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Constitutive Modeling and Impact Simulation of Random Carbon Fiber Polymer Matrix Composites for Automotive Applications

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ABSTRACT

Damage constitutive models based on micromechanical formulation and a combination of micromechanical and macromechanical effects were developed by the authors to predict progressive damage in aligned and randomly oriented carbon fiber polymer composites. The models are extended in order to account for the microcrack effect on the mechanical behavior of the composites. Progressive interfacial fiber debonding models are considered in accordance with a statistical function to describe the varying probability of fiber debonding. Finally, the complete progressive damage constitutive models are implemented into the finite element code DYNA3D to perform impact simulation of random fiber-reinforced composites for future use in advanced automotive materials. The implemented model is applicable for shell and solid elements in three-dimensional analysis, as well as axisymmetric elements in two-dimensional analysis. In addition, the numerical incorporation allows a prediction of the mechanical response of large composite structures under stress or during impact and eliminates the need for expensive, large-scale experiments.

INTRODUCTION

The goal of automotive engineers today is to provide lightweight, more fuel-efficient automobiles capable of greater crashworthiness. Carbon fiber composites, which are a new breed of high-strength materials, have attracted worldwide attention and hold great promise. The composites are lightweight, but are in general significantly more brittle when compared with other polymer composites. Thus, they are used in composites with a lightweight matrix (e.g., an epoxy resin composite). It is well known that organic matrix, fiber-reinforced composites are very susceptible to impact damage, especially at low velocities. Low-velocity impact can cause significant damage (i.e., delaminations and matrix cracks) inside the composites. Such damage is very difficult to detect and may cause significant reduction in the strength and stiffness of the materials. Lightweight, random carbon fiber polymer matrix composites have the potential to satisfy today's requirements. The applicability of these composites, along with their relatively low cost, makes them a strong candidate for automobile implementation.

Experimental investigations of fiber-reinforced tubes and cones indicate a wide range of material damage such as matrix crushing, delamination, and fiber breakage. The first two types of failure often show a rather slowly progressing failure with high-energy dissipation, while the last failure type might initiate catastrophic collapse of the entire structure with little dissipation of kinetic energy. It is noted that the microscopic failure behavior of the composites is still not completely understood. Nevertheless, several attempts towards the simulation of a composite crash have been made [1-3]. A more detailed review in failure of fiber-reinforced composites can be seen by Matzenmiller and Schweizerhof [2], Kutlu and Chang [4], and Meraghni and Benzeggagh [5].

Micromechanical approaches enable us to evaluate and predict local stress and strain fields in each constituent. Hence, the derivation of the constitutive equations in the form of a phenomenological parameter model from entirely micromechanical considerations is required to perform the rigorous analysis of composite structures. Such an approach is more justified in the case of composite materials reinforced with randomly oriented discontinuous fibers. Indeed, the microstructure of these materials, the complexity of damage mechanisms, and the diversity of their scenario significantly influence their overall properties. Furthermore, because of the natural tendency of the structure to acquire lower energy modes, both material and structural damage processes need to be well understood and be modeled to simulate and eventually design the desirable sustained crush of the component. Therefore, accurate analysis and simulation of the complete response of components and systems of random fiber composites are essential and require accurate micromechanical constitutive models.

While there is substantial research in the aerospace community on graphite-fiber laminated composites, there is very little information pertaining to the response of carbon-fiber composites during automotive impact-induced crash loading conditions. Furthermore, the predictive analytical and numerical tools required to accurately evaluate and design carbon fiber automotive structures for crush do not currently exist. In order to successfully develop these predictive tools, micromechanical damage constitutive models for random fiber-reinforced composites are developed. The complete progressive damage constitutive models are implemented into the finite element code DYNA3D to simulate the impact of random fiber-reinforced composites for future advanced automotive materials. More detailed information for complete progressive damage constitutive models, strain driven algorithms, micromechanical iterative algorithms for the progressive damage models, and three-dimensional return mapping algorithms can be found in our previous research [6].

