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Effect of strain rate on the mechanical properties of composites with a weak fibre/matrix interface

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Modelling studies have indicated a strong effect of the rate of deformation on the tensile strength of composites with a weak fibre/matrix interface. At high rates, the mode of deformation changes from a fibre pull-out to a fibre breaking mechanism typical of good adhesion composites. As a result, the mechanical properties become independent of those of the fibre/matrix interface. The model predictions are of great importance because they allow a straightforward identification of composites with poor fibre–matrix adhesion.

1. Introduction

In previous work [1], a stochastic model was developed based on Eyring's chemical activation rate theory [2] for the description of the mechanical properties of discontinuous fibre-reinforced composites. Within the framework of that model, both the fibre and the matrix are allowed to break during failure of the composite. The work was applied to a detailed study of the dependence of tensile strength on fibre aspect ratio, volume fraction and quality of adhesion at the fibre/matrix interface.

Recently, particular attention has been paid in the literature [3–6] to the effects of strain rate on damage accumulation in fibre-reinforced composites. These experimental studies have revealed a strong dependence of the mode of fracture on deformation rate in composites with weak fibre/matrix interface. In view of the obvious practical implications of those findings, the previously derived model [1] has been extended to a detailed investigation of their origin.

2. Model

The model has been described at length previously [1] and will be briefly summarized here. The composite is represented by a three-dimensional (x–y–z) lattice of sites which are linked by bonds having different elastic constants for the matrix and the fibre. The lattice is of the simple cubic type and typically comprises 200 nodes along the y-axis and 35 nodes in the transverse x- and z-directions. The Young's and shear moduli of the matrix and of the fibre are denoted by E_m, G_m and E_f, G_f, respectively. The fibres are oriented along the y-axis and have an aspect ratio, l/d, where l and d are length and diameter, respectively.

Adhesion between fibre and matrix was varied by decreasing the matrix shear modulus at the fibre–matrix interface. Thus, in the case of perfect adhesion, the matrix shear modulus at the interface equals G_m. For poor adhesion, characterized here by an adhesion factor equal to 0.1, that modulus equals $0.1 \times G_m$.

The lattice described above is strained along the y-axis at a constant rate of deformation and temperature, T. In the course of that process, bonds are broken according to the kinetic theory of fracture [2], i.e. at a rate

$$v = \tau \exp[(-U + \beta\sigma)/kT] \quad (1)$$

in which τ is the thermal vibration frequency ($\sim 10^{12}\ \mathrm{s}^{-1}$) and U is the activation energy. The latter is decreased by a linear function of the local stress, σ, with a proportionality factor, β, equal to an activation volume (for more details, see [7]). Values of U and β for the fibre and matrix were given previously [1]. For a temperature of 23 °C and a rate of deformation equal to $1\ \mathrm{min}^{-1}$, these values lead to tensile strength values of 5.6 and 0.56 GPa for the fibre and matrix, respectively [1]. The model is executed with the help of a stochastic Monte-Carlo process which visits the bonds at random and breaks a bond according to a probability

$$p = v/v_{max} \quad (2)$$

in which v_{max} denotes the rate of breakage of the mostly strained bond in the lattice. After each visit of a bond, the time t is incremented by $1/[v_{max}\, n(t)]$ in which $n(t)$ denotes the total number of unbroken bonds in the whole network (fibre and matrix) at time t. After a small time interval, δt, has elapsed, the bond breaking process is stopped and the network is strained by a small constant amount $\delta\varepsilon$ which is determined by the rate of strain $\dot{\varepsilon} = \delta\varepsilon/\delta t$. The lattice is then relaxed to its minimum energy configuration and the Monte-Carlo process of bond breakings is repeated for another time interval, δt and so on, until the sample breaks.

Note that the present work deals exclusively with the effect of strain rate, $\dot{\varepsilon}$, on fracture processes initiated by the external stress (Equation 1). The study in its present form does not consider possible effects of $\dot{\varepsilon}$ on the relaxation rates and values of the model parameters.

