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Fatigue failure in graphite fibre and glass fibre-polymer composites

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Our studies have established that unidirectional graphite fibre composites show excellent fatigue resistance with only a 20 to 30% decrease in strength with cycling. Fatigue failures invariably occurred on the surfaces undergoing compression and were identified by scanning electron microscope studies as resulting from matrix failure adjacent to local fibre buckling failure zones. In contrast, glass fibre composites showed a much larger (70%) loss in strength under cyclic loading. At intermediate lives, failure occurred by the growth of matrix microcracks followed by delamination, while at long lives, the applied stress levels were below the microcrack initiation stress and behaviour was characterized by crack nucleation processes.

These results have suggested a criterion for predicting high cycle fatigue strength which is based on the hypothesis that for failure to occur, the maximum applied effective cyclic strain in the composite must exceed a critical value which depends upon the fatigue response of the matrix material. The main assumption is that localized fatigue failures in the matrix are the predominant contributions to the ultimate fatigue failure of the composite.

1. Introduction

The use of fibre-reinforced materials in load bearing structural applications depends, to a large extent, on their ability to withstand cyclic loading. It is important for the designer of such an application to be well aware of the fatigue properties of these materials, especially their long-term fatigue strength. The designer, however, has at his disposal, a bewildering array of materials that consist of innumerable combinations of fibres and matrices in addition to lamination orders and orientations. Since a design based on static considerations will not necessarily survive in a fatigue environment, extensive testing of actual prototypes is often required before an optimum design is reached. There is, therefore, a well-established need for a logical procedure for characterizing fibre-reinforced materials to provide a rational basis for the selection of the fibre, matrix and lamination order that is required for survival in a given fatigue environment.

Cyclic deformation processes in composites are qualitatively different from those in metals and polymers and are characterized or affected by: (a) the presence of initial microscopic flaws and defects (voids, resin-rich areas, fibre breaks, etc); (b) the predominance of a variety of failure modes including matrix crazing or microcracking, individual fibre failures, debonding, delamination, etc, several of which may be present simultaneously at any given time prior to failure; (d) mixed mode crack growth; (e) failure initiation in either fibre or matrix depending upon the applied stress and fibre orientation; and (f) the differences in failure mechanisms that govern failure in the fibre from those that govern failure in the matrix or at the interface.

These factors make the study of fatigue failure in composites highly challenging. In this paper, the approach is to attempt an understanding of the mechanisms of fatigue failure in glass and graphite fibre-reinforced polymers. Once failure modes are characterized, a fatigue failure criterion is proposed which would also, it is hoped, apply to other combinations of fibre, matrix and orientation.

2. Experimental procedure

2.1. Glass fibre-reinforcement

Specimens were prepared by filament winding glass fibre roving in a mould which formed the final shape of the specimen, followed by vacuum impregnation with a degassed low-viscosity epoxy resin (Epon 815/TETA, The Shell Chemical Company). After a room temperature cure of 24 h, post-curing was done at 66° C for another 24 h. This procedure produced void-free specimens with good uniformity in fibre distribution [1]. The specimens were 6 in. (15.2 cm) long with a gauge length of 1.5 in. (3.8 cm) and a gauge cross-section of 1/16 in. x 3/8 in. (1.6 mm x 9.5 mm). The specimen shape (shown in Fig. 1a) was found to be satisfactory since it was successful in ensuring that failures occurred in the gauge section and not in the (wedge-shaped) grips. This can be a serious problem when highly anisotropic materials are fatigue tested along their maximum strength direction. Specimens were produced with fibre volume fractions of 0.16, 0.33 and 0.50. Fibre dispersion was achieved by carefully distributing the roving during filament winding.

In addition to the glass fibre–epoxy specimens, several tensile specimens of epoxy alone were made by vacuum casting the resin. The specimen shape corresponded to ASTM D738 Type I with a gauge length of 2 in. (5.1 cm) and a rectangular cross-section 0.125 x 0.500 in. (0.3 x 1.3 cm).

