MODELLING THE FRACTURE BEHAVIOUR OF CMCs

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A modified Rosen's model has been developed to predict the ultimate tensile load and fibre pullouts in CMC materials. The model assumes a Weibull distribution for the fibre strength and that the matrix is already cracked at the saturation density. The solid is discretised in fibre segments between matrix cracks. After fracture of one fibre segment, the load is redistributed among the intact neighbouring segments. After fibre load redistribution, the loads on the fibre segments are compared with the strength assigned to the individual fibre segments. New load redistributions are performed as required until no intact segment is loaded above its nominal strength. Then, a fracture path is explored through fractured or deactivated segments. The distance between the fracture path and the closest broken segment in the fibres provides the prediction of fibre pullout lengths.

INTRODUCTION

Ceramic materials have very interesting properties from a structural point of view but they are too brittle for most structural applications. Fibre reinforcements are used to increase their toughness. The introduction of fibres and interfaces between fibres and matrix allows new mechanisms of energy absorption to take place, namely: multiple matrix cracking, debonding of the fibre/matrix interface, and pullout of the reinforcing fibres.

The interest for micro-mechanically based models, able to predict the macroscopic behaviour of a composite, is obvious since the system of trial and error in developing ceramic matrix composite is expensive and afterwards testing is required. These models are not intended for engineering design, this is done by macroscopic models, see for example (1), but their results can be used as input for the behavioural models.

There are lots of analytical models (2-11). They necessarily work on very simple assumptions such as global load transfer, weak adhesion between fibres and matrix,... Also, most of them are concerned about the ultimate tensile stresses (8-11). This work presents a

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Simple micro-mechanical statistical model. A micro-mechanical model should be able to predict not only ultimate tensile stresses but also how the fracture morphology should be. One of the most characteristic features of the fracture of composite materials is the fibre pullout that can be observed on their fracture surfaces.

DESCRIPTION OF THE MODEL

The model describes the tensile behaviour of a composite material with a brittle matrix reinforced by fibres aligned with the loading direction. It will be assumed that the matrix is already cracked at its saturation density. The matrix crack spacing at saturation can be measured or derived (see, for example (2)). After saturation, all the load at the crack plane is carried by the fibres which are bridging the cracks in the matrix. The model also assumes a weak interface where the debonding energy is negligible small. The solid is divided into volume elements whose lengths are equal to the distance between two consecutive matrix cracks. Each volume element is discretised into a fixed number of fibre segments. For a description of related Rosen's models, see, for example, reference (12). The mechanical strength of the fibre segments is described by a Weibull distribution,

$$F = 1 - \exp\left[-\frac{l}{l_0}\left(\frac{\sigma}{\sigma_0}\right)^m \right] \tag{1}$$

where F is the cumulative failure probability of a fibre segment of length l subject to an axial stress σ. The exponent m, σ_0 and l_0 are parameters in the Weibull distribution function: the Weibull modulus, reference stress and length, respectively. The strength of the i-th fibre segment, σ_i, is assigned according to:

$$\sigma_i = \sigma_0 \left[-\frac{l}{l_0} \ln(1 - R_i) \right]^{1/m} \tag{2}$$

where l is the length of the fibre segment and R_i is a random number uniformly distributed between 0 and 1.

The load applied on the solid, once discretised in elements, is increased until producing the fracture of the weakest fibre segment. Once the fibre segment is broken, the load that this fibre was carrying through the matrix crack is transferred to the intact neighbouring fibres at the same matrix crack plane. The load transfer can be computed according to a global load share assumption or assuming some local load transfer: for example, all the load will be carried by the two closest intact fibres. If the model is applied to a single ply of fibres, the closest intact fibres to the left and to the right of the broken fibre will support one half of the load that was carried by the broken fibre. In the case of a regular disposition of fibres, in three dimensions, the six closest fibres will share the load drop in the broken fibre: each one with about 17% of it. It is possible to compute the load transferred to the intact fibres in a more accurate way, see (13, 14). Anyway, the present model actually considers the two extreme behaviours: global load transfer and highly localised.

