Composites: Part A 79 (2015) 127–137

Contents lists available at ScienceDirect

Composites: Part A

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

Micro-mechanical analysis of the effect of ply thickness on the transverse compressive strength of polymer composites

A. Arteiro [a], G. Catalanotti [a,b], A.R. Melro [c,d], P. Linde [e], P.P. Camanho [a,b,*]

[a] DEMec, Faculdade de Engenharia, Universidade do Porto, Rua Dr. Roberto Frias, s/n, 4200-465 Porto, Portugal
[b] INEGI, Universidade do Porto, Rua Dr. Roberto Frias, 400, 4200-465 Porto, Portugal
[c] ISVOUGA — Instituto Superior de entre Douro e Vouga, Rua António de Castro Corte Real, Apartado 132, 4520-181 Santa Maria da Feira, Portugal
[d] CIDEM/ISEP — Instituto Superior de Engenharia do Porto — Instituto Politécnico do Porto, Rua Dr. António Bernardino de Almeida, 431, 4249-015 Porto, Portugal
[e] AIRBUS Operations GmbH, Kreetslag 10, 21129 Hamburg, Germany

ARTICLE INFO

Article history:
Received 2 April 2015
Received in revised form 17 September 2015
Accepted 21 September 2015
Available online 28 September 2015

Keywords:
A. Laminates
B. Transverse cracking
C. Micro-mechanics

ABSTRACT

A micro-mechanical model is used to study the effect of ply thickness on constrained 90° plies subjected to transverse compressive loading (*in situ* effect). For cross-ply sublaminates with conventional, standard-thickness 90° plies, failure is dominated by fibre–matrix interface cracking and large localised plastic deformation of the matrix, forming a localised band in a plane that is not aligned with the loading direction. Ultra-thin plies show a dispersed damage mechanism, combining wedge cracking with ply fragmentation/separation. Moreover, a transverse crack suppression effect is clearly observed. To the authors' knowledge, it is the first time an *in situ* effect in transverse compression has been identified. When comparing the results of the micro-mechanical model with the predictions from analytical models for the *in situ* effect, the same trends are obtained. These results also show that, for realistic ply thicknesses, these analytical models can be considered fairly accurate.

1. Introduction

The spread tow thin-ply technology [1,2] is a recent technique which incorporates a pneumatic process where fibre tows are continuously and stably opened to produce flat and straight plies with dry ply thicknesses as low as 0.02 mm. Laminates made of these ultra-thin plies have the potential to suppress both microcracking and delamination before ultimate failure, which results in an increasing interest by the scientific and industrial communities in this technology [3–25].

The improved fibre dispersion resulting from tow spreading has motivated the study of the effect of ply thinness on the mechanical response of unidirectional (UD) laminae. Amacher et al. [21] performed an extensive experimental characterisation of the mechanical response of thin-ply laminates, including testing of UD carbon fibre-epoxy composites with ply areal weights ranging from ultra-thin to very high grades, all produced from the same batch of fibre and matrix. No significant influence of the ply thinness on the elastic and strength properties of the UD composites was observed, except for longitudinal compression, where an enhancement of approximately 20% in average was observed. This enhanced compressive strength is attributed to the more uniform microstructure of spread tow thin plies. Optical micrographs presented by Amacher et al. [21] show that the microstructure of high-grade composites is fairly inhomogeneous, with varying fibre volume fraction along the microstructure due to fibre rearrangement and resin flow during the low-viscosity phase of the curing cycle. This heterogeneous microstructure promotes instabilities at the constituent level that may result in premature compressive failure of the laminae [21]. As the ply thickness decreases, a better uniformity of the microstructure is achieved, becoming practically homogeneous for the lowest grades, delaying the instabilities that conduct to fibre compressive failure, therefore increasing the compressive strength.

When very thin plies are embedded in multidirectional laminates, the thin-ply effect becomes particularly important. In this case, the constraining effect imposed by the neighbouring plies delays damage propagation in the matrix [21,26], and the actual ply strengths are not only higher than those measured in UD coupons, but they reportedly increase with decreasing ply thickness [11,27–35].

Experimental studies have shown that the longitudinal compressive and transverse shear strengths are affected by the

* Corresponding author at: DEMec, Faculdade de Engenharia, Universidade do Porto, Rua Dr. Roberto Frias, s/n, 4200-465 Porto, Portugal. Tel.: +351 225081753; fax: +351 225081445.
E-mail address: pcamanho@fe.up.pt (P.P. Camanho).

http://dx.doi.org/10.1016/j.compositesa.2015.09.015
1359-835X/

presence of transverse microcracking [30]. If transverse microcracking is delayed, the strengths associated with other matrix-dominated failure mechanisms, such as wedge compressive fracture, may also increase due to the constraining effect imposed by adjacent plies. This can be addressed through the application of three-dimensional (3D) phenomenological failure criteria [36,37]. According to these models, when embedded in a multidirectional laminate, not only the transverse tensile and in-plane shear strengths (Y_T and S_L, respectively), calculated using e.g. the models proposed by Camanho et al. [33], but also the transverse compressive and transverse shear strengths (Y_C and S_T, respectively) are *in situ* properties. In addition, assuming that kink bands[1] are triggered by localised matrix failure in the vicinity of misaligned fibres [36,37], the *in situ* effect will also result in an increased longitudinal compressive strength in multidirectional laminated composites.

