis behaviour for representative loops at two stress levels.

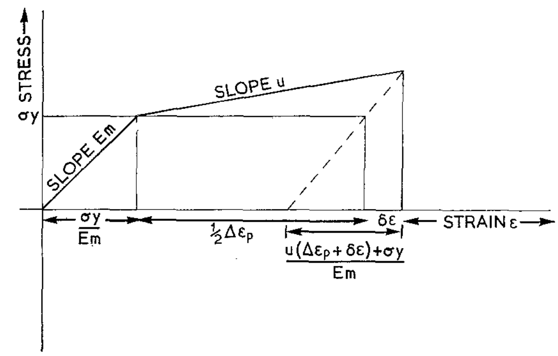


Figure 5 Schematic representation of matrix cyclic stress/strain curve.

at two stress levels are shown in 4) and is assumed to consist of two slopes, an initial linear region corresponding to elastic behaviour (no hysteresis loop) and a second linear region of a lower slope corresponding to plastic behaviour (where hysteresis loops form). Thus the stress in the composite σ_c may be divided into two parts: σ_c' at a composite strain of σ_y/E_m where the fibres and matrix behave elastically and σ_c'' above this strain where the fibres are elastic but the matrix plastic.

From fig. 5,

$$\sigma_c' = \frac{\sigma_y E_f V_f}{E_m} + \sigma_y (1 - V_f) \qquad (2)$$

where E_f and E_m are the moduli of fibres and matrix respectively, V_f the volume-fraction of fibres and σ_y the yield stress of the matrix and

$$\sigma_c'' = (\tfrac{1}{2}\Delta\epsilon_p + \delta\epsilon)\, E_f V_f + u(\tfrac{1}{2}\Delta\epsilon_p + \delta\epsilon)(1 - V_f)^* \qquad (3)$$

where u is the secondary slope of the cyclic stress/strain curve and $\delta\epsilon$ an increment of strain equal to $(\sigma_m/E_m) - (\sigma_y/E_m)$, σ_m is the maximum stress in the matrix.

*This is correct only at the tip of the hysteresis loop.

From fig. 5,

$$\delta\epsilon = \frac{\tfrac{1}{2}\Delta\epsilon_p u}{E_m - u} \qquad (4)$$

If $\Delta\epsilon_p$ is low and u small compared to E_m, $\delta\epsilon$ can be neglected. Since

$$\sigma_c = \sigma_c' + \sigma_c'', \Delta\epsilon_p = \frac{2\sigma_c - 2\sigma_y\,[(E_f/E_m)V_f + (1 - V_f)]}{E_f V_f + u(1 - V_f)} \qquad (5)$$

The use of the plastic-strain-range criterion as just discussed for metals with high rates of work-hardening will obviously involve considerable error because of the simplifying assumptions made in the stress/strain model.

4.2. Empirical Model

An alternative and more realistic approach is to make use of an empirical relationship found by Morrow and Tuler [17] between the plastic strain range and the applied stress for *unreinforced metals.* If the logarithm of the plastic-strain-range corresponding to the tips of pre-stabilised hysteresis loops is plotted against the logarithm of the stress range required to produce the loops, a straight line results, from which it is found that

$$\tfrac{1}{2}\Delta\epsilon_p = \epsilon_F \left(\frac{\sigma_m}{\sigma_F}\right)^{1/n'} = K\sigma_m^{1/n'} \qquad (6)$$

where $\Delta\epsilon_p$ is the plastic-strain-range, σ_m the applied stress, ϵ_F and σ_F are related to the true cyclic ductility and strength respectively, and n' is the work-hardening exponent of the cyclic stress/strain curve, shown to be about 0.15 for most of the metals investigated, regardless of their initial state, and K a constant equal to $\epsilon_F/\sigma_F^{1/n'}$.