If the fibre segments above and below the broken segment are within the matrix/fibre transfer length measured from the broken segment, they will also see their axial loads reduced, see Fig. 1. The matrix/fibre transfer length, d, is calculated according to a constant frictional stress model:

$$d_i = \frac{r\sigma_i}{2\tau} \qquad (3)$$

where r is the fibre radius, σ_i is the axial stress at the broken fibre segment and τ is the interfacial frictional sliding stress (assumed constant). Somehow, these segments where a load reduction occurs are immunised against fracture: they already supported a larger load without failure, now they only support the load that can be carried by friction from the matrix-embedded fibre end, thus it is improbable that they will fail. These segments are consequently deactivated as future candidates to fibre fractures. In particular, it should be noticed that with further increments of the load, the extension of the deactivated region will also increase, coupled with σ_i. In a rigorous model, the load drop in the segments within the transfer length should also be transferred to their intact neighbours at their respective matrix crack planes. The results that will be presented do not take into account this side effect.

After redistribution, the new loads on the intact fibre segments are checked against their assigned strengths. If the assigned strength is exceeded in any fibre segment, this segment is assumed to be broken and new load transfers are calculated. Once the load has been transferred to the other intact fibre segments, the model seeks for a crack path through broken or deactivated fibre segments from one side to the other of the test-piece. Deactivated fibre segments are also assumed to constitute a possible path for the fracture process because the fibres can be extracted from there to the nearest broken segment. A percolation algorithm is used to decide when the test-piece is broken and which is the fracture path. Depending on the size of the loading steps in the incremental procedure, one or several percolation paths can be found. The model selects locally the fracture path with less changes in the fracture plane, so that less volume of matrix has to be broken linking the different matrix cracks. Usually, a local minimum is enough to decide a unique fracture path. If no fracture path is found, the load is further increased, then increasing proportionally the load on the intact fibre segments. This will produce new fibre fractures and eventually a fracture path will be available.

RESULTS AND CONCLUSIONS

The described model has been used to simulate the fracture behaviour of a CAS matrix reinforced with continuous Nicalon™ fibres (β-SiC). The distribution of the strengths for Nicalon fibres was measured by Beyerlee et al. (15) on a SiC_f/CAS by measuring the fibre fracture mirrors on the fracture surface. The mean fracture strength was σ_0 = 2,000 MPa and the Weibull modulus is m = 3.6. The mean pullout length was 0.3 mm. Thus, in average, one fracture is found for 0.6 mm long fibre pieces. Therefore, for a fibre segment of length l, subject to an axial stress σ, the cumulative probability of fracture will be given by

$$F = 1 - \exp\left[-\frac{l}{0.6 \text{ mm}} \left(\frac{\sigma}{2{,}000 \text{ MPa}} \right)^{3.6} \right] \tag{4}$$

The mean Nicalon fibre radius is r = 7.5 μm and the frictional sliding stress at the interface, measured by nanoindentation techniques (16, 17), is τ = 9.5 MPa. The matrix crack spacing at saturation l = 97.6 μm was measured on the tensile side of a four point bending test-piece, with a $[(0^{o}/90^{o})_3]_s$ architecture, on the 0° plies (16). This measure matches the observations on tensile test-pieces with a unidirectional reinforcement (103 ± 15 μm) (15).

Figure 2 shows a mesh of 60 fibres bridging 60 matrix cracks. Different colours are used for the different strength levels assigned to the fibre segments. These strengths were randomly distributed according to the known Weibull distribution. Replacing the fibre strength parameters of Eq. (4) into Eq. (2), the strength of the i-th fibre segments is assigned as

$$\sigma_i = 2{,}000 \text{ MPa} \left[-\frac{0.6 \text{ mm}}{0.0976 \text{ mm}} \ln\left(1 - R_i\right) \right]^{1/3.6} \tag{5}$$

Figure 2 shows the axial stresses applied on the different fibre elements, the fibre segments that are broken or unloaded and the prediction for the fracture path.