2.2. Graphite fibre-reinforcement

Composites of PAN-based graphite fibres in a polyester matrix were obtained in the form of pultruded strip 0.125 in. x 1 in. (0.3 cm x 2.5 cm) and in lengths of 5 ft (1.52 m). Rectangular bend specimens were made by simply cutting the pultruded strip into 6.5 in. (8 cm) lengths (Fig. 1b). Two fibre volume fractions were obtained: 0.33 and 0.50.

All mechanical tests were performed in air at room termperature in electrohydraulic closed-loop systems. The glass fibre specimens were cycled axially in stress control in zero-tension while the graphite fibre specimens were fatigued in one-way bending in strain control in a four-point bending rig. In this mode, one surface of the beam specimen experienced a cyclic stress variation that was only in the compression regime, while the other surface was stressed only in tension. Failures could then be classified as tension or compression failures. In linear materials such as these, the

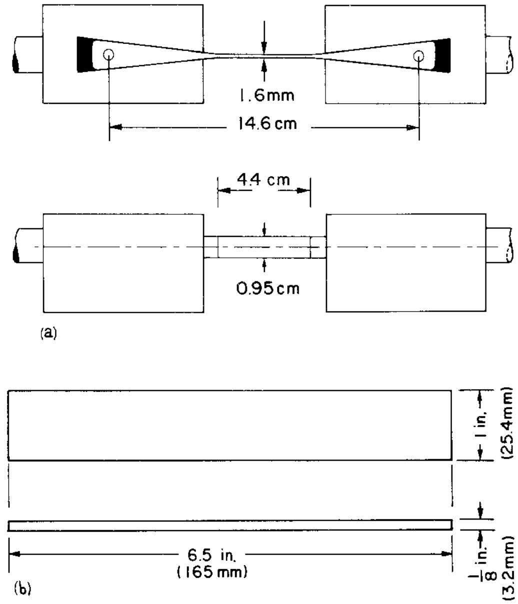

Figure 1 (a) Glass fibre–epoxy specimen and grips. (b) Graphite fibre–polyester bend specimen.

difference between stress and strain cycling is small especially if the same criterion for failure is used. The epoxy specimens, which are non-linear, were cycled in zero-tension in strain control.

The stiffness of the specimen was continuously monitored during cycling and the decrease in the stiffness during progressive fatigue failure was recorded as a function of the number of elapsed cycles. A 15% loss in stiffness was considered as failure. This generally occurred when at least one or more transverse cracks had propagated sufficiently into the specimen to cause delamination. In addition to providing a reasonable measure of the degree of fatigue damage, this criterion was also useful in evaluating composites for those applications in which a specified stiffness is the primary design requirement, as in chassis frames, helicopter and turbine blades, leaf springs, etc. In such components, a decrease in the stiffness, which can occur well before fracture, will quickly result in component failure.

The cyclic frequencies were kept low (2 to 4 Hz) to minimize heating. The actual frequency used for a particular strain amplitude was chosen so that a strain-rate of $2 \times 10^{-2}\,\text{sec}^{-1}$ was obtained with a triangular strain–time profile. This strain-rate was maintained for all the cyclic tests; therefore, high strain amplitude tests were conducted at lower frequencies and vice versa.

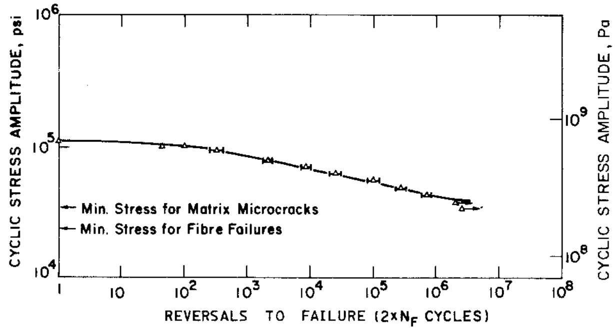

Figure 2 Log applied stress amplitude versus log reversals for 0.50 fibre volume fraction glass fibre composite.

3. Results and discussion

Results of the fatigue tests on the 0.50 volume fraction glass fibre epoxy are shown in Fig. 2 in which the logarithm of the cyclic stress amplitude is plotted as a function of the logarithm of the number of stress reversals (or twice the number of cycles). Also shown are the minimum stress values at which fibre failures and matrix microcracks were observed in a tension test. These were determined in an earlier study in which by illuminating the translucent specimen obliquely, the appearance of individual fibre breaks and microcracks could be detected through a microscope during the progress of each test [1].

