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Work of fracture of fibre-reinforced polymers

M. FILA, C. BREDIN, M. R. PIGGOTT
Department of Chemical Engineering and Applied Chemistry, University of Toronto, Toronto 5, Canada

The work of fracture has been measured by bending tests on notched specimens of graphite and glass fibre reinforced polyester resins. Fibre bundles were used to increase the effective fibre diameter and improve the uniformity of the fibre strength.

The results indicated that very tough specimens could be produced by these means (fracture surface energies of up to 11 kg/mm) and that toughness was determined by the strength, modulus and diameter of fibre bundles, as well as the volume fraction of fibre bundles. Failure occurred by fibre fracture close to the matrix fracture surface, and the fracture-surface energy appeared to result from the relative movement between fibre bundles and matrix as the fibres bridging the crack were stretched within the matrix. The work of fracture correlated well with the fibre-matrix interfacial stress, calculated from the observed stress transfer length.

1. Introduction

Since Cottrell's [1] suggestion in 1964 that long stress transfer length favours fracture toughness, a large number of theories of toughness have been discussed, but relatively little supporting data has been produced.

It seems generally agreed that poor fibre-matrix adhesion favours fracture toughness, but has the consequence of easy splitting of the composite parallel to the fibres, and loss of shear strength [2-4].

Elastic energy stored in the fibres has also been considered to be important, both with brittle [5] and ductile matrices [6]; in addition the work of fibre pull-out has been examined [7, 8]. A suggestion has been made for achieving moderate toughness without too much loss of shear strength [9], and it has been shown that ductile fibres can be used to produce tough composites [6, 10, 11].

This work describes an attempt to produce very tough composites using brittle fibres in a brittle matrix, by adopting the principles of a recent theory of toughness [12], and an attempt to verify the main predictions of the theory.

2. Experimental details

A fibre-reinforced polymer system was chosen because of the relative ease of fabrication of samples using polymer matrices. The matrix, chosen for its extreme brittleness, was a rigid copolymer of styrene and polyester, Reichold Polylite 31000.

Composite fibres of either glass or carbon were used for reinforcement since the theory indicated that for maximum toughness the fibres should be as strong and as thick as possible. Bundles of the fibres were impregnated with polymer, and strong fibrous rods with diameters in the region of 3/4 mm were obtained. The glass fibres used were Fiberglass of Canada type K891 glass rovings, bearing polyester compatible coupling agent in the size. The rovings contained 2050 individual fibres, each of about 13 μm diameter. The carbon fibres were Thornel 50 tow containing 1440 fibres, each of about 6.6 μm diameter. The properties of the fibres are given in Table I. A number of different grades of polyester were used for the impregnation, with the intention of varying the properties of the interfacial region between the fibrous rods and the rigid polyester matrix. The polymers used for impregnation were generally not so rigid as the matrix polymer, so that plastic flow of the impregnation polymer was possible at the surface of the fibre rods.

The fibre rovings were coated by dipping into a bath of the polymer. Excess was carefully removed, and the coating material was allowed to dry before curing at 80°C for 20 h. This produced a straight rod of material with about 70 vol. % of fibres.

The composites were produced by supporting the ends of 11 cm lengths of the above rods by threading them through the interstices of wire screen. This produced a regular array of parallel rods, the number of meshes per cm in the screen used determining the volume fraction of rods in the final composite. The pieces of screen used were about 2 cm × 2.5 cm and when loaded were placed in an oblong box 2 cm × 5 cm × 12.5 cm which had been sprayed with fluorocarbon release agent. The polylite 31000, containing 1% MEK peroxide initiator, was then poured in and allowed to cure 24 h under reduced pressure at room temperature. Curing was completed by heating to 40°C under argon for 3 h. Samples of the rigid polyester without reinforcement were also produced using the above procedure.

Notched bending tests, as described by Tattersall and Tappin [13] were used for determining the fracture toughness of the samples. This method was chosen because it provided a means of determining the work required to break the specimen completely, thus avoiding the difficulty of determining the position of the crack tip at any time (the difficulty arises because, with brittle matrices the matrix crack-front is normally in advance of fibre failure, so that fibres are left bridging the crack).