OVERALL ELASTOPLASTIC BEHAVIOR OF ALIGNED, FIBER COMPOSITES

First, an initially perfectly bonded, three-phase composite consisting of a matrix (phase 0) with bulk modulus k_0 and shear modulus m_0; aligned discontinuous, randomly dispersed, (prolate) spheroidal fibers (phase 1) with bulk modulus k_1 and shear modulus m_1; and aligned penny-shaped microcracks (phase 2) is considered. When spheroidal inclusions (discontinuous fibers and penny-shaped microcracks) are aligned, the composite as a whole is transversely isotropic. Subsequently, as loadings or deformations are applied, some fibers are partially debonded (phase 3). These partially debonded fibers are regarded as equivalent, transversely isotropic inclusions. A partially debonded fiber can be replaced by an equivalent, perfectly bonded fiber that possesses yet unknown transversely isotropic moduli. The transverse isotropy of the equivalent fiber can be determined in such a way that (a) its tensile and shear stresses will always vanish in the debonded direction, and (b) its stresses in the bonded directions exist because the fiber is still able to transmit stresses to the matrix on the bonded surfaces. Penny-shaped microcracks are assumed to remain at the same microcrack density during the deformations.

Unidirectionally aligned, penny-shaped microcracks can be regarded as the limiting case of unidirectionally aligned spheroidal voids with the aspect ratio $a \to 0$. That is, one can collapse one axis of a spheroidal microvoid to recover a penny-shaped microcrack. In our derivation, fibers are assumed to be elastic prolate (a1>a2=a3) spheroids that are initially perfectly bonded in the matrix. Penny-shaped microcracks are regarded as oblate (a1<a2=a3) spheroids and remains at the same volume fraction during the deformations. Furthermore, fibers and penny-shaped microcracks are assumed to be aligned.

The exterior-point Eshelby's tensor of an ellipsoidal inclusion can be derived by introducing a unit outward normal vector (shown in Figure 1) at a matrix point $\boldsymbol{x}$ on the new imaginary ellipsoid surface.

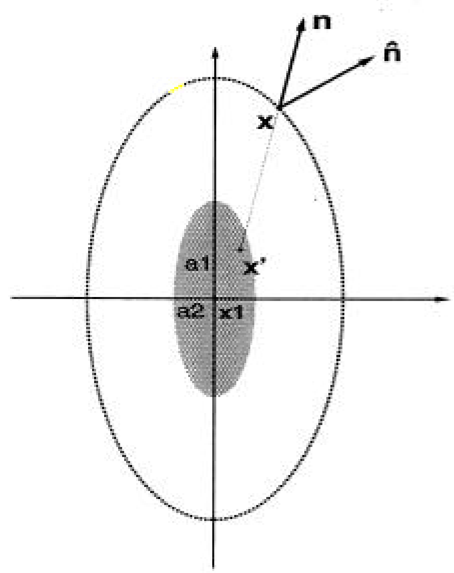

Figure 1. Schematic of an imaginary ellipsoid surface and its unit outward normal vector.

With the help of the exterior-point Eshelby's tensor of an ellipsoidal inclusion, the effective elastic stiffness tensor of aligned, fiber composites can be explicitly derived as

$$(C_*)_{ijkl} = \bar{F}_{ijkl}(t_1, t_2, t_3, t_4, t_5, t_6) \qquad (1)$$

where the fourth-rank tensor $\bar{F}$ and the parameters $t_1, ..., t_6$ are given in our previous research [7].

We now consider the overall elastoplastic responses of progressively debonded, microcrack-weakened, fiber-reinforced composites that initially feature perfect interfacial bonding between fibers and the matrix in two-phase composites. In what follows, the von Mises yield criterion with isotropic hardening law is assumed. Nevertheless, an extension of the present framework to a general yield criterion and a general hardening law is possible.

