concentrations, simultaneously with α' and traces of β' phases, but only when copper pistons were used. At two other concentrations α' and β' were present.

As regards efficiency of quenching our experiments allow us to range the four splat-cooling devices in the following order:

(1) "Gun" technique (9.5×10^7 K sec^{-1})

(2) "Mill" technique (5.0×10^7 K sec^{-1})

(3) "Two-piston" technique (5.0×10^7 K sec^{-1})

(4) "Levitation" technique with copper pistons (1.3×10^7 K sec^{-1})

(5) "Levitation" technique with hard-metal pistons (3.1×10^6 K sec^{-1}). In brackets are given the rates of cooling calculated using approximations similar to those used by Blétry [12].

In comparing the efficiency of quenching we should not forget that in special cases certain methods of splat-quenching are more convenient, e.g. the levitation method for obtaining samples of high purity; hard-metal pistons for quenching alloys with a high melting point, and the "gun" method produces samples which are very suitable for transmission electron microscopy investigation. In all these cases quenching efficiency may not be the most important quality of the device.

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Failure analysis of unidirectional glass-reinforced brittle matrix composites by the fault tree technique

In a recent paper by Masters *et al.* [1], a fault tree technique was used in a qualitative way for the identification of failure mechanisms occurring in laminated composite materials. That analysis reflects the fact that in composites there is a wide variety of overlapping fracture micromechanisms whose individual contributions are often shielded by the complexity of the general fracture process. The fault tree technique is applied in order to divide the general fracture process into chains of basic independent events to which a related probability of occurrence is theoretically assigned. This analysis is a bridge between a macroscopic physical event (the failure of the material) and its microscopic origins, and is used qualitatively to examine the static tensile failure of a fibrous composite laminate.

The present paper proposes a quantitative fault tree for failure of unidirectional glass fibre-reinforced brittle matrix composites, and it is a particular case of the general fault tree presented in [1]. The fault tree proposed here assigns a theoretical expression for the energy absorbed at each step of the chains of basic events, and an expression for the relative probability of occurrence of the two chains. Moreover, two fundamental parameters are underlined in our analysis: the interface strength (τ_i); and the fibre condition, that is, whether or not weak points are present in the fibre. Piggott, in a previous work [2], has

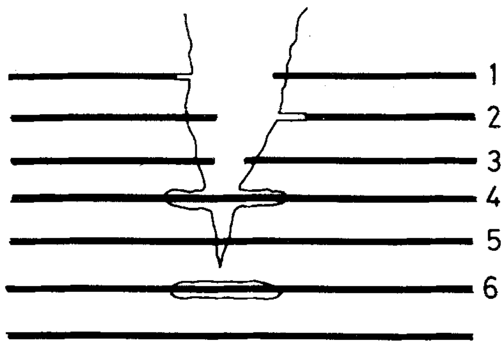

Figure 1 Crack propagation in a unidirectional fibrous composite.

already pointed out the importance of these two factors.

In the present analysis illustrated by Figs. 1 and 2, failure is assumed to occur first in the brittle matrix and to propagate toward the fibres. When a fibre is encountered by the crack front, two mechanisms may occur. In the case of a strong interface, the crack front propagates directly across the fibre, resulting in brittle composite failure (chain I). In the case of a weak interface, debonding at the interface may take place even ahead of the crack front (see fibres 4 and 6 in Fig. 1), leading to a debonding energy contribution calculated by Outwater and Murphy [3]. The debonded fibres may break preferentially in the crack plane for perfect fibres [2] (chain II), or away from it according to the flaw distribution in imperfect fibres. There, the broken fibres are reloaded [4] before they are pulled-out of the matrix [5] (chain III).

The fault tree (Fig. 2) describes the fracture of an ideal composite. In a real composite, it is possible that some or all of these processes occur simultaneously, so that the total work of fracture is given by a sum of two contributions, assuming that the interface strength is weak:

$$\gamma_{tot} = \gamma_{tot}^{III} f(n) + \gamma_{tot}^{II}\,[1 - f(n)], \quad (1)$$

where $f(n)$ is the percentage of flawed fibres, among n fibres. Substituting the expressions for γ_{tot}^{III} and γ_{tot}^{II} (from Fig. 2) into Equation 1, and rearranging, we get for a real composite the total work of fracture:

$$\gamma_{tot} = \gamma_m V_m + \gamma_f V_f + \gamma_i \frac{l_c}{r_f} V_f + \gamma_{deb} + f(n)\,(\gamma_{po} + \gamma_{rel} - \gamma_{deb}). \quad (2)$$

