JOURNAL OF MATERIALS SCIENCE 7 (1972) 289-294

The Fracture Energy and Acoustic Emission of a Boron-Epoxy Composite

J. FITZ-RANDOLPH, D. C. PHILLIPS*, P. W. R. BEAUMONT, A. S. TETELMAN
Materials Department, School of Engineering and Applied Science, University of California, Los Angeles, California, USA

The fracture surface energy (γ) of a boron fibre-epoxy resin composite has been measured by three different techniques: work of fracture, linear elastic fracture mechanics, and compliance variation. Significant differences were obtained by the different methods. The compliance data were analysed to give γ at different stages of crack propagation. It was observed that γ decreased as the crack entered the material and that this variation of γ could be correlated with the pull-out length of fibres and acoustic emission generated during fracture. The fracture surface energy is explained in terms of a debonding model.

1. Introduction

The ease with which a crack can propagate in a solid can be specified by means of the fracture surfaced energy (γ). This is defined as the minimum energy required to create unit area of fracture surface. It varies from 2×10^{-2} in. lb/in.^{2} for a brittle material such as glass, through about 10 in. lb/in.^{2} for woods, to values in excess of 10^3 in. lb/in.^{2} for some tough metals. As well as being a comparative material parameter, γ can sometimes be used as a design parameter to predict the strength of a component in the presence of pre-existing flaws through use of the principles of linear elastic fracture mechanics† [1]. Two important questions in composite technology at the present time are: how do composite materials compare in toughness with other materials–and can toughness be used with linear elastic fracture mechanics to predict the strength of structural components?

Fracture surface energy may be measured in several ways which are different in principle [2]. The work of fracture technique measures the average energy per unit area (γ_F) absorbed during the total separation of the two halves of a fracture specimen [3]. The change of specimen compliance technique can give an energy value for smaller incremental stages of crack growth than the work of fracture technique; we call this value γ_C [4]. Finally, linear elastic fracture mechanics, where it is applicable, gives a value which is obtained from the maximum load borne by a cracked specimen and thus is determined by the initial stages of crack growth; we call this γ_I [1]. When a material undergoes a process such as plastic deformation, phase transformation or fracture, stored energy may be released in the form of acoustic emission. This emission is usually a transient phenomenon, energy being emitted in discrete pulses which are characteristic of the microstructural process. In this note we shall describe the results of the three different types of γ measurement on an aligned fibre boron-epoxy composite during fracture on a plane perpendicular to the fibres. We shall describe an observed relationship between fracture energy and acoustic emission, and we shall also comment on the possible physical processes which gave rise to the values of γ obtained.

2. Experimental Procedure

The boron-epoxy composite was manufactured by the United Aircraft Corporation and consisted of 66 vol % of aligned fibres in Union Carbide ERLA 2256-0820 epoxy resin, cured at 250° F (120° C). The fibres had a mean diameter of 0.004 in., a mean fracture stress of 450 ksi and a Young's modulus of 55×10^6 psi.

All mechanical testing was carried out in

*Present address: Materials Development Division, AERE, Harwell, Berks, UK
†γ is equivalent to $Gc/2$ in LEFM terminology.

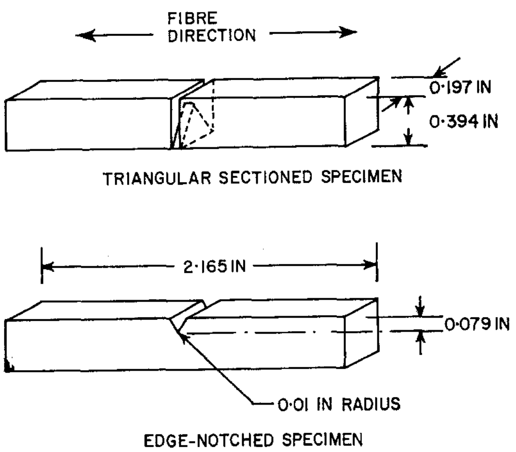

Figure 1 Edge-notched and triangular section specimens.

three-point bending on an Instron machine at a cross-head speed of 0.05 in. min^{-1}. Two types of specimen were used, single edge-notched and triangular section, fig. 1. In each case the fibres were initially parallel to the long axis of the bending beam.