To get the total stress/strain behaviour we must add to this equation the elastic component of the strain ($\tfrac{1}{2}\Delta\epsilon_e$) to obtain the total strain in one direction, $\tfrac{1}{2}\Delta\epsilon$.

We have that

$$\tfrac{1}{2}\Delta\epsilon = \tfrac{1}{2}\Delta\epsilon_e + \tfrac{1}{2}\Delta\epsilon_p \qquad (7)$$

and

$$\tfrac{1}{2}\Delta\epsilon_e = \sigma_m/E_m \qquad (8)$$

where E_m is the elastic modulus of the matrix. Substituting equations 6 and 8 in equation 7 we have that

$$\tfrac{1}{2}\Delta\epsilon = \frac{\sigma_m}{E_m} + K\sigma_m^{1/n'} \qquad (9)$$

This cyclic stress/strain relationship is similar to the idealised model previously assumed for the matrix and corresponds with maximum stress and strain that any prestabilised hysteresis loop attains.

To apply equation 9 to the behaviour of the composite material we assume the usual rule of mixtures [1]:

$$\sigma_c = \sigma_f V_f + \sigma_m (1 - V_f) = \tfrac{1}{2}\Delta\epsilon E_f V_f + \sigma_m (1 - V_f) \qquad (10)$$

where σ_c, σ_f and σ_m are the stresses in the composite, fibre and matrix respectively, E_f the modulus of the fibres and V_f the volume-fraction of fibres.

Substituting equation 9 into 10 we have that

$$\sigma_c = \sigma_m \left(\frac{V_f (E_f - E_m)}{E_m} + 1 \right) + K\sigma_m^{1/n'} E_f V_f \qquad (11)$$

Thus if we can find the value of σ_m corresponding with a value of σ_c this can be substituted back into equation 9 to obtain $\frac{1}{2}\Delta\epsilon_p$; it is assumed that the cyclic stress/strain behaviour of the matrix is unchanged by the presence of the fibres.

There are, however, two difficulties. The first is that no simple solution exists for a polynomial of this type and that the solution requires the use of a computer. The second is that the equations are extremely sensitive to the value taken for σ_F since this is raised to the power of 1/0.15.

This difficulty may be overcome by evaluating $\epsilon_F/\sigma_F^{1/n'} = K$ as a constant from stress/strain data on a composite with the relevant matrix. From equation 10 we have that

$$\sigma_m = \frac{\sigma_c - \frac{1}{2}\Delta\epsilon\, E_f V_f}{(1 - V_f)}$$

Then from an experimental plot of σ_c versus $\Delta\epsilon$, σ_m values may be obtained and substituted in equation 11 to obtain a value for K which (if found to be constant) may then be used to obtain a value for σ_m at any stress σ_c in the composite.

Using this procedure curves were produced from monotonic stress/strain data on a precycled aluminium/silica specimen* where it was found that the results obtained were indistinguishable from those of the simple model. Thus for composites with matrices having low rates of work-hardening the simple model is by far the most convenient to use, but the use of the empirical model may be of great value for composites with matrices having high rates of work-hardening.

4.3. Experimental Determination of the Plastic-Strain-Range

If the volume-fraction of fibres and the elastic moduli of the fibres and matrix are known it is possible to calculate the plastic-strain-range from a knowledge of the applied stress and secant modulus of the composite, taken from measurements on hysteresis loops [14].

We have that

$$E_c^s = E_f V_f + E_m^s (1 - V_f) \qquad (12)$$

where E_c^s and E_m^s are the secant moduli of the composite and matrix respectively

and $$E_m^s = \frac{\sigma_m}{\Delta\epsilon} = \frac{\sigma_m}{(\sigma_m/E_m) + \Delta\epsilon_p}$$

Substituting for E_m^s in equation 12,

$$E_c^s = E_f V_f + \frac{\sigma_m}{(\sigma_m/E_m) + \Delta\epsilon_p} (1 - V_f) \qquad (13)$$

