PHILOSOPHICAL TRANSACTIONS OF THE ROYAL SOCIETY A MATHEMATICAL, PHYSICAL & ENGINEERING SCIENCES

Fatigue Processes in Fibre Reinforced Plastics

M. J. Owen

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Fatigue processes in fibre reinforced plastics

By M. J. Owen
Department of Mechanical Engineering, University of Nottingham, Nottingham, U.K.

[Plate 1]

Glass reinforced plastics have been available for 40 years and there is now a substantial literature describing their properties under fatigue loading. Designing to avoid incipient damage leads to very uncompetitive designs. Some damage must be tolerated and although there is growing interest in damage studies, as yet there is no fully accepted design method for g.r.p. Individual items appear to be devised on an *ad hoc* basis. With the advent of new fibres such as carbon, boron and Kevlar it should be possible to make more rapid progress using the experience gained for g.r.p. The properties of the newer fibres and possible developments in polymer matrices are reviewed against the background knowledge of the fatigue of metals and g.r.p. Micromechanics studies can help to achieve an understanding of the failure behaviour.

Introduction

Engineers have learned through bitter experience that a large proportion of engineering failures can be attributed to fatigue, i.e. material failure produced by repeated or random service loads, each of which is less than the single load that would be expected to produce failure. Although the fundamental processes involved must differ, the macroscopic phenomenon of fatigue has been observed in almost all structural materials: metals, concrete, bitumen, wood, plastics, etc. When new materials come along engineers usually compare and contrast their properties with current materials and seek to apply them by familiar design methods. The compilation of property data and the development of improved design techniques are very expensive and inherently slow. Thus, new materials, if they are to be successful, i.e. profitable to their manufacturer, must usually find relatively large volume applications in either non-critical or undemanding situations.

It is not the author's task to provide an exhaustive literature survey of fatigue failure mechanisms in fibre composites. Instead it is proposed to try to draw attention to the manner in which composites fail under repeated loading, how this differs from metals, what traps there may be and what improvements should be sought. To accomplish this it is proposed to review briefly the salient facts about fatigue of metals and g.r.p., to consider the micromechanics of fibre composites and then to consider the newer fibres.

Fatigue of metals

Repeated loading studies of small metal samples and components have resulted in the definition of the stress–log life (S–N) curve for complete separation of the specimens (Frost *et al.* 1974). The use of replicates demonstrates that there is considerable scatter in fatigue life which tends to be worse in small components. Notches (discontinuities of shape) have a marked effect on fatigue life and are generally almost fully effective stress concentrators especially for higher strength materials. Fatigue of metals is usually surface initiated and any factor that affects the surface condition, such as corrosion, fretting, surface roughness, or surface stress arising from machining, working or polishing processes, is likely to have a marked effect. Studies of failure processes show that metal fatigue can be separated into initiation and propagation mechanisms and that the initiation mechanism in small samples may occupy as much as 90% of the life. Up to this stage changes in tensile strength and modulus etc. are negligible. Propagation usually takes the form of a single rapidly propagating crack normal to the maximum principal tensile stress. Propagation does not normally occur under compression or in the compressive part of a load cycle. Superficially the fatigue fracture surface has a brittle appearance characteristically different from the usual ductile or semi-ductile failure.

In recent years it has been realized that fatigue data and design methods based on small samples are inadequate for larger structures where a significant number of material or manufacturing defects are virtually inevitable. Under these circumstances defect growth and crack propagation become the dominant behaviour. New methods of crack growth prediction based on fracture mechanics concepts have been developed. Without these it would be difficult to understand failures in large structures and perhaps impossible to assess the significance of flaws.

Fatigue of g.r.p.

G.r.ps have been available for about 40 years and the first fatigue results were published about 25 years ago (Boller 1952). Progress has been slow and there is still no generally agreed design method. Initially, small samples were tested as if they were metals. Specimens were cut with their axes parallel and perpendicular to the principal material axes and it was overlooked that they were not only anisotropic but also inhomogeneous on both the macro- and micro-scales. Load cycles were applied until fracture of the specimen occurred. It was not always clear in early publications whether specimens were subjected to constant load or constant deformation testing. Several years passed before damage processes were first described (Throckmorton *et al.* 1963). With multidirectional reinforcements (both random and systematic arrangements) under both static and fatigue loading, the first signs of damage appear to be separation between the glass filaments and resin matrix where the filaments lie perpendicular to the principal tensile stress. This damage, unlike the initiation process in metals, is found throughout the stressed region of the material. The subsequent development of damage depends to a large extent on the construction of the material and the mode of loading. In low glass content materials with random fibres, fibre and resin separation is followed by the development of resin cracks perpendicular to the maximum principal stress (Owen & Dukes 1967). Under repeated loading, damage is intensified until there is considerably more damage than at failure under monotonic loading (Owen & Howe 1972). The progressive development of damage leads to reductions in modulus (Smith & Owen 1968) and tensile strength (Owen & Howe 1972). Eventually, in small samples further intensification becomes localized and separation occurs. In orthogonally plied non-woven materials, debonding spreads quickly from one ply interface to the next before spreading along the ply interface and among the aligned fibres (Broutman 1964). There may be a considerable delay before the aligned fibres are affected. In spite of the observations of progressive damage accompanied by deterioration in other properties it is still the usual practice to present data in the form of an S–N curve or as a derived constant life diagram which gives no indication of the degradation of the material (Owen 1970).

