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The fracture energy of a glass fibre composite

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The fracture energy of a glass fibre-polyester composite has been measured by work of fracture (γ_f) measurements on bending beams, and by linear elastic fracture mechanics analyses (γ_i) of the bending beams and edge-notched tensile plates. It was found that for the bend specimens $\gamma_i < \gamma_f$. The work of fracture, γ_f, displayed a strain rate dependence, but there was no such dependence of γ_i. It is postulated that γ_i is determined by a debonding mechanism while γ_f is the sum of a debonding mechanism plus a pull-out contribution. The edge-notched tensile plate experiments showed that γ_i obtained from thick plates was less than that obtained from side-grooved plates, and that in each case there was a dependence of γ_i on crack size.

1. Introduction

The fracture surface energy (γ) of a material may be measured by several methods which are totally different in principle. Two such methods are work of fracture and linear elastic fracture mechanics techniques. The work of fracture [1] measurement requires a specimen to be broken in a completely controlled manner so that all the stored elastic energy in the specimen-machine system is absorbed in the creation of fracture surface. The fracture surface energy obtained in this way is given by the integrated area under the load-deflection curve divided by twice the specimen cross-section and is termed the work of fracture (γ_f). Linear elastic fracture mechanics techniques [2] are based on the Griffith criterion for unstable crack propagation. Fracture occurs by rapid crack propagation under decreasing loads when the release of stored elastic energy during failure is greater than the energy required to create fracture surface. The fracture surface energy in that case (γ_i) is obtained from the maximum load at failure (P) through analytical solutions which relate γ_i to P, initial crack length, specimen geometry and material elastic constants ($\gamma_i \equiv G_{Ic}/2$ in LEFM terminology).

A comparison of the γ values obtained by these methods is useful for determining whether LEFM is applicable to the failure of the material, and if it is, for providing information about different stages of crack propagation [3]. This note describes the results of such measurements on a composite of randomly oriented glass fibres in a polyester matrix.

2. Experimental details

The material studied was a random glass fibre composite, consisting of 15 vol % of 10 μm diameter 'E'-glass fibres in a polyester matrix. It was made in the form of plates, 3.5 mm thick, and was isotropic in the plane of the plate. The fibres were sufficiently long to be regarded as continuous. Considerable scatter was obtained in the fracture energy measurements, and this is a reflection of the local inhomogeneity in fibre distribution displayed by this type of material.

All the mechanical testing was carried out on an Instron machine. Works of fracture were determined by fracturing, in three-point bending, notched bars as shown in fig. 1. A typical load-deflection curve is also shown and it can be seen that fracture was almost totally controlled. After a small amount of rapid crack propagation the crack speed was determined by the rate of movement of the Instron cross-head. Fig. 2 shows the LEFM specimens, which were doubled edge-notched and fractured in tension. Two types were initially employed, constant thickness (3.5 mm) and side-grooved (1.5 mm), to determine the effect of specimen thickness. The results obtained with the side-grooved specimens differed quite substantially from those obtained

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with the constant-thickness tensile-specimens and 3.5 mm thick-bend specimens. Therefore, in order to determine whether side-grooving was responsible for this effect, some measurements were also carried out on tensile specimens which were thinned homogeneously to 1.5 mm and also on homogeneously thinned, 1.5 mm bending beams. The notches in all the specimens were cut with a jeweller's saw and received no additional sharpening.

3. Experimental results

The maximum loads required to break both the work of fracture specimens and the tensile

