DAMAGE MECHANISM AND EFFECTIVE PROPERTIES OF MULTI-MATERIAL BOUNDARY LAYERS

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Keywords: interface, failure, homogenization, gradient-damage, large strain

Abstract

In this contribution the inelastic behavior of hybrid composites is considered. The composites consist of an aluminum and a fiber reinforced polymer component. We study different local phenomena and their effects on the overall interface characteristics, e.g. the failure of the polymer material and the surface roughness. Since there is a large separation in the length scales of the surface roughness, which is in the micrometer range, and conventional structural components, we employ a numerical homogenization approach to extract effective interface parameters. The description of the large strain behavior of the polymer material is based on a gradient-enhanced finite strain elastic-plastic damage model.

1. Introduction

The fulfillment of the high demands on innovative lightweight structures requires the combination of classical lightweight metals with fiber reinforced polymers (FRP) in hybrid components. The performance of the overall structure is limited by their damage and failure behavior and the contact zone between both materials is often identified as the weak point. Recent interlocking joining approaches show promising benefits to connect the materials in a non-destructive manner [1]. In this contribution, the failure behavior of an aluminum component and an FRP is considered on the microscale [2]. The microsection of the interface zone is shown in Fig. 1, which exhibits different inhomogeneities, e.g. the rough metal-polymer interface and embedded fibers.

A finite element model is presented to investigate the interaction of a rough aluminum-polymer interface with randomly distributed glass fibers and their influence on the overall interface properties. The microscopic model accounts for cohesive failure of the bulk material, where the large strain behavior of the considered epoxy matrix is described by an elastic-plastic material model with progressive damage evolution. The application of numerical homogenization techniques allows to derive effective traction-separation relations from the micromechanical simulations [3].

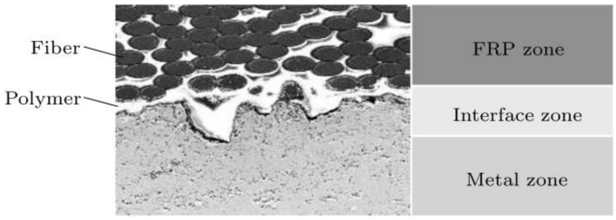

Figure 1: Microsection of the interface zone of a fiber reinforced polymer.

2. Constitutive models

In the micromodel, three different materials have to be described. In a finite strain framework, the elastic response of these materials is based on a hyperelastic material model, where the effective KIRCHHOFF stress tensor

$$\hat{\boldsymbol{\tau}} = 2\frac{\partial\hat{\Psi}}{\partial\mathbf{b}^{\mathrm{e}}} \cdot \mathbf{b}^{\mathrm{e}} \tag{1}$$

is related to the specific free HELMHOLTZ energy $\hat{\Psi}$, with $\mathbf{b}^{\mathrm{e}}$ the elastic left-CAUCHY-GREEN deformation tensor. The chosen OGDEN [4] ansatz for $\hat{\Psi}$ enables the flexible approximation of the elastic response of many metals and polymers. Since the considered epoxy matrix exhibits the lowest strength in the material mix, no damage of the aluminum and glass fiber compounds is expected. In contrast, the epoxy material shows a nonlinear inelastic material behavior with different strength values for tensile and compressive loads [5]. Therefore, a gradient-enhanced elastic-plastic damage model was developed, based on the parabolic yield criterion proposed by Tschoegl [5]

$$f = 6\hat{J}_2 + 2(\sigma_{\mathrm{c}} - \sigma_{\mathrm{t}})\hat{I}_1 - 2\sigma_{\mathrm{c}}\sigma_{\mathrm{t}}, \tag{2}$$

which enables the modelling of pressure dependent yield. In (2) $\hat{J}_2$ represents the second invariant of the deviatoric part of the effective KIRCHOFF stress, $\hat{I}_1$ the first invariant of the effective KIRCHHOFF stress and $\sigma_{\mathrm{c}}, \sigma_{\mathrm{t}}$ the compressive and tensile yield strength, respectively. The framework of finite plasticity used here is based on the multiplicative split of the deformation gradient $\mathbf{F} = \mathbf{F}^{\mathrm{e}} \cdot \mathbf{F}^{\mathrm{p}}$ into an elastic and plastic part. The plastic yield is governed by the non-associated evolution of the plastic deformation rate tensor

$$\mathbf{d}^{\mathrm{p}} = \gamma\frac{\partial\Phi}{\partial\hat{\boldsymbol{\tau}}} \quad \gamma \geq 0,\ \Phi \leq 0,\ \gamma\Phi = 0 \tag{3}$$

with the plastic potential Φ according to the model in Melro et al. [7]. The algorithm for the static multiplicative plasticity used here, is based on the classical return mapping scheme proposed by Simo [8]. After a certain amount of plastic yielding, damage onset and evolution are expected. Hence, an isotropic damage model based on the concept of effective stresses

$$\boldsymbol{\tau} = (1 - D)\hat{\boldsymbol{\tau}}, \tag{4}$$

is employed to model a progressive damage evolution. A scalar damage variable D changes exponentially between 0 and 1 and reduces the effective KIRCHHOFF stress tensor. To prevent the typical mesh dependency of softening materials, the gradient-enhanced damage formulation proposed by Peerlings et al. [9] is employed to guarantee mesh independent energy dissipation and damage localization. Therefore, two field equations have to be solved numerically: The momentum balance and the HELMOLTZ-type equation

$$\overline{\kappa} - l^2\nabla^2\overline{\kappa} = \kappa, \tag{5}$$

with κ the local source term, driven by the equivalent plastic strain and $\overline{\kappa}$ the nonlocal counterpart, leading to a damage propagation $D(\overline{\kappa})$. Similar to the parabolic yield criterion in (2), a pressure sensitive damage activation function separates the pure elastic-plastic and the active damaging range [7]. A schematic uniaxial stress-stretch curve for the proposed polymer model is shown in Fig 2a).