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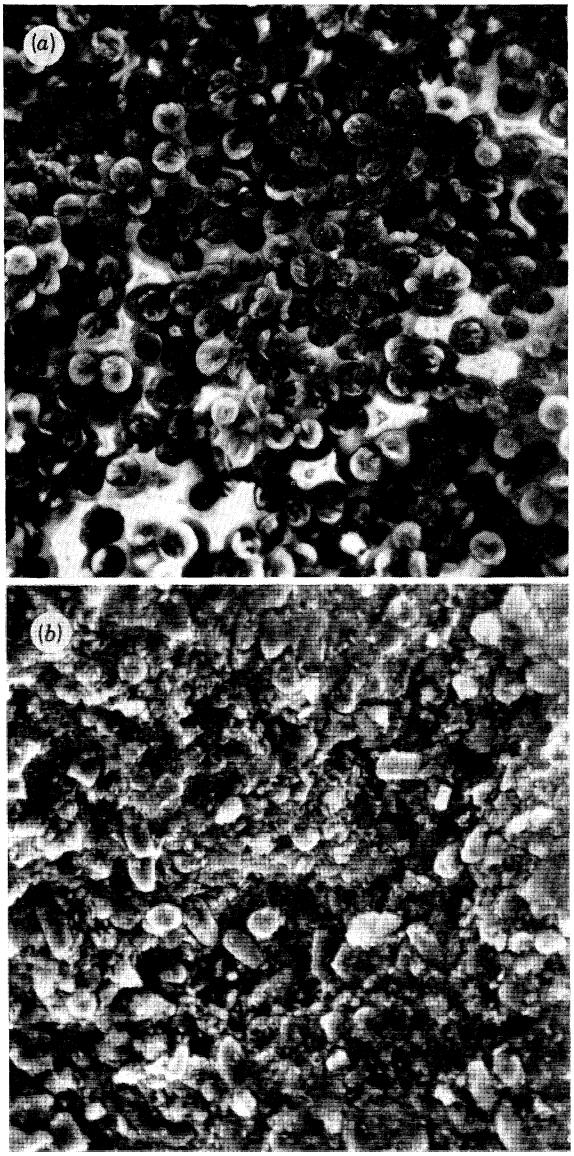


FIGURE 1. Fracture surfaces for high modulus surface treated carbon fibres in a wet lay-up epoxy resin system (Morris 1971): (*a*) static tension (×500); (*b*) reversed stress fatigue (×450).

In spite of the fact that many workers have published optical and scanning electron micrographs to illustrate both static and fatigue damage there seem to have been few, if any, systematic studies aimed at quantifying damage, distinguishing damage produced in various modes, or identifying fundamental damage processes. Quantifying damage may well assist the development of cumulative damage rules. Detailed descriptions of damage would be of assistance in failure analysis and identification of fundamental processes should assist in development of improved materials. Fracture surfaces of fatigue specimens are often covered with fine debris not present on the static failure surface. Figure 1 (Morris 1971) illustrates this for a carbon fibre reinforced plastic.