3. Results and discussion

Fig. 1 shows the effect of strain rate, $\dot{\varepsilon}$, on the stress–strain curves for the case of perfect adhesion between the fibre and matrix. The two sets of curves are for different values of fibre aspect ratio ($l/d = 1$ and 20). In both sets, an increase in rate $\dot{\varepsilon}$ leads to an increase in strain and stress values at composite failure. This is due to the fact that, the higher the rate, the shorter the time interval a stressed bond spends at a certain stress level and the smaller the probability for bond breaking at that level. Our observations in Fig. 1 are similar to those obtained from application of a similar model to perfectly ordered and oriented polymer filaments [7]. The initial modulus is found to be independent of $\dot{\varepsilon}$ [7] because our approach, in its present form, is limited to a description of the effects of $\dot{\varepsilon}$ on fracture processes alone. Note also that, as observed previously [1], the stress–strain curves at high $l/d = 20$ show two distinct regions. Our present work thus indicates that this feature is characteristic of high aspect ratio curves, irrespective of the strain rate. In all cases, the transition between those two regions is found to coincide with tensile failure of the matrix at fibre ends [1].

Figs 2 and 3 summarize our results for the dependence of tensile strength on fibre volume fraction, v_f, at three different rates: $\dot{\varepsilon} = 10^3$, 10^2 and 1 min^{-1}. The two sets of dotted lines (a and b), extending across the graph, represent limiting behaviours expected for infinitely long fibres [1]. Lines within each set are for the three different rates being studied, the highest rate ($\dot{\varepsilon} = 10^3\ \text{min}^{-1}$) corresponding to the highest slope. At very small v_f, the strength, σ, is expected to be controlled by the matrix and to decrease as $\sigma = \sigma_m (1 - v_f)$ (lines b), in which σ_m is the strength of the pure matrix at the corresponding rate. At higher v_f, a crossover is expected to a fibre dominated regime $\sigma = \sigma_f v_f + \sigma_m^* (1 - v_f)$ (lines a) where σ_f is the strength of the pure fibre at the appropriate rate $\dot{\varepsilon}$, whereas σ_m^* is the stress on the matrix at the breaking strain of the fibre. Inspection of the figures shows that our data for finite l/d ($l/d = 1$, Fig. 2 and $l/d = 20$, Fig. 3) also seem to follow two different regimes, as previously observed [1]. At low v_f, our strength values follow the set of lines b, therefore indicating that composite fracture is matrix dominated. At a critical v_f whose value decreases with larger l/d, the data increase with v_f and follow a regime that is fibre dominated (lines a). The results of Figs 2 and 3 which concern the rate dependence of our data, indicate that

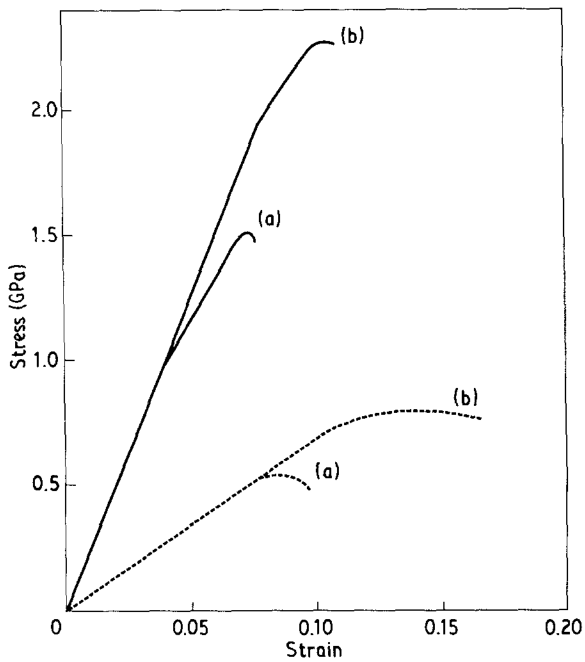

Figure 1 Effect of strain rate, $\dot{\varepsilon}$, on stress–strain curves: (a) $\dot{\varepsilon} = 1\ \text{min}^{-1}$; (b) $\dot{\varepsilon} = 10^3\ \text{min}^{-1}$. The two sets of curves are for different aspects ratios: l/d = (- - -) 1, and (——) 20. The figure is for the case of good adhesion between fibre and matrix (adhesion factor = 1) with $v_f = 0.44$.

