Crack Propagation in Carbon Fibre Composites

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1. Introduction

The fracture surface energies ($\gamma = \frac{G_{IC}}{2}$) of carbon fibre composites, measured by the work of fracture technique, can be quite high. For example, values of between 3 kJm^{-2} and 30 kJm^{-2} are commonly obtained. These values are frequently used to calculate, through the Griffith-Irwin failure criterion, that opening mode fracture will occur by the unstable propagation of cracks greater than $\sim$0.5mm in size. This interpretation need not necessarily be correct for several possible reasons.

Firstly, the Griffith-Irwin criterion may not be applicable to the composite. In a fibre composite fracture does not always occur by propagation of an existing crack because the stress intensifying effect of the crack may be removed by shear at the crack tip. However, even if crack propagation does occur, it has been suggested that on theoretical grounds the Griffith-Irwin criterion will not be the controlling mechanism but that a statistical variation of strength approach may be more suitable. Secondly, if the Griffith-Irwin approach is applicable, the fracture surface energy obtained from work of fracture measurements (γ_F) may not be the appropriate value for fracture initiation.(1) The work of fracture technique is popular because of its experimental simplicity. However, it measures the total amount of energy required to separate the two halves of a broken specimen during bending. When a composite exhibits large amounts of fibre pull-out, it is arguable that much of the work done during the work of fracture experiment occurs after the propagation of the crack and that fracture initiation, i.e. the immediate decrease in load, is determined by a smaller energy (γ_i). Alternatively, during bending, the broken fibres spanning the crack are subjected to bending stress concentrations which might cause them to break into smaller pull-out lengths than in a tensile test, and thus bending experiments might underestimate the fracture surface energy appropriate to tensile tests. Thirdly, the work of fracture test method relies on fracture occurring in a controlled manner. In ceramics it is necessary for crack propagation to be totally controlled during the measurement - any period of unstable crack propagation results in incorrectly high γ values even if the fracture is ultimately controlled(2). During the fracture of conventional work of fracture specimens of carbon fibre composites, there is usually a period of unstable crack propagation and this could result in an overestimate of the fracture surface energy. This paper will consider some of these problems for crack propagation perpendicular to fibres in carbon fibre composites.

2. Preliminary Results

The following measurements have been carried out on balanced $0^o/90^o$ cross-laminate plates of $60^v/o$ of surface treated and untreated type I (high modulus) and surface treated type II (high strength) carbon fibres in epoxy resins: centre-cracked tensile (CN), tapered cantilever (TC) and Δ-section work of fracture (WF) (fig. 1). Cracks were introduced with a jeweller's saw and sharpened with a scalpel. Mean γ values for the surface treated materials are shown in Table 1. These were calculated from the CN specimens using a linear elastic fracture mechanics analysis, and from the TC specimens from the experimental variation of compliance with crack length(3). The untreated fibre composites were notch-insensitive, and arm failure prevented use

of the TC technique. The surface treated materials were notch-sensitive but the γ values varied slightly with crack size in the CN tests as shown in fig. 2.

Bend tests have also been carried out on similar material containing $60^v/o$ of aligned fibre using rectangular section flexural specimens and Δ-section WF specimens. Table 2 and fig. 3 show typical load-deflection curves and calculated 'works of fracture' obtained from two strain rates and Charpy impact.

WF measurements have been carried out on specimens of a different composite material, carbon fibre reinforced glass (CRG). With CRG it is possible to get a totally controlled fracture during bending, using a circumferentially notched bar. Partially controlled failures may also be obtained using single edge-notched bend specimens, the degree of control increasing with pre-crack size. Fig. 4 shows the variation in γ_F with pre-crack size and the variation in crack control during the test. Measurements of micro-structural features in the fracture surface have shown that the high values at low crack sizes are not due to fracture processes within the material but are anomalously high due to energy losses in the testing system.