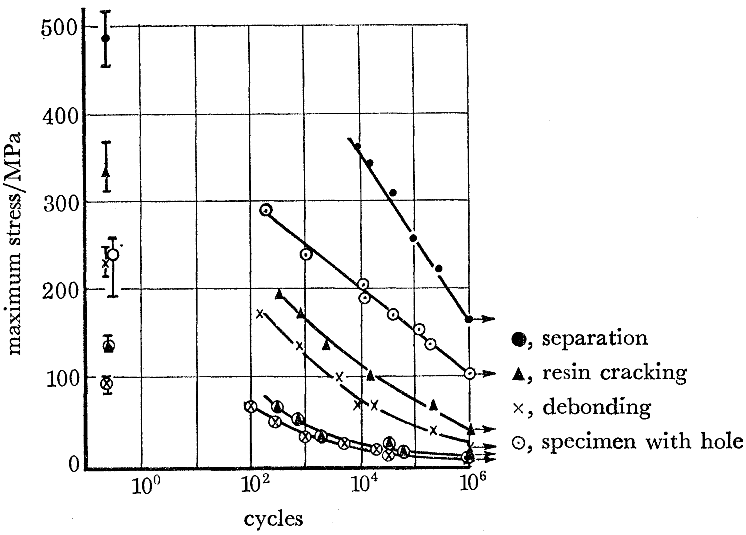


FIGURE 2. Zero-tension fatigue results for a glass fabric reinforced polyester resin composite showing the onset of debonding, resin cracking, and separation stages both with and without a hole in the specimens. Static strength ranges are shown at 0.25 cycles (Owen & Bishop 1973).

Progressive damage in g.r.p. is potentially of considerable importance to the designer, and figure 2 (Owen & Bishop 1973) compares fatigue curves for a fabric reinforced material at the onset of debonding damage, resin cracking, and separation of the specimen both with and without a hole. In the presence of a hole some debonding damage occurs at only 2% of the mean static ultimate strength of the material. The work illustrates quite clearly that each of the significant stages of damage is itself subject to the fatigue phenomenon. A design that is based on less than 2% of the u.t.s. is likely to be hopelessly uneconomical, and methods of avoiding or tolerating damage must be developed.

MICROMECHANICS

Consider a model material consisting of two sets of orthogonal fibres (figure 3). Under uniaxial loading, one set is loaded parallel to the fibres and the other set perpendicular to the fibres. Some years ago micromechanics studies were in vogue and these can contribute to an understanding of the observed modes of failure (Dukes 1968).

Consider the aligned fibre group. Under axial load, strain compatibility requires the common extension or compression of the system. Under tension there is the possibility of either fibre failure or resin failure according to the balance of properties. Similarly under compression either the matrix may yield or the fibres buckle. However, under axial loading differences between the elastic properties of the fibres and resin also give rise to micro-stresses in the transverse plane. Figure 4*a* shows the effect of compressive axial strain for several fibre moduli and one resin system at 65% volume fraction. Figure 4*b* shows the signs of the principal stresses in the matrix for an axial compressive load. All the stresses are reversed for a tensile load. There

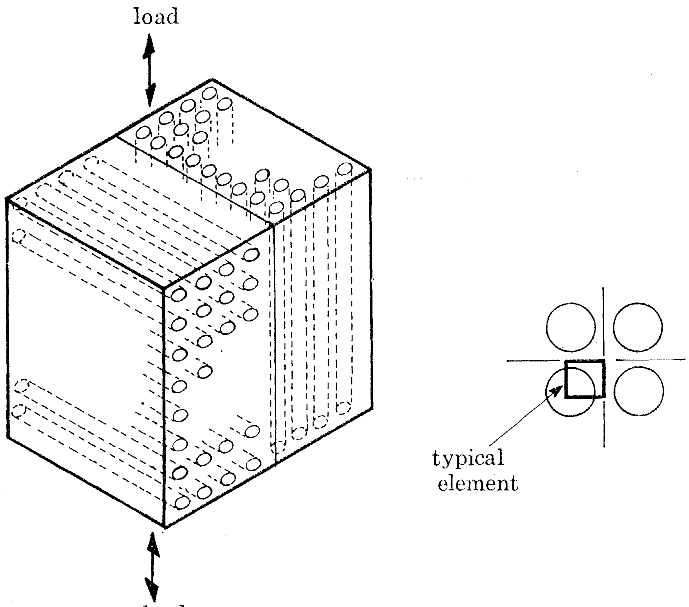


Figure 3. Two layers of an orthogonally cross-plied model fibre reinforced composite. The inset shows the fundamental element for microstress analysis (Dukes 1968).

