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Fatigue of boron-aluminium and carbon-aluminium fibre composites

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Reversed bending fatigue studies have been performed on boron-aluminium and carbon-aluminium fibre composites. The superior fatigue resistance of the boron composites arises partly from the use of an alloy matrix, but mainly from the more advanced development stage attained in the manufacture of the boron composite. (The carbon-aluminium composites contained many broken fibres and relatively weakly-bonded interfaces which provided easy paths for the propagation of fatigue cracks.)

Although matrix fatigue is the primary type of fatigue damage sustained by these composites, some evidence was found for induced fibre failure in boron-aluminium composites under high stress/low cycle conditions.

1. Introduction

The fatigue behaviour of boron fibre-aluminium alloy and carbon fibre-aluminium composites has been studied. Both composites are similar in that the fibre reinforcements are continuous, have about the same fibre strength and elastic modulus and are linearly elastic to failure. The main differences are the fibre diameter (boron ~ 125 µm and carbon ~ 8 µm) and the strength of the aluminium matrix. However, there are a number of other differences, particularly in the degree of technological development of the composites, which makes direct comparison very difficult.

The early part of the work on the boron-aluminium composite was previously reported in [1]. Before describing the fatigue results, some comments are made on the various mechanisms of fatigue in continuous fibre composites.

2. Fatigue failure mechanisms

A number of fatigue failure mechanisms are possible in a continuous fibre metal matrix composite. These may be categorized for convenience as (a) matrix fatigue, (b) interfacial fatigue, and (c) fibre fatigue, although all these mechanisms may operate simultaneously. In addition, as discussed below, there are various ways in which each mechanism may operate. Perhaps the simplest mechanism is matrix fatigue which occurs when the matrix is subjected to cyclic straining above its fatigue strain limit, but the fibres do not fail. For a composite with a relatively soft ductile matrix the plastic strain range endured, Δe_p, is a useful indication of the degree of matrix damage. On this basis the predicted life of the unreinforced matrix is given by [2]

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

where N_f is the number of cycles to failure and C a constant. In the composite it can be shown that [3] for a tension/compression cycle

$$\Delta e_p = \frac{2\sigma_c - 2\sigma_y[(E_f/E_m)V_f + (1 - V_f)]}{E_f V_f + U(1 - V_f)} \qquad (2)$$

where σ_c is the stress applied to the composite, σ_y the cyclic work hardened yield stress of the matrix, V_f the fibre volume fraction and U the effective modulus of the yielded matrix. From Equations 1 and 2 the stress in the composite for some specified degree of initial matrix cracking may be found, but this does not predict when the composite will fail because this depends on the ability of the fibres to deflect matrix cracks and to hold the damaged matrix together, and also on the failure criterion taken. Other theories for matrix failure based on the cumulative cyclic

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plastic strain have been given by Kelly and Bomford [4].

If the total plastic strain range in the matrix is low, the plastic strain will be concentrated in small regions within the matrix and the total strain endured by the matrix is a more useful criterion of matrix damage [5]. If the matrix is stressed mainly below its yield stress then the total matrix strain e_m is given approximately by

$$e_m = \frac{2\sigma_c}{E_f V_f + E_m(1 - V_f)} \quad (3)$$

In this case the matrix life, before the appearance of a first large crack, can be approximately obtained from the S/N curve or from some other empirical criterion of fracture. Of course in both cases we are assuming that the presence of the fibres does not affect the fatigue mechanisms in the matrix. This assumption will be in error if there is a Poisson's ratio difference between the fibres and the matrix [6]. The error will be small for large fibre spacings but may be considerable if the spacing is small. If there are broken fibres in the composite these will act as fatigue-strain concentrators [7] and produce local fatigue damage in the matrix.

Failure is very difficult to define for a composite undergoing matrix fatigue. Perhaps the most obvious criterion is the appearance of the first fatigue crack in the matrix [8]. This criterion is very limiting (unless the potential application demands it, e.g. in a pressure vessel) and the residual fatigue strength of the material would not be allowed for. Criteria based on the ability of the composite to continue to act as a composite (such as flexural or compressive stiffness) may be more realistic for the majority of applications.

A similar situation to matrix fatigue occurs when the mechanism involves interfacial failure, although this is much more difficult to characterize quantitatively as it cannot be related to known properties of the fibre or matrix. Various interfaces may be important in composites, particularly under fatigue conditions, including: (a) the fibre-matrix interface (possibly including a reaction zone), (b) inter-foil interfaces within the matrix (e.g. boron-aluminium), (c) interfaces between coatings previously appplied to fibres (e.g. silica-aluminium [3]), (d) intergranular interfaces in fibre coatings (e.g. carbon-aluminium, see below), and (e) weak matrix grain boundaries (e.g. carbon-nickel [18]). Clearly, as with matrix failure, if the fibres do not fail before the test is stopped because the assumed failure criterion is reached, the composite will be in a fail-safe condition.

Conversely, when fibre failure occurs the composite is not fail-safe and failure is very easily defined since the material rapidly separates into two pieces. This mechanism can occur either as a result of simply exceeding the fatigue strength of the fibres (unlikely at low levels of stress for a composite with a high volume fraction of ceramic fibres) or from the mechanical interaction between fibre and matrix. Fibre failure by mechanical interaction between the fibre and matrix may occur in two ways. Firstly, fatigue cracking of the matrix may occur and because of the resulting stress-concentration the resulting cracks may propagate directly through the fibres. In this case the fatigue properties of the composite will be limited to the fatigue strain limit of the matrix and this would be lower than the properties obtained from a composite failing by matrix fatigue because the fibres would not be capable of supporting the cracked matrix. The situation is of course complicated by the presence of broken fibres (produced either during composite manufacture or during testing) which would concentrate the strain in the matrix and result in matrix cracking below the expected level.

The second possibility is that the intense shear in the matrix around a broken fibre or notch due to local load transfer could induce fibre failure before the cyclic strain was sufficient to crack the matrix. This would correspond to the observations made by Cooper [9] on the fracture of copper/tungsten wire composites. Matrix cracks would then develop at a slower rate than fibre failure. This mode would be expected when the composite was tested at stresses near to the fibre tensile or fatigue strength limit.

In both of these mechanisms the local stress that can be induced in the fibre will depend on factors such as the strength of the fibre/matrix interface, the modulus, yield stress and work hardening capacity of the matrix, and the fatigue strength, volume fraction and diameter of the fibres. Quantitative evaluation of these factors would be of immense value in the design of fatigue resistant composites.

The actual failure criterion accepted can be a function of the method of testing if the material undergoes matrix fatigue [10]. For instance, in a tension/tension test the most likely criterion will be complete separation of the composite. In fact the composite may undergo matrix fatigue at much lower stresses than that required for fibre failure. This fact would be very obvious in, say, a tension/compression or a flexural fatigue test as it would not be possible to maintain the required stress level at the same level of deflection. Thus in this type of test the change in flexural or compressive modulus would be the most likely criterion. On this basis it must be assumed that tension testing produces the most optimistic values of fatigue strength.

3. Materials tested

3.1. Boron-aluminium

The boron-aluminium composite was purchased from the Harvey Aluminium Company, Torrance, California, USA in the form of a 1 mm thick plate. This contained about 47% boron fibres (~ 125 μm diameter) in a 6061 aluminium alloy matrix (1% Mg, 0.6% Si, 0.25% Cu, 0.25% Cr, Balance Al). The composite was prepared by laying the boron fibres between sheets of the aluminium alloy and then consolidating by hot-pressing. The composites were fatigue-tested with the matrix either in the as-received non-heat-treated or in the heat-treated condition (500°C for 1 h followed by water-quench, then aged at 170°C for 6 h). Fatigue samples were cut from the plate in the as-received condition and were contoured by grinding. Before testing, the surfaces were lightly ground with wet medium grade silicon carbide paper. A detailed account of the mechanical properties of this composite together with some of the early fatigue results, is given in [1].