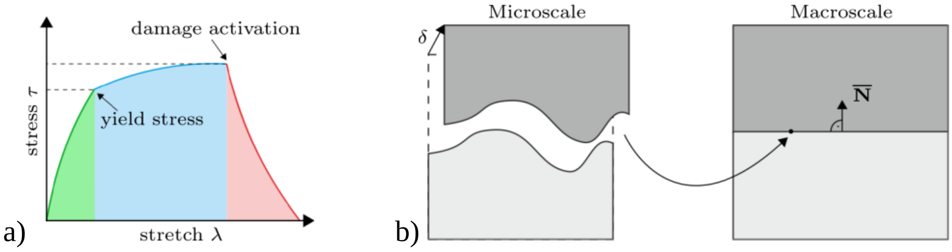

Figure 2: Schematic uniaxial response of elastic-plastic damage model in a) and homogenization scheme of an interface section in (b).

3. Homogenization

A homogenization scheme to extract effective interface properties of inhomogeneous material layers was proposed by Alfaro et al. [3] and already adopted in e.g. [2] and [10]. In this work a large deformation formulation is employed, which was published by Hirschberger et al. [11]. Based on the HILL-MANDEL energy condition, an appropriate criterion for interface considerations can be formulated

$$\bar{\mathbf{t}}_c \cdot \delta\bar{\boldsymbol{\delta}} = \frac{1}{b}\int\limits_{V_0} \mathbf{P} : \delta\mathbf{F}\mathrm{d}\Omega \tag{6}$$

where the virtual work on the macroscale done by the cohesive traction $\bar{\mathbf{t}}_c$ and the separation $\delta\bar{\boldsymbol{\delta}}$ equals the volume average of the work done on the microscale by the first PIOLA-KIRCHHOFF stress tensor $\mathbf{P}$ and the deformation gradient $\delta\mathbf{F}$. The width b of the micro domain is used for the averaging step. With the formulation of the macroscopic separation in terms of the displacement of one control node of the micro domain $\bar{\boldsymbol{\delta}} = \boldsymbol{\delta}$ (see Fig. 2b) and incorporation of hybrid boundary conditions [11], the expression of the macroscopic traction in terms of the microscale quantities is reached:

$$\bar{\mathbf{t}}_c = \langle\mathbf{P}\rangle \cdot \overline{\mathbf{N}} \tag{7}$$

Hence, the macroscopic effective traction vector results from the projection of the volume average of the microscopic first PIOLA-KIRCHHOFF stress tensor onto the macroscopic interface plane described by the normal unit vector $\overline{\mathbf{N}}$.

4. Simulation

In this section the presented constitutive models are used in combination with the homogenization scheme to study the failure behavior of an FRP-aluminum joint. The investigated representative volume elements (RVEs) are shown in Fig. 3, where the roughness profile between metal and polymer was generated with a random field algorithm. Afterwards fibers were placed with a random fiber algorithm. The material models are parametrized to mimic an aluminum compound connected to a glass fiber reinforced epoxy matrix with parameters taken from [7,12].

Two RVEs were investigated under shear loading conditions: In case one with fibers and case two without fibers to study the influence of their presence near the interface. The homogenized traction-separation curves are shown in Fig. 3. After an initial elastic range in both cases, a plastic regime follows with subsequent damage initiation and propagation to full failure. The presence of fibers near the interface leads to an increase of the effective stiffness and an embrittlement of the failure behavior. With absence of fibers the polymer material can yield in a more unconstrained manner over a wide range with a late imitation of damage in contrast to case one, where the plastic deformation localizes in an early stage.

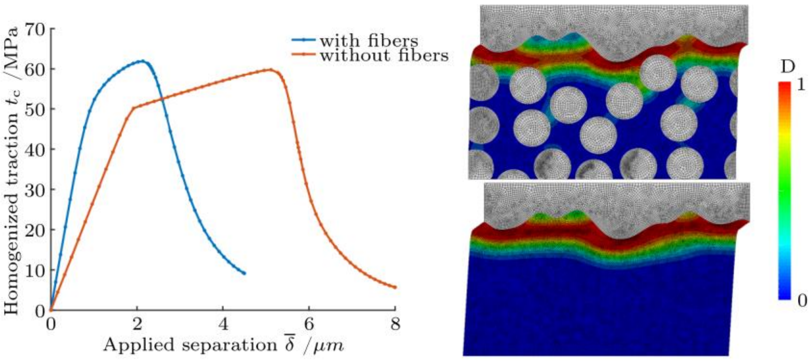

Figure 3: Homogenized traction-separation curves under shear loading (left) and contour plots of the damage variable D for RVEs with and without fibers (right).

4. Conclusion

The presented method enables the modelling of interface failure and the extraction of effective traction-separation relations of multi-layer interfaces. It was applied to analyze FRP-metal hybrid interfaces with consideration of cohesive failure of the polymer material. Numerical simulations show the influence of random fiber inclusions near the rough interface on the damage localization and the effective traction-separation curves. In addition to the diffuse cohesive failure of the polymer material, discrete adhesive failure of the local interfaces will be considered with cohesive zone models in the future.

Acknowledgments

The present project is supported by the German Research Foundation (DFG) within the Priority Programme (SPP) 1712. This support is gratefully acknowledged.

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