A PHYSICALLY BASED ANALYTICAL TENSILE MODEL FOR CARBON-CARBON COMPOSITES

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The unique thermostructural properties of carbon-carbon composites has afforded an opportunity to achieve significant improvements in reentry vehicle and rocket nozzle performance. The increasing use of carbon-carbon composites has created the need for an analytical model to predict composite properties. Such a model, if it is to be of use to the processor as well as the designer, must be physically based using identified microstructural characteristics with micromechanical principles to predict composite stress-strain behavior. The objective of this effort was directed towards developing an analytical model for composite tensile behavior. The development of such a model combined with experimental verification can provide a means of predicting optimum microstructural features, effect of defects, and failure criteria. Therefore, the model can provide guidance for materials research and development as well as inputs to structural analyses.

The analysis and modeling of carbon-carbon composites requires an understanding of the influence of many composite microstructural variables: filaments, matrices, filament-matrix interactions, construction parameters and processing procedures. Consequently, the initial steps in this study were to identify the microstructural characteristics and to determine the mode and sequence of constituent failures. The microstructural characteristics of interest, filament and matrix orientation and microcracking, have been described in Reference 1. The crack propagation studies were conducted in the scanning electron microscope to identify the microstructural parameters that control crack initiation and propagation (Ref. 2). Previous work on carbon-carbon composites have shown that the crack path was controlled by the highly oriented lamina in the matrix and that fracturing occured not at the fiber-matrix interface but within the matrix (Ref. 3). Model input parameters were obtained from these experimental measurements or observations in order to eliminate the necessity for estimating parameters.

The ability to predict the uniaxial tensile properties of a multidirectional composite by a physically based analytical model reguires the input of unit cell properties. This in turn requires the development of longitudinal and transverse unidirectional property models. A longitudinal model has been developed and reported previously (Ref. 4). This model could predict the experimental stress-strain behavior of carbon-carbon composites manufactured by high and low pressure procedures. In addition, the significant effect of bent filaments on strength was first predicted by the model and then confirmed by crack propagation tests in the SEM.

Through the longitudinal modeling efforts and microstructural analyses, the exceptionally high and difficult-to-explain modulus of composites fabricated by low pressure procedures was attributed to the highly aligned graphitic planes laying parallel to the filaments. A similar problem exists in attempting to explain the extremely low transverse modulus of such unidirectional composites. One mechanism proposed to explain the experimental data was that the majority of the filaments are not bonded to the matrix (Ref. 5). However, analyses of the fracture faces of transverse tension specimens indicated that the majority of the filaments had matrix bonded to them. Consequently, other mechanisms must be sought. One, following along similar lines, is that circumferential microcracks exist in the sheath surrounding the filaments. Another mechanism to be discussed deals with modeling the composite based on the microstructural features of the matrix phase.

The transverse model consists of filament, sheath and bulk matrix containing porosity arranged in series with equal stress in each. On the basis of the photomicrograph and sketch of a transverse view of a unidirectional composite in Figure 1, it is suggested that the sheath can be taken as well aligned graphite planes containing 90-deg kinks across the entire composite cross section. This description, which is idealized in Figure 2, is completely consistent with the longitudinal model already developed.

The sheath modulus was calculated using the equation for a kink along 100% of the sheath length:

$$E_{ts} = E_t \left\{ \frac{3\pi}{4} \left(\frac{R}{K} + 1 + \frac{E_t}{2G_t} \right) \right\}^{-1}$$

where R = kink radius
E_t = single crystal Young's modulus
G_t = single crystal shear modulus
$K = R - T_s / \ln((R + T_s/2)/(R - T_s/2))$
T_s = Sheath thickness

For a Thornel 50 filament with a 6 μm diameter and sheath thickness of 0.5 μm, the calculated modulus was 0.10×10^6 psi. The transverse modulus of a Thornel 50 fiber has been measured at 0.65×10^6 psi (Ref. 5).

A number of approaches exist for calculating the modulus of the bulk matrix which was assumed to be a porous isotropic material. An approach by Greszczuk calculates the average stress and strain in an isotropic elastic body containing a square array of cylindrical pores (Ref. 7). A second approach, by Mackenzie, calculates the shear modulus by applying a homogeneous shear to a large sphere containing a pore (Ref. 8).

