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Numerical simulation of strain-rate dependent transition of transverse tensile failure mode in fiber-reinforced composites

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ABSTRACT

This study numerically simulates strain-rate dependent transverse tensile failure of unidirectional composites. The authors' previous study reported that the failure mode depends on the strain rate, with an interface-failure-dominant mode at a relatively high strain rate and a matrix-failure-dominant mode at relatively low strain rate. The present study aims to demonstrate this failure-mode transition by a periodic unit-cell simulation containing 20 fibers located randomly in the matrix. An elasto-viscoplastic constitutive equation that involves continuum damage mechanics regarding yielding and cavitation-induced brittle failure is used for the matrix. A cohesive zone model is employed for the fiber–matrix interface, considering mixed-mode interfacial failure. For the results, the relationship between failure modes and the strain rate is consistent with the authors' previous studies.

1. Introduction

Polymer composite materials have been widely used, especially in the aerospace field, because of their high specific strength and modulus. A transverse crack, what we call first-ply failure, sometimes causes considerable damage and leaks, and can trigger fatal delamination. A precise estimate of the initiation of transverse cracking is necessary to improve reliability of composite utilization. In order to predict the initiation of transverse cracking, it is necessary to understand the characteristics of transverse failure of unidirectional composites. In our previous work [1], we studied the transverse failure of unidirectional composites, experimentally revealing that the failure mode depends on the strain rate (Fig. 1). At high strain rates, the failure of the composite exhibits an interface-failure dominant mode, while a matrix-failure dominant mode appears at low strain rates. The paper concluded that the strain-rate dependence of the interface strength is much less than that of the matrix strength.

There have been numerous works on transverse failure in composite materials. Hashin [2] presented criteria in which a quadratic interaction between stress invariants on the failure plane is considered. Puck and Schurmann [3] introduced the effect of internal material friction into Hashin's criteria. In addition, Davila et al. [4] proposed LaRC03 criteria and compared the above three criteria sets with experiments conducted in the World Wide Failure Exercise. Totry et al. [5] compared these criteria with a numerical analysis by periodic unit-cell (PUC) simulation, i.e. the representative volume element method. In order to simulate composite failures microscopically, they modeled individual fibers and matrix and adopted the Mohr–Coulomb criterion for matrix yielding. PUC has been the focus of many researchers in this decade because of the relatively lower computational cost; modeling of each individual fiber within the specimen of interest is impossible in terms of the computational cost. Among these researchers, Vaughan and McCarthy et al. [6] pointed out that both the matrix properties and interface properties affect the failure characteristics of composites. In fact, Hobbiebrunken et al. [7] and Canal et al. [8] presented experimental evidence of interface debonding in composite failure by in-situ observation. Thus, transverse failure of polymer composite materials consists of matrix failure and/or interface failure.

The following articles are worthy of special note in terms of matrix failure. Asp et al. [9] presented a scheme of critical dilatational deformation in the first quadrant of the bi-axial failure envelope for polymers, which physically suggests that cavitation or crazing occurs in polymer materials. Based on this concept, Gosse and Christensen [10] proposed a strain invariant failure theory (SIFT). SIFT agreed with the experimental results of Asp et al. [9] and was employed in the element-failure method developed by Tay et al. [11]. Canal et al. [12] also implemented PUC simulation, considering the cavitation-induced brittle failure with an elasto-visco-elastic constitutive equation. The above-mentioned articles deal with wholly static failure; however, the failure of polymer matrix is usually strain-rate-dependent [13–22]. In order to discuss

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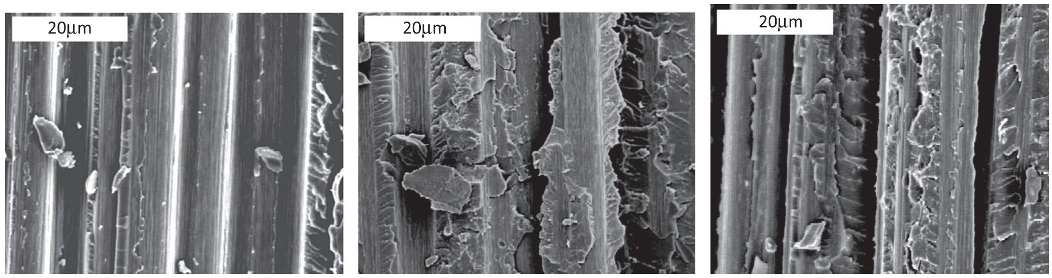

(a) High-strain rate (4.0x 10^{-4}) (b) Intermediate-strainrate (c) Low-strainrate (4.0x10^{-6})

Fig. 1. SEM photograph of failure surface for transverse tensile test with various strain rate: Interface failure appears at high-strain rate and matrix failure appears at low strain rate.

strain-rate-induced failure-mode transitions as in Fig. 1, rate-dependent failure characteristics should be taken into account; but such articles have been a few [23–27].

In this study, the characteristics of transverse failure in a unidirectional composite that depend on the strain rate are numerically simulated. We implement a periodic unit-cell simulation for the failure-mode transition caused by varying the applied strain rate. The unit cell consists of 20 fibers and the surrounding matrix. The fiber is assumed to be an elastic body, and the matrix employs an elasto-viscoplastic constitutive equation based on damage mechanics regarding yielding and cavitation failure. A cohesive zone model is adopted for the fiber–matrix interface. We determine the parameters used in the elasto-viscoplastic constitutive equation and cohesive element by comparing the failure mode, the specimen strength, and the stress–strain curve of the simulation results with experimental results [1].