3.2. Carbon-aluminium

Composites were prepared by hot-pressing carbon fibre tows which had been pre-coated with aluminium by chemical vapour deposition [11, 12]. The chemical vapour deposition process is based on the thermal decomposition of tri-isobutyl aluminium $(C_4H_9)_3Al$, and will be described in detail elsewhere. Two types of fibres (designated as types A and B) were used in the investigation and details of these are given in Table I; in all cases the fibre diameter was about 8 μm.

Hot-pressing of the coated fibre tows to form a composite was carried out in vacuum. The effects of the hot-pressing conditions on the mechanical properties obtained are reported in [11]. Briefly, it was found that the modulus values obtained for the composites with fibre volume fractions in the range 20 to 40% were close to the rule of mixtures but that the tensile and flexural strength levels were below the rule of mixtures. The highest strengthening efficiency based on the mean fibre strength was obtained for type A fibres which gave ~ 550 MN m^{-2} at ~ 30% fibre volume fraction; this is ~ 80% of rule of mixtures. For the fatigue experiments the main hot-pressing conditions used were a temperature of 550°C at a pressure of 77.3 MN m^{-2} for a time of 1 h; these conditions were found to be necessary to get near to full density in the composites without causing carbide formation and were one of the main conditions previously investigated [11]. However, because of the relatively poor alignment in the type A fibre tows, pressing at this pressure resulted in considerable fibre breakage which was thought to be the reason for the low tensile strength obtained. Thus a subsidiary series of experiments was carried out with coated type A fibre pressed at 600°C, 7.73 MN m^{-2} for a time of 0.5 h. This condition considerably reduced the mechanical fibre damage but was found to induce some chemical interaction between the fibres and the matrix. This suggests that there is an interaction between temperature and pressure, since compatibility studies have shown [13] that there should be no carbide formation with temperature alone after 0.5 h.

TABLE I Carbon fibres used.

Type	Form	Modulus (GN/m^2)	UTS (GN/m^2)	Comments
Courtaulds (HMT) (graphitized) (A)	Continuous 10^4 filament tow	351	2.0	Surface-treated as-received
Rolls-Royce material (carbonized) (B)	Continuous 10^4 filament tow split from continuous sheet	185	2.13	Surface-treated in hypochlorite. Very good alignment of fibres.

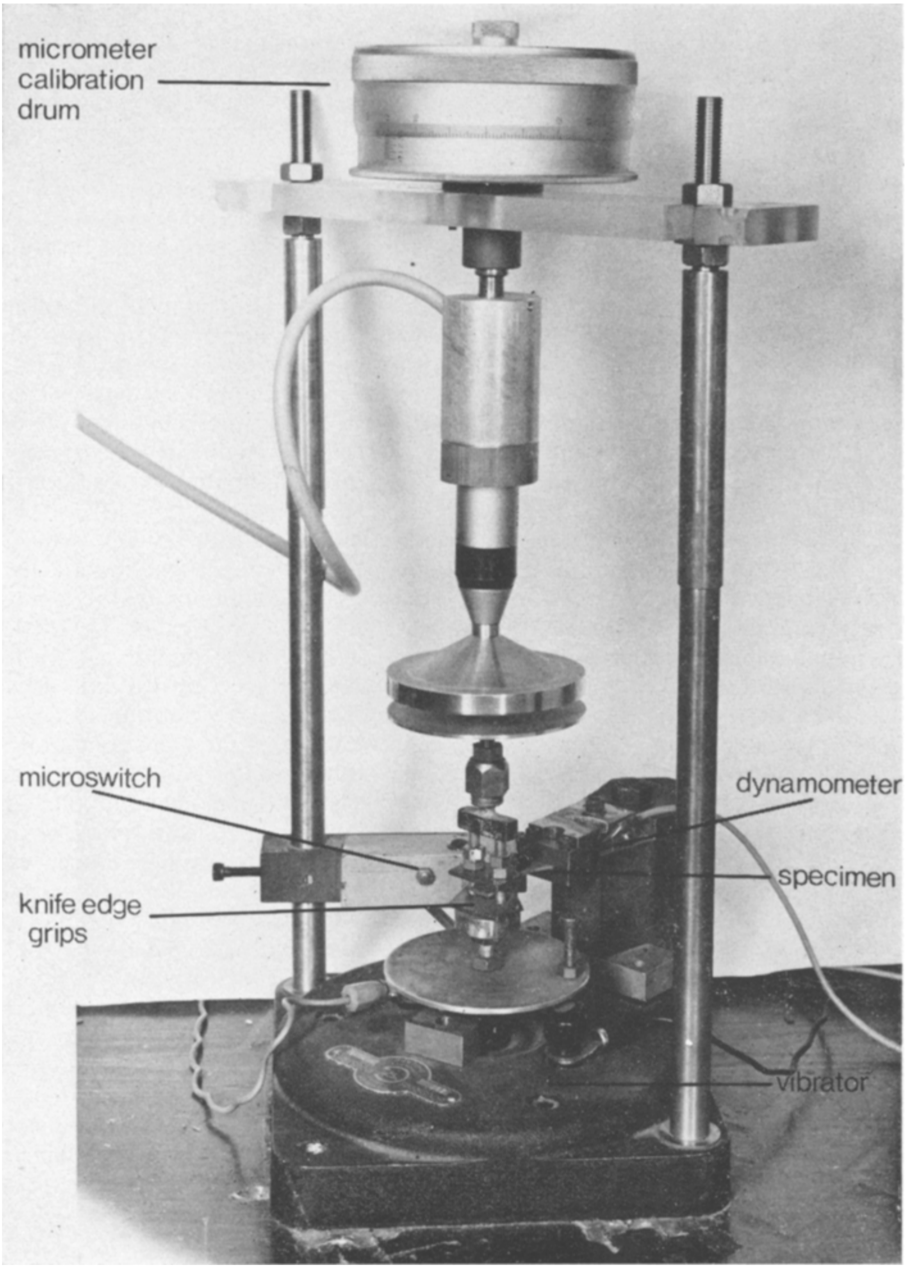

Figure 1 Fatigue machine.

The fibre damage in composites prepared from type B fibres should have been minimal, because of the high degree of fibre alignment, but this did not apparently result in an improved strengthening efficiency in the composite [11].

4. Fatigue evaluation

4.1. Fatigue-testing

Fatigue-testing was carried out in a reversed bending apparatus used earlier [3, 7, 10] based on an electromagnetic vibrator, Fig. 1 (used at 50 Hz).

The specimens were fatigued at constant nominal stress, measured from a resistance strain gauge dynamometer, and the deflection was continuously monitored, using a capacitance bridge. Most of the specimens underwent changes in flexural stiffness during the test and periodic adjustment of the power input into the vibrator was usually necessary to maintain constant stress*. A curve of deflection versus cycles at the applied stress level was produced for each specimen tested. It is possible, alternatively, to use the fatigue machine to maintain constant deflection in the specimen and measure the decay in stress although this was not done in these tests.

To define failure, for specimens which did not actually fracture, the curve of deflection versus number of cycles was extrapolated and the life measured from the point at which the curve was vertical, corresponding to zero stiffness in the material. This allows a reasonably consistent criterion of failure to be used without damaging the specimen too seriously for subsequent metallographic observations.

High-temperature tests (250° C) were carried out on some boron-aluminium samples using a similar fatigue apparatus. In these tests the specimens were self resistance-heated, using a welding transformer. Temperature measurement was made from a thermocouple wired onto the narrowest point (maximum stress) of the sample and the sample and thermocouple were insulated at this point with asbestos string. This method was not particularly reliable because of electrical contact problems but in most cases control to within $\pm 5°C$ was possible. Non-heat-treated specimens were used and were pre-conditioned in the fatigue machine at 250° C for 1 h before testing.