Wagner, O'Rourke, Armstrong measured the tensile and shear moduli and Poisson's ratio of an isotropic bulk graphite with a 15 to 31% range in porosity (Ref. 9). The results of this study were compared to Greszczuk and Mackenzie assuming Poisson's ration independent (ν) and dependent ($\nu = Fn(V_p)$) on porosity (Fig. 3). The tensile modulus of a coal tar pitch bulk matrix was obtained from the experimental data of Eitman, Greszczuk, and Jortner (Ref. 5). The tensile modulus of the pitch with 40% porosity was 0.52 x 10^6 psi. Using Mackenzie's equation and assuming a Poisson's ratio of .24, E_o was calculated to be 2.48 x 10^6 psi.

The effective composite transverse modulus (E_e) for representative Thornel 50-pitch composites is shown in Table 1. Porosity was found to have a smaller effect than anticipated. Decreasing the porosity by one-half resulted in only a 13% increase in modulus. These calculated moduli compared favorably to the experimentally measured valves of 0.15 to 0.30 x 10^6 psi. (Ref. 5).

Another mechanism must account for the lower range of these experimentally measured moduli. Micrographs and x-ray radiographs of these composites indicated that extensive macro/microcracking was present. Therefore, the displacements due to the presence of cracks must also be considered. Based upon the work of Tada, Paris, and Irwin (Ref. 10), the displacements associated with a crack of length 2a in a body of width 2b and height 2h was calculated. Assuming the composite is uniform body with average properties, the effective transverse composite modulus was calculated for a composite specimen (0.70 x 0.25 x 0.25 inch) with various a/b ratios (Ref. 11). It is seen from Table 1 that the scatter in the experimental measurements can be predicted by these analyses.

The combination of microstructural analyses to characterize composite structure and crack propagation modes, and micromechanical principles can be used to develop a physically based composite model. The transverse model being developed can predict the experimental modulus of unidirectional composites only when the matrix is taken as a highly aligned sheath containing kinks.

Acknowledgments

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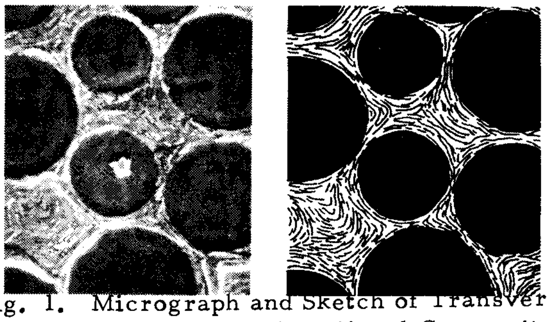

Fig. 1. Micrograph and Sketch of Transverse Section of Unidirectional Composite

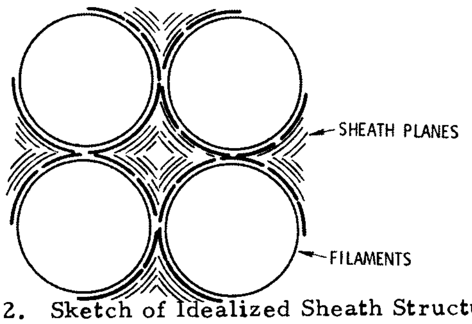

Fig. 2. Sketch of Idealized Sheath Structure

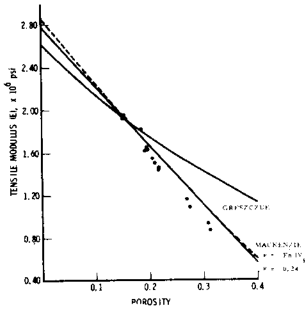

Fig. 3. Effect of Porosity on Tensile Modulus

Table 1. Predicted Transverse Moduli

Porosity	Matrix Modulus (10^6 psi)	Fracture (%)	a/b	Compsoite Modulus (10^6 psi)
15	1.77	3	0	0.20
10	1.99	8	0	0.23
			0.20	0.21
			0.50	0.18
			0.75	0.15
5	2.24	13	0	0.26

Filament Modulus = 0.65 x 10^6 psi
Sheath Modulus = 0.10 x 10^6 psi
Filament Fraction = 60%
Sheath Fracture = 22%