Rearranging equation 10 and allowing for the complete hysteresis loop

$$\sigma_m = \frac{\sigma_c - V_f \Delta\epsilon E_f}{(1 - V_f)} = \frac{\sigma_c - (V_f\, \sigma_c/E_c^s)\, E_f}{(1 - V_f)} \qquad (14)$$

Rearranging equation 13

$$\Delta\epsilon_p = \sigma_m \left\{ \frac{(1 - V_f)}{E_c^s - E_f V_f} - \frac{1}{E_m} \right\}$$

and from equation 14, substituting for σ_m,

$$\Delta\epsilon_p = \frac{\sigma_c (1 - V_f\, E_f/E_c^s)}{(1 - V_f)} \left[\frac{(1 - V_f)}{E_c^s - E_f V_f} - \frac{1}{E_m} \right] \qquad (15)$$

Measurements of the plastic-strain-range were made from strain-gauged constant-stress reversed-bending fatigue specimens of aluminium reinforced with silica or with stainless steel fibres [14]. The development of hysteresis loops in aluminium reinforced with stainless steel fibres is shown in fig. 6; in general the measured values agreed fairly well, but were slightly lower than those calculated from the models described. The main disadvantage of this method was that due to the strain limitation of the gauges and, in bending, the tendency of the

*It has been shown [17] that the cyclic stress/strain curve and monotonic stress/strain curve on precycled metals are similar.