4.2. Post-failure analysis

Surface observations were made at a low magnification to find the main path of crack propagation. Metallographic studies were carried out on taper sections and in some cases on cross sections near to the fracture. The boron-aluminium was found to be particularly difficult to polish and this greatly restricted the information that could be obtained about the fatigue mechanism in this material. Selected fractures were examined using the scanning electron microscope. Finally most of the boron-aluminium specimens were examined by micro-radiography

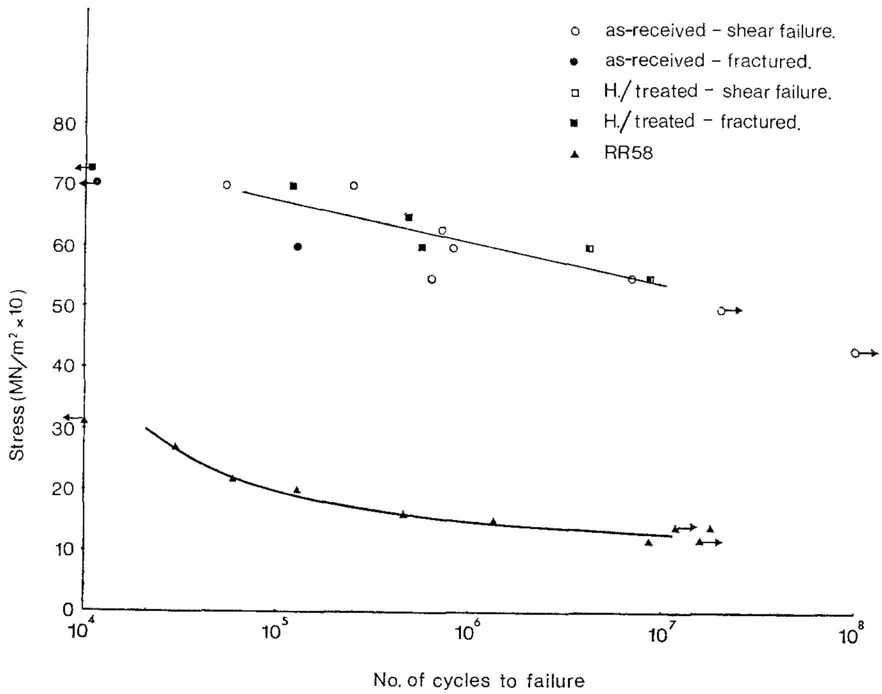

Figure 2 Plot of cycles to failure versus stress for boron fibre reinforced aluminium composites tested at ambient temperature. A curve for aluminium alloy RR58 is plotted for comparison.

*This is because the electromagnetic vibrator develops constant power rather than constant forces. For conventional metals no adjustment is required.

(using X-radiation at 30 kV with a beryllium window) in an attempt to find the density of fibre breaks away from the main fracture.

5. Experimental results

5.1. Boron-aluminium

The ambient temperature fatigue results (previously reported in [1]) are plotted in Fig. 2. This shows that a definite fatigue effect exists and this appears to be independent of the condition of the matrix.

However, various types of fracture modes occurred, which appeared to be dependent on the condition of the matrix and the stress amplitude. Examples of these modes, which are shear, part shear and no shear, and the corresponding microradiographs are shown in Fig. 3a, b and c. In the specimen which failed with no shear, Fig. 3c, the region of fracture on the surface of the specimen can be seen to correspond with large scale fibre damage in the microradiograph, whereas in the specimen which failed mainly in shear, Fig. 3a, only isolated areas of fibre damage can be seen. This is also the case for the sample which failed by part shear and the lack of correspondence between the apparent severe surface damage and the microradiograph is rather surprising. It would appear that only in the case of Fig. 3c was the failure entirely due to fibre fracture.

The change in normalized deflection (instantaneous deflection/initial deflection) during fatigue in some typical samples failing in these various modes is shown in Fig. 4a and b. These plots show that the large-scale deflection changes only occur in samples undergoing matrix fatigue, e.g. shear and part shear. In samples which failed with no matrix shear, fracture was sudden and accompanied by a loud click, usually during readjustment of the load.

It is important to note that the onset of matrix shear results in a higher than nominal maximum stress in the specimen, as can be seen from Fig. 3a. This effectively means that at the low-endurance end the non-heat-treated samples fail at a higher stress than the specimens which were heat-treated.

Longitudinal taper sections through these samples revealed a number of modes of matrix fatigue cracking. In Fig. 5a the initiation of a matrix crack from an apparently pre-cracked fibre can be seen and in Fig. 5b matrix fragmentation in the longitudinal shear zone. No evidence of matrix cracks actually propagating through fibres was found but the difficulties in metallographic preparation made the information available from these sections very limited. Cross

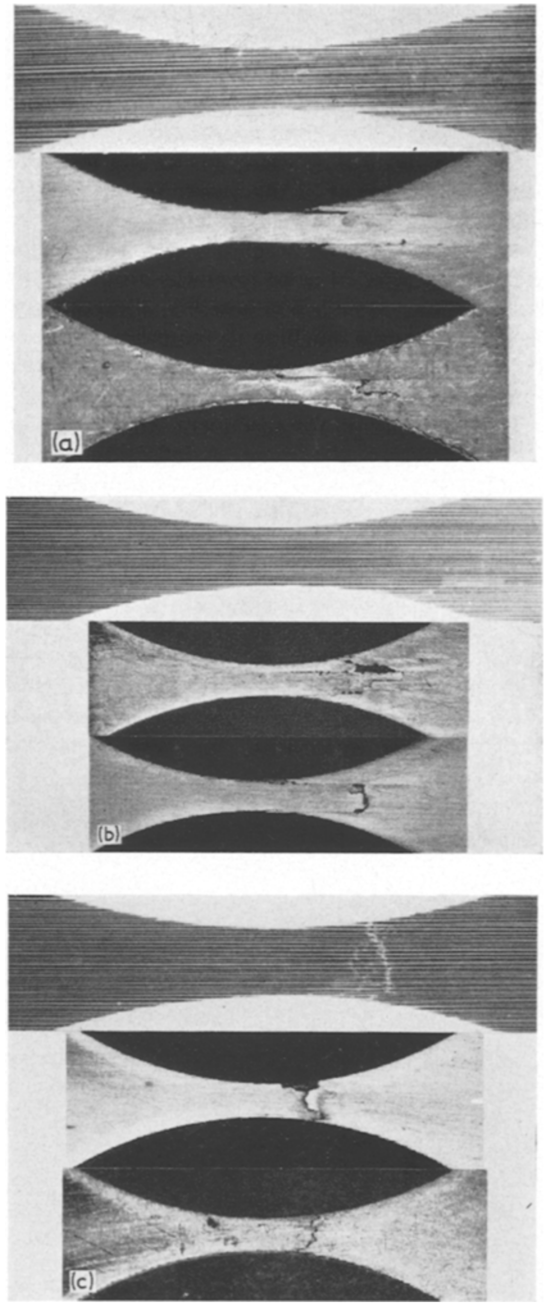

Figure 3 Macrophotographs and corresponding radiographs of failed boron-aluminium composites: (a) Non-heat-treated, tested at 550 MN/m^2 for $\sim 6 \times 10^6$ cycles. (b) Heat-treated, tested at 550 MN/m^2 for $\sim 9 \times 10^6$ cycles. (c) Heat-treated, tested at 650 MN/m^2 for $\sim 5 \times 10^5$ cycles.