Acoustic emission during fracture was monitored by means of Dunegan Research Corporation apparatus. This uses a piezoelectric transducer to detect the emission from a deforming specimen and convert stress wave pulses to electric pulses. The electric pulse is an analogue of the acoustic pulse, as the transducer operates at its own resonant frequency. Acoustic emission above the threshold of the transducer can be amplified and recorded either as a count rate or, by means of a totaliser, as a sum of counts. Count rate and sum of counts recordings produce a pattern of energy release characteristic of the observed microstructural process. In the present work, the acoustic emission was detected with a 150 kHz transducer, $\frac{3}{4}$ in. in diameter, which was attached to the sample with a silicon gum and then taped in place.

3. Experimental Results

A typical load versus deflection curve of an edge-notched specimen is shown in fig. 2 together with its acoustic emission recording. The fracture surfaces were characterised by short fibre pull-out lengths of the order of a few fibre diameters, and the maximum load at failure had an average value of 1220 lb $\pm$ 25 lb. Sih, Paris, and Irwin [5] have shown that the stress intensity factors calculated from isotropic elasticity theory are applicable to elastically anisotropic materials fracturing in the opening mode in a principle symmetry direction. Accordingly the K_c values have been calculated from the analysis due to Gross and Srawley, given by Brown and Srawley [6]. In anisotropically elastic media, K_c is related to the critical strain energy release rate G_c by

$$G_c = K_c^2 \left(\frac{a_{11}\, a_{22}}{2}\right)^{1/2} \left[\left(\frac{a_{22}}{a_{11}}\right)^{1/2} + \frac{2a_{12} + a_{66}}{2a_{11}}\right]^{1/2}$$

where the a_{ij} are components of the elastic compliance matrix. The a_{ij} were obtained theoretically from formulae and data given by Tsai [7], and this analysis led to values of $\gamma_I (= G_c/2)$ of 190 in. lb/in.^{2}

Fig. 3 shows, schematically, the load versus deflection curve and acoustic emission recording of a triangular section specimen. A step-wise series of load drops and reloadings was observed as the crack grew intermittently, and larger load drops were observed as the crack proceeded into

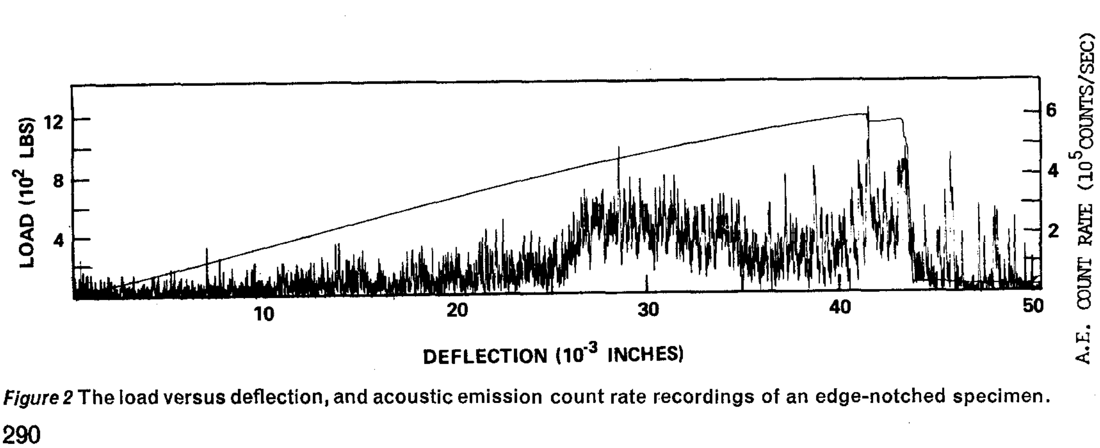

Figure 2 The load versus deflection, and acoustic emission count rate recordings of an edge-notched specimen.

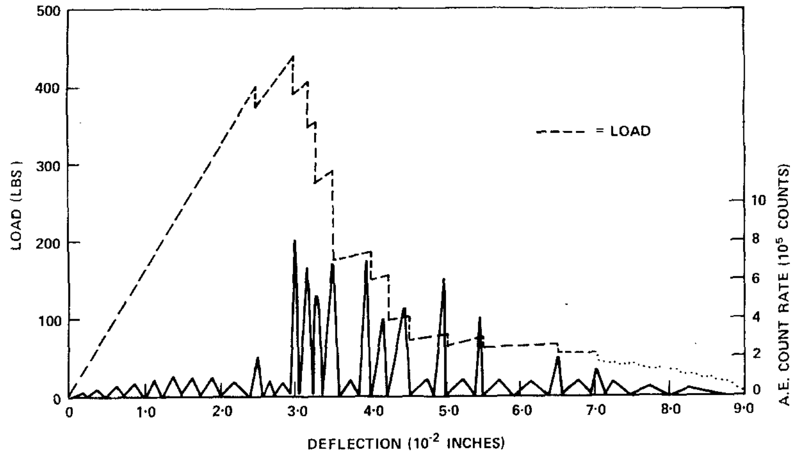

Figure 3 A schematic representation of the load versus deflection and acoustic emission count rate recordings of a triangular section specimen.

