JOURNAL OF MATERIALS SCIENCE 13 (1978) 195–201

Constrained cracking in glass fibre-reinforced epoxy cross-ply laminates

A. PARVIZI, K. W. GARRETT*, J. E. BAILEY
Department of Metallurgy and Materials Technology, University of Surrey, Guildford, UK

Specimens of 90° cross-ply glass-reinforced epoxy resins were tested in tension parallel to the direction of reinforcement in the outer plies. The thickness of the inner ply was varied and cracking constraint was observed at small thicknesses. At large inner-ply thicknesses the specimens showed uniform transverse cracking, and at very low inner-ply thicknesses this transverse cracking could be suppressed completely prior to total specimen failure. Fracture toughness values were determined on transverse unidirectional laminates of the same volume fraction. It was found that the cracking constraint observed can be accounted for, using the theory of Aveston and Kelly.

Introduction

In a recent paper, Garrett and Bailey [1] have described the occurrence of systematic transverse cracking of the inner ply in three-layered cross-ply laminates in which a tensile load is applied parallel to the fibres in the outer plies. The transverse cracks increase in number as the applied load is increased and the crack spacing approaches a limiting value which is dependant on the thickness of the inner ply.

This paper presents further results on the effect of inner-ply thickness on the apparent failure strain of the inner ply. In the work of Garrett and Bailey [1], the effective inner-ply thickness was investigated over only a limited range. The experiments reported here were aimed at investigating the possibility of constrained cracking of the kind first observed in unidirectional composites when the failure strain in the matrix is apparently increased in the composite: see, for example, Majumdar [2], Cooper and Sillwood [3], and Aveston, Mercer and Sillwood [4]; this phenomenon has been predicted theoretically by Aveston, Cooper and Kelly [5], and Aveston and Kelly [6]. The concept of constrained cracking may have technological significance, since it should be possible to design composite laminate structures in which cracking prior to ultimate failure is completely suppressed.

In this paper, experimental results are presented for glass fibre/epoxy resin cross-ply laminates, which show that crack constraint does occur at small transverse-ply thicknesses. Furthermore, by carrying out measurements of the fracture toughness for the transverse laminate, it has been possible to show that close agreement is obtained with existing theoretical ideas.

2. Experimental procedure

0, 90, 0 cross-ply and unidirectional laminates were made from epoxy resin (Shell Epikote 828 cured with 80 phr Epikure NMA and 0.5 phr BDMA) reinforced with "E" glass fibre rovings (Silenka 1200 tex). The glass rovings were wound onto open metallic frames of about 2 mm thickness and then the laminate was built up over an aluminium sheet by stacking up a sufficient number of frames in each direction. Fibres were wetted thoroughly by the liquid resin after each frame was laid down and the air entrapped was expelled by using a hot air blower. The laminate was eventually covered with a glass sheet to ensure a smooth surface and the excess resin squeezed out by applying sufficient pressure onto the laminate. Curing took place at 100° C for 3 to 4 h, followed by 3 h of post-curing at 150° C.

Cross-ply laminates were made, as illustrated in Fig. 1a, with a transverse-ply thickness $2d$ ranging

*Now at Federal University of Ceara, Brazil.