The specimens were prepared for the test by cutting off the ends, which contained the wire screens, and grinding off the excess matrix to obtain a rod of about 2 cm square section. Finally the section of the specimen was reduced to triangular shape at the centre by removal of about 0.5 mm thickness of material, using a 0.02 in. thick circular cutter in a milling machine. The fracture surface shown in Fig. 1 illustrates the triangular section at the centre, obtained by cutting away material from the top right-hand and top left-hand corners.

The samples were broken under three-point loading in an Instron tensile machine. The supports were 5 cm apart, and the crosshead speed was 0.05 cm/min.

The fibre bundles were also tested in the Instron. They were initially mounted on cardboard frames, using epoxy resin. When correctly aligned in the Instron the cardboard frame was cut so that the load could be applied to the fibres, and they were then pulled at a crosshead speed of 0.1 cm/min.

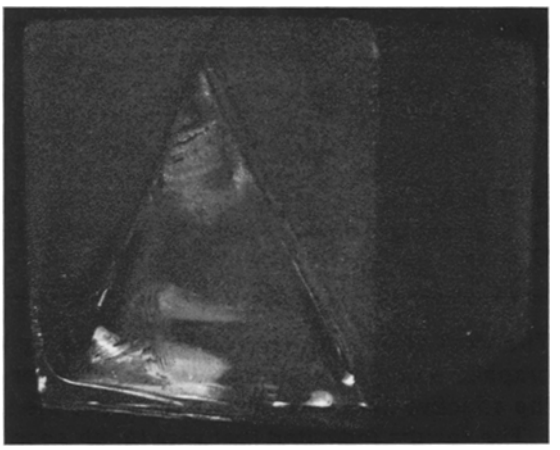


Figure 1 Typical fracture surface of unreinforced polyester. Block is 2 × 2 cm.

3. Experimental results

3.1. Fibre bundle strengths

The properties of the fibres and bundles are summarized in Table I. It was found to be difficult to avoid premature fracture of the impregnated carbon-fibre bundles in the tensile tests owing to their extreme brittleness and stiffness. In the case of the glass the strength of the individual fibres appeared to be increased as a result of their incorporation in impregnated bundles.

TABLE I Properties of fibres and bundles

Fibre type	Glass	Carbon
As-received fibre strength (σ_i, kg/mm^2)	154	230
As-received fibre modulus (E_i, kg/mm^2)	7000	44000
Coated bundle diameter (d, mm)	0.76	0.28
Volume fraction of fibres in bundle (V_b)	0.78	0.70
Calculated bundle modulus (V_bE_i)	5500	31000
Calculated bundle strength ($V_b\sigma_i$)	120	160
Measured bundle strength (σ_u, kg/mm^2)	135 ± 11	155 ± 5

3.2. Work of fracture of unreinforced polyester

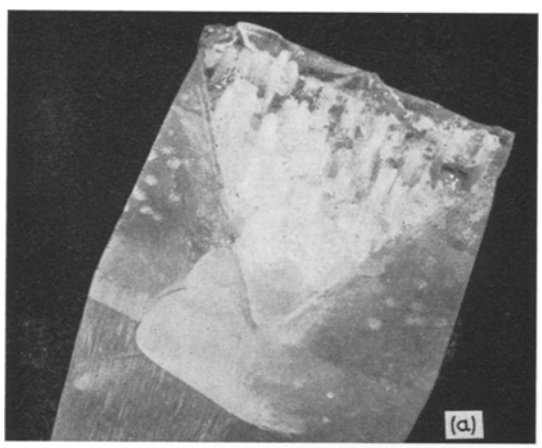
These samples behaved completely brittlely.

Once the specimens started to crack, the crack propagated completely through the reduced section at very high speed, much too fast for any work of fracture to be recorded. A typical fracture surface is illustrated in Fig. 1.

3.3. Work of fracture of glass-reinforced specimens

The crack proceeded quite quickly through the matrix with most of these specimens, leaving the fibres bridging it. The fibres finally failed close to the crack faces (Fig. 2), and the work of fracture was quite large, up to 11 kg/mm.

During the tests it was observed that the fibre bundles, which were almost invisible initially, developed a white appearance close to the crack faces (Fig. 3a). This can be seen more clearly in the case of a specimen having only one fibre bundle (Fig. 3b), and occurred on both sides of the crack to the same extent.