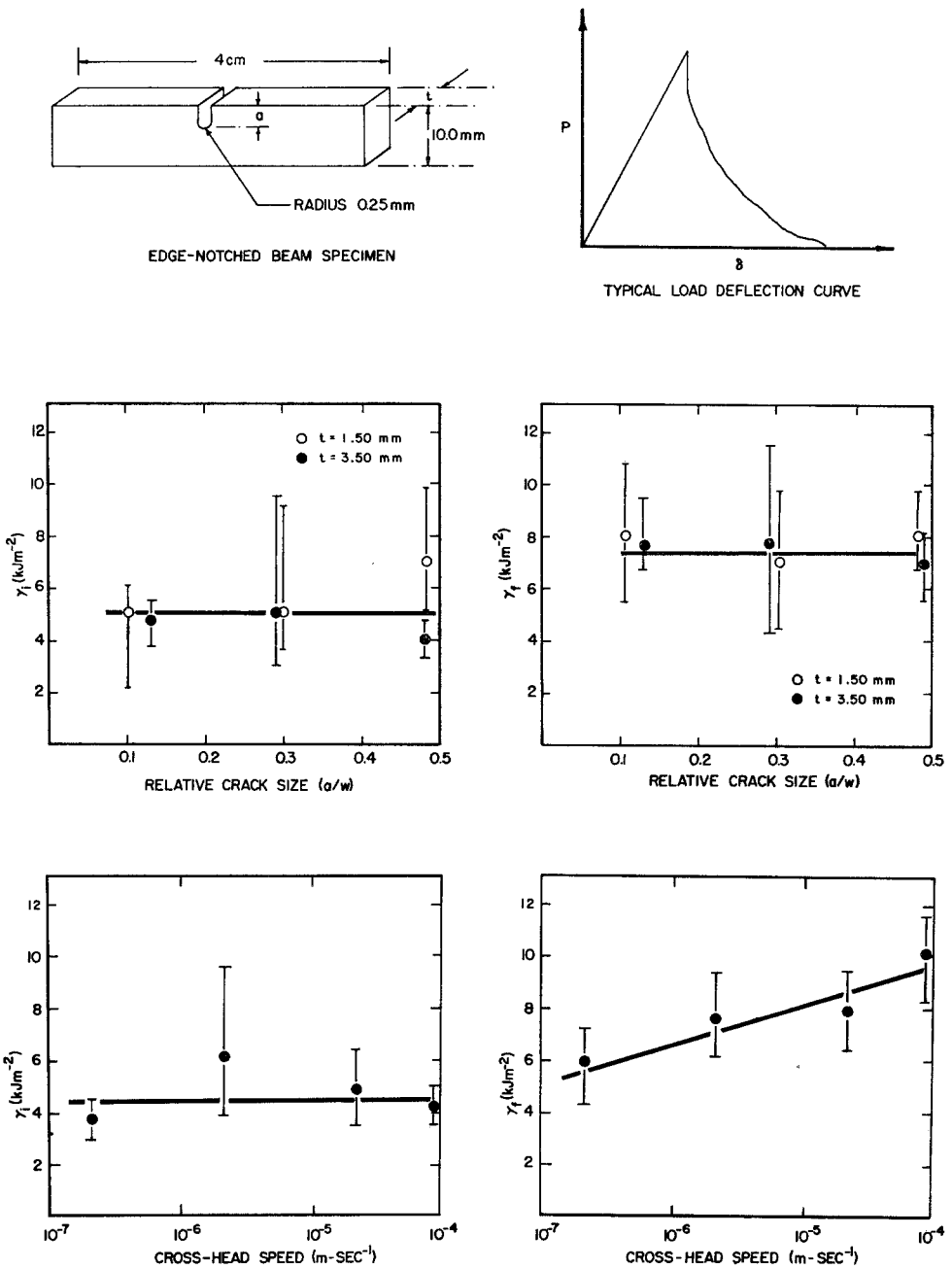

Figure 1 Bend specimens and data.

specimens were used to calculate the critical stress intensity factors (K_{Ic}) using the analytical formulae given by Srawley and Brown [4]. The fracture energies were calculated from the critical stress intensity factors by the standard plane strain relationship for isotropic materials,

$$2\gamma_i = G_{Ic} = \frac{(1 - \nu^2)}{E} K^2_{Ic} \qquad (1)$$

using a value of Young's modulus (E) of 6.3 GNm^{-2} which was determined experimentally, and assuming a value of ν of 0.25. Fig. 1 shows the γ_i and γ_f values obtained as a function of strain rate, crack size and specimen thickness from the bending experiments. Fig. 2 shows the γ_i values obtained as a function of crack size and specimen thickness from the tensile tests. There was no variation of γ_i with strain rate in the latter.

The results of the bend tests show that the mean γ_f and γ_i values are independent of crack size and specimen thickness, with the γ_f values at 7.5 kJm^{-2} being greater than the γ_i values at 4.5 kJm^{-2}. In contrast to the crack size independence of the bend-test values, the tensile tests show a clear variation of γ_i with crack size. The thick tensile plates display values varying from a minimum of 1.3 kJm^{-2} to a maximum of 4.3 kJm^{-2}, while the side-grooved tensile plate values vary from 1.5 kJm^{-2} to 9 kJm^{-2}. The thin tensile plate data, at a relative crack size of 0.3, span the range between the thick and side-grooved plates. The microstructures of the fracture surfaces of the thick, thin and side-grooved plates were examined optically, and the fibre contents determined quantitatively by burn-off measurements. There were no differences in fibre volume fraction or distribution between the fracture surfaces. All the tests showed that γ_i was independent of strain rate, but γ_f increased with increasing crack propagation speed.