Three regions of interest may be noted in Fig. 2. At low cycles, the applied cyclic stress was high enough to be sufficiently within the fibre flaw strength distribution so as to produce a high fibre failure density in the first cycle. Failure then occurred by the coalescence of local fibre breaks which then propagated in a few reversals to connect other such regions until the specimen failed catastrophically. Although failure is fibre-controlled, since fibre breaks occur at random throughout the volume of the material in sufficiently large numbers to cause incipient failure, stress reversals were necessary to cause actual separation within a few cycles.

At intermediate lives, the logarithm of the applied stress was proportional to the logarithm of the number of stress reversals, or

$$\Delta\sigma = K(2N_\mathrm{f})^b \tag{1}$$

where $b = -0.1$. In this region, microcracks were initiated and propagated during cycling until delamination began. The few fibre breaks that were formed during the first cycle did not increase in number with cycling and were not very important in influencing fatigue behaviour. However, the matrix microcracks appeared to play an important role, since their nucleation and growth during cycling resulted in fibre failure at the crack fronts followed by delamination and eventual failure.

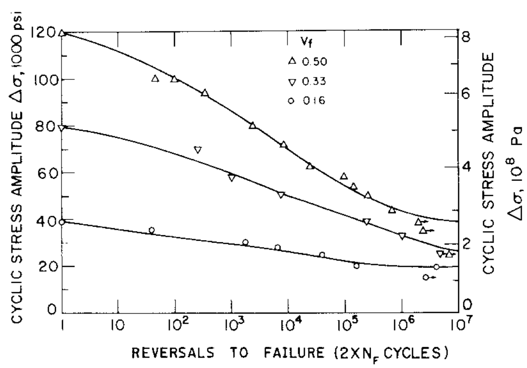

Figure 3 Cyclic stress aplitude versus log reversals for 0.50, 0.33 and 0.16 fibre volume fraction glass fibre composites.

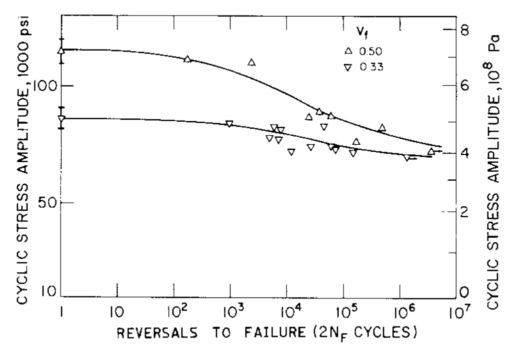

Figure 4 Cyclic stress amplitude versus log reversals for 0.50 and 0.33 fibre volume fraction graphite fibre composites.

In the high cycle region (greater than 10^6 reversals), the applied stress was approximately that of the microcrack initiation stress. The few specimens tested in this region did not fail. No microcracks were observed initially; however, after several cycles, well into each test, a few were observed to have nucleated. When the test was stopped (at 2×10^6 reversals), the specimens were otherwise undamaged.

Similar fatigue responses were obtained in specimens with 0.33 and 0.16 fibre volume fractions. These are shown in Fig. 3 (together with the 0.50 fibre volume fraction data) in which the stress ordinate is linear in order to spread the data. In each case, the stress levels at which runouts (no failure) occurred corresponded to the stresses required for microcrack initiation.

Results of the bend fatigue tests on the graphite fibre composites are shown in Fig. 4. Since the cyclic stress–strain curves for these materials (unidirectionally reinforced and stressed parallel to the reinforcement) are practically linear, the strain–life curves were converted to the stress–life curves in Fig. 4 using the longitudinal elastic modulus. Very little reduction in strength with cycling is seen; the strength at 10^6 cycles is about 70 to 80% of the static strength. These results are quite unlike the results for glass fibre composites which exhibit fatigue strengths at 10^6 cycles which are only 30 to 40% of the static strength. All fatigue failures in the graphite fibre composites occurred on the compression side of the specimen. Owen and Morris have also noted in their studies on carbon fibre composites that fatigue failures were always initiated on the compression surfaces [2].