In a previous work [22], a micro-mechanical finite element (FE) model of a composite sublaminate was proposed to study the mechanics of the *in situ* effect observed on constrained plies subjected to transverse tensile loading. This micro-mechanical model consisted of a representative volume element (RVE) of a 90° thin lamina in-between two homogenised 0° plies. Random distributions of carbon fibres in an epoxy matrix, statistically equivalent to real distributions, were analysed using a 3D computational micro-mechanics framework, with a special focus on the elastic–plastic and damage constitutive behaviours of the matrix and on the response of the fibre–matrix interface. Varying the 90° ply thickness, it was possible to assess its effect on the mechanical response of laminated composites.

The objective of this paper is to investigate whether there is an *in situ* effect for transverse compressive loads as predicted by some failure criteria [36,37]. In the absence of experimental information, a micro-mechanical model is used to study the *in situ* effect observed on constrained plies subjected to transverse compressive loading. Also, a comparison between the results of the micro-mechanical model and the predictions from the analytical models for the *in situ* effect based on Linear Elastic Fracture Mechanics (LEFM) [33] and phenomenological 3D failure criteria [36,37] is presented with the aim of validating both modelling strategies.

2. Micro-mechanical constitutive and finite element modelling

An FE model of the thin-ply sublaminate is generated, consisting of three main parts: a micro-mechanical RVE of the 90° ply, two adjacent homogenised 0° plies, and the interfaces between the 90° lamina and the homogenised plies [22]. The RVE of the 90° ply includes a discrete representation of the fibres, matrix and interfaces between fibres and matrix.

A random distribution of carbon fibres is generated using a modification of the algorithm proposed by Melro et al. [39], with imposed fibre continuity along the faces perpendicular to the *yy*-direction for implementation of periodic boundary conditions (PBCs)—see Fig. 1, where the *xx*-direction coincides with the longitudinal (fibre) direction of the discretised transverse ply (normal to the surface of the page), the *yy*-direction coincides with the in-plane transverse direction of the discretised transverse ply (horizontal direction), and the *zz*-direction coincides with the out-of-plane (through-the-thickness) transverse direction (vertical direction). Unlike 3D random RVEs with PBCs for analysis of UD composites [39,40], in the present study the entire thickness of the discretised transverse ply is explicitly represented, so its effect in the response of the sublaminate can be taken into account. The faces of the discretised transverse ply perpendicular to the *zz*-direction (top and bottom faces) will be connected to the homogenised outer plies, and, therefore, in the generation of the RVE, fibres are not allowed to intersect these faces.

In the present study, the individual carbon fibres are considered transversely isotropic and linear-elastic [22]. Following Melro et al. [41], the diameter of the carbon fibres is the same in the entire RVE.

The epoxy resin is modelled using the plastic-damage material model proposed by Melro et al. [40]. This material model, implemented as an UMAT user subroutine [42], includes a paraboloidal yield criterion, defined as a function of the stress tensor and of the compressive and tensile yield strengths, and a non-associative flow rule, which allows for a correct definition of the volumetric deformation in plasticity. Fig. 2 shows the hardening curves used in the plasticity model for both tension and compression. A thermodynamically consistent isotropic damage model, defined by a single damage variable, is used, where damage onset is defined by a damage activation function similar to the paraboloidal yield criterion, but using the final compressive and tensile strengths of the epoxy resin. To avoid damage localisation, the computed dissipated energy is regularised using the characteristic length of the FE and the fracture toughness of the epoxy resin [43].

The interfaces between fibres and matrix are modelled using cohesive elements [22,41,42], defined by a bilinear traction-separation damage law. The onset of cohesive damage is mode dependent, and is defined by the corresponding strengths in mode I and mode II. The rate of damage progression is controlled by the critical energy release rate under pure and mixed modes, according to the Benzeggagh–Kenane (BK) law [44].

Different 3D RVEs of the thin-ply sublaminate are generated [39]. These include different random fibre distributions for the same RVE's size, which are analysed to assess the effect of microstructural randomness. The width of the RVEs of the thin-ply sublaminates along the *yy*-direction (see Fig. 1) is 0.200 mm. This was defined to ensure the representativeness of the different RVEs analysed in the present work; in fact, it is important to note that the width of the RVEs should be defined in such a way that any diffuse damage, which might occur before a transverse crack has grown entirely through the thickness of the ply, can be captured [22]. In a compromise between the computational cost of the proposed models and the results obtained, it was observed that an RVE width of 0.200 mm was adequate to capture the diffuse damage occurring on the thinner transverse plies. However, due to the enormous computational cost of these models, the RVEs of the sublaminates with transverse ply thicknesses above or equal to 0.100 mm were modelled to accommodate approximately a single transverse crack, as in Ref. [22], reducing the total width of the RVEs to 0.120 mm. As the transverse ply gets thicker, the size of the models becomes so large that they cannot be handled. On the other hand, it was observed that, for such transverse ply thicknesses, the diffuse damage, before transverse cracking grew through the thickness, was very limited (similarly to what was observed in Ref. [22] for transverse tension). Therefore, defining the width of the RVE such that a single transverse crack could be captured was sufficient to study the damage morphology and predict failure of the thicker transverse plies. By reducing the RVE width, from 0.200 mm to 0.120 mm for transverse ply thicknesses of 0.100 mm and above, was enough to capture transverse cracking and keep the computational cost of the models in reasonable values. It is recognised that the size of the RVE may affect the predicted material response, particularly during softening [41]; however, the analysis of the RVEs with thicker transverse plies is still important to accurately address the causes of matrix transverse compressive failure as the ply thickness changes.

The length of the micro-mechanical models (*xx*-direction) is kept constant and approximately equal to two times the average

[1] Crack-like type of failure occurring in laminae subjected to compressive loadings in the fibre direction [38].