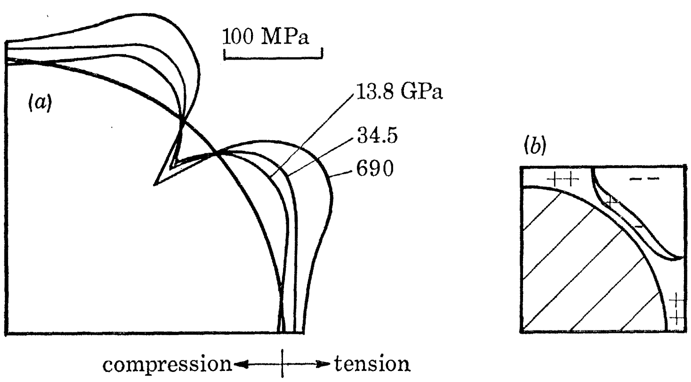


Figure 4. (*a*) Polar diagram showing the radial stress at the fibre–matrix interface caused by a 1% compressive axial strain on the model composite with 65% volume fraction of fibres. Fibre moduli as shown and matrix modulus 3.45 GPa (Dukes 1968). (*b*). Signs of the principal stresses in the matrix for the same model (Dukes 1968).

are several important observations: under axial load the matrix is under triaxial stress and some of these stresses are tensile whatever the axial stress; the interface is under both tension and compression in a closely spaced fibre array; the tensile stresses are more severe under compressive axial load.

Now consider the transverse fibre group. Figure 5*a* shows the effect of transverse tensile loading on a 65% volume fraction glass–resin system. Again it will be noted that around part of the interface there are compressive stresses, although of smaller magnitude than the maximum tensile stress. A sign change in the applied stress would reverse the stresses at the interface and again compressive stress produces some tensile interfacial stress. Figure 5*b* shows the variation in maximum stress concentration and strain concentration with fibre modulus. For isotropic fibres there is no great increase in these stress concentrations for fibres that are stiffer than glass. For anisotropic fibres with a low transverse modulus the stress and strain concentration factors may be significantly lower.

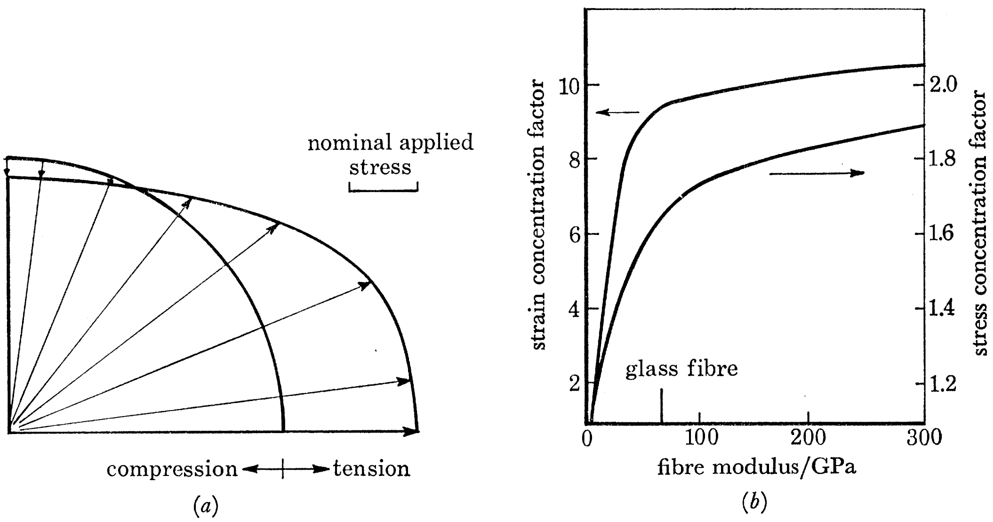


FIGURE 5. (*a*) Polar diagram showing the radial stress at the fibre–matrix interface for transverse loading on the model composite with 65 % volume fraction of fibres. Glass fibre in a matrix of modulus 3.45 GPa (Dukes 1968). (*b*) Maximum stress concentration at the interface and maximum strain concentration in the matrix for the model composite under transverse loading with various fibre moduli. 65 % volume fraction of fibres and matrix modulus 3.45 GPa (Dukes 1968).

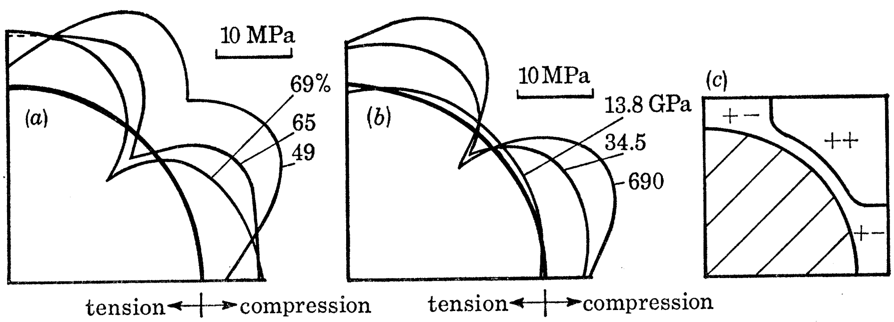


FIGURE 6. (*a*) Polar diagram showing the radial stress at the fibre–matrix interface caused by a 10 °C temperature fall for several volume fractions of glass fibre in a matrix of modulus 3.45 GPa (Dukes 1968). (*b*) As (*a*) for 65 % volume fraction and various fibre moduli and matrix modulus of 3.45 GPa (Dukes 1968). (*c*) Signs of the principal stresses in a matrix of modulus of 3.45 GPa caused by a fall of temperature (Dukes 1968).