An effective yield criterion is derived based on the ensemble-volume averaging process and first-order effects of eigenstrains due to the existence of spheroidal fibers and microcracks. The effective yield criterion, together with the assumed overall associative plastic flow rule and hardening law, constitutes the analytical foundation for the estimation of effective elastoplastic behavior of ductile matrix composites. By collecting and summing all the current stress norm perturbations produced by any typical perfectly bonded fiber, any typical partially debonded fiber, and any typical penny-shaped microcrack, and averaging these over all possible locations, the ensemble-averaged current stress norm at any matrix point can be derived as

$$< H >_m (x) = \bar{\sigma} : \bar{T} : \bar{\sigma} \tag{2}$$

where $\bar{\sigma}$ is the mecroscopic stress and the positive fourth-rank tensor $\bar{T}$ is defined as

$$\bar{T} \equiv P^T \cdot T \cdot P = \bar{F}_{ijkl}(\bar{T}_1, \bar{T}_2, \bar{T}_3, \bar{T}_4, \bar{T}_5, \bar{T}_6) \tag{3}$$

and the fourth-rank tensor P and the parameters $\bar{T}_1,\ldots,\bar{T}_6$ are given in our previous research [7]. A more detail explanation of the elastoplasic stress-strain relationship for partially debonded, four-phase, aligned fiber-reinforced composites can be found in our previous research [7].

OVERALL ELASTOPLASTIC BEHAVIOR OF RANDOMLY ORIENTED, FIBER COMPOSITES

Consider composite models in which spheroidal fibers with an aspect ratio of α (the ratio of length to diameter) are uniformly dispersed and randomly oriented in two- or three-dimensional space. The average process over all orientations upon governing constitutive field equations is performed to obtain the constitutive relations and the overall yield function for randomly oriented composites.

The local axes of an inclusion are denoted by the primed coordinate system and the fixed or material axes by the unprimed one. With no loss in generality, we let axis *1'* be the symmetric axis of the spheroid and *3'* lie in the *2-3* plane. Denoting Q_{ij} as the directional cosine between the *i*-th primed and *j*-th unprimed axes, we have

$$x'_i = [Q_{ij}] x_j \tag{4}$$

where the transformation matrix for two- and three-dimensional orientation have the form of

$$[Q_{ij}] = \begin{bmatrix} 0 & \cos\theta & \sin\theta \\ 0 & -\sin\theta & \cos\theta \\ 1 & 0 & 0 \end{bmatrix}; \quad \text{for 2-D orientation}$$

$$[Q_{ij}] = \begin{bmatrix} \sin\theta\sin\phi & \cos\theta & \sin\theta\cos\phi \\ \cos\theta\sin\phi & -\sin\theta & \cos\theta\cos\phi \\ \cos\phi & 0 & -\sin\phi \end{bmatrix}; \quad \text{for 3-D orientation} \tag{5}$$

in which θ is the angle between x_1 and x_1' and ϕ is the angle between x_3 and x_3'. Any second-rank tensor (e.g., stress tensor) can be transformed as

$$\sigma'_{ij} = Q_{ik} Q_{jl} \sigma_{kl} \tag{6}$$

The symbols $(\cdot)^{2D}$ and $(\cdot)^{3D}$ are used to define the two- and three-dimensional orientation averaging process, respectively, as

$$\begin{aligned} (\cdot)^{2D} &\equiv \int_0^{\pi} (\cdot)\, P(\theta) d\theta \\ (\cdot)^{3D} &\equiv \int_0^{\pi}\int_0^{\pi} (\cdot)\, P(\theta,\phi) \sin\theta \, d\theta d\phi \end{aligned} \tag{7}$$

where $P(\theta)$ and $P(\theta,\phi)$ are the probability density functions. In the special case of uniformly random orientation, we have $P(\theta) = 1/\pi$ and $P(\theta,\phi) = 1/(2\pi)$.

Accordingly, the constitutive relations and overall yield function of two- and three-dimensional random fiber reinforced composites can be derived by performing the orientation averaging process over all orientations upon the governing field equations and overall yield function for aligned-fiber orientations given in the previous section.

PROGRESSIVE INTERFACIAL DEBONDING

Progressive interfacial debonding may occur under increasing deformations and influence the overall stress-strain behavior of randomly oriented, discontinuous fiber-reinforced composites. After the interfacial debonding between the fibers and the matrix, the debonded fibers lose the load-carrying capacity in the debonded direction and are regarded as partially debonded fibers. Within the context of the first-order (noninteracting) approximation, the stresses inside fibers should be uniform. For convenience, following Zhao and Weng [8,9], we employ the average internal stresses of fibers as the controlling factor. The probability of partial debonding is modeled as a two-parameter Weibull process. See Ju and Lee [10] for more information.