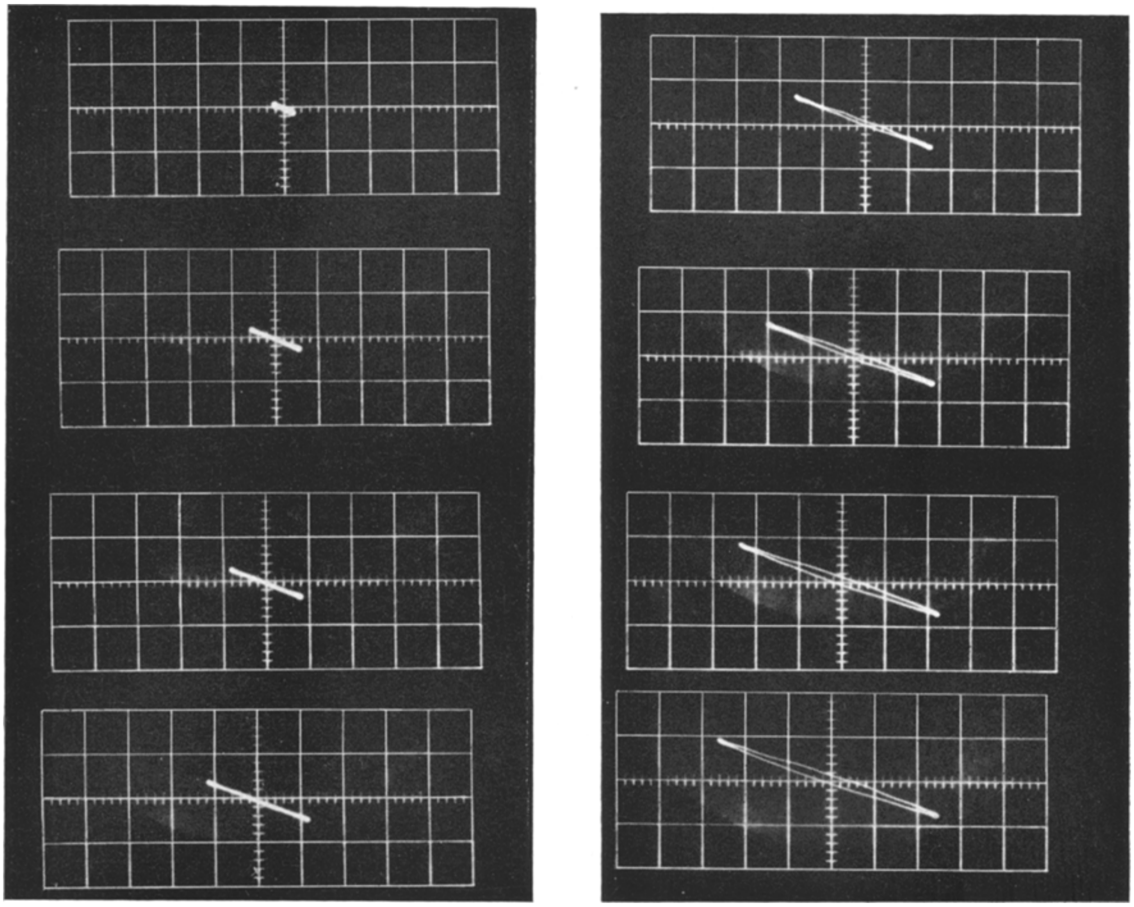


Figure 6 CRO photographs showing the development of dynamic hysteresis loops with increasing stress for an aluminium/stainless steel composite with 25% volume-fraction of fibres. Maximum stress approximately 17.5 kg/mm^2 (25×10^3 psi) first sign of open loop approximately 10.5 kg/mm^2 (15×10^3 psi).

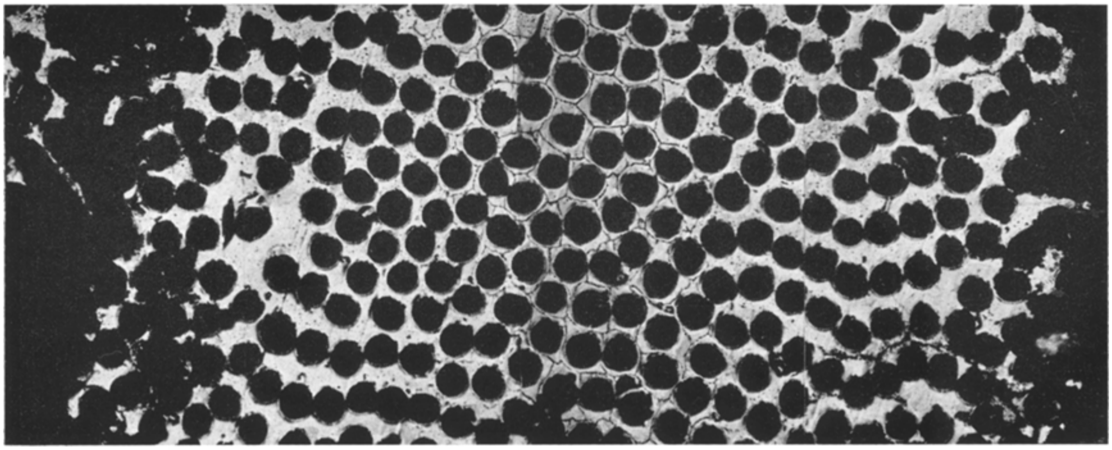
specimens to shear along the stress axis at high stresses, large extrapolations to the fatigue conditions were required.