Thermal stresses arising from matrix curing are inevitable in fibre reinforced composites. For an isolated fibre in resin this would result in a radial compressive stress at the interface accompanied by hoop tension in the resin. However, for closely spaced fibre arrays the situation is more complex, as shown by figures 6*a*, *b* and *c*. Tensile interfacial stresses are possible in closely spaced fibre arrays and are likely to affect the fatigue process.

When the orthogonal plies are bonded together, uniaxial loading gives rise to slight additional loadings in the separate plies because of the requirements of strain compatibility. These additional loadings become shear forces at the edges of the composite.

The micromechanics analyses show that for any closely spaced fibre array embedded in an elastic matrix either axial or transverse loading or a temperature change gives rise to complex micro-stresses. A simple orthogonal crossply subjected to axial loading gives rise to numerous failure possibilities depending on the balance of strengths. These include fibre tensile failure, matrix tensile failure, matrix yielding in compression, fibre buckling in compression, interface tensile failure and crack propagation in the matrix. It does not necessarily follow that failure under static and fatigue loading will be identical since this depends on the effect of fatigue on the separate mechanisms. If fatigue loading involves both tension and compression in the cycle there is the possibility of interaction, e.g. interfacial tensile damage during compression spreading as resin cracks during tension and subsequently permitting fibre buckling in compression.

In orthogonally cross-plied composites subject to axial loading the micro-stresses are governed by the axial strain, i.e. by the load and modulus of the aligned fibres. The stiffer the fibres the smaller will be the axial strain for a given load. Most commercial resins have a strain range at least compatible with the extension to failure of elastic fibres. The lower the modulus of the fibres the greater is the axial strain; hence the greater will be the transverse strain on the cross-aligned group and hence the greater will be the interfacial stress and strain in the matrix. This effect may be reduced if the fibres themselves are anisotropic, i.e. if they have a lower effective modulus in the transverse direction.

It is often suggested that a matrix resin with a greater strain to failure would give an improved performance in fatigue. Greater strain to failure usually means a greater area under the stress–strain curve and improved toughness. However, if the first stage of failure is by separation between the filaments, a matrix improved strain to failure only reduces the interfacial stress if there is a corresponding increase of matrix modulus. Most resin modifications also produce a small reduction in resin modulus and hence the debonding effect tends to be worse. Owen & Rose (1970) modified the strain to failure of a typical orthophthalic polyester resin by adding polypropylene maleate adipate and increased the nominal strain to failure from 2% to greater than 50%. The critical stress intensity factor was more than doubled and resin cracking was eliminated under static tensile loading. However, under fatigue loading it reappeared at less than 1000 cycles and there was no improvement in the long-term fatigue properties. Debonding was not suppressed under static or fatigue loading. The important point here is that improved fracture toughness does not necessarily imply improved crack growth resistance under repeated loading. This appears to have been almost entirely overlooked in the development of new matrix materials. A closely related point is that the interfacial strength is itself subject to a fatigue process (figure 2). A major improvement in the fatigue strength of g.r.p. systems would appear to result from a modification to the interfacial behaviour which would reduce the fatigue effect coupled with an improvement in the crack growth resistance of resin systems. If these benefits could be produced with roughly similar static strengths there should not be a reduction in composite toughness.

New fibres

It is now eight years since the first fatigue data were published for type 1 c.f.r.p. (Owen & Morris 1970). Under zero-tension or tension–tension loading for almost any resin matrix with or without fibre surface treatment fatigue tests produce an S–N curve which is almost horizontal and scarcely dips below the static strength scatter band. Flexural tests on unidirectional specimens at that time revealed that failures always occurred on the compressive side of the specimens, apparently by progressive fibre buckling (Morris 1971). Shortly afterwards static and fatigue tests revealed that the compressive strength was substantially lower than the tensile strength and that the *S–N* curve under zero-compression loading was similar in form to the tensile curve (Morris 1971). Further exploration by constant amplitude fatigue testing with varying mean stresses showed that the fatigue effect was greatest at low tensile mean stresses (figure 7). This early work was confirmed by Beaumont & Harris (1971) and has been extended to type 2 (Bevan 1977) and type 3 (Sturgeon *et al.* 1976) fibres. The fatigue effect is slightly greater for the less rigid type 2 and type 3 fibres. Undirectional boron and Kevlar fibre composites also show