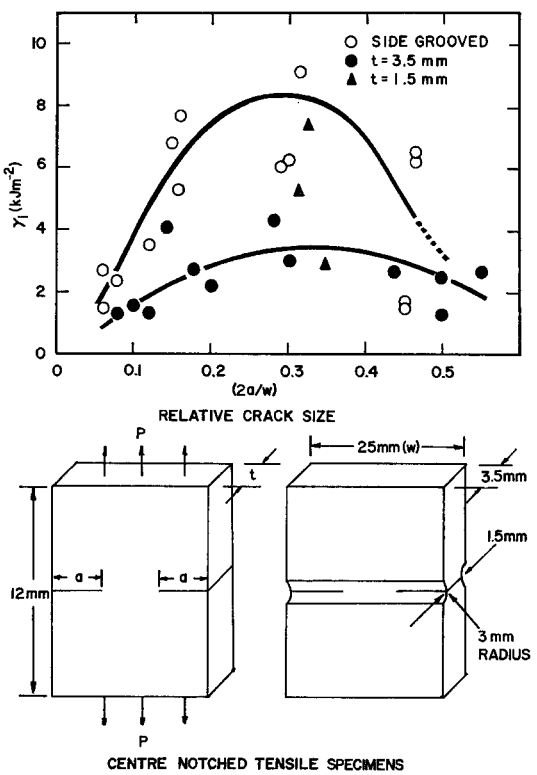

Figure 2 Tensile specimens and data.

During the loading of the double edge-notched tensile specimens, it was observed that a feature, somewhat similar in appearance to a plastic zone in a metal, developed prior to fracture. Before the material was loaded, it was translucent. During loading of the tensile specimens the material at the crack tip became increasingly more opaque at loads greater than about a half the ultimate strength. The diameter of this opaque zone at the crack tip was approximately 1 mm and increased in size only slightly as the load was increased. At the point of unstable fracture, the opaque zone suddenly extended into the specimen to a depth of about 2 mm on either side of the fracture surfaces. These features are illustrated in fig. 3. The fracture surfaces displayed protruding, broken fibres which extended in length up to the depth of the opaque zone.

4. Discussion

γ_i is obtained from the initial stage of crack propagation, under conditions where the unstable crack is moving rapidly and independently of the applied strain rate. γ_f is the fracture surface energy averaged over all stages of crack propagation during fracture, and obtained in these experiments at crack propagation rates approximately equal to the loading rate of the testing machine. The bend specimen data show that γ_i is less than γ_f. This cannot be explained merely in terms of different crack propagation speeds as γ_i is determined by rapid crack motion while γ_f increases with increasing crack velocity so as to increase the divergence over the values presented here. This suggests that they are controlled by two different processes.

Two theories which have been postulated to explain the fracture surface energy values of

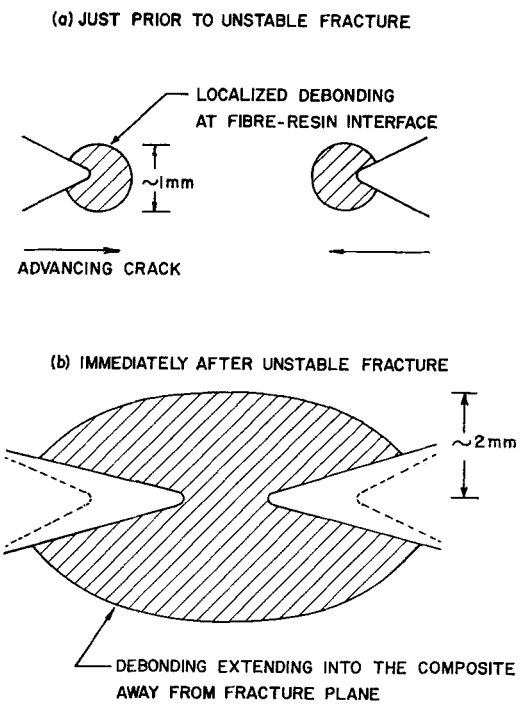

Figure 3 Schematic representation of debonding during loading of tensile specimens.

