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High-fidelity micro-scale modeling of the thermo-visco-plastic behavior of carbon fiber polymer matrix composites

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ABSTRACT

An experimentally validated micro-scale analysis of the visco-thermo-mechanical behavior of polymer matrix composites under different loads is proposed. A new constitutive law for the matrix material is developed taking into account the pressure dependence of the material as well as strain-rate and temperature dependence. Capturing the matrix behavior under multi-axial stress states is concluded to be essential to accurately predict the composite material behavior, even when considering simple load cases such as transverse compression and/or shear. Without any calibration procedure at the composite level, good agreement with the experimental data is observed for different loading conditions, including strain-rate dependency.

Using this validated micro-scale model, a three-dimensional simulation of the formation of a kink band under longitudinal compression of the composite is conducted. A new evidence at micro-scale is found supporting the hypothesis that shear stresses transferred between fibers and matrix are particularly important in the formation of the kink band.

1. Introduction

There is an increasing demand for accurate predictive models of the inelastic and fracture behavior of fiber reinforced polymer matrix composites due to the wide spread of applications where these materials are being used, especially in the transportation industry. However, accurately modeling the thermo-mechanical behavior of these materials remains a difficult challenge due to various factors, namely: (1) the matrix material is usually a highly cross-linked glassy polymer (e.g. epoxy resin) that shows rate- and thermal-dependent inelastic response [1]; (2) the high strength and brittleness of the fibers associated to the high fiber volume fraction of the composite (typically around 60%) leads to large three-dimensional strain concentrations in the matrix [2]; (3) the fiber distribution is non-uniform [3]; and (4) the manufacturing process introduces thermal residual stresses [4] and different defects such as voids, fiber waviness and fiber misalignment [5].

These factors limit the applicability of analysis models based on the homogenized behavior of the composite, denominated here by macro- and meso-scale models. Macro-scale models deal with the entire laminate as homogeneous continuum and therefore are capable of doing structural analysis and preliminary design and optimization [6,7]. Meso-scale models improve the predictive capabilities of the macro-scale models by analyzing individual plies as a homogeneous continuum [8–15] and by modeling the delamination between the different plies via cohesive models [16–18]. The recent work of Camanho et al. [19] presented a meso-scale model that coupled a fully three-dimensional damage model with a new plasticity law developed by Ref. [20] that led to accurate predictions for non-trivial experimental tests.

Since the macro- and meso-scale models rely on the description of the polymer composite as a homogeneous continuum, they inevitably lead to complex constitutive and damage laws that require a large number of experimental inputs [9,10,7]. Furthermore, the analyses obtained at these scales do not provide enough detail about the mechanical processes that explain the inelastic behavior of the material. This motivated efforts for the development of micro-scale models, where the highest level of detail in modeling the different failure mechanisms of polymer composites under different load cases has been reached. Significant progress was obtained by using different constitutive models for each constituent in the composite. Since fibers are very brittle, the constitutive law can be assumed as transversely isotropic linear

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elastic. However, the polymeric matrix constitutive law is significantly more challenging, as highlighted by several authors [21–23,4,24–26]. In fact, the polymeric matrix can exhibit strong nonlinear behavior under general load cases, as well as thermal and strain-rate dependence.

In this article the authors present a micro-scale analysis that will be shown to lead to accurate predictions of the behavior of the composite material. The presented model is validated experimentally not only at the level of the single constituents but also at the composite level by comparing its homogenized response with published experimental data. This contrasts with previous works where direct comparison with experimental data was not shown. A new constitutive model for polymer matrix is proposed and validated in Sections 2 and 3. Section 4 discusses the micro-scale model for composite materials and the validation of the predicted response for a carbon fiber polymer matrix composite. Section 5 is dedicated to the prediction of a particularly complex failure mechanism of composite materials under longitudinal compression: the formation of a kink band.

2. Constitutive model for polymer matrix

2.1. A modified paraboloid yield criterion

The motivation for developing a new constitutive law for the matrix material stems from the fact that even the simplest loading conditions applied to the composite produce general stress states that are far from the uniaxial loading conditions under which the bulk matrix material is typically tested [1]. Previous works used constitutive models that are known to be unable to describe such general stress states, as the Mohr–Coulomb elasto-plastic model [21–23,4], or the Drucker–Prager model [24–26]. The inadequacy of these models was pointed out by different authors [2,27,28], who proposed the paraboloidal criterion as an alternative.

Here an extension to the paraboloidal model is proposed so that the predictive ability of the micro-scale analyses is improved. The classical paraboloidal yield criterion [29,30] can be written as:

$$f(\boldsymbol{\sigma}, \sigma_{Y_C}, \sigma_{Y_T}) = 6J_2 + 2(\sigma_{Y_C} - \sigma_{Y_T})I_1 - 2\sigma_{Y_C}\sigma_{Y_T} \tag{1}$$

where σ_{Y_C} and σ_{Y_T} are the absolute values of compressive and tensile yield stresses, $I_1 = Tr(\boldsymbol{\sigma}_{ij})$ is the first stress invariant, and $J_2 = \frac{1}{2}\boldsymbol{s}_{ij}\boldsymbol{s}_{ij}$ is the second invariant of the deviatoric stress tensor $\boldsymbol{s}_{ij}$.