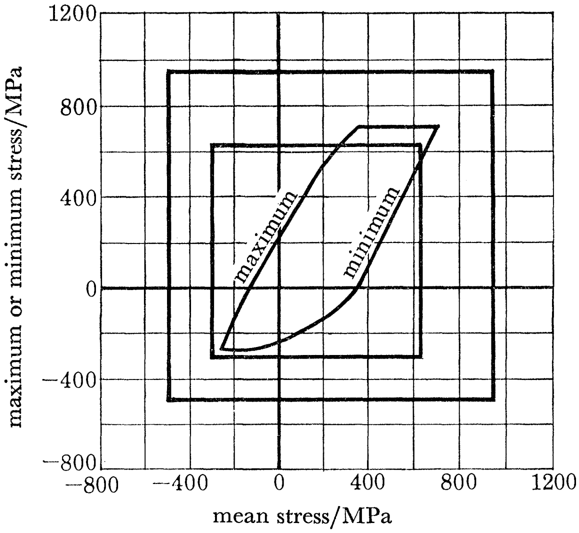


FIGURE 7. Modified Goodman diagram showing the failure envelope for 10^6 cycles life and its relationship with the static tensile and compressive strength ranges. Unidirectional type 1 surface treated fibre in an epoxy resin prepreg system at approximately 65 % volume fraction (Morris 1971).

exceptionally good fatigue properties under zero-tension loading (Miner *et al.* 1975). Later work with orthogonally cross-plied c.f.r.p. (Owen & Morris 1971) and more complex lay-ups $0 \pm 45°$ (Sturgeon *et al.* 1976; Schütz & Gerharz 1977) show very similar behaviour with both the static and fatigue strengths depending almost entirely on the proportion of fibre aligned with the loading axis. However, the $\pm 45°$ layers show considerable cracking. Donat (1970) presented fatigue results for $0 \pm 45°$ boron–epoxy material and found similar fatigue curves with very small slopes. He did not comment on failure mechanisms.

Early type 1 fibres were without surface treatments and the tensile and tensile fatigue failures had a brushlike appearance. After the introduction of surface treatment the static and fatigue tensile failure surfaces for both uniaxial and orthogonal cross-plies assumed a brittle appearance but without any significant change in strength (Owen & Morris 1970). In both cases it had to be assumed that individual fibres failed and these affected neighbouring fibres to form fractures. With high coefficients of variation on the individual fibre strengths fibre efficiencies were low and it is not difficult to assume a progressive mechanism of fibre failures.

It was soon realized that the fatigue behaviour of c.f.r.p. under axial loading was not representative of applications where the stresses are complex and generally vary in space in a manner that cannot be matched by changes in fibre orientation. Accordingly interlaminar shear and torsion tests were introduced in order to apply loads to the fibre interface and matrix. It rapidly became clear that these modes had a severe effect. Interlaminar shear specimens show an increasingly severe effect as the static interlaminar shear strength is improved (Owen & Morris 1972) (see figure 8). Torsion of rods produces severe cracking and loss of stiffness (Beaumont &

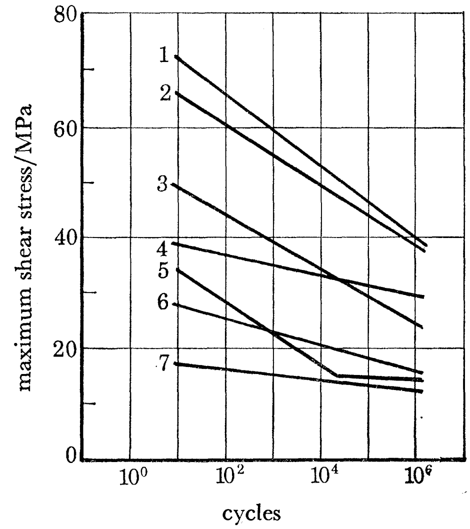


Figure 8. Interlaminar shear stress fatigue of various glass and carbon fibre reinforced plastics (Owen & Morris 1972): curve 1, cross-plied type 2 surface treated carbon fibre; curve 2, unidirectional 'S' glass; curve 3, cross-plied 'S' glass; curve 4, unidirectional type 1 surface treated carbon fibre; curve 5, unidirectional 'E' glass; curve 6, cross-plied type 1 surface treated carbon fibre; curve 7, cross-plied type 1 carbon fibre without surface treatment.