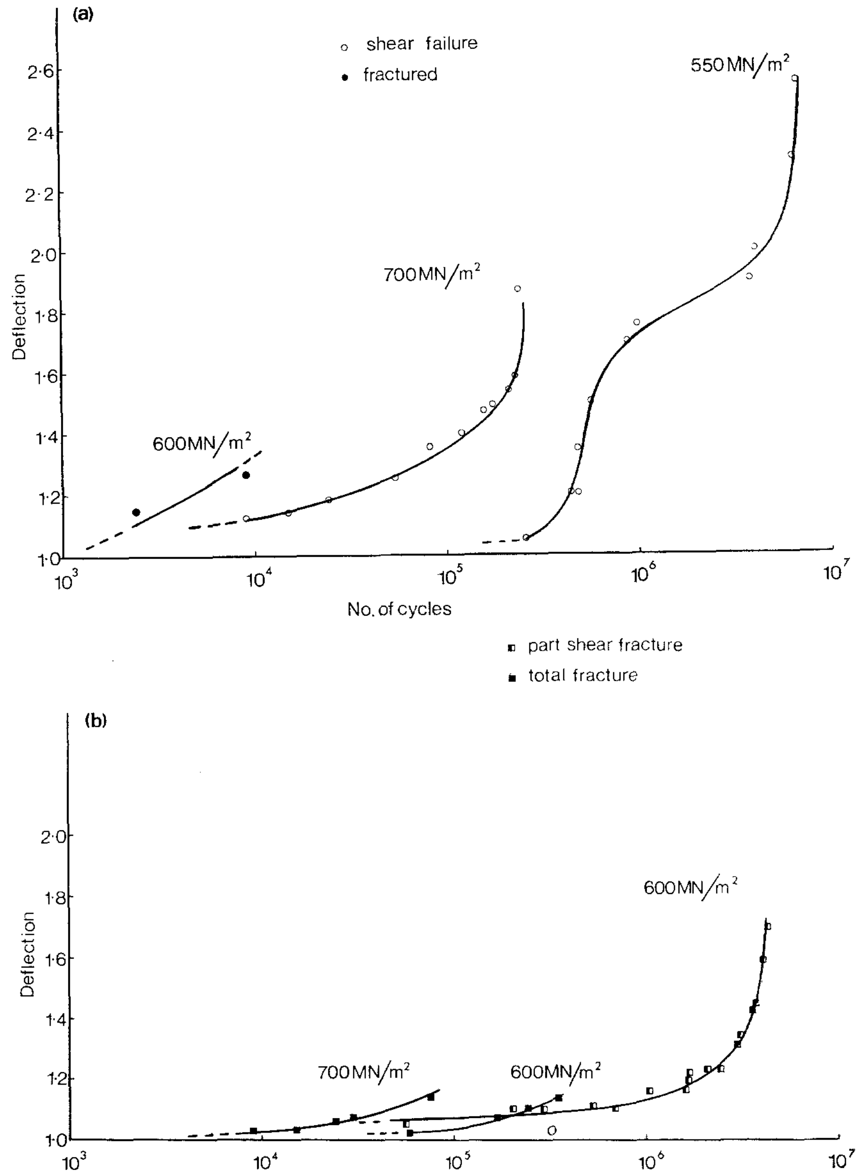

Figure 4 Plot of normalized deflection versus fatigue life for the boron-aluminium composite: (a) Non-heat-treated. (b) Heat-treated.

sections taken just behind the main fracture revealed the ability of fatigue stressing to delaminate the material through the foil/foil pressing boundary, Fig. 6a and to propagate transverse shear cracks through the matrix and the fibre/matrix interface, Fig. 6b. These observations confirm the scanning electron fractography previously described [1] where it was also found that the fibre fractures were generally surface initiated and the matrix showed very little ductility. A further example of this behaviour is shown in Fig. 7 which is taken from an area close to the maximum stress surface.

The results for the fatigue tests carried out at 250° C are plotted in Fig. 8. It can be seen that a very marked fatigue effect is present and that the fatigue properties fall considerably below the values at ambient temperatures for lives above 10^5 cycles. Although there is no experimental evidence to prove this, it seems likely that the

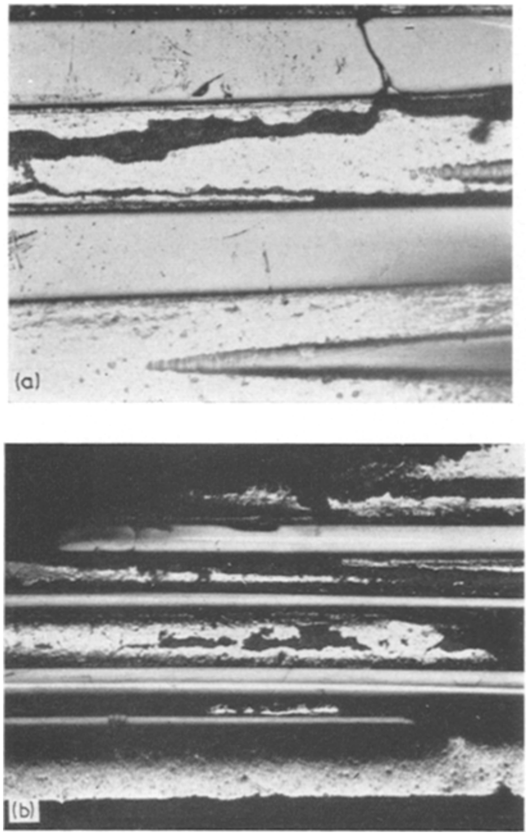

Figure 5 Micrographs of longitudinal taper sections through a non-heat-treated boron-aluminium specimen, tested at 700 MN/m^2 for 6×10^4 cycles (Fibre diameter 125 μm), (a) Initiation of a matrix crack from a pre-broken fibre. (b) Matrix fragmentation.

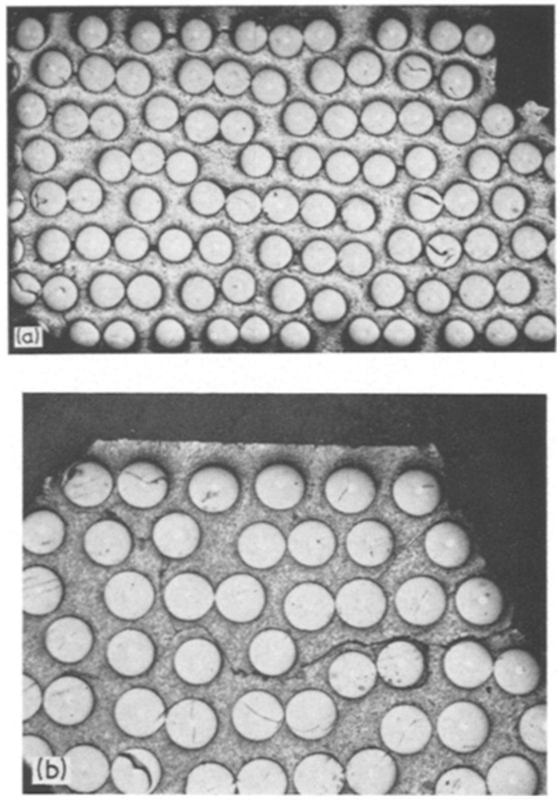

Figure 6 Micrographs of cross-sections through boron-aluminium specimens taken near to fracture zone. (Fibre diameter 125 m.) (a) Delamination in a heat treated specimen tested at 720 MN/m^2 for $< 10^4$ cycles, and (b) Transverse shear crack propagation in a non-heat-treated specimen tested at 600 MN/m^2 for $\sim 10^3$ cycles.

high temperature fatigue curve would extrapolate above the ambient temperature result at high stress amplitudes. In only one case was actual fracture of the specimen observed; in most other specimens the main damage was matrix fragmentation and in some cases partial removal. An example of this is shown in Fig. 9a and a very severe case of matrix damage with some associated fibre damage is shown in Fig. 9b in a long endurance specimen. Although these samples were not investigated metallographically the main cause of failure would appear to be matrix fatigue.

5.2. Carbon-aluminium
5.2.1. Type A fibres
The fatigue results for carbon-aluminium composites with type A fibres (fibre volume fraction in the range 33 to 38 %) are plotted in Fig. 10. This shows that a very strong fatigue effect exists for the main series of samples hot pressed at 550° C and that the results for the samples pressed at 600° C are generally at a lower level of stress and highly scattered. Some typical normalized curves of deflection versus number of cycles for the 550° C series of specimens tested at various stress levels are shown in Fig. 11. In most cases failure occurred by complete separation of the sample and examples of the various fracture modes are shown in Fig. 12. These, as can be seen, show different degrees of shear depending on the stress amplitude used and are rather similar to the heat-treated boron-aluminium composites.