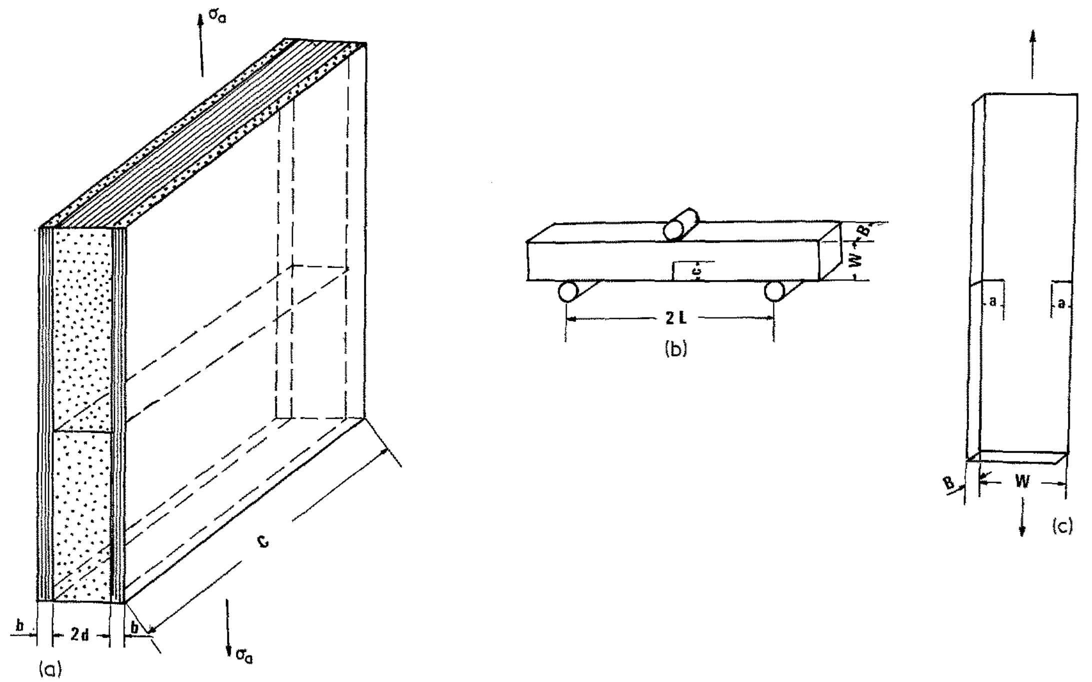

Figure 1 Illustration of specimen models: (a) 0, 90, 0 cross-ply tensile specimen (b) work of fracture specimen (c) double-edge-notched tensile specimen.

from 0.1 to 4 mm, whilst the longitudinal-ply thickness b, remained constant at about 0.5 mm on either side of the laminate. The fibre volume fraction of laminates was also kept constant at an average value of 0.55. Parallel sided specimens of dimensions 220 mm x 20 mm were cut from each cross-ply laminate and the specimen ends were reinforced with GRP tabs to prevent their premature failure at the grips. Specimens were tensile tested on a TTD model Instron machine at a cross-head speed of 0.5 mm min^{-1} (2.5 x 10^{-3} min^{-1} strain rate). The strain was recorded by electrical resistance strain gauges attached to the specimen and connected to the Instron recorder via a suitable bridge circuit. An acoustic emission transducer was also fixed to each specimen to give additional information during failure.

Tensile tests were also carried out on the unidirectional laminates, parallel and perpendicular to the fibre axis. Specimen preparation and test procedure were similar to those for the cross-ply laminates.

The fracture surface energy of transverse unidirectional laminates (crack running parallel to the fibres), having the same fibre volume fraction as cross-ply laminates, was measured using both work of fracture (WF) and linear elastic fracture mechanics (LEFM) methods.

The WF specimens were 4 mm x 6 mm rectangular bars of 60 mm length (45 mm span) with the edge notches cut into them by a thin blade diamond saw, see Fig. 1b. The specimens were fractured under slow three-point bending. The typical load–deflection curves obtained, Fig. 2, show the quasi-controlled type of crack propagation in WF specimens. The tests were carried out on a floor model Instron machine at a cross-head speed of 0.5 mm min^{-1} (~ 7.5 x 10^{-3} min^{-1} strain rate). Work of fracture, γ_F, was determined by dividing the area under the load–deflection curve (total energy absorbed) by twice the fracture cross-sectional area.

The LEFM specimens were doubled edge notched specimens (DEN) cut from 2 mm thick unidirectional laminates having dimensions 15 mm x 50 mm (see Fig. 1c). These specimens were fractured under tensile loadings carried out on a TTD model Instron at a cross-head speed of 0.2 mm min^{-1} (~ 4 x 10^{-3} min^{-1} strain rate). The fracture surface energy of initiation, γ_I, was calculated from the maximum load P_f required to fracture both the tensile specimens and the work of