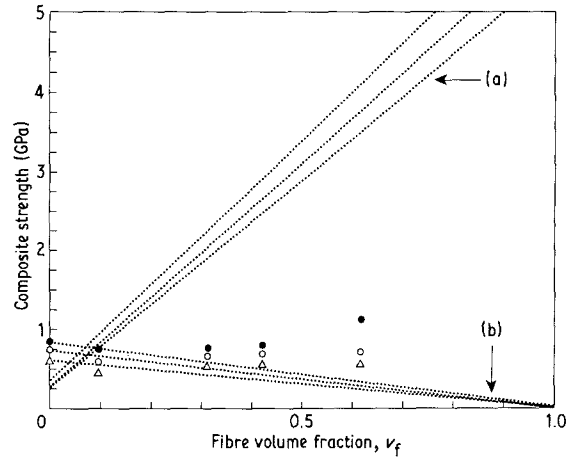

Figure 2 Dependence of tensile strength on fibre volume fraction, v_f, for $l/d = 1$ at different rates: $\dot{\varepsilon}$ = (●) $10^3\ \text{min}^{-1}$, (○) $10^2\ \text{min}^{-1}$, (△) 1 min^{-1}. The two sets of dotted lines a and b extending across the graph represent the following equations: (a) $\sigma = \sigma_f v_f + \sigma_m^* (1 - v_f)$; (b) $\sigma = \sigma_m(1 - v_f)$, see text. Lines within each set are for the three different rates being studied, the highest rate ($\dot{\varepsilon} = 10^3\ \text{min}^{-1}$) corresponding to the highest slope.

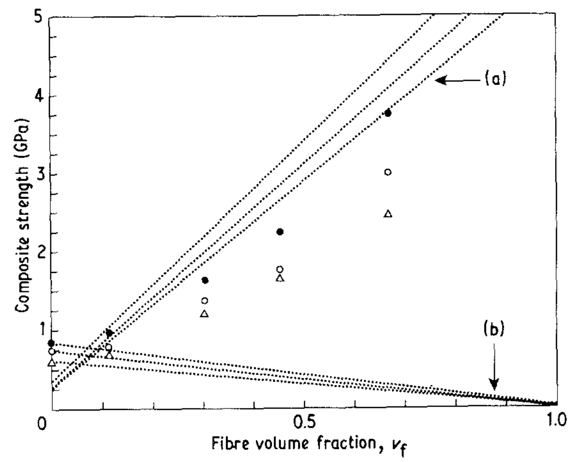

Figure 3 As Fig. 2 but, for $l/d = 20$.

the increase in strength with $\dot{\varepsilon}$ at constant v_f is comparable to that found in the two limiting regimes, a and b. In other words, the rate dependence of composite strength can be entirely attributed to that of σ_f and σ_m for the pure fibre and matrix.

All the results of Figs 1–3 were for the case of perfect adhesion between the fibre and matrix. We now turn to the case of an adhesion factor equal to 0.1 for which the matrix shear modulus at the interface with the fibre is only 10% of G_m. The rate dependence of the stress–strain curves for those composites is given in Fig. 4 for $v_f = 0.45$. The figure is qualitatively similar to Fig. 1 obtained for good adhesion, except for the presence of an abnormally large increase in strength and elongation at break at $\dot{\varepsilon} = 10^3\ \text{min}^{-1}$ for $l/d = 20$ (——, Curve b). That observation constitutes the crux of the present study and will be elaborated upon in the remainder of this paper.

