Modelling Fracture in Fibrous Microstructures

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Abstract

This work describes some complementary studies directed towards micromechanical modelling and simulation of the statistical fracture process in composites with fibrous microstructures. A few studies involve combining efficient computational stress analyses and piezospectroscopic measurement techniques to quantify interface deformation around a single break in model composites. It is shown how estimated interface parameters can be used to predict activity around more complex break arrangements in much larger composites. The final studies involve incorporating these experimentally refined stress analyses into large-scale simulation for statistical predictions and subsequent analytical modelling of composite fracture.

Introduction

Predicting the fracture tensile strength of fibrous composites begins with the recognition that failure evolution entails a complex interaction between statistical fiber strength, localized matrix and interfacial deformation, and the micromechanics of stress transfer around randomly located fiber breaks and defects. In this paper, comprehensive computational mechanics and failure models for a composite consisting of continuous fibers (or elongated grains or platelets) are described. Basically a representative computational failure model, capable of simulating randomly occurring and evolving damage, needs two components: (i) quick calculation of the stresses and displacements as they change in time around any arrangement of fiber fractures and (ii) experimental characterization of the microstructural and constituent properties, both statistical and deterministic. Its ultimate goal is to aid in high reliability design, that is, minimizing failure probabilities (e.g. 1 failure out of 10^9) for a given lifetime (e.g. 30 years) with knowledge of material microstructure and phase properties.

Computational Mechanics Techniques

With regards to the first component, these computational composite stress analyses must (i) be multifiber, (ii) be realistic and fast, and (iii) account for dominant matrix and interface failure mechanisms, which may change in time due to matrix creep, fiber break progression, fiber degradation, and fiber-matrix debonding. Our current research involves developing efficient computational mechanics techniques (CMTs) to calculate elastic, inelastic, and time-dependent stresses around multiple

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fiber breaks. To date, these techniques are built upon the Hedgepeth (1961) shear-lag model for 2-D and 3-D multifiber composites, assuming that the fibers carry all the tensile load, and the matrix only transmits shear between the fibers. In these CMTs, computation time is tied to the amount of damage and not the composite volume, and therefore, can potentially be 1 to 2 orders of magnitude more powerful and practical beyond current state of the art.

From previous work (Beyerlein et al. 1996-8), the multifiber, shear-lag model shows promise in modelling the essential physics of random break propagation, accompanied by matrix plasticity, creep, and debonding. For instance, using an elastic CMT, called break influence superposition (BIS), Beyerlein et al. (1996) demonstrated that for a crack consisting of over 50 breaks, the fiber tensile stresses calculated from BIS achieved excellent agreement with Mode I fracture mechanics solutions for an orthotropic material down to the scale of one fiber diameter. Another key development is the quadratic influence superposition (QIS) technique (Beyerlein and Phoenix 1996). QIS allows for localized matrix yielding, interfacial debonding and frictional sliding around multiple, arbitrarily located fiber breaks.

**Experimental validation**

With regards to the second component, two recent developments, Micro-Raman spectroscopy (MRS) and fluorescence piezospectroscopy, have greatly improved micromechanical measurement of fiber and interface stresses around breaks. With 2 μm spatial resolution, these techniques are capable of mapping fiber axial strain profiles produced by CTE mismatch and breaks in multifiber composites. The latter can measure the distribution of axial stresses in alumina-containing materials (He and Clarke), and MRS can measure the distribution of strains in carbon and aramid fibers embedded in a transparent matrix (Wagner et al.). In two recent studies (He et al., Beyerlein et al. 1998b), these two measurement techniques were combined with QIS to develop a methodology for determining in situ interfacial strength parameters, such as the yield stress, \( \tau_y \), and frictional sliding resistance, \( \tau_d \), in Fig. 1.

In Fig. 2, QIS was used to interpret fluorescence tensile stress measurements along isolated broken alumina fibers in an Al matrix-fiber composite. This analysis led to estimates for the in situ yield stress, \( \tau_y = 98.5 \) MPa, associated yield

![Fig. 1 Elastic-plastic-debond matrix-interface constitutive relationship in shear.](image)

![Fig. 2 Stress profiles of a broken alumina fiber under a far field fiber stress of 875 MPa mapped by the piezospectroscopic technique in He et al. and by QIS.](image)
Fig. 3 Comparison of the QIS predicted and MRS mapped normalized strain profiles along a broken graphite fiber and within a multifiber composite.