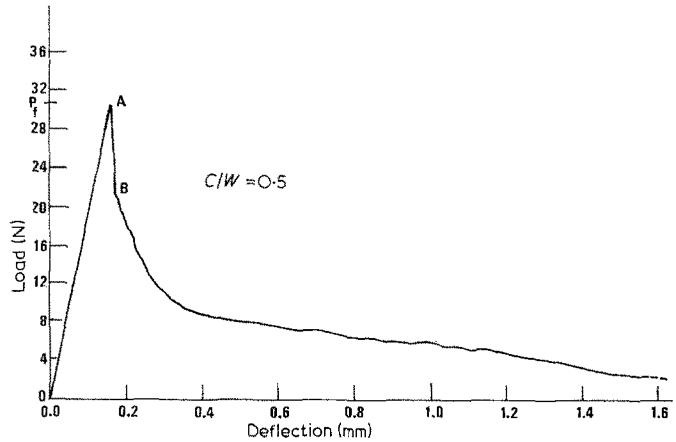

Figure 2 Typical load–deflection curve obtained for work of fracture specimens showing the quasi-controlled manner of crack propagation.

fracture specimens by using the following equations [7]. DEN specimen;

$$\gamma_{\rm I} = \frac{P_{\rm f}^2}{2EWB^2}\left(\tan\frac{\pi a}{W} + 0.1\sin\frac{2\pi a}{W}\right), \quad (1)$$

WF specimen;

$$\gamma_{\rm I} = \frac{P_{\rm f}^2 L^2}{2EB^2W^3}\left(31.7\frac{c}{W} - 64.8\frac{c^2}{W^2} + 211\frac{c^3}{W^3}\right). \quad (2)$$

3. Experimental results

The stress–strain curves obtained for the cross-ply laminates with comparatively large transverse-ply thicknesses were very similar to those described elsewhere [1]; a characteristic "knee" was observed at a well defined strain ϵ_{tu} which was associated with the onset of transverse cracking in the inner ply. The cracking phenomena could be observed visibly and detected by acoustic emission. However, in the experiments presented here, the transverse cracking behaviour changed markedly as the inner-ply thickness decreased below about 0.4 mm.

Fig. 3 shows the observed variation of ϵ_{tu} with inner-ply thickness; above about 0.4 mm, ϵ_{tu} is constant having a value of 0.6 ± 0.1%. These laminates exhibited typical multiple cracking in which the crack-spacing decreased with increasing applied stress and the specimen was eventually covered with evenly spaced cracks. The crack-spacing was also found to decrease with decreasing inner-ply thickness. No delamination between transverse and longitudinal plies was observed except for the 4 mm thick laminate, in which delaminating cracks appeared at both sides of the transverse cracks soon after they were formed. The observed value of 0.6 ± 0.1% for ϵ_{tu} is to be compared with the experimental value of 0.5 ± 0.1% obtained for the transverse failure strain of single unidirectional laminates.

With comparatively thick inner-ply specimens the majority of the cracks spanned the whole of the inner ply and propagated instantaneously. As the inner ply thickness was reduced below ~0.4 mm, changes occurred in the nature of the cracking phenomena. Small "edge cracks" appeared at the edges of the inner ply at a strain of about 0.8%. These "edge cracks" multiplied and grew very slowly to varying lengths; some eventually spanned the inner ply with increasing load. This transition from instantaneous to slow crack growth occurred at a thickness of 0.25 mm.

It is now necessary to distinguish between the strain ϵ_{tu} at which transverse cracks spanning the whole inner ply were first observed and the strain ϵ_{ti} at which edge cracks first appeared. At large transverse ply thicknesses $2d$, ϵ_{tu} and ϵ_{ti} had similar values as the cracking phenomenon was instantaneous; as $2d$ decreased below 0.4 mm ϵ_{tu} increased more rapidly than ϵ_{ti}. Only ϵ_{tu} has been plotted in Fig. 3.

When the inner-ply thickness was reduced to 0.15 mm, transverse-cracking was not observed prior to the complete failure of the specimen, i.e. $\epsilon_{tu} > \epsilon_{cu}$, where ϵ_{cu} is the composite fracture strain. However, edge cracks were observed at $\epsilon_{ti} \approx 1.0\%$ but their growth under increasing strain was so slow that they did not exceed ~1 mm length when the specimen failed at ϵ_{cu}. As the inner ply thickness decreased to 0.1 mm, not even the "edge cracks" were observed. Fig. 4 shows the multiple cracking, transition state, edge cracks and complete crack suppression in cross-ply specimens.