Assuming that the Weibull statistics govern, we can express the cumulative probability distribution function of the fiber debonding (damage), P_d, at the level of hydrostatic tensile stress as

$$P_d[(\bar{\boldsymbol{\sigma}}_m)_1] = 1 - exp\left[-\left(\frac{(\bar{\boldsymbol{\sigma}}_m)_1}{S_o}\right)^M\right] \tag{8}$$

where $(s_m)_1$ is the hydrostatic tensile stress of the fibers, the subscript denotes the fiber phase, and S_o and M are the Weibull parameters.

FINITE ELEMENT IMPLEMENTATION FOR IMPACT SIMULATION

To describe the various phenomena taking place during impact, it is necessary to characterize the behavior of materials under impact loading conditions. The characterization involves not only the stress-strain response at large strains and different strain rates, but also the accumulation of damage and the mode of failure. Such complex material damage behavior under dynamic loading is difficult to describe in analytical models. In numerical simulations, constitutive models of nearly any degree of complexity can be incorporated into the code.

The constitutive models derived are implemented into the nonlinear finite element code using a user-defined material subroutine to simulate the dynamic inelastic behavior and the progressive damage of the composite materials. While the implicit integration is chosen with an automatic time increment because it is not restricted to mildly nonlinear deformations or short response times, the explicit method is computationally attractive when the response time of interest is within an order of magnitude of the time it takes for a stress wave to travel through the shell thickness. Accordingly, the developed damage models are implemented into the explicit finite element code DYNA3D for impact simulation required a very small time step.

The methodology used in this work is based on the well-known strain-driven algorithm in which the stress history is to be uniquely determined by the given strain history, mainly because of its computational efficiency in the framework of explicit time integration computer program DYNA3D. The two-step operator splitting methodology is also adopted here to split the elastoplastic loading process into the elastic predictor and the plastic corrector.

The implemented model is applicable for shell and solid elements in three-dimensional analysis, as well as axisymmetric elements in two-dimensional analysis. The main advantage of the model compared with most computational damage models is that the damage evolution, giving material degradation, is fully coupled with the constitutive equation by incorporating probabilistic micromechanics appropriate to the constitutive models.

NUMERICAL SIMULATIONS AND DISCUSSION

To illustrate the elastoplastic behavior of randomly oriented, microcrack-weakened, progressively debonded composites containing discontinuous fibers, our damage model considering microcrack and progressive interfacial debonding is presented numerically for the case of random, glass-fiber epoxy resin composites. The material properties and aspect ratio of fibers involving these simulations are E_0=3.0GPa, n_0=0.35, E_1=72GPa, n_1=0.17, and a =20.0. In addition, to implement the proposed probabilistic micromechanics based on the Weibull function into the present constitutive models, we need to estimate the values of the Weibull parameters S_o and M. For simplicity, we assume the Weibull parameters to be S_o =27.25*σ_y and M=250. Figure 2 shows the effect of microcrack density and fiber debonding on the mechanical behavior of the composites. It clearly shows that the mechanical behavior of the composites is strongly dependent upon the microcrack density. Figure 3 exhibits the evolution of debonded fiber volume fraction as a function of the uniaxial strain. It is found that the composite with low microcrack density is stiffer, but the influence of damage on the stress-strain response of the composite is faster because of quick damage evolution.

