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Factors Affecting Fracture Behaviour of Composite Materials

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Abstract: A review of the various factors affecting the critical energy release rate (G_c) for composite materials in terms of mode-I and mode–II delamination are given. It is noted that resin toughness is particularly significant in determining the composite resistance to mode-I delamination when the neat-resin toughness is in the range of 100-400 J m^{-2}, moderately significant for G^m_{IC} in the range of 400-2000 J m^{-2}, and fairly insensitive to neat-resin toughness with G^m_{IC} greater than 2000 J m^{-2}. The resistance to mode-II delamination is generally less sensitive to neat–resin toughness than is for mode-I delamination. In addition, other factors such as, composite structure, through-thickness reinforcement, moulding methods, specimen thickness, specimen defect type, rate of testing and moulding temperature are also reviewed and discussed.

Key words: Interlaminar fracture toughness, Composite materials, Delamination, Crack growth.

Introduction

Laminated fibre reinforced plastic composites are particularly susceptible to failure by delamination initiation and growth as a result of a combination of compressive and bending stresses caused by the delaminated plies as they buckle out of plane. Structural polymer composites exhibit complex fracture behaviour. It is therefore important to understand specifically how the properties of the fibre reinforced composites and also the various factors affect fracture behaviour, particularly damage caused by delamination or interlaminar fracture[1-12]. This paper reviews the various factors that affect the delamination resistance of fibre composites under the mode-I and mode-II which are believed to represent the critical strain energy release rate for delamination of composites.

Effect of matrix toughness

The matrix toughness plays an important role in the interlaminar fracture behaviour of composites. It is reported by Bradley [13] and Jordan et al [14] that decreasing the yield strength of the matrix increases delamination fracture energy by increasing the size of the plastic deformation or non-linear viscoelastic zone ahead of the crack tip, resulting in greater load redistribution away from the crack tip and hence more crack-tip blunting. However, there is a relatively low efficiency of translation of a high matrix G^m_{IC} value into the delamination fracture energy of a composite, due mainly to constraint provided by the fibres in the confined spaces between the reinforcing plies which restricts the size of the plastic deformation zone [15-17]. Thus, a decrease in fibre V_f can result in a smaller degree of crack tip constraint, giving larger deformation/damage zone sizes and, consequently, higher composite G_C values in either mode-I and mode-II. In contrast, very efficient translation of G^m_{IC} into delamination fracture energy is observed in composites made with brittle matrices, due to the full development of the small plastic deformation zones which occur in these matrices and, in addition, failure mechanisms such as interfacial debonding and fibre bridging can also contribute significantly to G_{IC}.

Bradley [13] reported that increasing matrix G^m_{IC} above an initial value of 70 J m^{-2} resulted in significant increases in composite G_{IC}, but that further increases in matrix G^m_{IC} (> 400 J m^{-2}) resulted in a poor translation of G^m_{IC} values into composite G_{IC}. Similar results were reported by Russell and Street [15], in that the incremental increase in composite G_{IC} that resulted from increasing G^m_{IC} above 400 J m^{-2} was much smaller than the incremental increases observed as G^m_{IC} was increased from 100 to 400 J m^{-2}, and suggested that there may be little improvement in the composite G_{IC} values for increases in G^m_{IC} above 2000 J m^{-2}. Bradley [13] and Russell and Street [15] also reported that the composite mode-II (G_{IIC}) values are less sensitive to G^m_{IC} values than the composite G_{IC}, in that a two times variation in G_{IIC} and a ten times variation in G_{IC} resulted from a hundred times increase in matrix G^m_{IC} (from 80 to 8100 J m^{-2}). Bradley [13] reported that for tougher resins, this difference in behaviour between the composite mode-I and mode-II fracture energy is minimised for the G^m_{IC} greater than 6000 J m^{-2}, when the ratio of composite mode-II to mode-I fracture energy approaches 1.0.