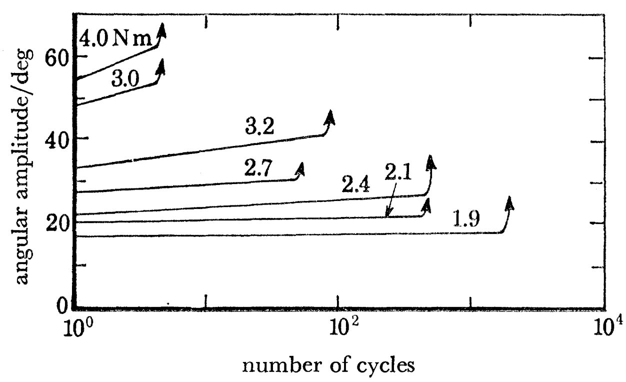


Figure 9. Variation of angular amplitude during constant torque amplitude cycling of c.f.r.p. rods (Phillips *et al.* 1978).

Harris 1971), especially in degrading environments. Recent work (Phillips *et al.* 1978) for g.f.r.p., K.f.r.p. and c.f.r.p. has confirmed similar behaviour for all three types of material in torsion. Phillips *et al.* show very clearly in their work that under constant angular deflexion there is a sudden reduction in torque, or alternatively under constant torque amplitude, a sudden increase in angular amplitude (figure 9). They interpreted this as a runaway crack propagation in the matrix defining the end of the useful life of the composite.

Discussion

It is clear that fibre reinforced plastics fail by progressive damage throughout the stressed region of small samples. Fatigue damage results in a progressive reduction in other properties and the conventional *S–N* curve is inadequate for conveying this information. The major weaknesses of f.r.p. are the fibre/matrix interface and the matrix itself. The strength of both of these is subject to the fatigue phenomenon. High modulus fibres (especially if they are less rigid in the transverse direction) give excellent fatigue properties in a composite when stressed in tension in the principal fibre direction. Under reversed stresses the fatigue effect is slightly greater due to interactive mechanisms. The fatigue effect is relatively severe under shear loading and it is made worse by environmental attack.

The fatigue properties of the new fibre composites can be made to look very good or very bad according to the mode of loading. Both of these views represent a trap and the truth lies in-between. Engineering components such as vehicle springs and transmission shafts are being developed which are fatigue resistant. The developoment work is expensive and its success represents a considerable commercial advantage; consequently there will be little detailed information about the techniques involved.

Fatigue damage has been described but this description neither is systematic, nor does it represent fundamental understanding. Published descriptions are as yet inadequate to form the basis for failure analysis and, as yet, understanding has not pointed the way to improvements. Part of the trouble here lies in the separate interests of fibre and matrix manufacturers and the fact that the composite is often made by the user.

References (Owen)

Beaumont, P. W. R. & Harris, B. 1971 *International Conference on Carbon Fibres, their Composites and Applications*, paper 49, pp. 283–291. London: The Plastics Institute.

Bevan, L. G. 1977 *Composites* **8**, 227–232.

Boller, K. H. 1952 *U.S. Department of Agriculture Forest Products Laboratory Report* no. 1823.

Broutman, L. 1964 19*th Annual SP Conference*, section 9*c*. Chicago.

Donat, R. C. 1970 *J. Compos. Mater.* **4**, 124–128.

Dukes, R. 1968 Ph.D. thesis, University of Nottingham.

Frost, N. E., Marsh, K. J. & Pook, L. P. 1974 *Metal Fatigue*. Oxford: Clarendon Press.

Miner, Lit, Wolffe, R. A. & Zweben, C. H. 1975 *Composite reliability*. ASTM STP 580, pp. 550–559.

Morris, S. 1971 Ph.D. thesis, University of Nottingham.

Owen, M. J. 1970 *Composites* **1**, 346–355.

Owen, M. J. & Bishop, P. T. 1973 *J. Phys. D* **6**, 2057–2069.

Owen, M. J. & Dukes, R. 1967 *J. Strain Analysis* **2**, 272–279.

Owen, M. J. & Howe, R. J. 1972 *J. Phys D* **5**, 1637–1649.

Owen, M. J. & Morris, S. 1970 *Mod. Plast.* **47**, no. 4, 158.

Owen, M. J. & Morris, S. 1971 *International Conference on Carbon Fibres, their Composites and Applications*, paper 51, pp. 292–302. London: The Plastics Institute.

Owen, M. J. & Morris, S. 1972 *Plast. Polym.* **40**, 209–216.

Owen, M. J. & Rose, R. G. 1970 *Mod. Plast.* **47**, no. 11, 130–138.