5. The Effect of Elevated Temperature

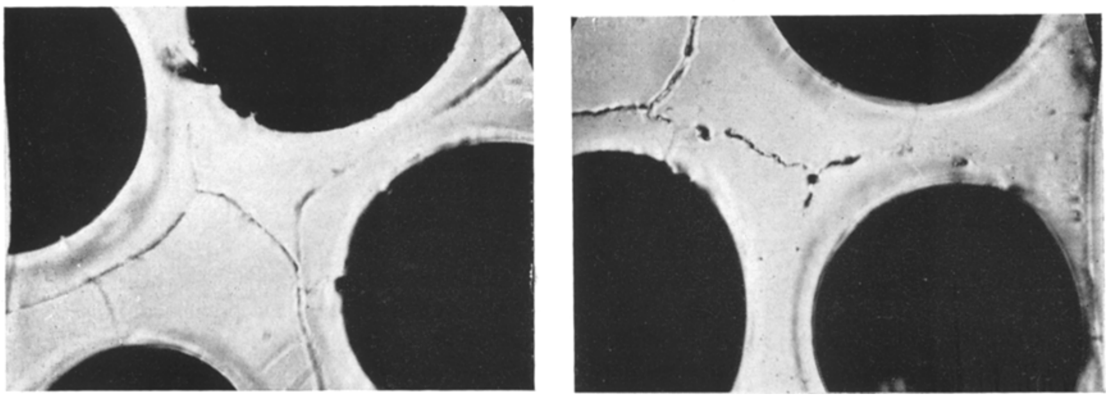
Since fibre-reinforced metals will be primarily intended for high temperature applications, the fatigue behaviour under these conditions will be of paramount importance.

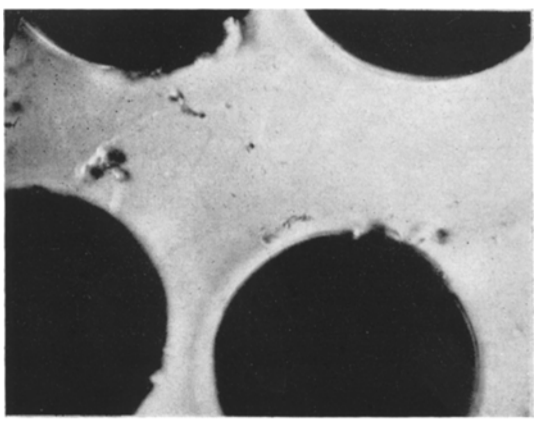
Fatigue tests at temperatures up to 350° C were carried out on aluminium/silica specimens using the reversed plain-bending rig as described in [11]; the specimens were heated directly by the passage of a heavy current and were controlled to $\pm$ 5° C of the set temperature.

With increase in temperature, matrix damage gradually changed from transgranular to intergranular, and much recrystallisation of the original hot-pressed structure occurred. Some other very interesting observations were made on the effect of cyclic plastic deformation on the composite structure at elevated temperature. The method of composite fabrication [11, 18] (prior fibre-coating and hot-pressing) leads to the presence of boundaries (discontinuities) between the precoated fibres in the composite; these presumably consist of areas of aluminium oxide. These are very difficult to remove by normal hot-pressing techniques if fibre damage is to be avoided [18]. The coating boundaries are however readily removed by fatigue at elevated temperature. This is shown in fig. 7a which is a cross-section through a fatigued specimen; coating boundaries are only present in the low-stress neutral axis region. At a higher magnification, fig. 7b, the gradual removal of the coating boundaries away from the neutral axis is shown. This observation may have



(a)





(b)

Figure 7 Cross-sections through a specimen fatigued at 350° C at a nominal stress of 11 kg/mm^2 (7 tsi) showing the removal of the coating boundaries away from the neutral axis [(a) × 75; (b) × 675].

important implications for developing improved methods of hot-pressing composites: for instance, pressing with applied ultrasonic vibration.

The results of the fatigue tests are given in fig. 8 as plots of the fatigue life versus temperature at three levels of stress*. At the lower stress levels the reduction in fatigue properties is not large. Since the plastic-strain-range of the matrix would not be greatly increased with an increase in temperature and the ductility of the matrix to fracture possibly even increased, this is not altogether surprising. However, if serious fibre-weakening had occurred as a result of diffusion induced by cyclic stress at high temperature, then the results would have been far more dependent on temperature than was in fact observed.

6. Alloying the Matrix

The method of composite production made any

*Experimental determinations of plastic-strain-range were made on some specimens at low stresses [14], as discussed in section 4.3. However, the extrapolation to the fatigue stress was considered to be too large to be accurate.