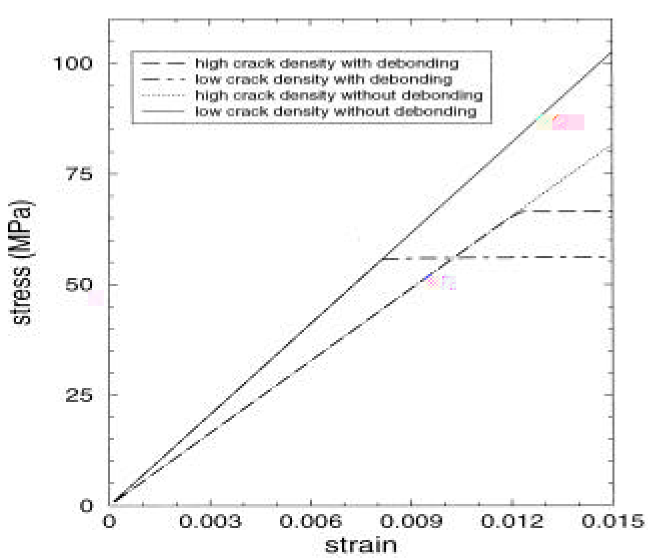

Figure 2. Effects of microcrack and fiber debonding on the stress-strain behavior of composites.

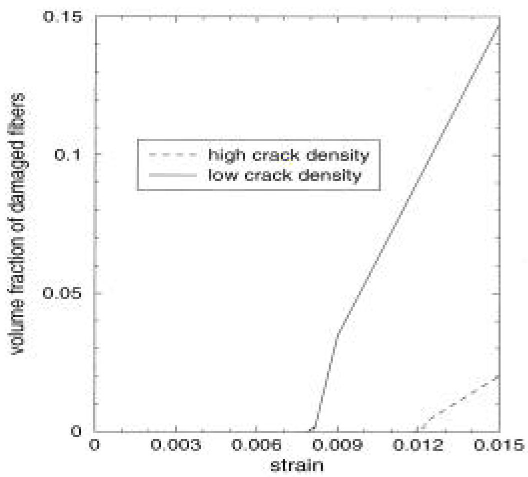

Figure 3. The predicted evolution of damaged fiber volume fraction corresponding to Figure 2.

To further address the damage evolution during the impact, we perform a numerical simulation for a cantilever composite beam under impact loading. In the simulation, the beam is fully clamped at the support and is subjected to a load at the free edge as shown in Figure 4. For brevity, we employ the same material parameters for the composites as those used in Figures 1 and 2. Here, the Weibull parameters for interfacial debonding and the initial volume fraction of fibers are assumed to be S_o =25*σ_y, M=4, and f_1=0.5. Time-history plots for the damage index, indicating the volume fraction of damaged fibers, during impact are shown in Figure 5. Figure 6 shows the displacement in the z-direction at the free edge, indicating vibration relative to the equilibrium position under impact. The simulation is used for the identification of model characteristics for solving boundary value problems.

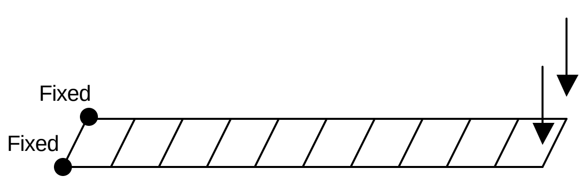

Figure 4. A cantilever beam subjected to impact loading.

Finally, preliminary composite tube impact simulations are carried out to examine whether the computational model is able to predict the experimentally obtained target response during impact. The scope of the simulations is to construct the force-displacement curve for the composite target based on the numerical results and to examine in detail a direct comparison with the drop tower test for composite tube. The drop tower test will be performed by Automotive Composites Consortium (ACC) and be provided for comparison with the simulated results.

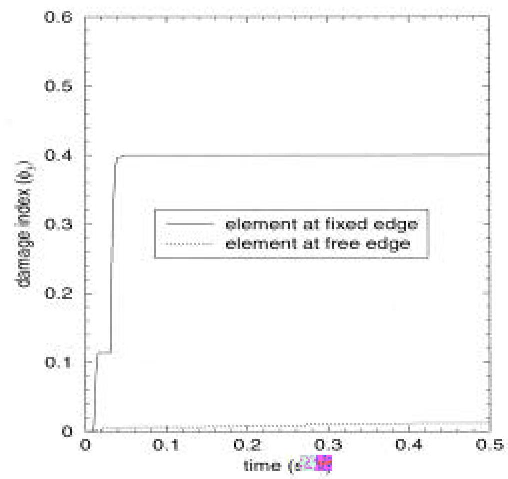

Figure 5. Damage index of elements at free and fixed edges during impact.

