Numerical assessment of failure mechanisms due to tensile loading in angle-ply laminates

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ABSTRACT: A mesoscale finite element model is presented for simulating the failure behavior of E-glass/epoxy angle-ply laminates under tensile loading. The effective laminate properties are determined from the properties of ply constituents, i.e. fiber and matrix by using numerical homogenization technique. Two different interfiber failure mechanisms leading to matrix cracking are reproduced in the simulations by using appropriate constitutive equations. The predictions from this model are compared with experimental data available in the literature, and are found to be in good agreement.

KEY WORDS: polymer-matrix composites, finite element analysis, numerical homogenization,

1. Introduction

The main advantage of micromechanical and mesomechanical approaches for studying damage mechanisms in composite laminates in regard to macroscopic analysis is that they can accurately reproduce the initiation and evolution of damage because the individual constituents of the ply such as fiber, matrix and the interface between them are explicitly included in the models. The mesomechanical approach is based on an assumption that the structure of the whole laminate can be idealized at the ply scale as a periodic array of repeated unit cells [1,2].

2. Mesoscale finite element model

As shown in Fig.1, a unit cell of an angle-ply laminate consists of three rhombohedrons which represent three successive plies with various fiber orientations. The structure of an $\pm\theta$ angle–ply laminate can be viewed at the ply scale in the skew coordinate system (x_1^s–x_2^s–x_3^s) as a periodical array of plies containing a periodic subarray of fibers. Six different ply orientations θ = 15°, 30°, 45°, 55°, 60° and 75° are considered.

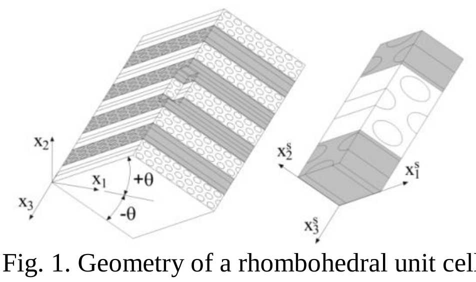

Fig. 1. Geometry of a rhombohedral unit cell

Periodic deformation of the rhombohedral unit cell is controlled by macroscopic strains by using the following boundary conditions [3]

$$u_i = \varepsilon_{ij} x_j + u_i^{\mathrm{p}} \qquad (1)$$

where ε_{ij} are the components of the applied strain in the skew coordinate system. The components of macroscopic stress corresponding to the applied strain can be calculated from

$$\bar{\sigma}_{ij} = \frac{1}{V}\int_V \sigma_{ij}\,\mathrm{d}V \qquad (2)$$

where V is the volume of the rhombohedral unit cell.

The plastic deformation of polymeric materials is highly sensitive to the hydrostatic pressure and plastic flow of these materials can exhibit plastic dilatancy. To address this requirement, the Drucker–Prager plasticity model [4], which incorporates the linear dependence on the hydrostatic stress, is used. In terms of the first invariant of stress I_1 and the second invariant of the deviatoric part of stress J_2, the yield function is given as

$$\mathrm{f} = (\mu\, I_1 / 3) + \mathrm{sqrt}(J_2) - k\,, \qquad (3)$$

where μ is the pressure sensitivity factor, k is the flow stress of the material under pure shear.

For the fiber/matrix interface failure, a cohesive zone model is employed, in which the constitutive equations of the interface relate the normal σ_n and tangential τ_t cohesive tractions to the normal u_n and tangential u_t displacement jumps and a scalar damage variable d, through [5]

$$\sigma_n = k_n u_n (1-d), \quad \tau_t = k_t u_t (1-d), \qquad (4)$$

where k_n , k_t are initial contact stiffnesses in the normal and tangential direction, respectively. The scalar damage variable d represents the loss of stiffness and it is a function of both displacement jumps. To define the completion of fracture in the cohesive zone model, a linear energy criterion is used [5]

$$(G_n / G_n^{\mathrm{c}}) + (G_t / G_t^{\mathrm{c}}) = 1 \qquad (5)$$

where G_n, G_t denote energy release rates for mode I fracture and mode II fracture, respectively and G_n^{c} and G_t^{c} correspond to the interfacial fracture energies.