The first impression obtained from longitudinal taper sections through a failed specimen (550° C) was that matrix fatigue cracks could propagate directly through the fibres. An example is shown

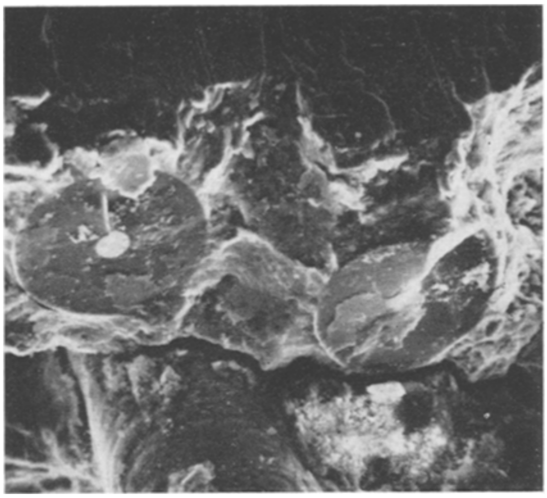

Figure 7 Fractograph taken from a point near to the maximum stress surface in a heat-treated boron-aluminium composite sample tested at 600 MN/m^2 for 7×10^5 cycles, showing surface initiated fibre failure and interlaminar cracking ($\times$ 280).

in Fig. 13. But with closer examination, Fig. 14a and b, it can be seen that in almost every case matrix cracks have propagated through fibres which appear to have been pre-broken, probably during the hot-pressing process or at the fibre/matrix interface, particularly when the fibres are misaligned. Furthermore it was found that the matrix fracture surface in the fatigued composite was very granular in places compared with one fractured either with no fatigue or in a very few cycles. An extreme example, taken from scanning electron microscope observations on two specimens taken from the same hot-pressed blank, is given in Fig. 15a and b. The matrix behaviour of the type shown in Fig. 15b would normally be found in a tensile sample pressed at too low a temperature and pressure whereas the matrix in Fig. 15a has the appearance of a normal tensile fracture.

Convincing evidence of the ability of unbroken carbon fibres from this series to deflect matrix cracks is shown in Fig. 16. In this case the cracks must pass around the fibre/matrix interface.

In the specimen with type A fibres which had been hot-pressed at 600°C a number of areas where direct matrix crack propagation had apparently occurred were found. An example is shown in Fig. 17 although even here a few of the fibres do appear to have been fractured prior to fatigue. The behaviour of these composites is complicated by the presence of particles of aluminium carbide at the fibre matrix interface.

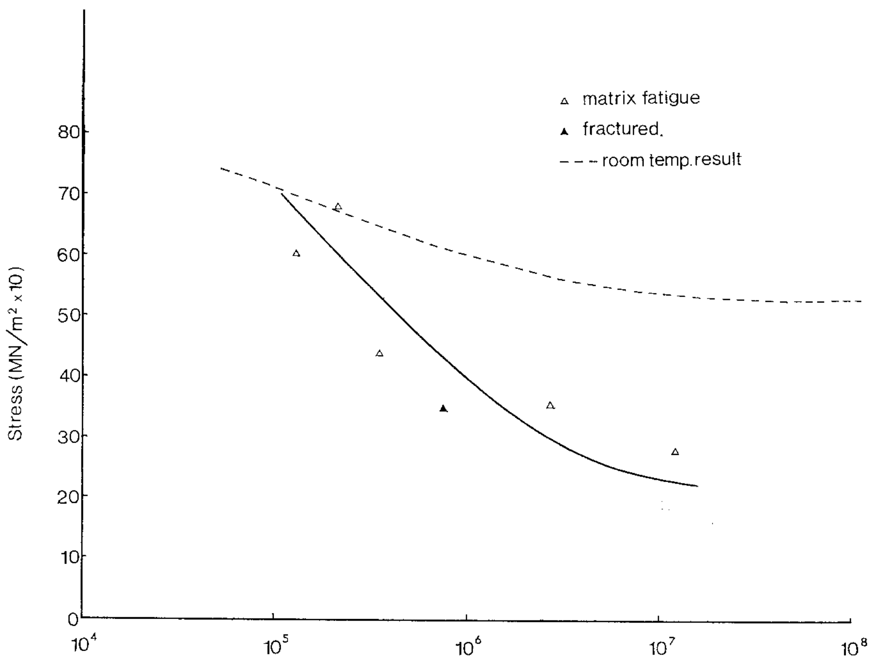

Figure 8 Plot of cycles to failure versus stress for boron-aluminium composites tested at 250°C.

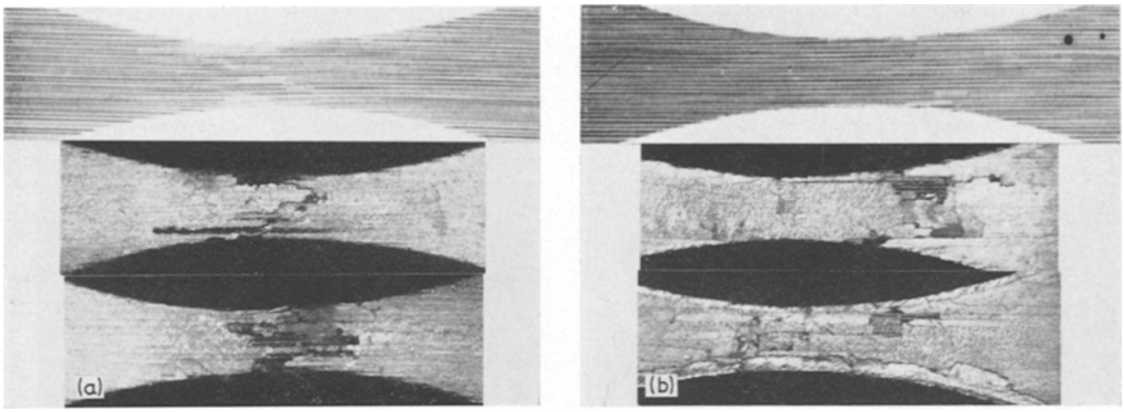

Figure 9 Macrographs of boron-aluminium samples fatigued at 250°C after: (a) 10^7 cycles at a stress of 300 MN/m^2 ($\times$ 9), (b) 3×10^6 cycles at a stress of 370 MN/m^2 ($\times$ 9).

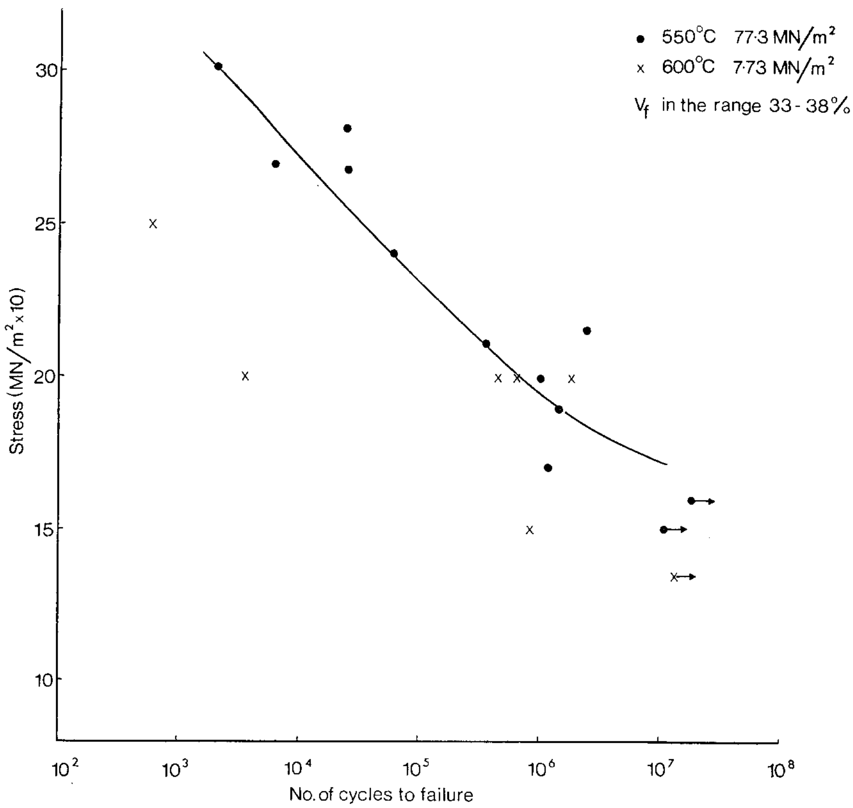

Figure 10 Plot of cycles to failure versus stress for aluminium reinforced with type A carbon fibres.

