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Micromechanical study on the origin of fiber bridging under interlaminar and intralaminar mode I failure

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Abstract

Fiber reinforced polymers (FRP) subjected to mode I fracture show important toughening due to the development of large scale bridging (LSB). Experimental studies of this phenomenon in unidirectional carbon/epoxy laminates using double cantilever beam specimens, demonstrate important differences in R-curve response for inter- and intralaminar fracture. Post fracture observation of composites cross-section pointed out dissimilar fiber bundle size and shape, as the main origin of their differences. In the present paper, representative volume elements with the composite constituents, based on the actual material microstructure, and homogenized 2D finite element models were developed to study the effects of microstructure on the first stage of damage leading to LSB development in carbon/epoxy composites under mode I fracture. The differences between inter- and intralaminar fracture were investigated along with the influence of fiber dispersion and the presence of interply and intraply resin-rich zones. The numerical simulations captured different microcrack morphologies for inter- and intralaminar fracture, supporting the experimental observations, while parametric studies showed the influence of the microstructure in the formation of LSB. In particular, fiber dispersion within a ply and resin rich zone between plies play significant roles in mode I fracture and can be used to control toughening mechanisms in FRPs.

Keywords: Polymer-matrix composites (PMCs); Computational micromechanics (RVE); Intralaminar; Interlaminar; Microstructure; Fiber bridging & fracture toughness.

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1. Introduction

Fiber Reinforced Polymers (FRPs) are nowadays extensively used in applications where outstanding mechanical properties (stiffness and strength) are required in combination with weight optimization. Certainly, the study of failure in FRPs is still a challenging problem [1, 2, 3], despite all current knowledge about these materials. Indeed, the accurate prediction of the failure envelope of composite materials and structures has been an elusive task due to the complexity of the failure mechanisms involved [4] and for that reason an important amount of resources and time have been invested in understanding the micromechanisms involved in damage accumulation and failure under different loading conditions [5, 6, 7, 8]. So far, the material strengths corresponding to isolated or combined failure modes have been predicted, with reasonable accuracy especially for UD laminates [9, 10, 11, 12], including non-homogeneous microstructures [13]. However, the problem becomes even more challenging when dealing with fracture, and crack propagation, of FRPs due to their anisotropy and the complex microcracking evolution and coalescence with combination of fiber-matrix interface, matrix and fiber failure.

Among the different types of fracture in FRPs, mode I is perhaps the most intensively studied in terms of toughness and crack propagation due to the low critical values at initiation. Nevertheless, important toughening phenomena are present, like fiber bridging where single fibers or fiber bundles are created and pulled out during the fracture process. This fracture mode can be tested using Double Cantilever Beam (DCB) specimen geometry. The bridging zone developed during such tests often has a length comparable to specimen's size, leading to the so-called Large Scale Bridging (LSB). Therefore, the resistance curves (R-curves), relating the Energy Release Rate (ERR) and crack growth increment, a , change according to geometrical parameters like specimen's thickness [14, 15, 16, 17]. Therefore, it cannot be considered as a material property [17, 18, 19] and advanced techniques, such as the inverse-numerical approach based on reliable strain acquisition along the bridging zone by means of fiber Bragg grating sensors, provided a robust scheme of bridging tractions evaluation [20, 21]. This method, was also used by [18] and [19] to characterize the traction separation relations in unidirectional laminates of different thicknesses for interlaminar and intralaminar fracture respectively, with LSB phenomena. The comparison of the aforementioned studies reveals

an important effect of the microstructure (inter- and intralaminar crack propagation) in the fracture response. Micro/mesoscale observations on those studies report different bridging bundle morphology in terms of their size and population. Individual bridging fibers are found in interlaminar fracture [18], together with a few bundles of elongated section, as well as in [22] for another carbon-epoxy system. On the other hand, in intralaminar fracture completely different morphology is reported [19] containing larger bundles and a rather irregular fracture cross-section, analogous to the behavior of inter-intralaminar fracture reported in [23]. While differences in inter- and intralaminar fracture for the same material have been reported in the literature, the origin of these differences is still unclear. Thus, the objective of this paper is to investigate the effect of the microstructure on inter- and intralaminar fracture using experimentally obtained microstructures and computational micromechanics.