Figs 5 and 6 show the rate dependence of composite strength at different fibre volume fractions, v_f. The results are again similar to those obtained previously (Figs 2 and 3) except for the exceptionally large strength values observed for $l/d = 20$ at high rates $\dot{\varepsilon} = 10^3\ \text{min}^{-1}$. Those values are comparable to those obtained for good adhesion (Fig. 3) and they cannot be explained solely on the basis of an increase in σ_f and σ_m with rate.

These abnormally high strength values are further studied in Fig. 7 for a constant $v_f = 0.45$. The figure describes the strain dependence of the fraction of broken matrix bonds at two different rates ($\dot{\varepsilon} = 1$ and $10^3\ \text{min}^{-1}$). At low rates, the matrix starts to fail in tension near 2% strain (see continuous curve on left). Further investigation indicates that those early matrix breaks occur near fibre ends. Once all the ends have debonded from the matrix, the curve levels off and further straining leads to high shear stress concentration at the fibre/matrix interface near fibre ends. Near 6% strain, the matrix then starts to fail in shear and the fibres progressively debond starting from their ends. A typical illustration of that mode of deformation is shown in Fig. 8a. Observations similar to those reported here have been made by Yuan *et al.* [5] for the case of short glass fibre-reinforced PVC composites with poor adhesion.

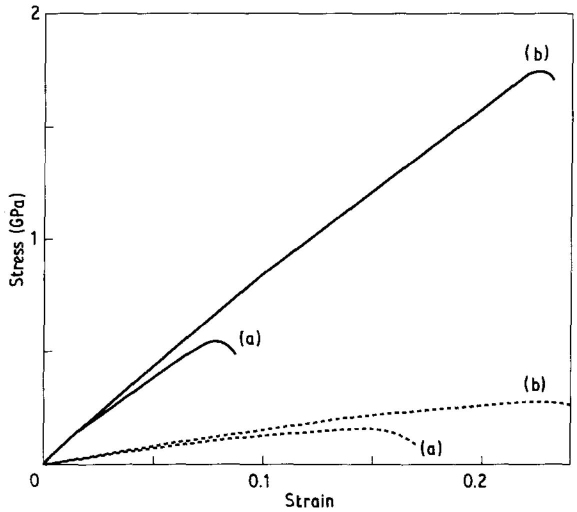

Figure 4 Effect of strain rate, $\dot{\varepsilon}$, on stress–strain curves: (a) $\dot{\varepsilon} = 1\ \text{min}^{-1}$, (b) $\dot{\varepsilon} = 10^3\ \text{min}^{-1}$, for the case of poor adhesion between fibre and matrix (adhesion factor = 0.1) with $v_f = 0.44$. Notation is the same as for Fig. 1.

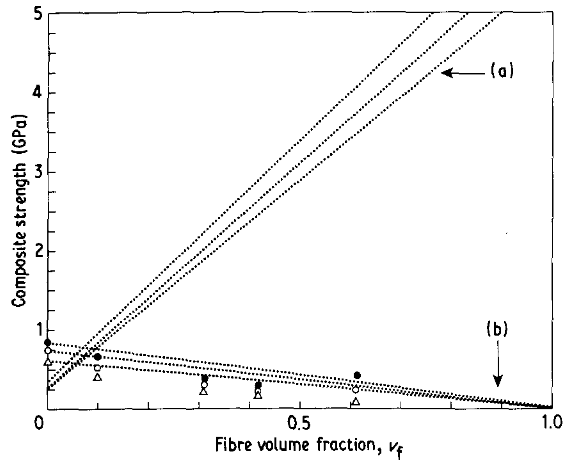

Figure 5 Dependence of tensile strength on fibre volume fraction, v_f, in the case of poor adhesion for $l/d = 1$. Notation is the same as for Fig. 2.