Fig. 4 Comparison of QIS predicted and MRS mapped normalized strain profiles along a sized graphite fiber containing three breaks and $\epsilon = 0.79\%$. Dashed and dashed-dotted profiles are QIS predictions for $\epsilon = 0.33\%$ and 98\%, respectively.

zone lengths (extending on the order of fiber diameters), and fiber and matrix stresses and displacements. Similar Fig. 3 compares the fiber tensile strain along a graphite fiber containing one break mapped by MRS and QIS. Observations suggested that the asymmetric response of the interface above the break was due to defects in the plasma fiber coating. As indicated, we estimated an $\text{in situ} \tau_\text{f}$ of 40±1 MPa. Once the limit debond shear strain, $\gamma_\text{d}$, was reached, interfacial sliding was resisted by frictional shear $\tau_\text{s}$, estimated at 10±0.2 MPa, over the debond zone. Important results were increasing $\tau_\text{f}$ with decreasing interfiber spacing and concomitant decreases in peak stress concentration factor, yet increases in overload lengths.

One strength of QIS is that these parameters can be used to calculate the interfacial damage progression from any number and arrangement of fiber fractures under increasing strain. As a test, a sized fiber, containing 3 close breaks along its length and spaced 1 fiber diameter from its two neighboring fibers, was analyzed. Based on prior MRS-QIS estimates for sized fibers containing one break, it is assumed that $\tau_\text{f} = 30.00$ MPa. Given applied strain, $\epsilon = 0.79\%$, and spacing between breaks, QIS calculated the extent of yielding emanating from each break and corresponding fiber strain profile. As shown in Fig. 4, QIS predictions show reasonably good agreement with MRS measurements. Also QIS predicted asymmetries in the yield zone lengths growing from each break, naturally resulting from the unequal break spacing.

Computational and probability modelling

Objectives of a probabilistic composite failure model are to (i) relate microstructure to macroscopic failure, particular to temperature and loading and (ii) require constituent data, deterministic or statistical, which can be obtained by experiment. For instance, much experimental evidence shows that the strength of most fibers is well represented by a Weibull distribution, $F(\sigma)=1-\exp\{- (\sigma/\sigma_0)^\gamma\}$, where $F(\sigma)$ is the failure probability at applied stress $\sigma$ or lower; $\gamma$, the shape parameter; and $\sigma_0$, the scale parameter. The latter parameters can be obtained quite rapidly, even for small diameter graphite fibers.

For illustration, recent studies (Beyerlein and Phoenix 1997), which incorporated BIS into Monte Carlo simulation to model the randomly evolving fracture process in a
composite containing Weibull fibers, are described. Composite laminae contained an initial notch of \( N \) broken fibers \((N \leq 51)\) and Weibull fibers with \( \gamma \) ranging from 3 to 30 (from high to low variability in strength). To best isolate the effects of statistical fiber strength and provide insight, it is important to know the 'deterministic' response for the analogous system. For the same \( N \)-notched composite reinforced with single strength \( \sigma_0 \) fibers, the fracture strength is \( \sigma_0/K_N \), where \( K_N = \sqrt{\pi(N+1)/2} \) for \( N > 0 \).

Analytical probability models and useful approximations were developed to study the isolated effects of statistical variation in fiber strength on failure mode, crack growth, and statistical strength distribution. Despite the fact that only planar elastic crack growth was considered, variability in fiber strength was shown to provide toughening by altering the progression of fiber fractures up to instability. Incorporating these features, various probability models for crack growth, including microcracking, small notch growth, and fracture resistance, were developed.

Though computer simulation reduces the need to perform hundreds of costly, time consuming experiments, analytical expressions coupled with fundamental experimental data, is the most practical route. Accordingly, the simulation results guided in the development of expressions for the statistical fracture strength. In particular, for large \( N (N > 20) \), the composite failure probability \( P_N(\sigma) \) with an initial \( N \)-sized notch at applied stress \( \sigma \) or lower was determined to be approximately,

\[
P_N(\sigma) = \exp\left\{ -\frac{2(N+1)}{m} \left( s^*\gamma \exp\left\{ -m(s^*\gamma) \right\} \right) \right\}, \quad 0 \leq s^* < \infty \tag{1}
\]

where \( m = 1, 2 \) correspond to upper and lower bound approximations, respectively, and \( s^* = K_N \sigma_0/\sigma_0 \). From (1), notched composite strength \( \sigma \) at probability level \( P \) is

\[
\sigma/\sigma_0 = \left( \frac{2}{\sqrt{\pi(N+1)}} \right) \left( \frac{1}{m} \ln\left( \frac{N+1}{C} \right) \right)^{1/\gamma} \quad \text{as} \quad N \to \infty \tag{2}
\]

where \( C = (\gamma 2)(-\ln(P)) \). For instance, for the median strength, \( P = 0.5, C = 0.35\gamma \).

The common objectives of all the works presented here were two-fold: to develop methods (i) for quantifying the in situ interface deformation parameters and (ii) providing realistic input and stress analyses for large scale composite deformation and fracture models. With respect to realistic modeling of composites, these CMTs are currently being extended to 3D (Landis and McMeeking) and also to reflect the tensile load carrying capacity of the matrix.

References