In this work, DCB experiments for the same carbon-epoxy reported in [18, 19] were used to reproduce inter- and intralaminar fracture R-curves. The postmortem analysis of the material cross-section revealed major differences in the microstructure in terms of fiber dispersion, potentially related with the formation of bridging bundles. To investigate the effect of the microstructure on different loading conditions (inter- or intra- laminar), or even for the same ones, as captured with the current experimental results, a thorough micromechanical study of the cross-sectional microstructure was performed and implemented in computational models using parametric Representative Volume Elements (RVEs) and homogenized microstructures. These models aim to explore the influence of the microstructure during the first stages of the development of LSB and define the parameters that cause bundle formation with an optimal shape for the highest toughening effect.

2. Experimental methods

2.1. Material and sample preparation

In this work, the effects of two experimentally extracted microstructures on the R-curves are examined. The composite material is the carbon/epoxy pre-impregnated (prepreg) system SE-70, manufactured by Gurit that was received in two batches and at different times. These batches are referred to as "batch 1" and "batch 2", respectively. While the fracture response of specimens from batch 2 is investigated in the present work, the fracture resistance of

similar specimens from batch 1 is reported in [18, 19] and compared with the results of the present studies. In order to evaluate the mode I interlaminar and intralaminar toughness of the composite, DCB specimens were cut out from two different laminates, coming from the same prepreg roll (batch 2). The plates were produced by hand-lay-up and manufactured using standard autoclave procedure applying a uniform pressure of 3 bar. The material was cured at a maximum temperature of 78 °C for a total curing time of 16 h, following the recommendations provided by the supplier as it was also done in [18, 19] for batch 1.

During manufacturing of the first laminate, devoted to the interlaminar fracture specimens, a release film (13 μm thickness) was inserted in the mid-plane of the laminate to create a 60 mm long starter crack (Figure 1a). A thick aluminum cap was set on top of the plate to improve the pressure distribution during curing and provide a homogeneous composite plate thickness. Since the nominal ply thickness, t_p , is around 200 μm the resulting laminate thickness after curing was $H = 4 \pm 0.05$ mm, for a total of 20 stacked plies and the final nominal dimensions of the panel were 280x200 mm². Stripes of about 10 mm from the outer edges were cut-off the plate to avoid the undesired effects resulting from the manual lay-up (higher porosity, poor fiber alignment, loss of flatness etc.). Small sections of each plate were examined under an optical microscope, after curing, showing a high quality composite material with good ply consolidation (void content around 1%). The DCB specimens for interlaminar fracture were produced following the ASTM Standard D5528-13 [24]. Therefore, specimens with a rectangular cross section of 25x4 mm² were cut out from the plate (Figure 1b).

A second plate, with thickness of 6 mm was produced for intralaminar fracture testing. This plate was manufactured by stacking 50 carbon/epoxy plies and using the same standard autoclave procedure and curing cycle as in the first one. Similar molding technique with interlaminar fracture specimens was used with an aluminum plate. Again, regions close to the outer edge were discarded. The final dimensions of the panel were 360x200x(10 $\pm$ 0.06) mm³. Due to the lack of standardized methods, the geometry of the intralaminar-DCB specimens was chosen based on the feasibility of their fabrication. In this work, beams with a rectangular cross-section of 10x6 mm² were cut out from the plate using the same procedure followed by the authors in previous works [19, 25]. Since the thickness of the composite panel corresponds to the width of the intralaminar-DCB specimens, the plate thickness was set to $B = 10$ mm

(Figures 1c and 1d). The total height of the DCB, H , was set to 6 mm to provide a bending stiffness comparable to the one of the interlaminar specimens. In this case, the 60 mm initial crack was introduced using a diamond wire saw (130 μm nominal diameter). The length, L , of intralaminar DCB specimens was set to 340 mm, large enough to discard any influence of the specimen's end at the crack tip during the fracture test.