In the present study progressive fatigue failure in graphite fibre composites was observed to occur in the following stages:

(a) initiation at a zone of fibre buckling;

(b) localized delamination at the failure zone;

(c) formation of a surface transverse crack with a width that depended upon the applied stress, and subsequent propagation of the crack into the specimen; and

(d) delamination or mode II crack propagation parallel to the fibres.

An earlier scanning electron microscopy study of fatigue fracture surfaces showed that the fracture morphology of the fibres, which failed at the crack initiation zone on the compression surface, was similar to that which is generated when a single fibre is fractured in bending [3, 4]. The conclusion was that failure occurred as a result of fibre local buckling during cyclic compression.

When all the data obtained in this investigation are plotted in terms of the applied cyclic *strain* versus the number of stress reversals to failure, a clearer picture of fatigue emerges. This is shown in Fig. 5 in which the results for the epoxy matrix are also plotted. All the data for the glass fibre composites fall within a band which has a strain width of about 0.002. When compared with the fatigue response for the epoxy matrix, one notes that between 10 and 10^6 reversals, for a given cyclic strain amplitude, the composite outlasts the matrix by about two orders of magnitude. Beyond 10^6 reversals, however, the points for composite and matrix approach each other and almost coincide. The significance of this observation is that for long-life survival, the cyclic strain amplitude in the composite should be less than or equal to the matrix endurance strain (defined here as the fatigue strain at 10^6 reversals). Since glass fibre has a low modulus (as compared with graphite), the strain at static failure is large. Cycles at strain levels less than the static failure strain but greater than the matrix endurance strain, results in matrix fatigue failure followed by composite failure. The data in Fig. 5 inidicate that at strain levels of 0.0075 or less, the composite (or the matrix) will not fail at 10^6 cycles.

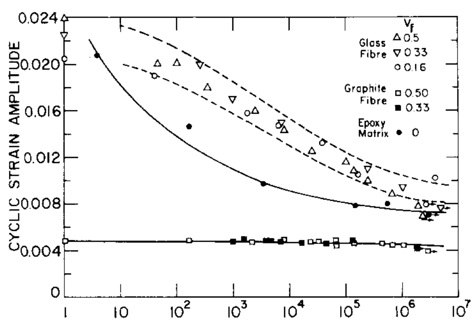

Figure 5 Cyclic strain amplitude versus log reversals for the epoxy matrix and glass fibre and graphite fibre composites.

With graphite fibres, however, a different mechanism operates. Because of their high modulus, the strain level at which static fracture occurs is only 0.005 which is well below the endurance strain of the matrix. The matrix should not then fail during cycling and the mechanisms of fatigue failure should depend upon the flaw strength distribution of the buckling fibres. The narrow range of strain in which failure occurs appears to support this conjecture. When the applied strain level is close to the static compressive failure strain, the condition for buckling instability is reached and failures occur at the weakest flaw locations on the most highly stressed fibres.

Analytical models that predict the static compressive strength of composites have met with only limited success. The one-dimensional model by Rosen [5] gives the compressive strength for shear mode buckling as

$$\sigma_c \cong \frac{G_m}{1 - V_f} \qquad (2)$$

where G_m is the shear modulus of the resin. This equation predicts a compressive strength that is larger than measured strength values by about a factor of two. In the present case, for G_m = 120 000 psi* the strength for 0.33 and 0.50 fibre volume fractions from Equation 2 are, respectively, 180 000 and 240 000 psi as compared to 89 000 and 120 000 psi obtained experimentally.