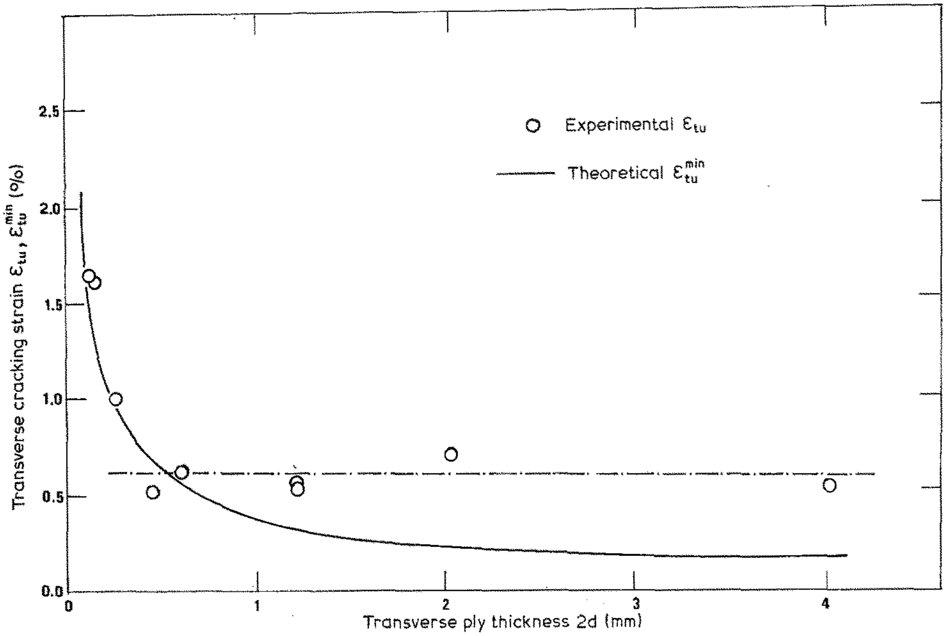

Figure 3 Values of ϵ_{tu}^{min}, as a function of ply thickness 2d (from Equation 13) and experimental values of ϵ_{tu} for various ply thicknesses in glass fibre/epoxy 90° cross-ply laminates. The horizontal line depicts the limiting values of ϵ_{tu} for large inner-ply thicknesses.

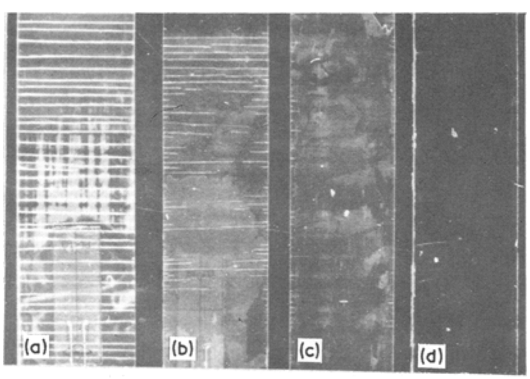

Figure 4 Cross-ply laminates as the thickness of the inner-ply $2d$ is reduced; (a) transverse multiple cracking ($2d = 1.20$ mm), (b) transition from instantaneous to slow crack growth ($2d = 25$ mm), (c) edge cracks ($2d = 0.15$), and (d) complete crack supression ($2d = 0.10$). All specimens have been loaded up to 1.4% strain.

In view of the appearance of "edge cracks" further studies are being made on specimens deliberately designed to avoid them.

The values of the fracture surface energy of initiation γ_I and the work of fracture γ_F obtained from the bend tests and the values of γ_I obtained from the tensile tests are shown in Fig. 5. The values of γ_I obtained from both tests are found to be identical and are independent of crack size, indicating the applicability of the LEFM to the material under test. The close fit of the experimental results and Equations 1 and 2, is shown in Fig. 6; these equations have been fitted to determine a γ_I value of $120 \pm 30\,\mathrm{J\,m^{-2}}$.