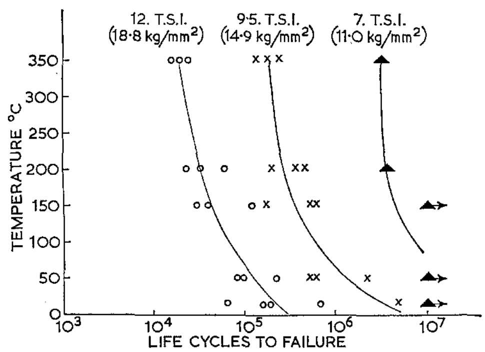


Figure 8 Temperature versus log cycles to failure for aluminium/silica composites tested various stresses.

significant alloying of the aluminium matrix difficult [19]. It is not possible to present any useful results here, but a discussion on the possible effects of alloying is thought worth while.

Alloying the matrix would be expected to reduce the strain range endured by the matrix at a given stress in the composite. Whether or not this gives an improvement in the fatigue properties would depend on (i) the resistance of the fibres and the fibre-matrix interface to the increased stress at the tip of matrix fatigue cracks, and (ii) the resistance of the matrix to the smaller plastic-strain-range. Some idea of the improvement from point (ii) can be obtained by assuming that the same factors that influence the Coffin relationship equation apply here. Thus the alloying should not result in a disproportionate decrease in fracture strain in the matrix. Comparison between composites on this basis can be made using equations 1 and 5 if the fracture ductility of the alloyed and unalloyed matrices are known. The resistance of the fibres to the matrix fatigue cracks probably depends on the following factors. (*a*) The ratio between the yield stress of the matrix and fibres; since the matrix yield stress determines the maximum stress at the tip of a propagating matrix crack [4] and the yield stress or hardness of the fibres determines the resistance to this stress. (*b*) The interfacial bond strength between fibres and matrix. It has been shown that a tensile stress perpendicular to the applied stress exists near the root of a notch [2]. Under elastic conditions this stress can reach one-fifth of the maximum applied stress. If the fatigue strength of the fibre-matrix interface is less than the fatigue strength of the fibre surface on the basis of this ratio, for a plastic matrix, the interface will fail and the crack relatively harmlessly deflected. This situation is obviously highly desirable in composites with highly alloyed matrices. (*c*) The crystallographic relationship and degree of coherency between the fibres and matrix. This will determine the ease with which dislocations can pass across the interface and thus initiate damage in the fibres [20]. Thus ceramics, particularly amorphous glasses, would be more resistant to damage of this nature than metal fibres. However, it is also known that due to the surface-sensitive nature of these materials a small crack, even one of atomic dimensions, can be sufficient to cause failure. It was therefore gratifying to find that in the aluminium/silica system the fibres were apparently able to deflect a fatigue crack in the aluminium matrix [11] without being damaged.

7. Damping Behaviour

One other consequence of the repeated deformation suffered by the matrix of a fibre-reinforced metal is that the material will have a high damping capacity.

The high damping capacity of aluminium reinforced with silica fibres has been shown by both dynamic and static measurements [3]. Damping is only significant when the stress in the composite causes plastic deformation of the matrix.

The specific damping capacity is defined as P = energy lost during one cycle of loading/energy at the beginning of the cycle. If the loop is narrow, the energy at the beginning of the cycle is approximately the elastic energy in the system (OBC in fig. 9) and the energy absorbed by the area of the hysteresis loop is OABD; thus P = area OABD/area OBC. A plot of the specific damping obtained from measurements on tension cycling loops on this basis, for aluminium reinforced with silica fibres and with stainless steel fibres, is shown in fig. 10a where it is seen that a maximum value of damping is obtained.