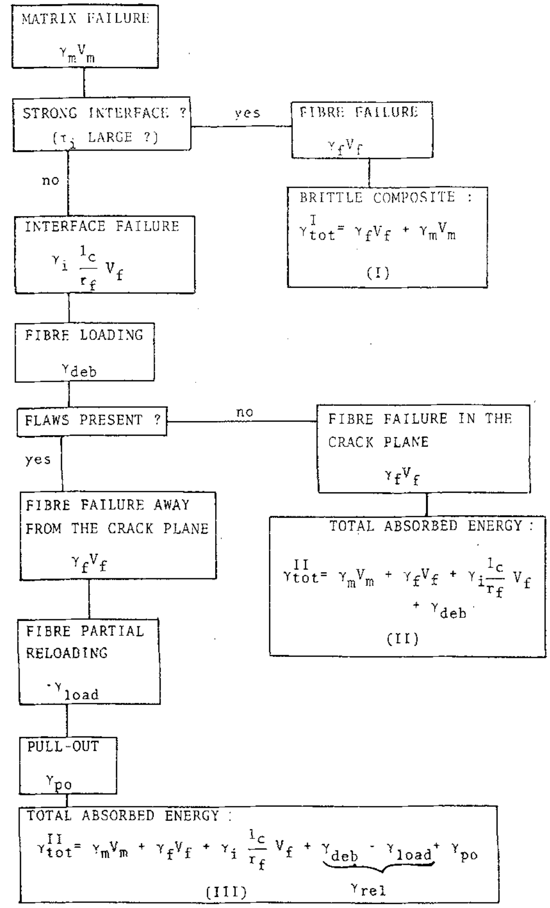

Figure 2 Fault tree for failure of a fibre-reinforced brittle matrix composite.

The coefficients V_f, V_m and $(l_c/r_f)\,V_f$ of γ_f, γ_m and γ_i, respectively, are obtained in each case by considering an energy (γ) times the new corresponding surface area, divided by twice the area of the general crack face, that is:

$$\gamma_f \frac{2\pi r_f^2}{2A}\frac{AV_f}{\pi r_f^2} \qquad \text{for } n \text{ fibres}$$

$$\gamma_m \frac{2(A - n\pi r_f^2)}{2A} \qquad \text{for the matrix,}$$

and

$$\gamma_i \frac{2\pi r_f l_c}{2A}\frac{AV_f}{\pi r_f^2} \qquad \text{for the interface,}$$

assuming that the debonded length of the fibres is l_c.

The proposed fault tree exhibits an interesting feature: when the interface is strong, the composite behaves in a brittle way and the work of fracture, γ_{tot}, is given by a rule of mixtures (RoM). Moreover, the results of Marston *et al.* ([6], Equations 10 and A9) appear to be a particular case of Equation 2 when $f(n) = 1$. When $f(n) = 0$, the results of Marston *et al.* are no longer applicable. Since in a real composite $f(n)$ varies between 0 and 1, their equation is only an approximation to the real work of fracture.

Another outcome of Equation 2 is that unflawed fibres yield lower work of fracture than flawed fibres, which is a commonly accepted result [7].

In a recent experimental work carried out in our laboratory [8], work of fracture values of 38.8 and $62.7\,\text{kJ}\,\text{m}^{-2}$ for $V_f = 0.37$ and 0.60, respectively, were found (testing conditions: three-point bending, room temperature, cross-head speed $0.05\,\text{cm}\,\text{min}^{-1}$, $5 \times 0.5 \times 0.5\,\text{cm}^3$ specimens). These results together with the data

$$\gamma_f = 5\,\text{J}\,\text{m}^{-2}, \gamma_m = 100\,\text{J}\,\text{m}^{-2}, \gamma_i \simeq \gamma_m,$$

$$E_f = 7.1 \times 10^{10}\,\text{Pa},$$

$$\sigma_f = 1.5 \times 10^9\,\text{Pa}, \text{from [8]},$$

and

$$r_f = 7 \times 10^{-6}\,\text{m}, l_c = 2.3 \times 10^{-3}\,\text{m}, \text{from [9]},$$

produce $f(n) \simeq 0.40$ for the two volume fractions considered. This means that approximately half the fibres are flawed and undergo pull-out. $f(n)$ can be evaluated by microscopy; alternatively, it can be related to the probability of having a flawed fibre in the vicinity of the crack plane. This probability itself can be related to other statistical characterizations of strength, as Weibull's distribution for instance, for which the probability that a fibre fails in the range from 0 to σ is calculated.

In summary, a fault tree analysis, which focuses on the fibre condition and on the interfacial strength, is presented for brittle matrix composites. The elements of the fault tree, i.e. the chains of basic events, are described qualitatively and quantitatively as a particular case of a general fault tree presented elsewhere. The basic events are the micromechanisms of energy absorption, and it is shown that brittle matrix composites may fail either in a completely brittle way (which is illustrated by an RoM) or in a much more energy dissipative way, where the term characterizing the deviation from the RoM is important.

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