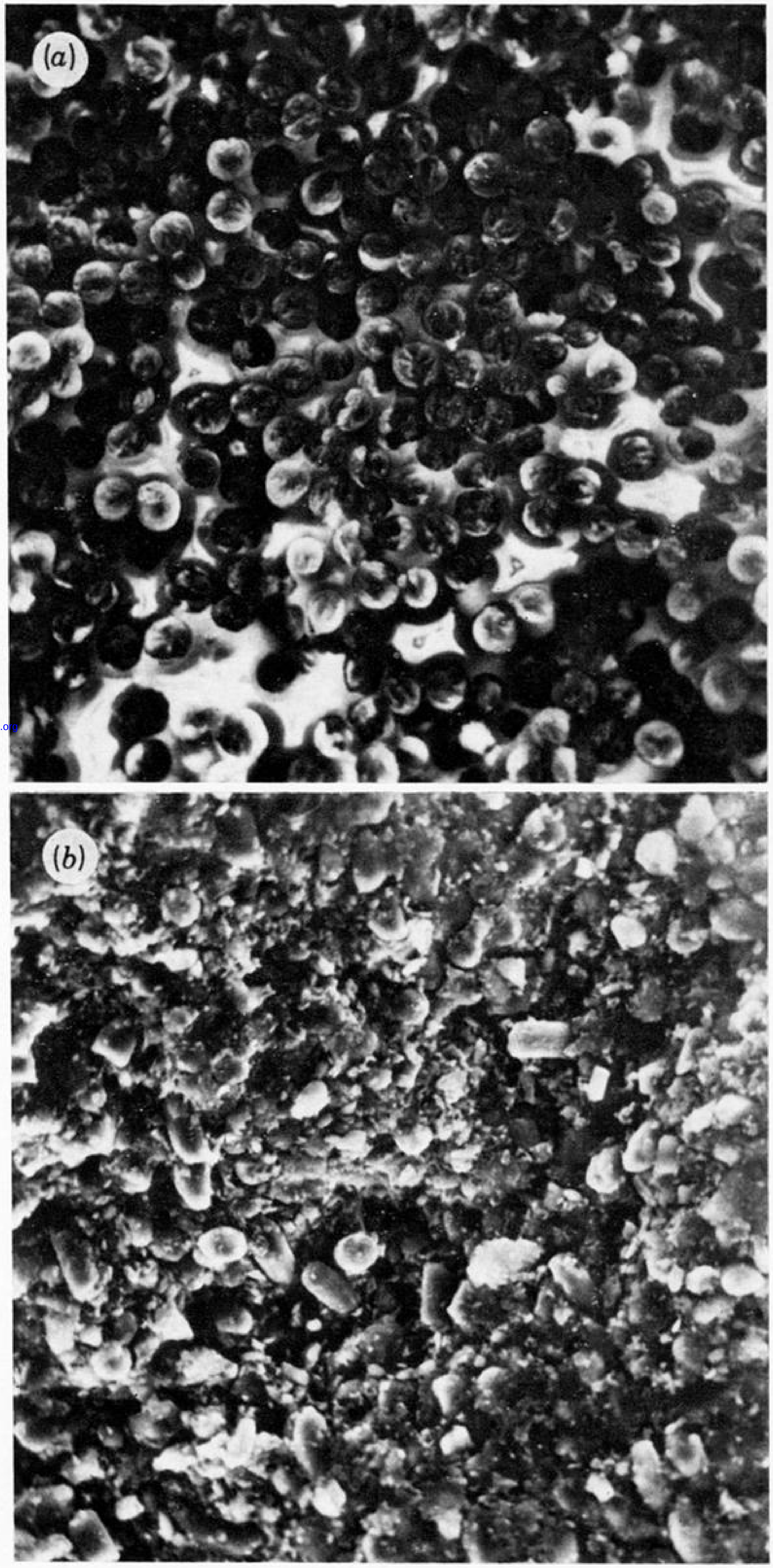
Phillips, D. C., Scott, J. M. & Buckley, N. 1978 Preprint 2*nd International Conference on Composite Materials*. Toronto, April 1978.

Schütz, D. & Gerharz 1977 *Composites* **8**, 245–250.

Smith, T. R. & Owen, M. J. 1968 6*th International RP Conference British Plastics Federation*, paper 27. London.

Sturgeon, J. B., McLaughlin, K. S. & Moore, B. B. 1976 *R.A.E. Technical Report* 76153, H.M.S.O.

Throckmorton, P. E., Hickman, H. M. & Browne, M. F. 1963 18*th Annual S.P.I. Conference*, section 14A. Chicago.



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