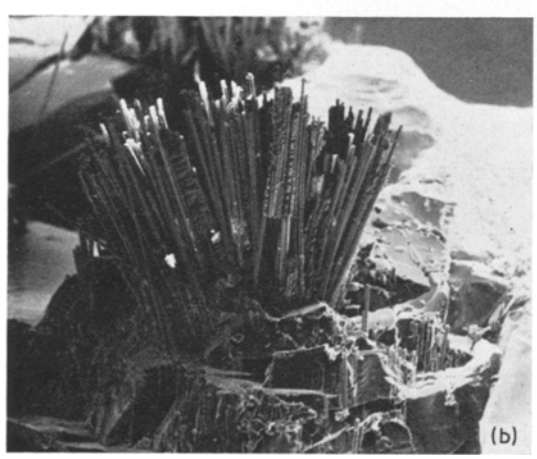
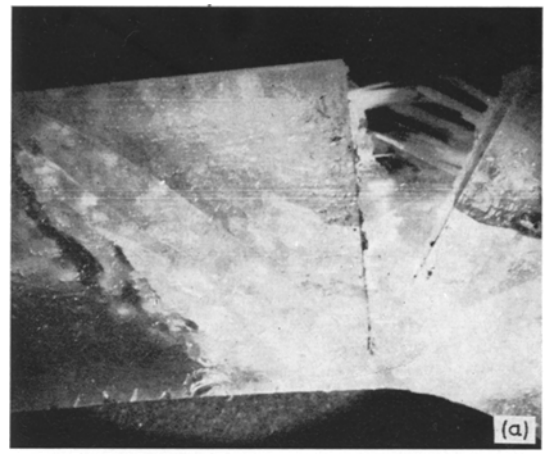


Figure 2 Fracture surface of glass-fibre bundle reinforced polyester (a) × 2.2, (b) × 80.



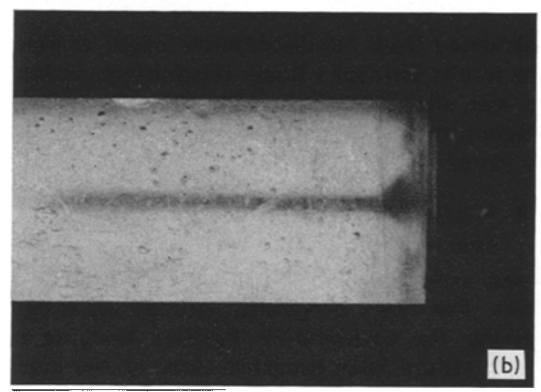


Figure 3 Opaque region produced when fibre-bundle failure occurs (a) for multiple-bundle reinforced specimen (b) single bundle.

The discolouration was taken to indicate that the fibre has been stressed, and this being so, the length of discoloured region should be related to the stress transfer length. Making the assumptions that the discoloured length is equal to the transfer length, that the fibre coating behaves as a rigid, perfectly plastic material, and that the bundle behaves as a single fibre, the shear flow stress of the fibre bundle-matrix interface region, τ_y, can be calculated from the formula $\tau_y = \sigma_u d/(4l)$ where σ_u is the fibre bundle strength, d the fibre diameter, and l the stress transfer length. (If the fibre bundle-matrix bond had failed, stress transfer could still take place by friction. In this case τ_y should be replaced by the frictional force, F.)

It was found that when the work of fracture, γ_f, was plotted against the reciprocal of τ_y

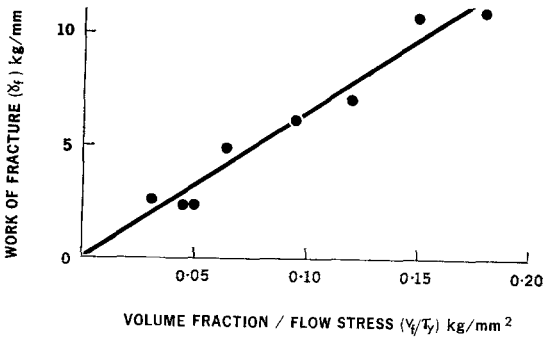


Figure 4 Variation of work of fracture with volume fraction/flow stress ratio for glass-fibre bundle reinforced polyester.

calculated from the discoloured length of fibre, the results satisfied a linear relationship, so long as due allowance was made for the volume fraction of fibre bundles, Fig. 4. The values for τ_y varied from 1.0 to 2.9 kg/mm^2.

