Composites Science and Technology 110 (2015) 196–204

Contents lists available at ScienceDirect

Composites Science and Technology

journal homepage: www.elsevier.com/locate/compscitech

Transverse cracking of cross-ply laminates: A computational micromechanics perspective

Miguel Herráez [a], Diego Mora [a], Fernando Naya [a], Claudio S. Lopes [a], Carlos González [a,b], Javier LLorca [a,b,*]

[a] IMDEA Materials Institute, C/Eric Kandel 2, 28906 Getafe, Madrid, Spain
[b] Department of Materials Science, Polytechnic University of Madrid, 28040 Madrid, Spain

ARTICLE INFO

Article history:
Received 10 July 2014
Received in revised form 5 February 2015
Accepted 8 February 2015
Available online 19 February 2015

Keywords:
A. polymer-matrix composites (PMC's)
C. Finite element analysis
C. Transverse cracking
C. Computational mechanics

ABSTRACT

Transverse cracking in cross-ply carbon/epoxy and glass/epoxy laminates in tension is analyzed by means of computational micromechanics. Longitudinal plies were modeled as homogenized, anisotropic elastic solids while the actual fiber distribution was included in the transverse plies. The mechanical response was obtained by the finite element analysis of a long representative volume element of the laminate. Damage in the transverse plies was triggered by interface decohesion and matrix cracking. The simulation strategy was applied to study the influence of ply thickness on the critical stress for the cracking of the transverse plies and on the evolution of crack density in $[0_2/90_{n/2}]_s$ laminates, with $n = 1$, 2, 4 and 8. It was found that the transverse ply strength corresponding to the initiation and propagation of a through-thickness crack was independent of the ply thickness and that the transverse strength of carbon/epoxy laminates was 35% higher than that of the glass fiber counterparts. In addition, the mechanisms of crack initiation and propagation through the thickness as well as of multiple matrix cracking were ascertained and the stiffness reduction in the 90° ply as a function of crack density was computed as a function of the ply thickness.

1. Introduction

Fiber-reinforced polymers are nowadays extensively used in engineering applications which require high specific stiffness and strength. They present several different physical failure mechanisms and transverse ply cracking (also denominated matrix cracking) is very often the first one to develop under the application of thermal or mechanical loads. Nucleation and propagation of matrix cracks do not normally lead to structural collapse but degrades very rapidly the laminate resistance to permeation and leakage, limiting the application of cracked laminates in pressure vessels, fuel tanks, etc. Moreover, transverse ply cracks induce interply delamination which may have more serious consequences from the structural viewpoint.

The mechanics of matrix cracking is now well established on the basis of extensive experimental campaigns and of analytical models. The first microcrack causes negligible changes in the thermo-mechanical response of the laminate, but the crack density (the number of cracks per unit length) increases with the applied strain until saturation, leading to material degradation in terms of a moderate to significant loss in the ply transverse stiffness and shear modulus. Extensive reviews of matrix cracking and its effect on the behavior of composite laminates are available in the works of Nairn [1] and Talreja and Singh [2].

The onset of matrix cracking and the crack density at saturation is dictated by a number of factors, including the quality of the material (which is controlled by fiber distribution as well as by the mechanical properties of the matrix and of the fiber/matrix interface) together with the laminate stacking sequence and thickness [1]. Parvizi et al. [3,4] studied the influence of the ply thickness on the strain-to-failure of the 90° layers on glass–epoxy cross-ply $[0/90]_s$ laminates by varying the relative thickness of the inner 90° layers with respect to the supporting 0° plies. They found that the thinner the internal layers, the higher the critical strain for crack initiation and the crack density. For sufficiently thin inner plies, matrix cracking could eventually be suppressed prior to the failure of the supporting layers which determines the final collapse of the specimen. In addition, the position of the 90° layers relative to the supporting plies also influences the critical strain and Nairn [1] showed that matrix cracking occurs at early stages in $[90/0]_s$ laminates, due to the lack of constrain on the external plies. Similar trends were reported in [5,6] for $[\pm\theta/90]_s$

* Corresponding author at: IMDEA Materials Institute, C/Eric Kandel 2, 28906 Getafe, Madrid, Spain.
E-mail address: javier.llorca@imdea.org (J. LLorca).

http://dx.doi.org/10.1016/j.compscitech.2015.02.008
0266-3538/

carbon–epoxy laminates, with different θ values. To take advantage of this phenomenon, Kawabe et al. [7] developed the tow-spreading technique which can be used to produce prepreg tapes with a thickness below one-third of that of conventional laminates, leading to thin ply laminates which present outstanding damage tolerance [8,9].

The mechanics of transverse matrix cracking in cross-ply laminates was analyzed by Dvorak and Laws [10] from a fracture mechanics perspective. They assumed the presence of an initial crack nucleus and computed the energy release rate for the crack propagation through the ply thickness and parallel to the fibers. Crack propagation occurs when the energy release rate is equal to the fracture energy G_c. In the case of uniaxial tension perpendicular to the fibers, the critical stress that leads to the through-thickness crack propagation, Y_T^{tt}, is given by

$$Y_T^{tt} = \sqrt{\frac{2G_c}{\pi \delta_c \Lambda_{22}}} \tag{1}$$

where δ_c is the crack length in the through-the-thickness direction ($<< t$, the ply thickness) and $\Lambda_{22} = 2(1/E_1 - \nu_{12}^2/E_2)$ where E_1, E_2 and ν_{12} stand for the longitudinal and transverse elastic modulus of the ply and the in-plane Poisson's ratio, respectively. Dvorak and Laws [10] assumed that the initial crack nucleus was controlled by the microstructural inhomogeneities in the composite and thus Y_T^{tt} was independent of the ply thickness.