The minimum distance along a fibre, from the fracture path to the corresponding broken fibre segment, provides the estimation of the fibre pullout length. Figure 3 collects the results obtained under different hypotheses for the load transfer: global load share and localised on closest neighbours, sharing a 50% or 17% of the broken fibre load. Figure 3 also shows the experimental results obtained by Beyerlee et al. (15) for the mean pullout length and the distribution of pullout lengths, as measured in this work, for SiC_f/CAS with a cross-ply architecture $[(0^{o}/90^{o})_3]_s$, at room temperature. The model correctly predicts the mean pullout length for the unidirectional reinforcement (as measured by Beyerlee et al. (15)). The shape of the distribution is reasonable when compared with the experimental measurements, on cross-ply structures, but clearly cross-ply architectures produce longer pullouts. The reason for that is that the initial cracks, in the 90° plies, impose to a great extent the fracture path, therefore the crack path obtained with the present model is no longer a minimum energy path when both plies at 0° and 90° are considered. As a result, longer pullouts are observed in 0° plies when 90° plies are introduced.

Figure 4 shows the model predictions for the ultimate tensile stress under the assumptions of global and local fibre load transfer, and the experimental results obtained by Beyerlee et al. (15) for the unidirectional reinforcement and in this work for the cross-ply architecture. Fibre volume fraction in the 0° direction is V_f = 0.175, as described elsewhere (18) (Beyerlee et al. results (15) were scaled to this fibre volume fraction). The global assumption seems to be rather optimistic. On the contrary, highly localised fibre load transfer provides a lower bound, but close to the experiments for the unidirectional reinforcement. 2D reinforcements seem to transfer load between these two extremes.

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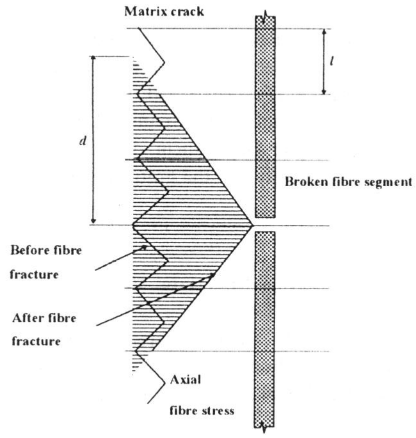


Figure 1. The effect of the fracture of a fibre segment on the adjacent fibre segments.

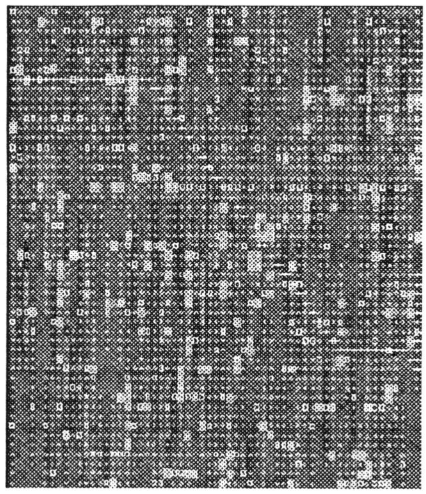
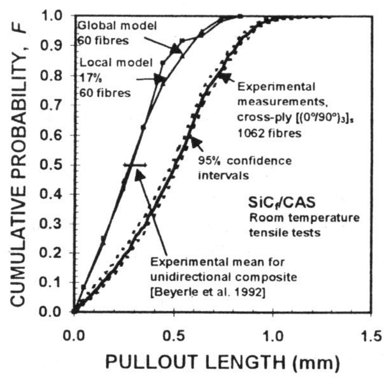


Figure 2. Results of a 2D computer simulation of the tensile behaviour of a SiC_f/CAS.



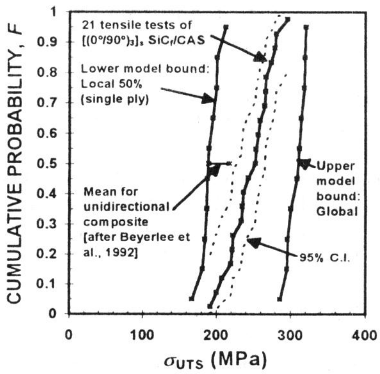


Figure 3. Cumulative distribution of fibre pullout lengths.

Figure 4. Distribution of σ_{UTS}. Fibre volume fraction in the 0° plies is $V_f = 0.175$.