Typical G_{IC} values for unidirectional carbon and glass fibre epoxy composites are reported [18-24] to be in the range 200-400 J m^{-2}, and 800-1700 J m^{-2} for toughened-epoxy composites. For epoxy composites reinforced with woven glass-fibre mats, typical values of G_{IC} are in the range 800-1000 J m^{-2}, even if the matrix is not modified [24]. Wang and Zhao [22] in a study of glass woven roving/epoxy composites with particulate-filled matrices, reported significant improvements in G_{IC} values from 800 J m^{-2} (unfilled matrix) to 1450 J m^{-2} and 1700 J m^{-2} for composites containing calcium sulphate whiskers and mineral fillers, respectively. Srivastava and Hogg [23] studied glass woven roving/polyester composites containing particles of polyethylene and aluminium tri-hydrate and reported little improvement in G_{IC} values, whereas the toughening effect of particles on G_{IIC} was significant, particularly for the polyethylene filled composites. Thus, the G_{IC} values of filled composites increased from 900 J m^{-2} (unfilled) to 1000 J m^{-2} to 1260 J m^{-2} for the composites containing aluminium tri-hydrate fillers and polyethylene particles, respectively, whereas G_{IIC} increased from 1250 J m^{-2} (unfilled) to 1450 J m^{-2} to 1850 J m^{-2} for composites containing aluminium tri-hydrate and polyethylene particles, respectively. The authors suggested [22-23] that the increased toughness was due to the reduction of matrix rich regions between the reinforcement plies, in which the added particles effectively enhanced the matrix performance by diverting the crack growth, plus debonding of the particles and pulling-out of the particles from the matrix, all of which act as energy sinks and increase the fracture energy.

Effect of composite structure

Davies and Bezeggagh [24] reported the interlaminar fracture behaviour of composites to be structure dependent, even for composites with identical fibre volume fractions, in that they tend to show different propagation behaviour if fibre distributions differ. Mode-I tests on carbon/epoxy (61% volume fraction) composites with even fibre distribution showed higher values of G_{IC} than composites with matrix-rich interlaminar regions. The higher value of the former was attributed to the formation of fibre bridging behind the crack tip, these bridges span the crack and result in higher delamination resistance.

Effect of through-thickness reinforcement

Guenon et al [25], in a study of carbon (T300)/epoxy (3501-6) composites with a 1% volume fraction of through-thickness fibres, reported a ten-fold increase in G_{IC} as a result of the transverse stitching across the laminate which held the reinforcing fibres together. Lalit and Yiu-Wing [26] studied similar carbon/epoxy composites with kevlar threads as through-thickness reinforcement, and reported that G_{IIC} increased from 1300 J m^{-2} (composite without through-thickness stitching) to 2350 J m^{-2}, due to the development of a bridging stitch-thread zone behind the crack tip. The authors reported that the increase in G_{IIC} was approximately 80% for a stitch density S_d = 4 st cm^{-2} and 3.5-fold for a stitch density S_d = 12 st cm^{-2}.

Effect of moulding methods

Interlaminar fracture energies obtained for specimens produced by different moulding methods can show different values [27]. Specimens produced by hand lay-up (HLU) can cause problems during fracture testing; cracks may deviate from the original crack plane invalidating the fracture mechanics approach, a lack of reinforcement symmetry may cause twisting and mixed mode loading, and the introduction of voids can lead to lower interlaminar fracture energies [24]. These problems may be resolved by producing specimens via resin transfer moulding (RTM), which tend to show higher interlaminar fracture energies than those produced by HLU. Sumpter et al [27] suggested that interlaminar fracture energy can be a function of both specimen geometry and manufacturing method. For example, HLU specimens showed a reduction in G_{IC} with increasing thickness while RTM specimens showed the opposite trend. The RTM materials exhibit twice the G_{IC} of the HLU materials (≈ 1200 J m^{-2}, compared to ≈ 650 J m^{-2} for HLU).