There are several reasons for this discrepancy. (a) The model assumes a nominal fibre volume fraction V_f resulting from uniform fibre spacing. In real composites, however, there are local variations in V_f which will reduce the value of σ_c. In a previous study, fatigue cracks in graphite fibre composites (after initiating at the compression surface) were observed to propagate in shear along resin-rich (or low V_f) zones [3]. It is possible, therefore, for the presence of local resin-rich areas to reduce the compressive strength as predicted in Equation 2. One cannot carry this argument too far, however, since Equation 2 is valid only for moderate fibre volume fractions (~ 0.3 to 0.6). (b) Use of the tangent shear modulus instead of the elastic modulus will further reduce σ_c. However, in the present case, the low strain at which failure occurs, places the stress state in the matrix well within the elastic region. (c) Hanasaki and Hasegawa have recently shown that initial curvature in the fibres can reduce drastically the theoretical compressive strength [6]. (d) Finally, graphite fibres are anisotropic in strength, possessing low transverse and shear properties. Elastic loop tests on individual fibres have shown buckled areas on the compression side of the loop indicating local microbuckling of the fibre structure [4]. Fibre failures initiated at strain levels (0.005 maximum for an individual fibre) much lower than the fracture strain or even the endurance strain of the matrix. These failures are dependent upon the distribution of flaws on the fibre surface. One type of flaw has been suggested as resulting from discontinuities in the crystallographic structure produced during carbonization of precursor material deposited on fibre surfaces as the fibres emerge from the spinneret during manufacture [7].

These factors suggest that the fatigue responses of graphite fibre composites are governed by the conditions that influence fibre buckling and by the flaw strength distribution. Fibre failure zones are formed during the first few cycles of the test at favourable locations such as, local fibre flaws, resin-rich areas and regions of initial fibre curvcurvature. These cracks then propagate during cycling and cause ultimate failure. The behaviour is very similar to that observed for glass fibre composites at very low cycles (<100 stress reversals) (Fig. 2), where the applied stress levels are high enough to cause significant fibre fractures during the first few cycles. In the case of graphite fibre composites, this behaviour extends throughout the fatigue range since at stresses below the static compressive strength, fibre buckling does not occur, the matrix experiences a stress level less than that required to initiate microcracking, and the composite does not fail in fatigue.

From the foregoing, a fatigue failure criterion for unidirectional composites stressed along the fibre direction may be postulated as follows:

$$\begin{aligned} \sigma_{ec} &= \epsilon_{em} E_c \qquad \text{when } \frac{\sigma_{fc}}{E_c} > \epsilon_{em} \qquad (3) \\ \sigma_{ec} &\approx (\sigma_{fc})_{min} \qquad \text{when } \frac{\sigma_{fc}}{E_c} < \epsilon_{em} \end{aligned}$$

where σ_{ec} is the composite long-life fatigue stress, ϵ_{em} is the matrix long-life fatigue strain, σ_{fc} is the (minimum) composite static fracture stress and E_c is the composite modulus. It should be noted that Equations 3 and 4 apply only to a single lamina under uniaxial stress parallel to the reinforcement axis. For angles other than zero between the applied stress and the fibre direction, the matrix is subjected to shear strain in addition to normal strains, and an equivalent strain should be calculated depending upon the particular yield theory that applies. This is the subject of an ongoing investigation the results of which will be reported at a later date.

4. Conclusions

(1) Glass fibre composites exhibited a significant loss of strength with cycling. At intermediate lives,

* 10^3 psi = 6.89 N mm^{-2}.

failure occurred by the growth of matrix microcracks followed by delamination, while at long lives the applied stresses were below the microcrack initiation stress and behaviour was characterized by crack nucleation processes.

(2) Graphite fibre composites showed excellent fatigue resistance with little loss in strength. The fatigue failures that were observed invariably occurred during the compression phase of the cycle and were attributed to local failure from fibre buckling.

(3) Improvements in the static and fatigue strengths of graphite composites should be possible by (a) careful control of fibre quality during production; (b) uniformly distributing the fibres in the matrix; or alternately, by increasing the fibre volume fraction in areas experiencing the highest compressive stresses, for example, at surfaces in bending; (c) increasing the resin shear modulus G_m, perhaps by the introduction of additional (non-fibrous) fillers; and (d) producing composites with well collimated straight fibres with a minimum of local orientation defects.

(4) A fatigue criterion for long-life survival is proposed which is based on the hypothesis that for failure to occur, the maximum applied effective cyclic strain in the composite must exceed a critical value which depends upon the fatigue behaviour of the matrix material. The main assumption here is that high-cycle fatigue failures in the matrix are the predominant contributions to the ultimate fatigue failure of the composite.

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