the increasing cross-section. A large acoustic emission burst was seen at every load drop and a smaller amount of emission was observed during reloading. Thus, only the large bursts were associated with crack growth. The fracture surface was confined to the notched area and no delamination occurred. Fibre pull-out lengths varied across the fracture face, being largest near the apex of the triangle where fracture commenced and decreasing towards the base of the triangle. Fig. 4 shows the variation of mean pull-out length with increasing crack depth.

The works of fracture (γ_F) were calculated by integration of the quasi-controlled load versus deflection curves of the triangular section specimens and ranged from 75 to 125 in. lb/in.^{2} with an average value of 100 in. lb/in.^{2}

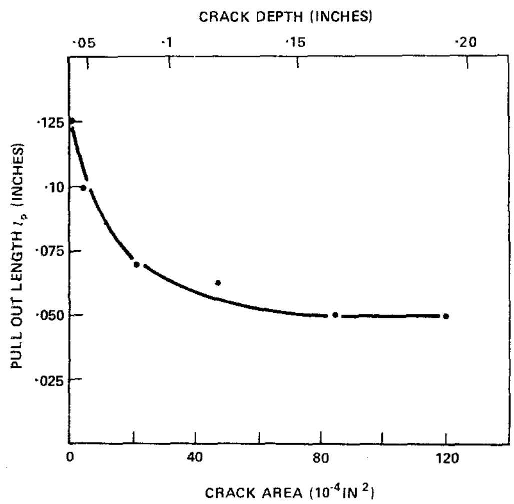

Figure 4 The variation of fibre pull-out length with crack area for triangular section specimens.

A compliance analysis [4] of the release of stored energy in an isotropic elastic plate during crack extension gives the strain energy release rate γ_c as

$$\gamma_c = \tfrac{1}{4} P^2 \frac{\partial c}{\partial a}$$

where P is the maximum load obtained during loading, c is the compliance of the system, and $\partial c/\partial a$ is the slope of compliance versus crack area curve. A compliance calibration for triangular section specimens was made by cutting the triangular area at successive 0.01 in. depth intervals and loading the specimen until the load versus deflection curve became linear. The compliance was then measured from the slope of the linear portion of the load versus deflection curve and is plotted as a function of crack area in fig. 5. A third order polynomial was fitted to the compliance versus crack area data using a least squares analysis and $\partial c/\partial a$ was calculated by differentiation of this polynomial. Using this calibration, it was possible to measure γ_c at various stages of fracture of the triangular section specimens. This was carried out by measuring the compliance of a breaking specimen at successive load drops by drawing a line from the origin through the point where reloading ended, and taking the inverse of the slope of that line, fig. 6. The crack area and $\partial c/\partial a$ were then obtained. Fig. 7 shows that the strain energy release rate values γ_c calculated by this compliance technique decreased from 450 to

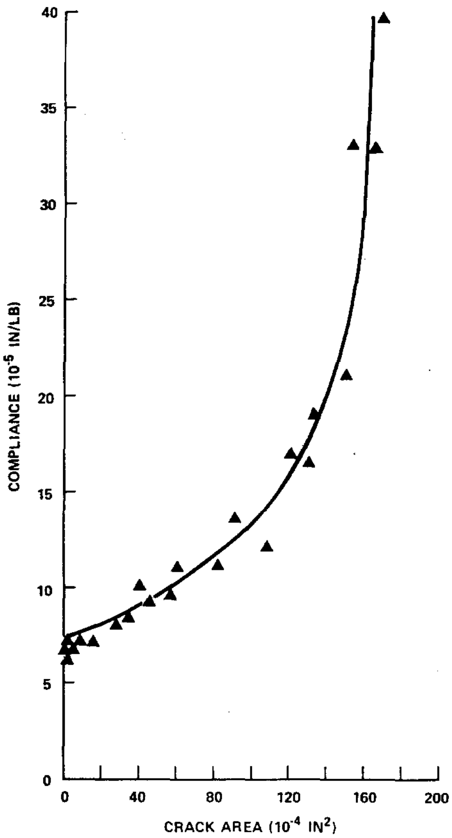

Figure 5 Compliance calibration curve.