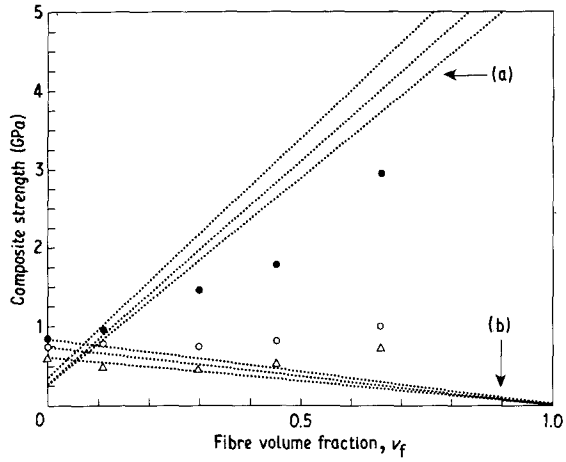

Figure 6 Same as for Fig. 5 but, for $l/d = 20$.

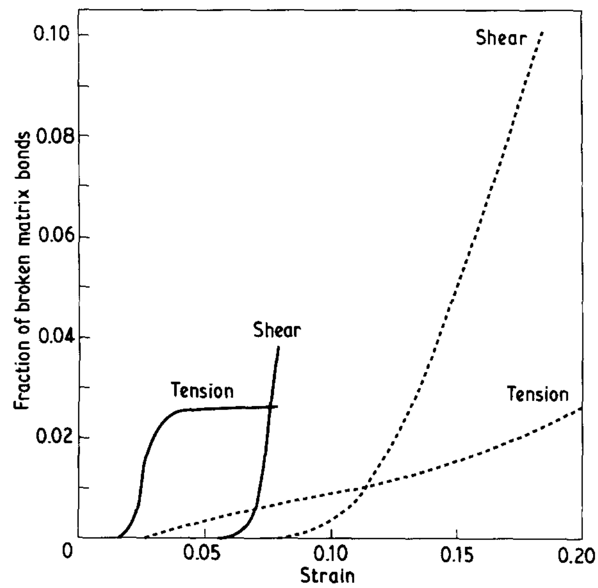

Figure 7 Dependence on strain of the fraction of broken matrix bonds for the case of poor adhesion at two different rates ($\dot{\varepsilon}$ = (——) 1, and (–––) $10^3\ \text{min}^{-1}$). The figure is for $l/d = 20$ and $v_f = 0.45$.