Effect of specimen thickness

Hojo and Aoki [28] investigated the effects of DCB thickness for carbon (AS4)/PEEK (APC-2) composites (3, 4, 5, and 8 mm thick) and measured initiation G_{IC}

values in the range of 1100 J m^{-2} to 1300 J m^{-2}, which were essentially independent of the specimen thickness. In contrast to the initiation values, the propagation G_{IC} values varied in the range of 1000 to 2000 J m^{-2}. For carbon (T800)/epoxy (3631) composites [40], the initiation G_{IC} values were lower (150 to 180 J m^{-2}) and were again independent of the specimen thickness, but in contrast to the carbon/PEEK composites the propagation G_{IC} values showed only small variations with thickness in the range of 180 to 200 J m^{-2}. Extensive fibre bridging was observed near the crack tip for the carbon/epoxy composites and it was reported that this appeared to minimise any effect of specimen thickness. Studies by Davies et al [29] on similar composite systems reported reasonably constant propagation G_{IC} values of $\approx$ 200 J m^{-2} for DCB specimen with thicknesses of 1.6, 3.2 and 5.2 mm for carbon/epoxy composites, but increased propagation G_{IC} values (1600, 1700 and 1900 J m^{-2}, respectively) with specimen thickness for carbon/PEEK composites. Davies et al [30] also reported the effects of DCB thickness (3 mm and 5 mm thick) for carbon (AS4)/PEEK (APC-2) and carbon (IM6)/PEEK (APC-2) composites, initiation G_{IC} values were once again essentially independent of the specimen thickness, whereas the propagation G_{IC} values showed increasing values of (AS4) 1540 J m^{-2} and 2400 J m^{-2} and (IM6) 2110 J m^{-2} and 3240 J m^{-2}, for the 3 mm and 5 mm specimens respectively. Thus, for most studies the results indicate a trend of increasing propagation G_{IC} values with increasing DCB thickness.

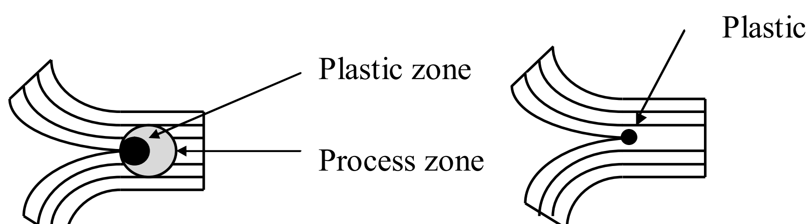
Davies et al [30] studied the effect of the width of carbon/epoxy DCB specimens tested in mode-I and found no influence of width on the propagation G_{IC} values of 250 J m^{-2} for specimens of 10, 15 and 20 mm wide. The same group also studied the influence of specimen width on ENF specimens tested in mode-II and reported different G_{IIC} values for the 20 mm specimens at the onset of non-linearity (300 J m^{-2}), at 5% increase in compliance (610 J m^{-2}) and at maximum load (650 J m^{-2}) compared with the 10 mm and 15 mm specimens which showed similar G_{IIC} values at the onset of non-linearity (400 J m^{-2}), at 5% increase in compliance (500 J m^{-2}) and at maximum load (600 J m^{-2}). In addition, increasing the distance between the supports from 80 to 106 mm for ENF specimens of 20 mm width gave higher G_{IIC} values at the onset of non-linearity of $\approx$ 450 J m^{-2}, but G_{IIC} values at maximum load remained unchanged at $\approx$ 650 J m^{-2}.

Effect of defect type

A study of starter film thickness and precracks by Davies et al [31] indicated that carbon/epoxy DCB specimens with thin aluminium foil (20 μm) starter films showed lower initiation G_{IC} values of 100 J m^{-2} compared to specimens with 40 μm thick aluminium foil (120 J m^{-2}) and 60 μm thick PTFE film (180 J m^{-2}). Mode-I precracking resulted in initiation G_{IC} values similar to those during propagation (200 J m^{-2}), but mode-II precracking gave initiation values close to that of the thinnest film. Mode-II ENF specimens precracked in mode-I showed values independent of the precrack length (1mm, 4 mm and 8 mm) with G_{IIC} of 500 J m^{-2}, whereas specimens with 20 μm and 40 μm thick aluminium foil gave higher G_{IIC} values of 1000 J m^{-2} and 1500 J m^{-2}, respectively.