2. Analytical procedures

2.1. Periodic unit-cell simulation

This study employs a 2D-unit cell as illustrated in Fig. 2, consisting of 20 fibers and a matrix in which the fiber array is randomly positioned. Note that this study only assumes this fiber distribution although simulated results might differ when the fiber distribution differs. The unit cell, which employs 6-node triangular elements, consists of approximately 100,000 elements. A plane strain condition is assumed. The fiber radius is 3.5 μm, and the fiber volume fraction is 55%. The assumed material properties for fiber and matrix are listed in the Table 1. Since the experiments depicted in Fig. 1 were carried out at 75 °C and the specimen was fabricated at 175 °C, we assign a temperature difference of negative 100 K to all nodes as an initial condition. Periodic boundary conditions are applied to all the corresponding edges in the unit cell. The fiber is modeled as an elastic body and is isotropic in this plane. Modeling for the matrix and interface is described in the following sections. For the unit-cell, strain and displacement are separated into global (specimen strain) and local components as specified in the equations below.

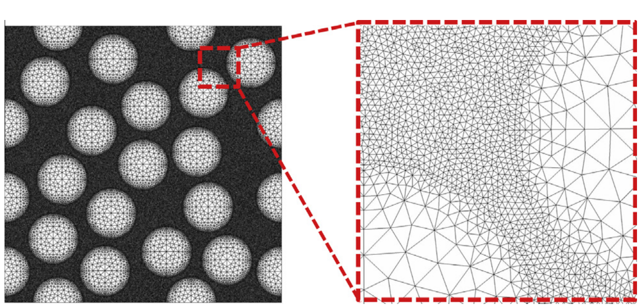

Fig. 2. Periodic unit cell with mesh and magnified one: triangle quadratic element is used, fine mesh is adopted for matrix part and near interface, no volume cohesive element is inserted at every interface. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)

Table 1
Material properties of carbon fiber and epoxy matrix.

Fiber axial Young's modulus E_Z	230 GPa
Fiber radial and hoop Young's modulus E_R, E_θ	14 GPa
Fiber longitudinal Poisson's ratio ν_{ZR}, $\nu_{Z\theta}$	0.35
Fiber cross-sectional Poisson's ratio $\nu_{R\theta}$	0.35
Fiber radius r_f	3.5 μm
Fiber's coefficient of thermal expansion for the axial direction a_Z	-1.1×10^{-6}/K
Fiber's coefficient of thermal expansion for the radial and hoop direction a_R, a_θ	10×10^{-6}/K
Matrix Young's modulus E_m	3.4 GPa
Matrix Poisson's ratio ν_m	0.31
Matrix's coefficient of thermal expansion a_m	60×10^{-6}/K
ΔT	-100 K

$$\boldsymbol{u} = \boldsymbol{u}_G + \boldsymbol{u}_L \tag{1}$$

$$\Delta\boldsymbol{\varepsilon} = \Delta\boldsymbol{\varepsilon}_G + \Delta\boldsymbol{\varepsilon}_L \tag{2}$$

Subscript G indicates the global component, and L indicates the local component. The macroscopic tensile load (specimen load) is controlled by the global component. The numerical implementation in the present study is based on Okabe et al. [28,29]; in addition, a cohesive zone model is taken into account.

2.2. Continuum damage mechanics model for matrix

The stress–strain response of the matrix resin is modeled by a parameter damage model suggested by Kobayashi et al. [30], in which the constitutive equation including the effect of damage is defined by the following equations.

$$\dot{\boldsymbol{\sigma}} = (1-D)\boldsymbol{C}^{v} : \dot{\boldsymbol{\varepsilon}} - (1-D)\frac{3\mu\dot{\bar{\varepsilon}}^{p}\cos\delta}{\bar{\sigma}}\boldsymbol{\sigma}' - \frac{\dot{D}}{1-D}\boldsymbol{\sigma} \tag{3}$$

$$\boldsymbol{C}^{v} = \frac{H}{H+3\mu}\left[\boldsymbol{C}^{e} + \frac{3\mu}{H}\left\{\frac{3\lambda+2\mu}{3}\boldsymbol{I}\otimes\boldsymbol{I} + 3\mu\frac{\boldsymbol{\sigma}'\otimes\boldsymbol{\sigma}'}{\bar{\sigma}^{2}}\right\}\right]$$

$$H = \frac{1}{1-D}\frac{\dot{\bar{\sigma}}}{\dot{\bar{\varepsilon}}^{p}k}, \quad \bar{\sigma} = \sqrt{\frac{3}{2}\boldsymbol{\sigma}' : \boldsymbol{\sigma}'}, \quad \dot{\bar{\sigma}} = \sqrt{\frac{3}{2}\dot{\boldsymbol{\sigma}}' : \dot{\boldsymbol{\sigma}}'}$$

Here, δ is a non-coaxial angle, k is a non-coaxial parameter, λ and μ are Lame's constants, and $\boldsymbol{\sigma}'$ is the deviatoric stress tensor. The equivalent plastic strain rate $\dot{\bar{\varepsilon}}^{p}$ is expressed in the following