Observing Eq. (1), one sees that this yield surface is determined uniquely by the two parameters σ_{Y_C} and σ_{Y_T}. However, this yield surface leads to inaccurate predictions for the response of the bulk matrix under pure shear case, as demonstrated by the results presented in Section 3.

The following modification to the paraboloidal yield criterion is proposed here to include the shear yield behavior:

$$f(\boldsymbol{\sigma}, \sigma_{Y_C}, \sigma_{Y_T}, \sigma_{Y_S}) = a\left(J_2^3 + bJ_3^2\right)^{1/3} + 2(\sigma_{Y_C} - \sigma_{Y_T})I_1 - 2\sigma_{Y_C}\sigma_{Y_T} \tag{2}$$

where the third deviatoric stress invariant $J_3 = \det(\boldsymbol{s}_{ij})$ and two inter-dependent parameters a and b are introduced. These parameters can be determined with only one additional experimental test.

The relationship between a and b can be determined from the uniaxial tensile test by considering $\sigma_2 = \sigma_3 = 0$ in Eq. (2) to obtain:

$$a = 6\left(\frac{27}{4b + 27}\right)^{1/3} \tag{3}$$

Since this modified paraboloidal yield criterion has only one additional independent parameter, the pure shear experimental test of the matrix is sufficient to calibrate the constitutive law. This is achieved by considering the pure shear stress state $\sigma_1 = -\sigma_2$ and $\sigma_3 = 0$ in Eq. (2) and relating the parameter b with the shear yield strength σ_{Y_S}:

$$b = \frac{1}{4}\left(\frac{(3\sigma_{Y_S})^6}{(\sigma_{Y_C}\sigma_{Y_T})^3} - 27\right) \tag{4}$$

The derivation of a and b can be found in Appendix A. The effect of the third invariant J_3 is modulated by the parameter b as illustrated in Fig. 1 and Eq. (5)

$$\begin{cases} b = 0 & \Rightarrow \quad \text{Classical paraboloidal model} \\ b > 0 & \Rightarrow \quad \text{Increase shear yield stress} \\ b < 0 & \Rightarrow \quad \text{Decrease shear yield stress} \end{cases} \tag{5}$$

It is also observed that the variation of b has no effect on the compressive, tensile and biaxial yield strengths. The added term has an influence on the shear yield strength, the asymmetry of the yield function and its evolution, which will be shown to lead to accurate predictions for the material response with particular impact on the predictions of the response for the composite material.

A non-associative flow rule [31] was used to prevent positive volumetric plastic strain under hydrostatic pressure

$$g = \sigma_{vm}^2 + \alpha p^2 \tag{6}$$

where $\sigma_{vm} = \sqrt{3J_2}$ is the von Mises equivalent stress, $p = 1/3I_1$ is the hydrostatic pressure, and the parameter α is given as:

$$\alpha = \frac{9}{2}\frac{1 - 2\nu_p}{1 + \nu_p} \tag{7}$$

with ν_p being the plastic Poisson's ratio that can be determined from the uniaxial tensile test. From the above results one can write the increment in plastic deformation via the flow rule that uses the plastic potential given in Eq. (6):

$$\Delta\varepsilon^p = \Delta\lambda\frac{\partial g}{\partial \boldsymbol{\sigma}} \tag{8}$$

with $\Delta\lambda$ being the plastic multiplier, subjected to the Kuhn–Tucker consistency conditions and to be updated via the return mapping algorithm [32].

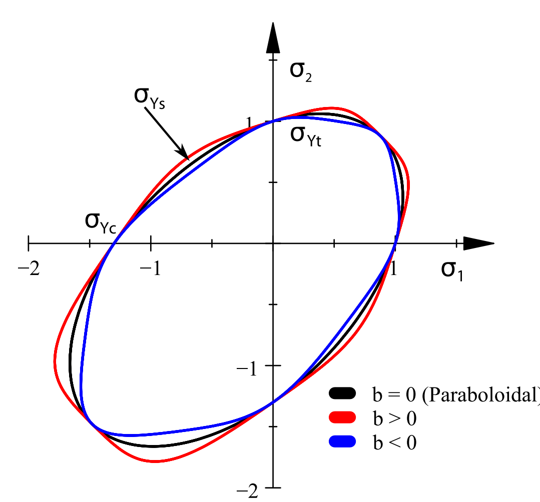

Fig. 1. Effect of the parameter b on the yield surface of the matrix constitutive model in principal stress space with $\sigma_3 = 0$.