5.2.2. Type B fibres

The fatigue results for the specimens produced from type B fibres are shown in Fig. 18. Although the fibre damage was minimal in these composites the fatigue properties can be seen to be inferior to the composites with graphitized fibres. Evidence of direct matrix crack propagation was difficult to find, although the existence of extremely flat regions in the fracture surface suggests that it does occur. This can be seen by

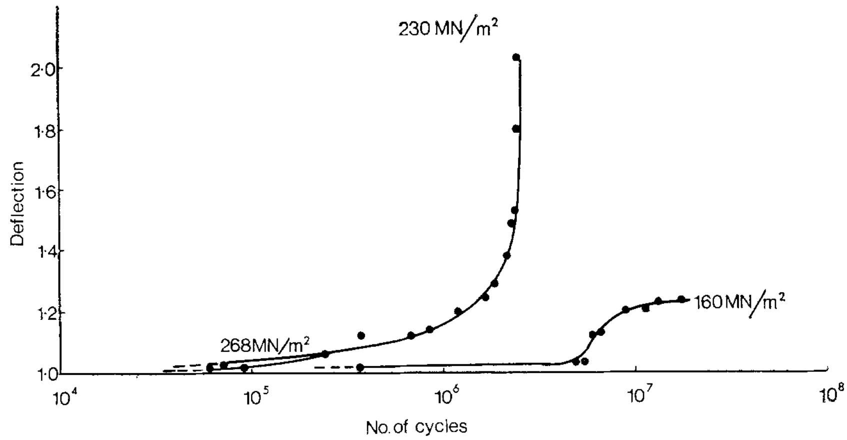

Figure 11 Plot of normalized deflection versus number of cycles to failure for composites of aluminium reinforced with type A carbon fibres.

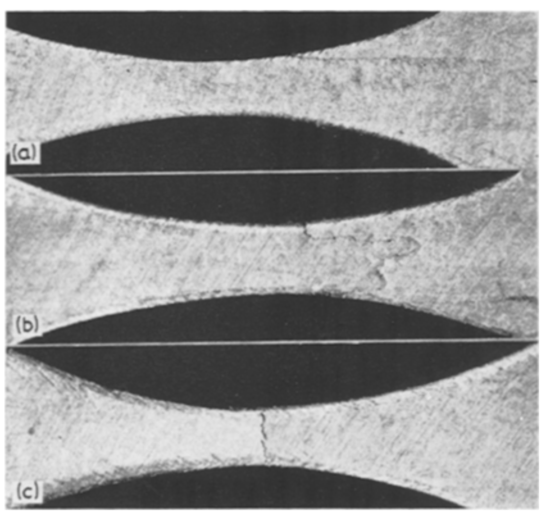

Figure 12 Micrographs ($\times$ 3) of carbon-aluminium composites (type A fibres) tested at: (a) 160 MN/m^2 for $\sim 10^7$ cycles (unfailed). (b) 220 MN/m^2 for 3×10^6 cycles. (c) 270 MN/m^2 for 6×10^3 cycles.

comparing a specimen fractured by static bending, Fig. 19a with an area taken from a fatigued specimen, Fig. 19b.

6. Discussion and conclusions

6.1. Boron-aluminium composites

It was shown that boron-aluminium composites could fail in reversed bending by a process involving fracture of the boron fibres [1]. This type of failure was found particularly in composites with matrices that had been heat-treated and which were tested at high stress levels. At lower stress levels and particularly in composites which had not been heat-treated a considerable amount of matrix fatigue in the form of tensile and shear cracks occurred and this resulted in the composites undergoing a gradual change in flexural stiffness until final failure; this mode of failure altered the geometry of the specimen so that these composites were under a higher maximum stress than the nominal value. Evidence of matrix fatigue cracking from pre-fractured fibres and of actual debonding between the alloy sheets used in producing the composites was also found.

No direct metallographic evidence of matrix cracks actually passing through fibres was seen in specimens failing primarily by fibre failure in reversed bending. However, previous work [1] on high-amplitude three-point bending fatigue, where the fracture surfaces were not damaged by repeated contact, showed that a high proportion of fibres failed by surface initiation rather than the more usual core initiated fracture. In addition there was also evidence of matrix fatigue cracks adjacent to surface initiated fibre fractures; see for example Fig. 15a, [1]. The circumstantial evidence also suggests that matrix initiated fibre failure can occur at stress levels substantially below that at which single fibres would be expected to fail in fatigue. For instance, the increased tendency for fibre failure to occur when the matrix is in its strongest condition suggests that the matrix influences the fatigue properties. This could either be by allowing a higher stress

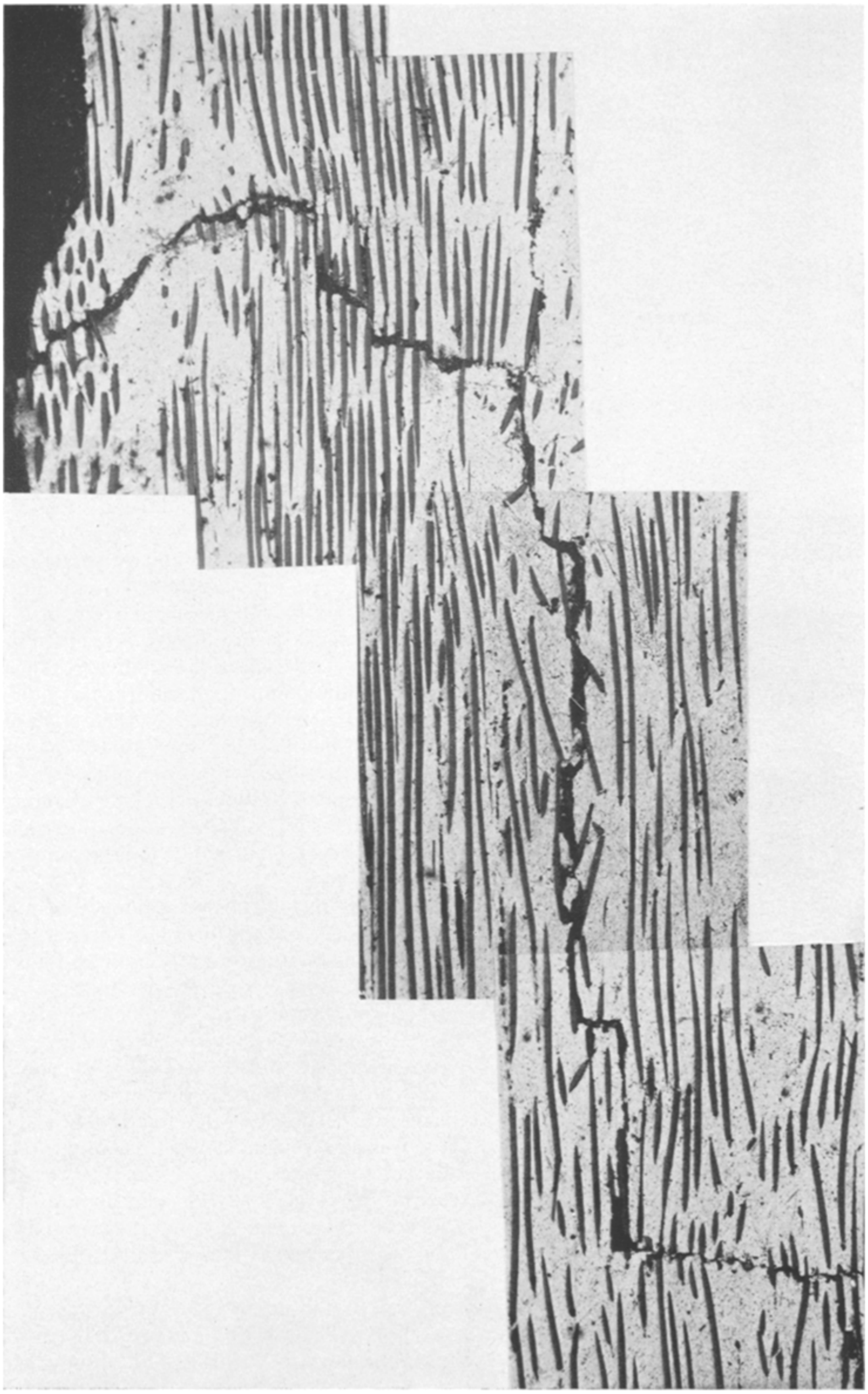

Figure 13 Micrographs of a longitudinal taper section through a carbon-aluminium composite after fatigue at a stress of 215 MN/m^2 for $\sim 3 \times 10^6$ cycles. (Fibre diameter 8 μm.)

