The energy of crack propagation in carbon fibre-reinforced resin systems

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The energy expended during controlled crack propagation in unidirectionally reinforced composites of carbon fibre in a brittle resin matrix has been evaluated in terms of the energy dissipated during fibre-snapping, matrix-cracking and fibre pull-out. The work of fracture, \( \gamma_f \), is found to depend principally on the frictional shear stress at the fibre/resin interface opposing pulling out of broken fibres. Differences in \( \gamma_f \) for carbon fibre/resin composites exhibiting a range of interfacial shear strengths and void contents have been explained with reference to variations in fracture surface topography of the fibrous composites. The effect of environment on properties of the interface and work of fracture was also investigated. The energy required to propagate a crack has been compared with the energy for fracture initiation, \( \gamma_i \), using a linear elastic fracture mechanics approach. It was found that fibre pull-out energy is the principal contribution to \( \gamma_f \), and \( \gamma_i \) is similar to the elastic strain energy release rate at the initiation of fracture of a brittle, orthotropic solid. For crack propagation parallel to fibres, \( \gamma_f \) and \( \gamma_i \) are similar and not unlike the fracture surface energy of the resin alone. The strength of the interface is important only in so far as it affects the value of \( \gamma_i \).

1. Introduction

An understanding of the events leading to initiation and growth of microcracks in composite materials under the combined action of stress and some active environment should indicate the important parameters for the design of a tough, notch-insensitive engineering material. Although the prevention of flaws and voids in materials is highly desirable, their presence is nevertheless an inherent feature of some types of composite. Quite apart from the existence of internal flaws in fibres, it is not unusual to find interfacial or matrix porosity in the form of pockets of air trapped during the fabrication process. This raises the question of how composites containing these and worse imperfections, such as cracks, will perform under load, particularly in the presence of an environment other than air.

The experiments described here have been carried out from a fracture mechanics viewpoint. We have measured the work required to fracture sound and porous materials. We have attempted to establish the nature of the major energy absorbing fracture processes in CFRP and to isolate the mechanisms which influence crack motion. The effect of moisture and oil on fracture mechanisms has also been investigated.

2. A fracture mechanics approach to the failure of fibrous composites

The resistance of a solid to microcracking, leading to the formation of macrocracks, can be evaluated quantitatively by determining the fracture "surface" energy of the solid. In isotropic materials, the release of stored elastic strain energy per unit area of fracture surface during initiation of microcracking, \( \gamma_i \), may be similar to the energy dissipated during its extension to macrocrack dimensions, leading to ultimate failure of the material. The relationship between \( \gamma_i \) and the ultimate strength of the cracked solid, \( \sigma_t \), is described by the classical equation given by Griffith [1]:

\[
\gamma_i = \frac{\sigma_t^2 (\pi c)}{2E}
\]  

(1)

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where \( c \) is the critical size of flaw, and \( E \) is the Young’s modulus of the solid.*

In composite materials, the multiplicity of microfracture processes which influence the formation and growth of microcracks makes the separation of their characteristic energy-release rates extremely difficult. The mechanism controlling the slow growth of a crack may be quite different from that influencing its rapid propagation and, as a consequence, the resistance of the composite to crack nucleation may not be the same as for growing cracks [2].

The fracture surface energy of a carbon fibre-reinforced polyester resin is much greater than the sum of the fracture surface energy of the fibre, \( \gamma_f \), and resin matrix, \( \gamma_m \) [3]. Additional energy must, therefore, be absorbed by other microfracture processes as the crack extends through the composite. Fracture mechanisms that have been proposed to account for this extra energy include: (1) relaxation of a fibre when it snaps, \( \gamma_r \) [4]; (2) debonding energy released during separation of fibre and resin, \( \gamma_d \) [5]; and (3) fibre pull-out energy, \( \gamma_p \), to overcome friction at the fibre/resin interface, during extraction of the fibre [6]. The total work to fracture the composite, \( \gamma_c \), is now considered as the sum of these fracture energies:

\[
\gamma_c = \gamma_f + \gamma_m + \gamma_r + \gamma_d + \gamma_p
\]

However, one or more of these microfracture processes may control different stages in the complete fracture of the composite, and the resistance of the composite to cracking will therefore vary as the crack extends.

Under certain conditions, the Griffith fracture concept may be applied to anisotropic materials and, by a further approximation, to the behaviour of fibrous composites. The problem of extending a crack in an orthotropic solid, has been considered by Sih, et al [7], and Wu [8]. Their calculations show that the usual Irwin-Westergaard expressions [9] describing the complete stress distribution about a crack tip must be modified by functions containing the orthotropic elastic constants of the material, but that the stress field intensities at the crack front are identical with those for an isotropic solid.

For an infinite plate containing a small crack, the stress intensification at the tip of the crack is given by:

\[
K = \sigma_{\infty} \sqrt{\pi c} \quad \text{(2)}
\]

where \( K \) is the stress intensity factor and \( \sigma_{\infty} \) is the stress applied to the plate. When the crack reaches a critical size, \( K \) also becomes critical, \( K_c \), and combining Equations 1 and 2:

\[
K_c = \sqrt{(2E \gamma_f)} \quad \text{(3)}
\]

\[
P_{\text{eff}} = \begin{pmatrix}
\frac{s_{11}}{E_{11}} & \frac{s_{12}}{E_{11}} & \frac{s_{16}}{E_{11}} \\
\frac{s_{12}}{E_{12}} & \frac{s_{22}}{E_{22}} & \frac{s_{26}}{E_{22}} \\
\frac{s_{16}}{E_{16}} & \frac{s_{26}}{E_{26}} & \frac{s_{66}}{E_{66}}
\end{pmatrix}
\]

Table 1 shows the independent elastic compliances of the orthotropic solid for a thin plate in plane stress.

Rearranging Equation 4:

\[
G_{1c} = \frac{K_{1c}}{K_{1c}} = \frac{1}{E_{\text{eff}}} = \left[ \frac{S_{11} S_{22}}{2} \right]^{-1} \left( \frac{S_{22}}{S_{11}} + \frac{2S_{12} + S_{66}}{2S_{11}} \right)^{-1} \quad \text{(5)}
\]

The reciprocal of the term on the right hand side of this equation can be considered as the “effective modulus” of the orthotropic solid for crack propagation in the \( x_1 \) \( x_3 \) plane. The \( S_{ij} \)

*In linear elastic fracture mechanics terminology \( \gamma_I = G_{1c}/2 \): Equation 1 refers, of course, to a state of plane stress.