This behaviour may be predicted approximately from considerations of the simple model [3]* (fig. 11). We have that the total elastic energy in the system is the sum of elastic energy in the fibres and the elastic energy in the matrix. From fig. 11 these are $\frac{1}{2}\sigma_f \epsilon V_f$ and $(2\sigma_y^2/E_m)$ $(1 - V_f)$ respectively, where σ_f is the maximum stress in the fibres, σ_y the yield stress in the matrix,

*Assuming zero rate of work-hardening in the matrix.

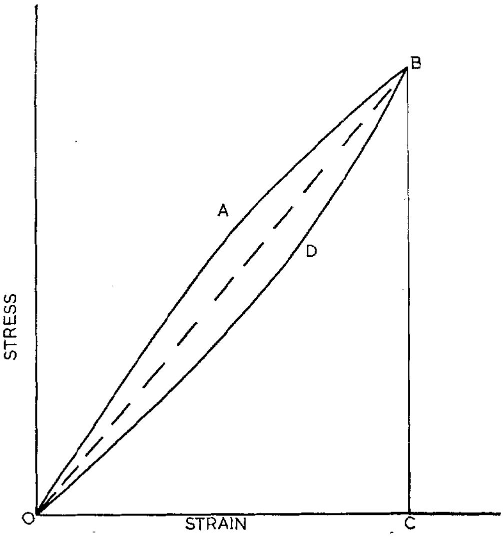


Figure 9 Schematic representation of damping obtained from a stress/strain loop.

E_m the modulus of the matrix, V_f the volume-fraction of fibres and ϵ the strain in the components of the composite. The energy absorbed is equal to the area of the hysteresis loop $2\Delta\epsilon_p\sigma_y(1 - V_f)$ where $\Delta\epsilon_p$ is the plastic-strain-range endured by the matrix.

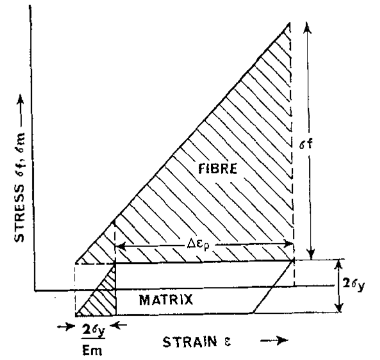


Figure 11 Model for damping capacity behaviour.

Then

$$P = \frac{2\sigma_y \Delta\epsilon_p (1 - V_f)}{\frac{1}{2}E_f\epsilon^2 V_f + (2\sigma_y^2/E_m)(1 - V_f)} \quad (16)$$

and since $\epsilon = \dfrac{2\sigma_y}{E_m} + \Delta\epsilon_p$

$$P = \frac{2\sigma_y \Delta\epsilon_p (1 - V_f)}{\frac{1}{2}E_f V_f \left[\Delta\epsilon_p + \left(\frac{2\sigma_y}{E_m}\right)^2\right] + \left[\frac{2\sigma_y^2 (1 - V_f)}{E_m}\right]} \quad (17)$$

This equation in fact predicts too high a value of damping capacity partially because in the