The γ_F values are generally higher than the γ_I values within the range of the crack size used in this work and decrease fairly rapidly as the notch depth increases. This was found to be associated with a change in the extent of the catastrophic crack propagation (i.e. the length of the line AB in the load–deflection curve in Fig. 2). For shallow notches, the crack propagation was almost catastrophic. As the notch depth increased, the slow propagation of the crack became more dominant, until for the very deep notches the crack propagation was largely controlled. The approach of γ_F values towards the γ_I values at large crack sizes is the result of more controlled crack propagation.

The Young's moduli of longitudinal and transverse unidirectional composites containing 0.55 volume fraction glass fibres have been measured and the values obtained are: $42 \pm 1\,\mathrm{GN\,m^{-2}}$ for the longitudinal Young's modulus E_l and $14 \pm 0.5\,\mathrm{GN\,m^{-2}}$ for the transverse Young's modulus E_t. A value of $5\,\mathrm{GN\,m^{-2}}$ has been deduced for the shear modulus G_t of the transverse ply using the equation given by Halpin and Tsai [8].

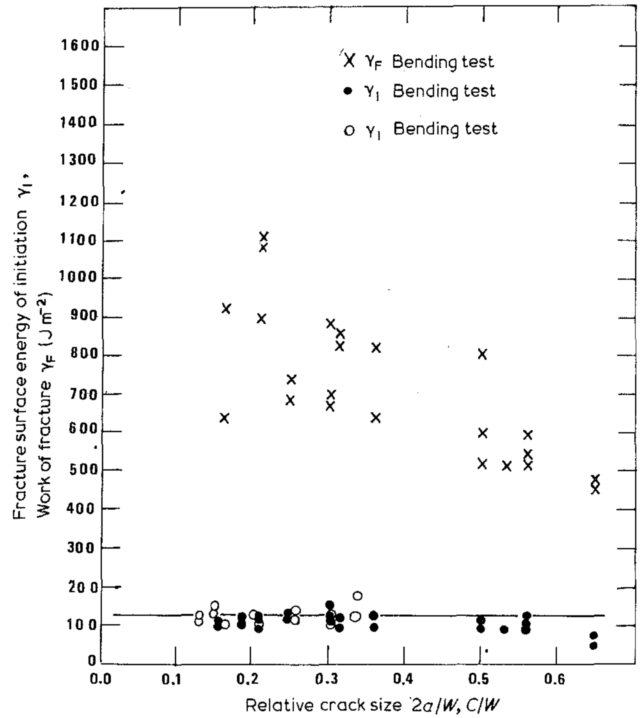

Figure 5 The fracture surface energy of initiation γ_I and work of fracture γ_F values of transverse unidirectional glass fibre/epoxy composites as a function of relative crack size ($V_f = 0.55$).

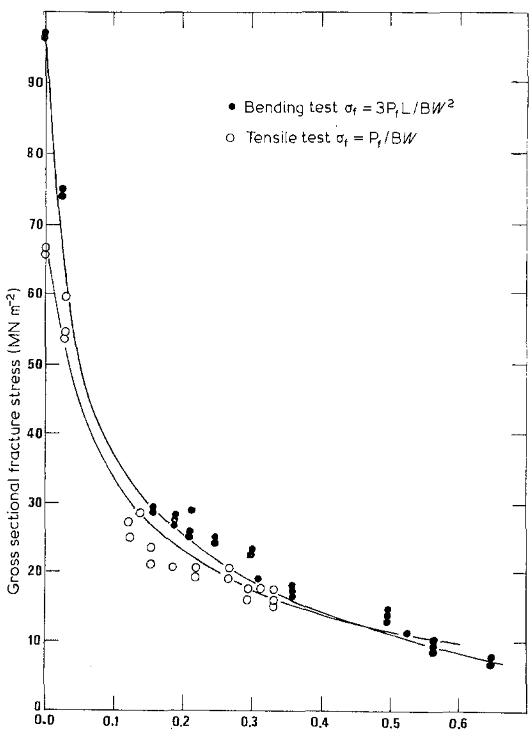

Figure 6 The cross-sectional applied stress as a function of relative crack size in glass fibre/epoxy transverse unidirectional composites. The solid lines are the theoretical values obtained by fitting Equations 1 and 2 to the results.