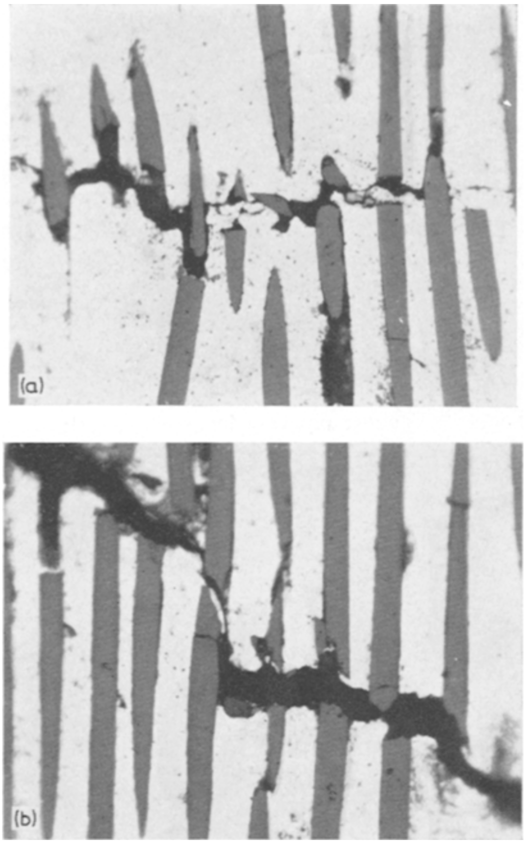

Figure 14 (a) and (b). Micrographs taken at two points along the crack path shown in Fig. 13. (Fibre diameter 8 μm.)

at the crack tip or alternatively by concentrating the shear stress near a randomly broken fibre, a process not involving the direct propagation of matrix cracks. Perhaps the most convincing evidence is the relationship between the expected failure strain of the matrix at the stress in the composite, obtained from Equation 2 (assuming that no macro-yielding of the matrix occurs) and the fatigue properties of the composite. The results for the heat-treated composite on this basis can be compared with the value for the unreinforced matrix taken from [14] for 6061 alloy in the T6 condition tested in rotating bending fatigue. This gives a life of 10^5 cycles at a strain of $\sim \pm 3 \times 10^{-3}$, 10^6 cycles at a strain of $\sim \pm 2.6 \times 10^{-3}$ and 10^7 cycles at a strain of $\sim \pm 2.3 \times 10^{-3}$ taking $E = 225$ GN/m^2.

Thus under the strain conditions where composite fracture actually occurs the correlation is very good, although the difference in testing method must make the comparison very tentative.

The alternative explanation of fibre fatigue must also be considered. For instance it has been shown that a definite fatigue effect exists with boron filaments [15] at a relatively high fraction of their ultimate fracture strain. In [1] it was shown that the failure strain of the composite in tension was of the order of 0.55, about twice the value obtained at 10^6 cycles in fatigue. On this basis it is suggested that fibre fatigue is not the reason for failure of the composite.

In general, the interpretation of matrix-controlled fatigue behaviour of the composite agrees with the comments made by Toth [16] for

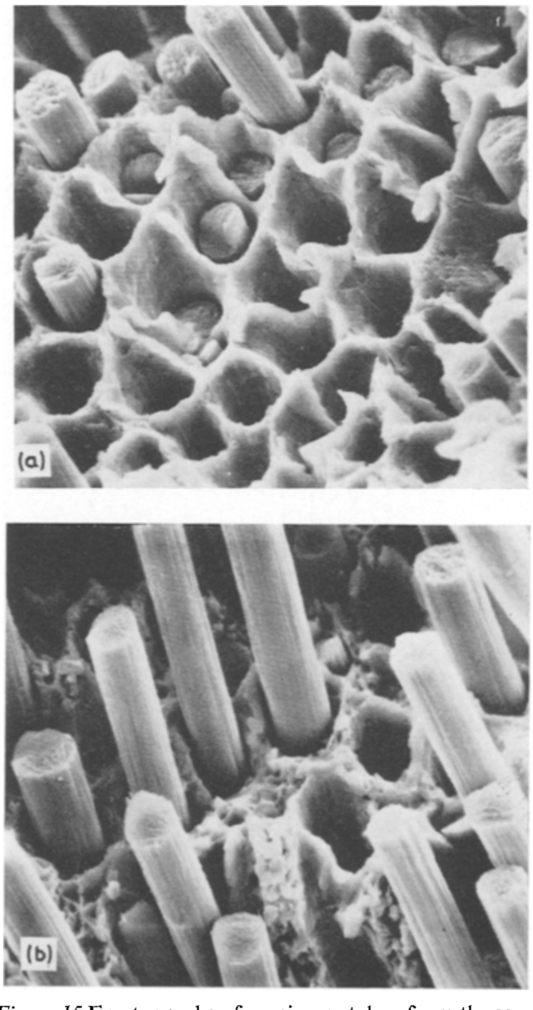

Figure 15 Fractographs of specimens taken from the same hot-pressed blank: (a) After $\sim 10^3$ cycles at a stress of 300 MN/m^2 ($\times$ 780). (b) After 2×10^4 cycles at a stress of 280 MN/m^2 ($\times$ 840).

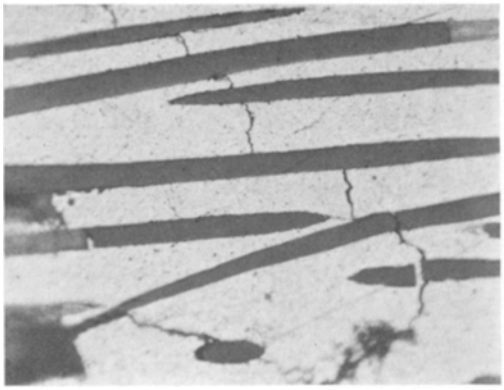

Figure 16 Micrographs of a longitudinal taper section through a carbon-aluminium specimen after fatigue at a stress of 210 MN/m^2 for $\sim 4 \times 10^5$ cycles, showing matrix cracks passing around fibres. (Fibre diameter 8 μm.)

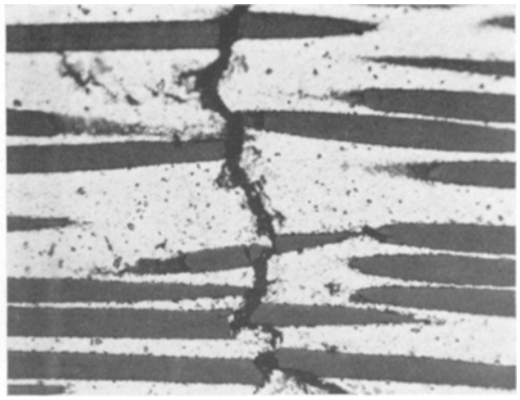

Figure 17 Micrographs of a longitudinal taper section through a carbon-aluminium specimen (hot-pressed at 600°C) after fatigue at a stress of 17 MN/m^2 for $\sim 2 \times 10^6$ cycles.