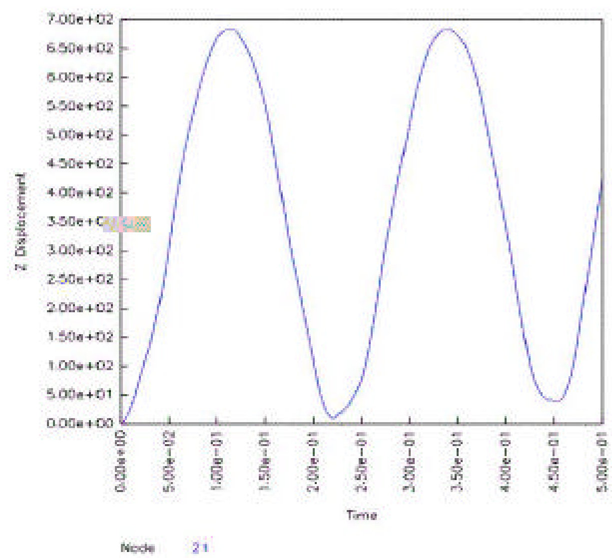

Figure 6. Displacement in the z-direction at the free edge during impact.

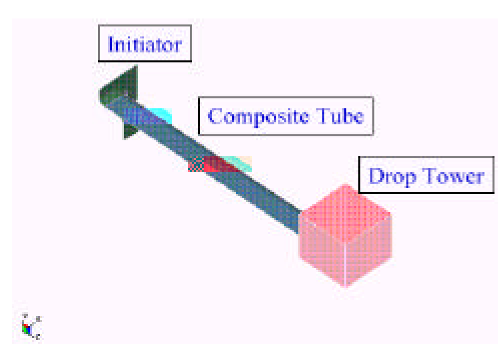

Figure 7. Schematic of composite tube impact simulation.

A schematic of the drop tower test is shown in Figure 7. In the simulation, the initiator is defined as a rigid material and is restrained in all translational and rotational degrees of feedom. The lands of the initiator lay in the *x-y* plane at *z*=0. A quarter of the drop table mass is represented by a single solid element. The centerline of the composite tube is the *z*-axis. The nodes on the top of the tube are tied to the bottom face of the solid element using a contact surface. There is 0.5mm of clearance between the tube and the vertical walls of the initiator. The nodes of the composite tube and solid element are given as an initial velocity of 8.6m/sec in the negative direction. The default DYNA3D shell element formulation based on Belyschko-Tsay theory is utilized. Based on the same material properties as those used in Figures 1 and 2, we depict our predictions in Figure 8. This figure exhibits a damage index during impact showing a severe damage at the contact surface.

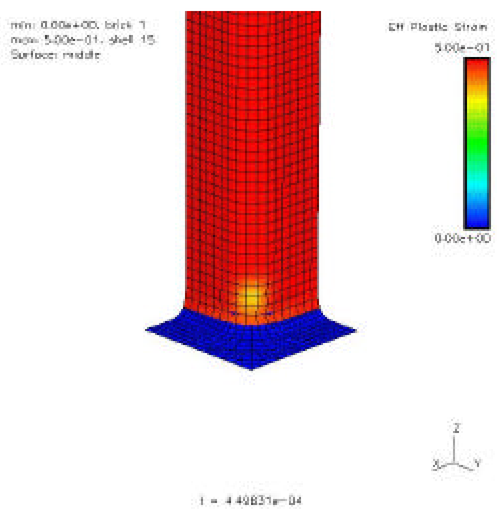

Figure 8. Damage index during impact showing a severe damage at the contact surface.

It should be noted that more work needs to be performed in order to draw a final conclusion from numerical results. Specifically, the developed damage model will be coupled with an element-failure algorithm to allow corner crack growth. The failed element will be removed from the model within an element-failure algorithm when a damage variable reaches its critical value. Furthermore, experimental work is needed to determine the model constants and damage parameters.

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