3. Results

In order to identify the failure mechanism of angle-ply laminates under tensile loading along x_2, an analysis of the local stresses in the unit cell models for different ply orientations is performed. For this purpose, it is assumed that the initiation of cracks in the matrix occurs if the maximum shear stress in the matrix or the first principal stress in the matrix goes beyond the corresponding ultimate strengths of the matrix

$$\tau_{\max} \geq \mathrm{S_m}, \quad \text{or} \quad \sigma_1 \geq \mathrm{Y_m}. \qquad (6)$$

Fig. 2 shows distributions of the equivalent plastic strain in the ±45 angle-ply laminate for critical loads. In the case of the imperfect interface, the critical load corresponds to a limit beyond which the fiber/matrix interface cracks are fully opened ($\varepsilon_y^c = 0.01$). In turn, in the case of the perfect interface, it corresponds to a limit beyond which the shear strength of the matrix is attained ($\varepsilon_y^c = 0.1185$). The most intense plastic deformation in the matrix is localized in shear bands which develop along planes containing the fiber direction. Thus, both models predict that cracking of the matrix takes place along the fiber direction within the shear bands.

(a)

(b)

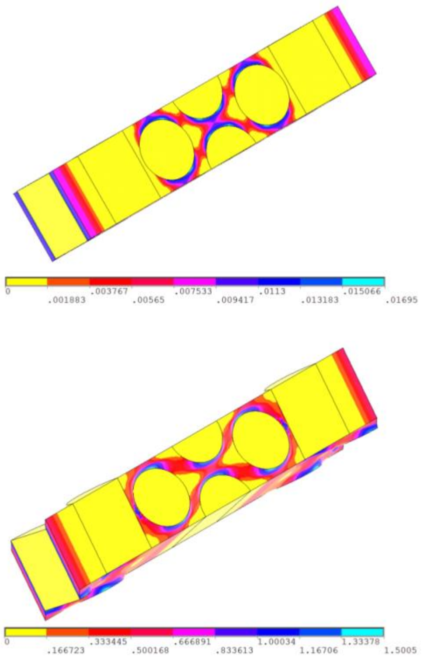

Fig. 2. Contour plots of ε_{pl} in the ±45^{o} angle-ply laminate at the critical stage of deformation from the mesoscale model with (a) imperfect and (b) perfect fiber/matrix interface conditions.

In order to check the reliability of the opening mode of fracture, an analysis of the first principal stress σ_1 and the equivalent plastic strain ε_{pl} in the matrix is performed by using the mesoscale model with the imperfect fiber/matrix interface. The distributions of these quantities at the fiber/matrix interface are presented in Fig.3 at three successive stages of deformation corresponding to the beginning of inelastic behavior, the limit load and the softening regime. It is interesting to note that the tensile stress in the matrix decreases at the softening regime and it is below the tensile strength (Y_m = 100 MPa). In contrast to the tensile stress, the plastic strain in the matrix increases constantly with increasing applied strain. This means that the condition for crack growth in the matrix under the opening mode of fracture cannot be satisfied.

In order to validate the shearing mode of fracture, an analysis of the maximum shear stress τ_{max} and the equivalent plastic strain ε_{pl} in the matrix is performed by using the mesoscale model with the perfect fiber/matrix interface. The distributions of these quantities at the fiber/matrix interface are presented in Fig.4 at three successive stages of deformation. It can be observed that both the shear stress τ_{max} and the plastic strain ε_{pl} in the matrix increase with increasing applied strain. Thus, the shear stress in the matrix may go beyond the shear strength of the epoxy matrix (S_m = 87 MPa).

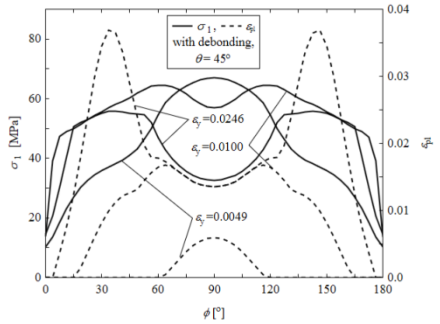

Fig. 3. Angular distributions of σ_1 and ε_{pl} at the fiber/matrix interface at three successive stages of deformation from the mesoscale model with imperfect fiber/matrix interface.

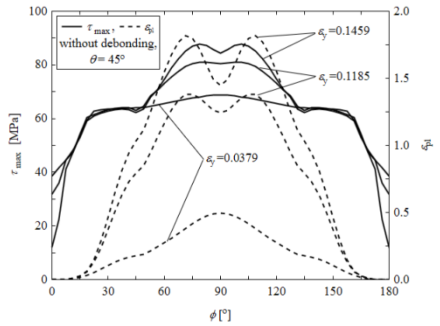

Fig. 4. Angular distributions of τ_{max} and ε_{pl} at the fiber/matrix interface at three successive stages of deformation from the mesoscale model with perfect fiber/matrix interface.

4. Conclusions

The efficiency of two types of mesoscale models, with and without interfacial debonding, was evaluated. The first type was found to reproduce well the tensile response of angle-ply laminates when plies are oriented at an angle larger than 45 degree with respect to the loading direction, and in turn, the second type was better suited to it when plies are oriented at angle less than 45 degree.

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