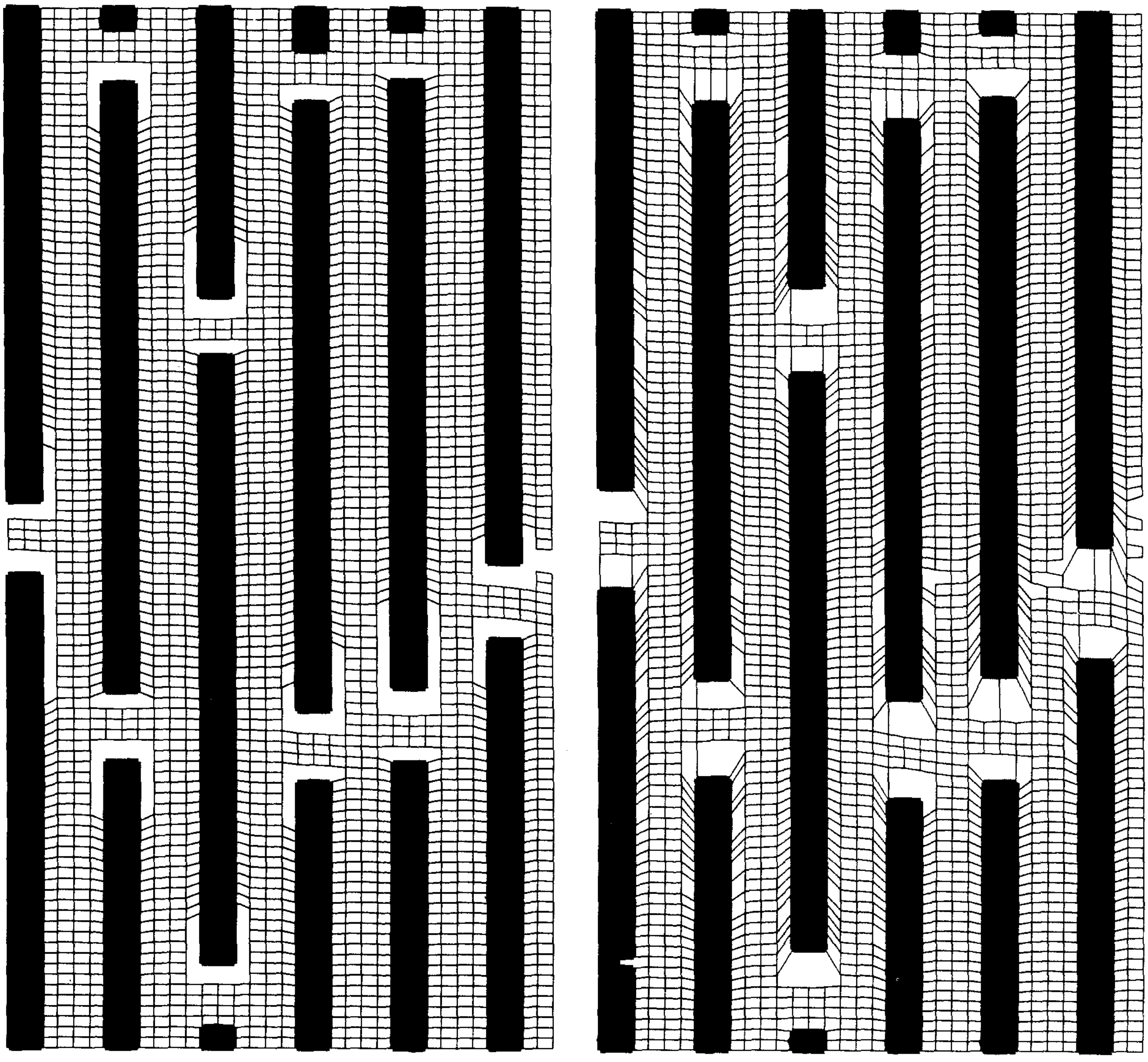

Figure 8 Typical modes of deformation at two different rates, (a) $\dot{\varepsilon} = 1\ \text{min}^{-1}$, (b) $\dot{\varepsilon} = 10^3\ \text{min}^{-1}$, for the case of poor adhesion. The parameter values are the same as in Fig. 7. The figure is for a longitudinal x–y plane passing through the centre of the lattice.

A quite different mode of deformation is obtained at high rates (see Fig. 7). For slow rates, the matrix starts to fail in tension at fibre ends near 2.5% strain. However, because the rate of deformation is high, the probability for matrix bond breaking at any stress level is quite low. Near 7% strain, only one-fifth of the matrix bonds near fibre ends have failed in tension (compare with the curve at low rate for which all fibre ends have debonded). As a result, the shear stress at the weak fibre/matrix interface is more uniform along the fibre with no large concentration being observed near fibre ends. Thus, when the fibre/matrix interface starts to fail in shear near 8% strain, those bond breaks occur more randomly along the fibre and no unzipping of the fibres from their ends is observed. This is clearly illustrated in Fig. 8b for which the number of broken matrix bonds at the fibre/matrix interface is the same as in Fig. 8a. Thus, fibre pull-out does not occur at high rates and composite properties become independent of those of the weak fibre/matrix interface. The composite material can now easily sustain high stress levels that are comparable to those obtained for good adhesion between fibre and matrix (compare Figs 3 and 6 for $\dot{\varepsilon} = 10^3\ \text{min}^{-1}$). Composite fracture occurs at a high 21% strain value and is associated with a substantial amount of fibre fracture (see also broken fibre, top right of Fig. 8b). The present observations are in complete agreement with the experimental studies of Kander and Siegmann [3].

To conclude, the modelling studies presented here indicate a strong effect of the rate of deformation on the tensile strength of composites with a weak fibre/matrix interface. At high rates, the mode of deformation changes from a fibre pull-out to a fibre breaking mechanism typical of good adhesion composites. As a result, the mechanical properties become independent of those of the fibre/matrix interface.

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