3.4. Work of fracture of carbon-reinforced specimens

These samples behaved similarly to the previous ones. The crack proceeded through the matrix leaving the carbon-fibre bundles bridging it. Fibre failure later occurred close to the crack faces (Fig. 5), but the work of fracture was never greater than 0.41 kg/mm. The black appearance of the carbon-fibre bundles did not permit any observation of transfer length. However, the work of fracture was proportional to the fibre bundle volume fraction, and the different types of material used for fibre bundle impregnation and coating had no consistent effect, Fig. 6.

4. Discussion

The fact that fibres could be seen bridging the crack before failure, the observation of the discoloured regions of about equal length on either side of the crack in the case of the glass-fibre bundles, and the final failure of the bundles close to the crack faces, strongly suggests that the mechanism discussed by Piggott [12] is operating.

Piggott considered the case of a material reinforced by parallel fibres of uniform strength. A crack propagating through the matrix does not immediately break the fibres, but stress transfer between fibres and matrix close to the crack faces eventually does. The stress-transfer process causes the fibres to have their highest stresses

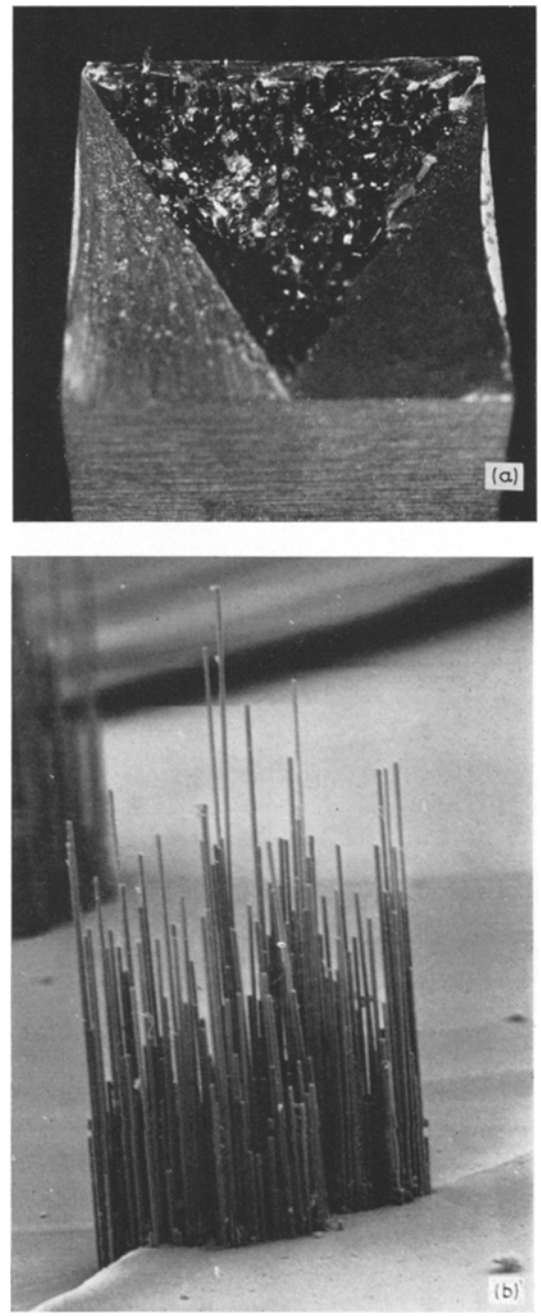


Figure 5 Fracture surface of carbon-fibre bundle reinforced polyester (a) × 2.6, (b) × 95.

where they bridge the crack. Thus, being of uniform strength, they should fail there.