4. Discussion

Aveston *et al.* [5] and Aveston and Kelly [6] have discussed cracking constraint in unidirectional fibre composites for the two cases when the fibres and matrix are considered to debond during multiple cracking in the matrix and when a perfect bond exists during this cracking. They argue that, for a specimen under conditions of constant load, a crack will not form unless

$$\Delta W > \Delta U_S + U_D + 2\gamma_m V_m \qquad (3)$$

where ΔW is the work done by the applied stress per unit area of composite, ΔU_S is the increase in stored energy per unit area of the specimen, U_D is the energy lost due to any dissipative processes present (e.g. sliding friction between debonded fibres and matrix), γ_m is the fracture surface work per unit area and V_m is the volume fraction of the matrix. For a fully elastic composite $U_D = 0$. The analysis predicts the conditions under which a crack can exist in the matrix by considering the energy state of the body just before the crack is formed, to that when the crack has completely severed the matrix. No consideration is given to the origin of the crack or the way it propagates through the matrix, but nevertheless the method has been very successful in predicting cracking constraint in unidirectional composites. These ideas can be extended to the cross-ply laminates investigated in this work using the theory of multiple cracking developed in an earlier paper [1].

It has been found that the interface between the longitudinal and transverse plies remains bonded during the cracking of the transverse ply and so the laminate is considered to behave in a fully elastic manner. In our specimens Inequality 3 becomes

$$\Delta W > \Delta U_S + 2\gamma_t \frac{d}{(d+b)} \tag{3a}$$

where γ_t is the fracture surface energy of the inner ply in a direction parallel to the fibres. Then following the procedure of Aveston and Kelly we assume that half of the work done by the applied stress is available to produce the 2 crack surfaces. Therefore

$$\Delta U_S = \tfrac{1}{2}\Delta W, \tag{4}$$

and so from Inequality 3a a crack will occur if

$$\tfrac{1}{2}\Delta W > 2\gamma_t \frac{d}{(d+b)} \tag{5}$$

where $2d$ is the transverse-ply thickness. Fig. 7 explains the reasoning behind Equation 4.

When the first crack occurs in the transverse ply at a strain of ϵ_{tu} an additional stress $\Delta\sigma$ is thrown onto the outer plies. $\Delta\sigma$ is given by

$$\Delta\sigma = \frac{d}{b} E_t \epsilon_{tu} \exp(-\phi^{1/2} y) \tag{6}$$

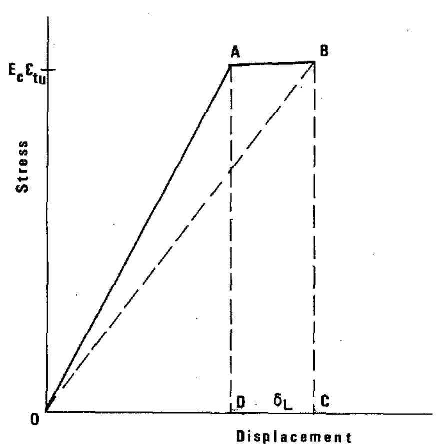

Figure 7 Specimen extension under constant stress resulting from first transverse crack. $\Delta W \equiv$ area ABCD (see Equation 4; $\Delta U_S \equiv$ area ABO $\equiv \frac{1}{2}$ area ABCD $\equiv \frac{1}{2}\Delta W$; thus energy available for formation of crack surfaces is $\Delta W - \Delta U_S = \frac{1}{2}\Delta W$.

where b is the longitudinal-ply thickness, E_t is the modulus of the transverse ply, E_l is the modulus of the longitudinal ply, y is the distance from the plane of the crack along the length of the specimen.

$$\phi = \frac{E_c G_t}{E_l E_t} \frac{b+d}{bd^2}, \tag{7}$$

where E_c is the modulus of the laminate, and G_t the shear modulus of the transverse ply parallel to the fibres.