fibre composites are debonding and fibre pull-out. Outwater and Murphy [5] have observed that, during the fracture of glass fibre-resin composites, the adhesive bond between fibres and matrix is destroyed for some distance on either side of the fracture surface. Where this occurs, the stored elastic energy in the debonded length of fibre immediately before its fracture cannot be redistributed in the composite when the fibre snaps. If it is assumed that the fibre-matrix interfacial shear stresses fall to zero on debonding, the following expression for the fracture surface energy due to this process may be derived,

$$\gamma_{\text{Debond}} = \frac{V_f \sigma_f^2 Y}{4E_f} \tag{2}$$

where V_f is the volume fraction of fibres, σ_f is their strength, E_f their Young's modulus and Y is the debonded length. Alternatively, Cottrell [6] and Kelly [7] have suggested a model in which the interfacial shear stress is maintained during fracture and fibre pull-out. The fracture surface energy in that case arises from the work done in extracting fibres against restraining shear stresses. Following Cottrell and Kelly it can easily be shown that the work done in pulling out a fibre of embedded length l_p against a constant restraining shear stress, τ is

$$W(l_p) = \pi r \tau \, l_p^2 \tag{3}$$

and the fracture energy of a composite due to pull-out is

$$\gamma_p = \Sigma_p N_p W(l_p) \tag{3a}$$

where N_p is the number of fibres per unit area of pull-out length l_p. This expression is a more general case of the equation given by Cottrell for a discontinuous fibre composite in which all the fibres are of length equal to the critical transfer length l_c.

$$\gamma_{\text{pull-out}} = \frac{V_f \sigma_f l_c}{24} \tag{3b}$$

The results presented here can be explained in terms of these models if it is assumed that on "debonding", the fibre-matrix interfacial shear strength decreases not to zero but to some much smaller finite value, τ'. The result on the debonding model of this assumption is to make equation 2 an overestimate of γ.

It is postulated that the initial crack motion is controlled by the debonding mechanism, with little or no contribution to γ_i from fibre pull-out. The opaque zone which develops during loading a tensile specimen is a region of debonded fibres and may be used to estimate the debonded length Y as approximately 4 mm. The strength of the fibres in the composite is quoted by the manufacturer to be approximately 2100 MNm^{-2}, and their Young's modulus is 72 GNm^{-2}. Equation 2 strictly applies to an aligned fibre composite. In a random fibre composite, not all the fibres will contribute to strengthening or toughening in a given direction. If it is assumed that approximately one-half of the fibres contribute to debonding, then inserting the above values into equation 2 results in a theoretical value of γ_{Debond} of 4.6 kJm^{-2}. This is remarkably close to the observed bend specimen γ_i values and, in view of the assumptions and approximations involved, the good agreement must be regarded as fortuitous. It is worth noting that equation 3b can be used to give a very approximate value for $\gamma_{\text{pull-out}}$ if it is assumed that the mean fibre pull-out lengths are approximately equal to $l_c/4$ (l_c then becomes a statistical concept). Equation 3b predicts a value of 250 kJm^{-2} which is very much greater than experimentally observed. As the crack propagates, fibres begin to pull out behind the crack tip as the fracture faces separate. Provided there is some interfacial shear stress τ'

between fibres and matrix, this will lead to another energy absorption term to add to γ_i, resulting in γ_f being increased above γ_i. In general this interfacial shear stress will be strain-rate dependent, because of the visco-elastic nature of the matrix, and will increase with strain rate. From equation 3 this will cause $\gamma_{\text{pull-out}}$ to increase with strain rate. A strain rate dependence of γ_f due to $\gamma_{\text{pull-out}}$ has been observed in other fibre composites also [8].

The results obtained from the bend tests are self-consistent, displaying no variation of γ_i or γ_f with crack size or specimen thickness. The results obtained from the tensile tests are different. Although the mean γ_i values are approximately the same as those obtained from the bend tests there is a variation with crack size and specimen thickness. The scatter in this data is quite large but there is clearly a maximum in the values at a relative crack size of approximately 0.3. Further, the side-grooved specimen values are higher than the constant thickness specimen data. It is not clear why there should be this variation of γ_i with crack length and specimen thickness in the tensile plates especially as there is apparently no such variation in the bend specimens. Similar variations of γ_i with plate thickness are observed in metals and are there due to differences in plastic zone size depending on whether plane stress (thin plate) or plane strain (thick plate) conditions exist. In metals there is also sometimes a variation of γ_i with crack-length because where the crack-tip stress field interacts with free surfaces, or with other crack stress fields, the plastic zone size is decreased over that of an isolated crack, resulting in a decreased γ_i. It is possible that similar effects might be responsible for variations in the debonded zone size, or affect the decrease in interfacial shear stress on "debonding". However, it is difficult to reconcile these ideas with the invariability of the bend test data. Clearly more work remains to be done to clarify this. However, taking the results as a whole, they are sufficiently promising to suggest that LEFM may be used to predict and explain the fracture of this type of material with the provision that care must be taken in the selection of the geometry of test specimens.

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