Piggott's equation for the work of fracture

$$\gamma_f = \frac{V_f\, d\, \sigma_u^3}{12\tau_y\, E}$$

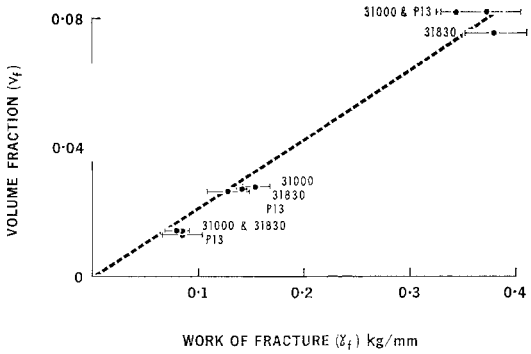


Figure 6 Effect of volume fraction of carbon-fibre bundles on work of fracture. The numbers refer to the type of resin used to coat the fibres, 31000 being the hardest and P13 the softest.

also indicates a linear relationship between γ_f and V_f/τ_y which appears to be substantiated for the case of the glass-fibre reinforced material in Fig. 4.

In addition the effect of fibre bundle modulus, E, on the fracture toughness can be shown to support the validity of the equation. The carbon-fibre bundle rods were very brittle, indicating that a good bond had been obtained between the fibres and the impregnating material. It is therefore quite likely that the slip between the carbon-fibre bundle rod and the matrix took place close to the interface between the bundle impregnating material, and the matrix. The stress at which this occurs should thus be similar to that in the case of glass. If we take the average value for τ_y for the glass reinforced samples, i.e., 1.7 kg/mm^2 and apply it to the carbon case, we can calculate the ratio

$$\frac{\gamma_f \tau_y E}{V_f d \sigma_u^3}$$

for carbon, and can also calculate this ratio for glass. The values are about 0.23 for carbon, and about 0.19 for glass. The agreement between these values does therefore support the inclusion of modulus in the equation for fracture toughness.

There is a discrepancy between theory and experiment however, in regard to the actual value of this ratio. Theory indicates a value of $\frac{1}{12}$ while the experiment yields a value between $\frac{1}{5}$ and $\frac{1}{6}$. The theory however was developed for single-fibre reinforcement and only takes into account the work in the matrix close to the interface between fibres and matrix, or in a fibre coating. Composite fibres may well suffer shear deformation in the impregnating material in the fibre-bundle transfer length, in addition to the work at the surface of the fibre bundle. This internal work could account for the discrepancy.

The works of fracture observed, having values up to 11 kg/mm, with only a volume fraction of the fibre bundles of 0.18 (which is a true fibre volume fraction of only 0.11) compare favourably with Harris *et al* [3] maximum value of about 3.8 kg/mm for a volume fraction of fibres of 0.4.

Although Harris *et al* found that their results agreed with Cottrell's [1] theory, so long as the observed pull-out length could be equated with flaw spacing in their fibres, it is useful to discuss their results in terms of Piggott's theory. The only unknown parameter in their work, τ_y (or frictional force, F), can be calculated from their experimental value of γ_f, since using Piggott's equation and Harris's data, $\gamma_f \tau_y$ comes to about 2.9×10^{-2} kg^2 mm^{-3}. On this basis, we would deduce that for the toughest materials $\tau_y \simeq 8.2 \times 10^{-3}$ kg/mm^2. Such a low value suggests that stress transfer must be occurring by friction, and that the bond between fibres and matrix fails easily. In this case the bond could well have insufficient strength to withstand the shear forces of the bend test used to measure interlaminar shear strength. For 0.4 volume fraction of fibres, only about $\frac{1}{4}$ to $\frac{1}{5}$ of any section parallel to the fibres consists of matrix (assuming hexagonal packing of fibres). Thus for complete fibre-matrix bond failure the interlaminar shear strength should be about $\frac{1}{4}$ to $\frac{1}{5}$ of that for no bond failure. Comparing the results for this yield about the right factor, indicating that their results were consistent with Piggott's theory.

The work of Harris *et al* clearly demonstrates the draw-back of using poor bonding as a means of getting great toughness. The work reported here suggests that there is an alternative, i.e. to use fibre bundles, controlling toughness by using appropriate materials to impregnate the bundles. This principle is applicable to all types of matrix; metal, ceramic or polymer, etc.

5. Conclusions

Fibre bundles can operate as reinforcing rods of great strength with little variation of strength along their lengths. They can be used to toughen very brittle materials to a high degree, and their effectiveness appears to depend largely on their strength, but depends also on their diameter,

which should be large, and modulus, which should be small. Fracture toughness may be controlled by impregnating the fibre bundles, with materials having suitable shear flow properties, before embedding them in a matrix.

Acknowledgement

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