When the additional stress is thrown onto the outer plies the laminate increases in length by δL (Fig. 7), given by

$$\delta L = 2\int_0^\infty \frac{\Delta\sigma}{E_l} \, dy. \tag{8}$$

Substituting Equation 6 into Equation 8 and integrating, we have

$$\delta L = \frac{2dE_t\epsilon_{tu}}{bE_l\phi^{1/2}}. \tag{9}$$

The work done by the applied stress σ_a at the strain when the crack forms is given by

$$\Delta W = \delta L \sigma_a. \tag{10}$$

Substituting Equation 9 and $\sigma_a = E_c\epsilon_{tu}$ into Equation 10, we have

$$\Delta W = \frac{2dE_tE_c\epsilon_{tu}^2}{bE_l\phi^{1/2}}. \tag{11}$$

Thus substituting Equation 11 into Equation 5 a crack will not be possible unless

$$\epsilon_{tu}^2 > \frac{2bE_l\gamma_t\phi^{1/2}}{(b+d)E_tE_c}. \tag{12}$$

The minimum value of the transverse failure strain ϵ_{tu}^{min} is given by

$$\epsilon_{tu}^{min^2} = \frac{2bE_l\gamma_t\phi^{1/2}}{(b+d)E_tE_c}. \tag{13}$$

It is possible to predict minimum values of ϵ_{tu} from Equation 13 as a function of inner-ply thickness $2d$. Values are known of all the required parameters; an average value of $120\,J\,m^{-2}$ determined by the fracture toughness tests have been taken for γ_t.

The values of ϵ_{tu}^{min} are shown as a function of ply thickness $2d$ in Fig. 3; also shown are the experimental values.

The experimental values for ϵ_{tu} at large values of $2d$ are appreciably greater than those predicted by Equation 13 but agreement between experiment and theory is remarkably close at values of $2d$ less than about 0.5 mm and the theory does therefore predict the observed cracking constraint very well. The theory also predicts complete constraint, i.e. $\epsilon_{tu} > \epsilon_{cu}$ for $2d < 0.1$ mm in close agreement with the experimental results. Equation 13 is essentially a thermodynamic condition and the results obtained seem to suggest that at large values of $2d$ other factors determine the failure strain, e.g. the mechanism of crack propagation. It should also be noted that during the onset of constrained cracking, small cracks (generally edge cracks) are observed to propagate slowly with increasing load; some of these finally span the whole cross-section of the inner ply. It would seem therefore that at these low-ply thicknesses it is the energetics that are controlling crack growth rather than the mechanism. The growth behaviour and origin of these "edge cracks" is the subject of further investigation.

5. Conclusions

Cracking constraint has been observed in 90° cross-ply laminates of glass fibre-reinforced epoxy resin with inner ply thicknesses less than about 0.4 mm. Complete suppression of inner-ply failure is observed at inner-ply thicknesses of 0.1 mm. In the thickness range from 0.4 to 0.1 mm small cracks (predominantly edge cracks) are observed which propagate slowly under a rising load, finally triggering total inner-ply failure. The onset of constrained cracking and its complete suppression can be accounted for remarkably well using the theory of Aveston and Kelly for the fully elastic case.

References

1. K. W. GARRETT and J. E. BAILEY, *J. Mater. Sci.* 12 (1977) 157.
2. A. MAJUMDAR, *Proc. Roy. Soc.* A319 (1970) 69.
3. G. A. COOPER and J. M. SILLWOOD, *J. Mater. Sci.* 7 (1972) 325.
4. J. AVESTON, R. A. MERCER and J. M. SILLWOOD "Composites Standard Testing and Design". Conference Proceedings. National Physical Laboratory, 8, 9 April (IPC Science and Technology Press, Guildford, Surrey, 1974) p. 93.
5. J. AVESTON, G. A. COOPER and A. KELLY, "The Properties of Fibre Composites" Conference Proceedings, National Physical Laboratory (IPC Science and Technology Press, 1971) p. 15.
6. J. AVESTON and A. KELLY, *J. Mater. Sci.* 8 (1973) 352.
7. J. E. SRAWLEY and W. F. BROWN, "Fracture Toughness Testing and its Applications". ASTM STP 381 (1964).
8. J. E. ASHTON, J. C. HALPIN and P. H. PETIT "Primer on Composite Materials: Analysis", (Technomic, Stamford, 1969).

Received 28 February and accepted 4 April 1977.