Effect of rate of testing

Gillespie et al [32] conducted mode-I interlaminar fracture tests on a carbon (AS4)/epoxy (3501-6) composite over a range of cross-head displacement rates (2.5, 25 and 250 mm min^{-1}), observing stable crack propagation at all rates and little variation in G_{IC}. Similarly, Smiley and Pipes [33] conducted mode-I DCB tests on carbon (AS4)/epoxy (3501-6) composites but at much higher testing rates (25 mm s^{-1} to 210 mm s^{-1}), finding that these brittle-matrix composites exhibited a stable, brittle mode of fracture under all conditions. Gillespie et al [32] also conducted tests on carbon (AS4)/PEEK (APC-2) composite at cross-head rates of 0.25, 2.5, 25 and 250 mm min^{-1}, and found that crack propagation was rate-dependent, which was attributed to plastic and viscoelastic effects in the process zone around the crack tip. This is shown schematically in Figure 1 [32].



B

Figure 1. Schematic diagram showing deformation process zone for a carbon/PEEK composite; (A) low rates, (B) high rates.

The authors[32-33] reported that at low rates, viscoelastic effects (i.e. time-dependent matrix deformation and microcrack formation) produced an upper bound on the size of the deformation process zone (external to plastic deformation zone, see Figure 1) at the crack tip. At the highest rates, viscoelastic effects were negligible and the deformation process zone tended to be coincident with the developing plastic zone (Figure 1.B).

The propagation G_{IC} values for these composites decreased with testing rate. This reduction in G_{IC} was attributed to a ductile-to-brittle transition of the polymer matrix in the deformation process zone with increasing testing rate. Blackman et al [34] conducted mode-I tests on carbon (AS4)/PEEK (APC-2) composite, reported no major decrease in G_{IC} upon increasing rate of test from 3.3 x 10^{-5} m s^{-1} to 15 m s^{-1}. For carbon (T400)/epoxy (6376C) composite [34], the value of G_{IC} remained insensitive upon increasing rate. Blackman et al [35] also conducted mode-II tests on the same composites over a range of cross-head rates from 1.67 x 10^{-5} m s^{-1} to 10 m s^{-1}. At a high rate of loading, a slight decrease in the value of G_{IIC} from 2500 J m^{-2} to 2000 J m^{-2} occurred for the thermoplastic composite and a slight increase in the corresponding values from 700 J m^{-2} to 800 J m^{-2} for the epoxy composite.

Effect of moulding temperature

Saidpour et al [36] conducted mode-II tests on unidirectional carbon/epoxy composites, reported higher G_{IIC} values of interlaminar fracture energy than mode-I [37]. The G_{IIC} values were well above 1000 J m^{-2} for composites moulded at higher temperature compared with medium and low temperature moulding composites. They reported that after the initial cure G_{IIC} values were fairly low for medium moulding composites. However, G_{IIC} for these materials, increased significantly after 200 ^{0}C post-cure reaching the similar values obtained for high temperature moulding composites. Saidpour et al [36] also reported that postcuring conditions had significant effect on fracture toughness energy giving the similar values obtained for high temperature moulding due to better phase separation for medium and low temperature moulding systems.

Conclusions

1. Resin toughness was seen to play a significant role in the interlaminar fracture toughness of composite materials.

2. Low efficiency of translation of resin fracture toughness into delamination for very ductile resins was the result of the constraint in the development of a larger plastic zone in the resin-rich area between plies by the fibres in the adjacent plies.

3. High delamination toughness observed in composites made with brittle resins was due to the full development of the small plastic deformation zones which occur in these matrices and, also due to the failure mechanisms such as interfacial debonding and fibre bridging that contributed significantly to G_{IC} values.

4. Composites made with HLU showed a reduction in G_{IC} with increasing thickness in comparison to composites made with RTM which showed the opposite trend due to less void content in these materials.

5. Studies indicated a trend of increasing propagation G_{IC} values with increasing thickness and width of the DCB specimens.
6. It was noted that G_{IIC} values can be affected by moulding temperature and post-cure of composite materials.

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