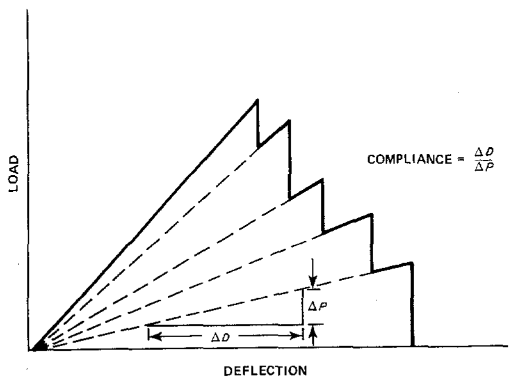

Figure 6 Schematic representation of the compliance calculation from the load-deflection curve of a triangular section specimen.

100 in. lb/in.^{2} as the crack proceeded into the specimen.

To determine the effect of the triangular section geometry on γ_c values, the same compliance technique was also used to calculate the

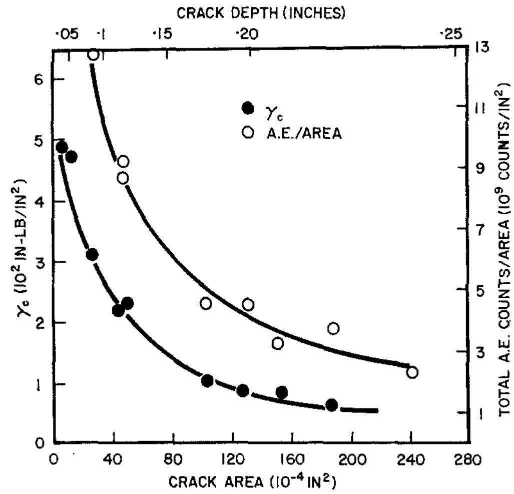

Figure 7 Variation of γ_c and $\sum AE$/area with crack depth.

fracture energy of plain resin specimens $(\gamma_c)_{resin}$. Fig. 8 shows that the fracture energy of the unreinforced resin was independent of crack penetration. This suggests that geometrical effects are not responsible for the variation of γ_c with crack position that is shown in fig. 7.

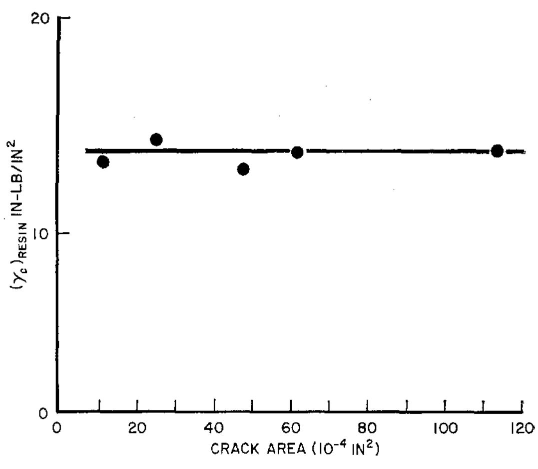

Figure 8 Variation of $(\gamma_c)_{resin}$ of a plain resin triangular section specimen.

4. Discussion

Table I shows the values of γ obtained by the three different techniques. Although there are significant differences between them, they do not differ greatly. The agreement between the linear elastic fracture mechanics value γ_l and the other values indicate that linear elastic fracture mechanics is suitable for describing the behav-

TABLE I Experimental and theoretical γ values.

Experiment	LEFM (γ_I)	190 in. lb/in².
	Work of fracture (γ_F)	100 in. lb/in².
	Compliance (γ_c)	450 to 100 in. lb/in².
Theory	Debonding	300 to 120 in. lb/in².
	Relaxation	200 to 80 in. lb/in².
	Pull-out	8000 to 4000 in. lb/in².

iour of the boron-epoxy system, although clearly more work remains to be done to verify this. The fact that γ_I is greater than γ_F might be due to the relative bluntness of the crack. The variation of the compliance analysis value γ_c with crack depth is surprising but the fact that the unreinforced material did not display the same variation suggests that it is not the test that is responsible. The variation of fibre pull-out length, across the fracture surface suggests that this is real material effect. It is, however, not clear why the pull-out length should vary.