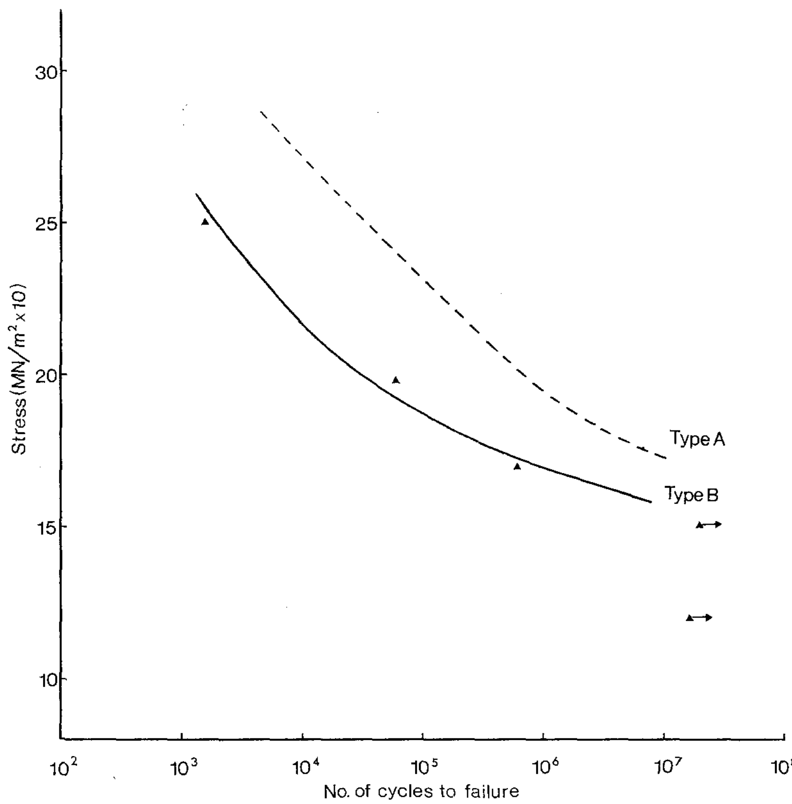

Figure 18 Plot of stress versus cycles to failure for a carbon-aluminium specimen produced from type B fibre.

his results on tension/tension fatigue for a similar boron-aluminium composite (although in this case the matrix was not heat-treated). Because matrix cracks were not observed to pass through fibres it was suggested that cyclic work hardening of the matrix near pre-fractured fibres resulted in filament fracture in adjacent regions.

It seems reasonable to conclude that the fatigue

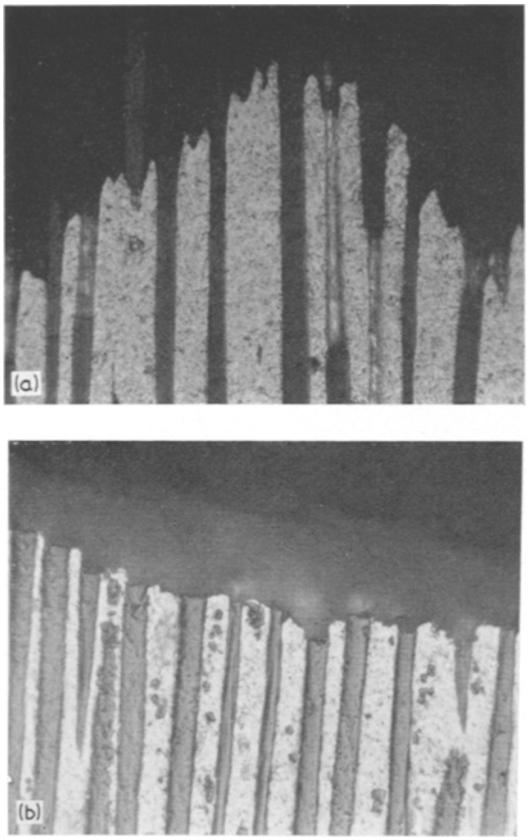

Figure 19 Micrographs of longitudinal sections through a carbon-aluminium composite testpiece (type B fibres): (a) Fractured in static bending. (Fibre diameter 8 μm.) (b) Fatigued at a stress of 170 MN/m² for $\sim 6 \times 10^5$ cycles. (Fibre diameter 8 μm.)

failure in heat-treated specimens is by matrix controlled fibre fatigue although the precise mechanism has not been identified.

At elevated temperatures it is suggested that the main cause of failure is matrix fatigue. The fatigue properties of the matrix would be adversely affected by the prolonged ageing under fatigue conditions where the effective diffusion conditions are representative of a considerably higher temperature. This effect was seen as fatigue-crazing of the matrix with pieces actually stripping from the surface of the fibres, which results in a considerable drop in fatigue properties. It is significant that under high-stress low-cycle conditions the fatigue properties were at least as good as for composites tested at ambient temperature, and fibre fatigue was not observed. This strongly suggests that cracks in the softer matrix cannot propagate through fibres.

6.2. Carbon-aluminium fibre composites

In the main series of samples tested (type A fibres hot-pressed at 550°C) fatigue failure appeared to be caused mainly by matrix crack propagation through pre-broken fibres. This mechanism is similar to that described in some earlier work on a model composite of aluminium reinforced with discontinuous stainless steel wires [7]. In addition it was found in this model that the fibre ends also initiated local matrix cracking. Although several examples of this effect were seen in the carbon-aluminium composite it was not as general and many examples of broken fibres in the high stress region which did not initiate matrix cracks were found.

Although fatigue fracture surfaces of normally ductile metals do have a brittle appearance, the granular appearance found in these carbon-aluminium composites is rather suspect. The ability of the fatigue process to cause failure of fairly strongly bonded interfaces in the matrix was shown in the boron-aluminium composites and in composites of aluminium reinforced with silica fibres [17]. This suggests that the ductile behaviour (when tested in tension) implied from observations [11] on composites produced under these fabrication conditions is suspect and that the matrix may not be fully bonded. A similar effect was found by Braddick during fatigue of carbon fibre reinforced nickel [18]. In this case the nickel matrix which had been made ductile by prolonged heat-treatment was found to revert to brittle behaviour in fatigue. Thus the fatigue strain limit of the matrix must be very limited in these composites and this together with the presence of broken fibres could account for the rather poor fatigue properties obtained.

The degree of fibre damage should have been considerably lower in composites hot-pressed at 600°C. This did not, however, result in improved fatigue properties. Again matrix fatigue cracks were observed to propagate through pre-broken fibres but in this case direct propagation through unbroken fibres was suspected. This may have been due to an increased fibre/matrix bond strength and even possibly fibre weakening caused by the chemical interaction between fibres and matrix at this temperature. Furthermore, the presence of the reaction product aluminium-carbide (Al_4C_3) may have reduced the fatigue strain capabilities of the matrix.

The poor fatigue properties obtained from composites obtained from type B fibres was very disappointing because fibre damage produced during composite manufacture should have been minimal in this case. However, under the conditions used (550°C, 77.3 MN m^{-2}, 1 h) the matrix was no better bonded than for the main type A fibre series and the lower modulus of the fibres (Table I) increases the strain range imposed on the matrix at a given stress in the composite, e.g. Equations 1 or 2. Furthermore, tensile tests on these composites [11] showed that the degree of fibre pull-out was much less than with composites prepared from type A fibres. This suggests that matrix fatigue cracking and direct crack propagation through fibres can occur at low stresses in the composite and is the reason for the poor fatigue properties. It is, however, hard to obtain unambiguous metallographic evidence for direct crack propagation.

It is unfortunately not possible to compare the boron-aluminium and carbon-aluminium composites on the basis of the differences in fibre diameter. This is because of the large number of complications associated with the behaviour of the carbon-aluminium composites. Furthermore, it is obvious that there is considerable room for improvement in the consolidation stage of these composites. Ideally the matrix bond must be improved without breaking fibres or causing fibre/matrix chemical interaction.

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