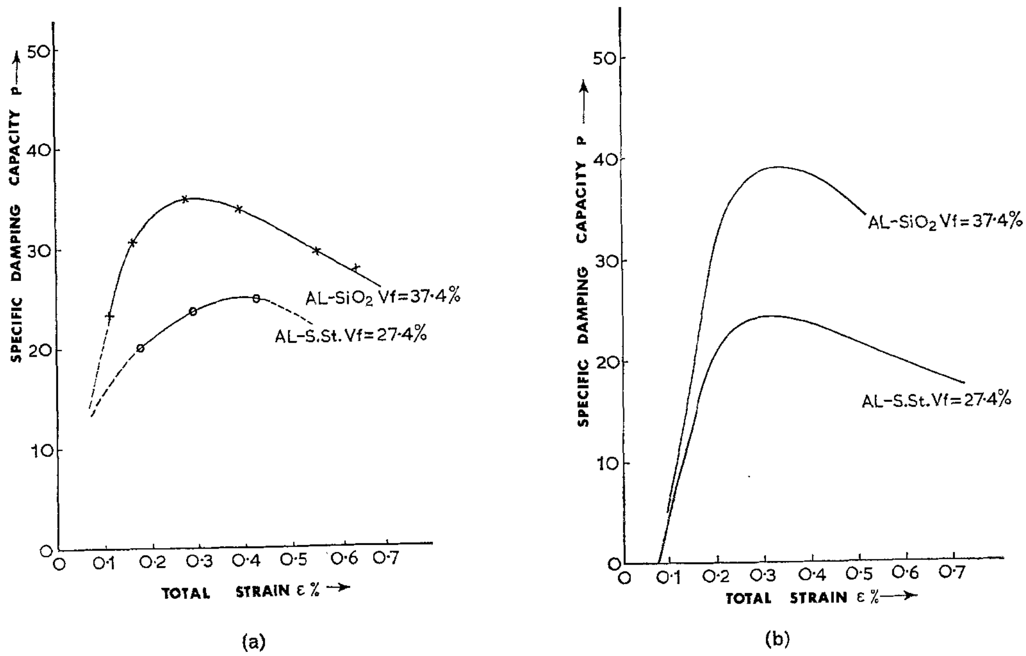


Figure 10 (a) Experimental plot of specific damping capacity versus strain for an aluminium/silica and aluminium/stainless steel composite. (b) Theoretical plot of specific damping for the two composites.

simple model the rectangle $2\sigma_Y \Delta\epsilon_p$ has been used to represent a loop. From measurements of the proportion of the area of the rectangle actually occupied by the loop a figure of 1.25 instead of 2 was used in the top line of equation 17. The resulting predicted curves are shown in fig. 11b. The reason for the maximum in the specific damping curves with stress is the more rapid build-up of elastic energy compared to loss of energy due to plastic deformation. Analogous curves can also be obtained for tension/compression cycling as in fig. 3.

Dynamic measurements [3] have in general confirmed these observations; at low stresses, the damping is similar to other low-damping materials and increases with increasing stress until it becomes greater than that of grey cast iron.

Under conditions of forced vibration, a high damping capacity is of little value and may cause damage through heating of the material; in a metal where damping is achieved by plastic deformation, fatigue damage will occur.

However, under conditions where resonant vibrations occur, a high damping capacity may be of great significance, for instance in aerodynamic machines such as a jet engine. In this case the resulting fatigue stress induced in the component is inversely proportional to the damping capacity [21]. Thus the material chosen for service under resonant conditions may not be the one with maximum fatigue strength but rather the one with the best combination of fatigue strength and damping capacity. This then is another advantage of having a plastically deforming matrix in a fibre-reinforced metal. Indeed, it is unlikely that a combination of such high strength and damping capacity could be obtained in any other metal.

8. Summary

It has been suggested that the matrix of fibre-reinforced metal components will undergo plastic deformation under normal service conditions. This will lead to fatigue damage. The effect of the fatigue damage depends to a very large extent on the method of fatigue testing. Providing that the fibres are not easily damaged by the matrix fatigue cracks, the theoretical plastic-strain-range endured by the matrix would appear to be a useful criterion of the fatigue damage produced in the composite by the applied stress. This is borne out by some experimental results on fibre-reinforced aluminium tested at ambient temperature in reversed bending. Methods for obtaining values for theoretical plastic-strain-range are given on the basis of (i) a simple stress/strain model, (ii) an empirical model, and (iii) experimental determination.

Fatigue at elevated temperature (up to 350° C) results in a change from transgranular to intergranular cracking but with no catastrophic fall-off in strength. The effect of alloying of the matrix is considered particularly with respect to some of the factors expected to influence the tendency to fibre failure.

It is shown that under conditions where the matrix undergoes plastic deformation, the composite will have a high damping capacity. A method of calculating the value of "specific damping" is discussed on the basis of a simple stress-strain model.

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