γ_c values calculated by the compliance method are on a per unit crack area basis. Converting total acoustic emission counts to the same basis allows a direct comparison of the effect of crack depth on γ_c and $\sum$AE/area. Fig. 7 shows that both γ_c and $\sum$AE/area decreased as the crack depth increased. Fig. 9 shows that when $\sum$AE/area is plotted against γ_c, a consistent relationship is obtained. The number of counts recorded by the DRC apparatus is a function of both the number of acoustic emission pulses and

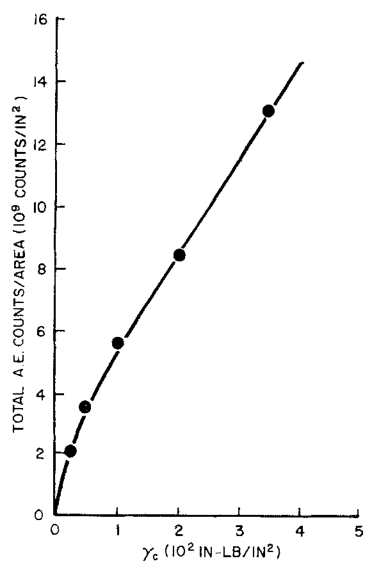

Figure 9 Variation of $\sum AE$/area with γ_c.

their amplitudes. However, because of the finite response time of the apparatus, it is also affected by the shape of the pulses and the frequency with which they are emitted. A more detailed correlation of microstructural processes with the physics of acoustic emission will therefore require much more experimental and developmental work. However, these preliminary results show that acoustic emission is directly related to the critical strain energy release rate and that this technique might be used as a means of monitoring fracture surface energies.

The high fracture surface energies of fibre composites arise from processes which are a consequence of their inhomogeneity. Two physical models which have received much attention are those of debonding [8] and fibre pull-out [9]. In the former, the fibre is considered to debond from the matrix over a length on either side of the matrix crack, so that when the fibre snaps the stored elastic energy in the debonded length cannot be redistributed in the rest of the composite, but is lost from the system. This leads to a fracture surface energy,

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

where Y is the debonded length, V_f is the volume fraction of fibres of strength σ_f and modulus E_f. It is not necessary for the fibre to debond for energy to be dissipated on fibre fracture, since when a fibre snaps, the stress at the broken end falls to zero and builds up over a distance $l_{c/2}$ from the end (where l_c is the critical transfer length [10]). If it is assumed that stress builds up linearly from the broken end it can easily be shown [11] that the stored energy lost from a fibre during its fracture gives

$$\gamma_{\text{relax}} = \frac{V_f \sigma_f^2 l_c}{6E_f}$$

In general, stress does not build up linearly from the end, so this estimate is likely to be rather too high. The fibre pull-out model requires that interfacial shear stresses be maintained while the fracture faces are being separated so that fibres pull out against a restraining force. The energy dissipated during the pull-out of discontinuous fibres of length equal to the critical transfer length l_c is given by [10].

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

This may be extended to the real case of a continuous fibre composite with a distribution of strengths by making the approximation that the lengths of fibre protruding from the fracture surface vary between zero and $l_c/2$ with a mean of $l_c/4$. Using the pull-out lengths to thus estimate l_c or the debonded length Y, theoretical estimates of the fracture surface energy may be made using the above models. Table I shows these values. The debonding or relaxation models give better agreement with experiment than the maintained shear stress during pull-out model. Each of the models leads to a decreasing fracture surface energy with increasing crack length because of the experimentally observed variation of pull-out length.

References

1. A. S. TETELMAN and A. J. MCEVILY, JUN., "Fracture of Structural Materials" (J. Wiley and Sons, Inc. New York, 1967).
2. R. W. DAVIDGE and G. TAPPIN, *J. Mater. Sci.* **3** (1968) 165.
3. H. G. TATTERSALL and G. TAPPIN, *J. Mater. Sci.* **1** (1966) 296.
4. H. T. CORTEN, "Micromechanics and Fracture Behavior of Composites," Chapter 2 of "Modern Composite Materials," Ed. Broutman and Krock (Addison-Wesley, Reading, Massachusetts, 1967) p. 45.
5. G. C. SIH, P. C. PARIS, and G. R. IRWIN, *Internat. J. Fracture Mech.* **1** (1965) 189.
6. W. F. BROWN JUN., and J. E. SRAWLEY, *ASTM*, STP No. 410 (1966).
7. S. W. TSAI, ONR/ARPA Association, Monsanto/ Washington University, (1968) HPC 68-61.
8. J. O. OUTWATER and M. C. MURPHY, 24th Annual Technical Conf., Reinforced Plastics/Composites Division, The Soc. of Plastics Industry, Inc., 1969.
9. A. KELLY, *Proc. Roy. Soc.*, **A 319** (1970) 95.
10. A. KELLY and G. J. DAVIES, *Metallurgical Rev.* **10** (1965) 1.
11. J. M. FITZ-RANDOLPH, M.S. Thesis, University of California, Los Angeles, 1971.

Received 15 October and accepted 26 October 1971.