The energy release rate for through-thickness propagation remains greater than the energy release rate for propagation parallel to the fibers (tunneling) until the crack approaches the interface. Assuming that the through-thickness crack has spanned the whole ply thickness, the critical stress for the continuation of crack propagation along the fibers, Y_T^l, under uniaxial tension perpendicular to the fibers is given by

$$Y_T^l = \sqrt{\frac{8G_c}{\pi t \psi_l}} \tag{2}$$

where ψ_l is a coefficient (of the order of the unity) that takes into account the constraint of the adjacent plies [11]. Transverse ply cracking requires an applied tensile stress higher than both Y_T^{tt} and Y_T^l. Whether $Y_T^{tt} > Y_T^l$ or *vice versa* depends on the size of the initial crack nucleus, δ_c and on the ply thickness t but it is evident that the transverse cracking stress is given by Eq. (1) in very thick plies. As the δ_c is unknown, Dvorak and Laws [10] related the *in situ* transverse strength of thick plies to the transverse tensile strength measured on an unconstrained unidirectional ply, Y_T, as:

$$Y_T^{tt} = 1.12\sqrt{2}Y_T \tag{3}$$

where the factor 1.12 accounts for the stress intensity magnification of a surface crack. In the case of very thin plies, the transverse strength was given by Eq. (2) and it was proportional to $1/\sqrt{t}$.

These predictions for the behavior of thick and thin plies were in good agreement with the experimental observations [3,5,6] but the model was not able to predict the critical ply thickness that separates "thin" from "thick" plies because δ_c is unknown. Moreover, the actual value of the critical stress for crack initiation, Y_T^{tt} in Eq. (1), cannot be predicted because is a function of the unknown size of the initial crack nucleus, while the validity of the *in situ* ply strength in Eq. (3) has not been ascertained [12]. Nevertheless, these limitations can be overcome by means of computational micromechanics, which can account for the influence of matrix, fiber and interface properties on the onset and development of matrix cracking. Computational micromechanics is emerging in recent years as a powerful tool to predict the influence of the constituent properties on the ply behavior under different loading conditions, including transverse compression [13], shear [14] and fracture [15–17]. This information can be used as input in multiscale modeling strategies aimed at predicting the laminate and component behavior [18]. Computational micromechanics is used here to analyze the effect of ply thickness on the onset and development of transverse ply cracking in cross-ply carbon/epoxy and glass/epoxy laminates.

2. Computational micromechanics strategy

The onset and development of transverse matrix cracking in cross-ply $[0_2/90_{n/2}]_s$ laminates was studied by means of the finite element analysis of a representative volume element of the material (RVE). The RVE is rectangular with length L = 10 mm and thickness $t = 2(\frac{n}{2}+2)t_0$ where t_0 is the thickness of a single ply, Fig. 1(a). t_0 was equal to 68.75 μm in the case of glass fiber composites and to 34.38 μm in the case of carbon fiber composites to account for the smaller radius of carbon fibers. The length of the RVE was long enough to compute accurately the increment in crack density upon deformation along the x direction. Obviously, this 2D model can only account for the through-thickness initiation and propagation of cracks. Simulation of crack tunneling will require a full 3D simulation which is out of the scope of this paper.

The 0° plies were assumed to be homogenized, transversally isotropic elastic solids, with equivalent effective properties, while the actual random fiber spatial distribution was included in the 90° ply, Fig. 1(a). The fiber radius was constant and equal to R = 9 μm for E glass fibers and to 4.5 μm for AS4 carbon fibers and the fibers were dispersed in the 90° ply using the modified random sequential adsorption algorithm [19]. This algorithm provides random fiber distribution within the RVE while imposing some limitations on the minimum distance between the fiber surfaces (>0.07R) and between the fiber surface and the ply edges (>0.1R) to avoid the presence of distorted finite element during meshing. The fiber volume fraction within the RVE was set to 65%. It should be indicated that the fibers intersecting the internal edges with the 90° plies were removed from the model to represent the typical matrix rich region between adjacent plies with different angle directions, Fig. 1(b), but it was ensured that the final volume fraction was always 65%. The RVE was discretized with generalized plain strain isoparametric four-noded quadrilateral elements (CPEG4 in Abaqus/Standard [20]). Special care was taken to obtain a good mesh discretization between fiber ligaments, as this is necessary to capture adequately the complex stress gradients in these regions, Fig. 1(c). In addition, four-noded isoparametric cohesive elements (COH2D4 in Abaqus/Standard [20] with thickness of 10^{-3} μm) were inserted at the fiber/matrix interfaces to address interface decohesion during the simulations.

Four different RVEs were generated, corresponding to n = 1, 2, 4 and 8, to analyze the influence of ply thickness on the mechanics of matrix cracking. The fiber distribution in a central slice of the RVE is shown in Fig. 2 for the different RVEs. The total length of the RVEs is always 10 mm along the x axis (not included in Fig. 2) while the thickness of the central 90° ply is nt_0.

Periodic boundary conditions were applied to the edges of the RVE to maintain the continuity between adjacent RVE's. They can be expressed as:

$$\begin{aligned} \vec{u}(0,z) - \vec{\delta_x} &= \vec{u}(L,z) \\ \vec{u}(x,-t/2) - \vec{\delta_z} &= \vec{u}(x,t/2) \end{aligned} \tag{4}$$

where $\vec{\delta_x} = (\delta_x, 0)$ stands for the imposed displacement vector along the x direction and $\vec{\delta_z} = (0, \delta_z)$ is computed from the condition that the average stresses along the through-the-thickness direction