3. Discussion

The CN and TC measurements on type I-epoxy matrix cross-laminate material yield γ values which agree well with each other. The CN calculation depended on an analysis based on the Griffith-Irwin criterion, while the TC calculation did not. Thus the agreement between the values is strong support for the Griffith-Irwin criterion being the controlling factor where crack propagation occurs perpendicular to the fibres in this type of material. Further, since the failure load during crack propagation in a TC experiment is determined by γ,(3) the shape of curves, similar to those in fig. 1, obtained during TC experiments shows (a) that there is no difference between initiation and propagation energies within experimental uncertainty, and (b) that γ_i of an 'as-cut crack' is the same as that of a 'natural' crack

γ_F of the type I-epoxy cross-laminate is a factor of three greater than the CN and TC values of γ, while those of the tougher type II-epoxy material agree well. This could mean either that for the type I-epoxy $\gamma_i < \gamma_F$ or that the work of fracture value is inaccurate as a material parameter. The CRG measurements showed that anomalously high values of γ_f were obtained when there was a period of unstable crack propagation, even when fracture was ultimately controlled. This can be used to explain the agreement between the γ_F values obtained from surface treated WF and flexural specimens in Table 2. A priori it would be expected that the γ_F values calculated from the flexural specimens would be greater than those obtained from the WF specimens because the former gave totally uncontrolled fractures, while the latter gave a stepwise controlled fracture. The agreement between the two implies that both overestimate γ. These preliminary results therefore suggest that in view of the agreement between γ_i and γ_F of the tougher type II material, the discrepancy between the values for the less tough type I is not so much due to different microfracture processes in the different stages of fracture but rather to the invalidity of the WF test for the more brittle material.

References

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(3) S. Mostavoy, P. B. Crosley and E. J. Ripling, J. Materials, 2 (1967) 661.

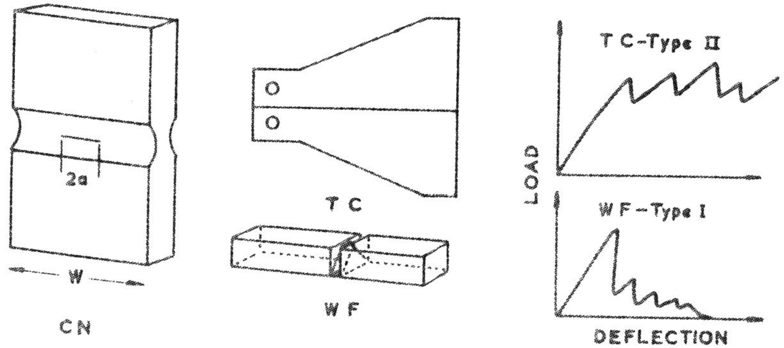

FIG. 1. SPECIMENS AND TYPICAL FAILURE CURVES

FIBRE	CN	TC	WF
Type I	0.22	0.17	0.77
Type II	—	13.0	14.8

TABLE .1. γ (kJm^{-2}) TREATED FIBRES CROSS-LAMINATES

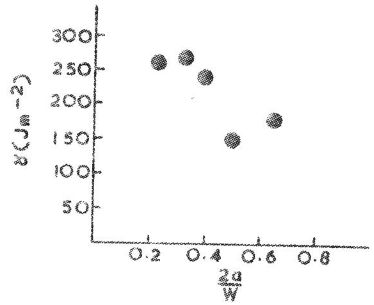

FIG. 2. CN DATA (Type I)

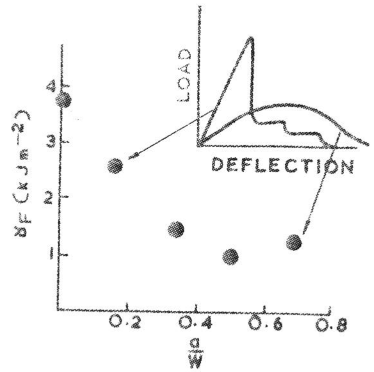

FIG. 4. CRG DATA (MEAN)

MATERIAL	SPECIMEN	LOW SPEED	HIGH SPEED	CHARPY
Type I	Δ	6.5	6.9	7.7
	F	6.4	8.0	7.3
Type II	Δ	25.0	30.5	34.4
	F	23.5	29.0	33.1

TABLE .2. MEAN γ_F VALUES (kJm^{-2}) ALIGNED FIBRES.

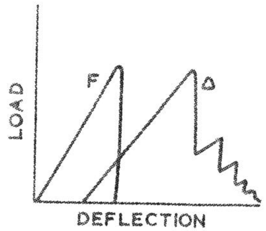

FIG. 3.