2.2. Mode I fracture testing

The DCB tests were carried out following the ASTM Standard D5528-13, under displacement control at a constant rate of 3 mm min⁻¹ in an Instron[®] 5848 MicroTester electromechanical testing machine, equipped with a 2 kN load cell. The load was applied through two pairs of plain bearings and forks, pinned to a set of stainless steel loading blocks glued on each arm of the DCB samples. To monitor the crack advance, one side of each specimen was painted in white and a fine ruler (1 mm step) was grated on it. The crack propagation was then tracked using a high resolution CCD camera with a data acquisition rate of 1 Hz, providing an accuracy of at least ± 0.25 mm.

The mode I ERR was initially computed combining the load-displacement curves and the crack advance, Δa , measurements using the Modified Compliance Calibration (MCC) method recommended in the ASTM Standard [24]. In addition, the relative rotation between the DCB arms, θ , was measured using a second high resolution camera, which focused two targets painted on two aluminum arms that were glued to the loading blocks [25]. This parameter, together with the applied load, P , and the specimen width, B , were used to evaluate the ERR based on the calculation of the J-integral as $J = P\theta/B$ [26]. While the methods proposed by the ASTM standards assume linear response, the method using the J-integral, which accounts for geometrical and material non-linearity, provides a more realistic measurement of the ERR. Here, differences of up to 8% were found between the two data reduction methods. Similar results are reported in [19, 22].

2.3. Microstructure analysis

In order to understand the influence of the microstructure during the onset of LSB, the composite cross section must be studied. To this end, a few tested interlaminar and intralaminar DCB specimens were transversely sectioned around the crack tip. The resulting

material blocks were then embedded into mounting resin and polished using a sequence of SiC grinding papers (P500-P2400) of decreasing roughness. Since the scope of this microscopic study was to characterize the fiber distribution and size, a very fine polishing was required. Therefore, three additional steps, including the addition of 9 μm , 3 μm and 1 μm grain size polycrystalline diamonds, were used.

After polishing, the cross section of the CFRP was observed using a Keyence VTX-5000 digital microscope. The resulting images were then analyzed using ImageJ open source software (Figure 3b). As a result, local and global fiber volume fraction ($V_f \approx 54\%$) together with fiber diameter distribution were obtained (Gaussian distribution, $6.7 \pm 0.1 \mu\text{m}$). This information is part of the input required for the generation of the finite element model explained later on.

3. Computational models

Assuming that the DCB test guarantees pure mode I fracture in the specimen with a crack front transverse to fiber's direction, a pure transverse tensile stress state is considered to be representative for the region of the composite material just ahead of the crack front (Figure 2). Therefore, crack formation can be analyzed by using a two-dimensional (2D) plane strain finite element (FE) model. Since we are interested in the influence of the material microstructure on the onset of LSB, the model was based on computational micromechanics and the microstructure of the unidirectional composite is idealized as a dispersion of parallel circular fibers distributed in the polymer matrix. This RVE model spans a length equivalent to ~ 8 plies of the laminate.

A second model was also developed to explain some aspects of crack propagation during intralaminar failure where the composite cross-section is represented as a homogenized linear elastic transversely isotropic material. The main crack, corresponding to the symmetric fracture plane, is approximated using a cohesive contact and a set of cohesive planes is introduced to simulate crack initiation and propagation within a ply, located at the longitudinal symmetry plane (Figure 4). Both models are explained in detail in the next paragraphs.

3.1. RVE model concept

The input required to generate the RVE comes from the microscopic analysis of the material cross section. The fiber volume fraction, V_f , and the fiber diameter distribution are used here. According to the most common techniques, the first step required to set-up a micromechanical model based on an RVE is the generation of a fiber distribution statistically equivalent to the real one. To this end, several strategies are available in the literature [27, 28], with the Random Sequential Absorption (RSA) [29] and the Nearest Neighbor Algorithm (NNA) [30] being the most popular ones, due to the ease to achieve high fiber volume fraction. However, both algorithms experienced serious problems when dealing with microstructures spanning several plies within a laminate.

For a given V_f the RSA performs quite well generating homogeneous dispersion of fibers of the same diameter, even if a relative large number of fibers and/or high V_f is involved [31]. However, this is not the case for the composite material investigated here. The Gurit-SE70 presents a non-homogeneous microstructure, which means that the local V_f varies within the ply. On top of that, important resin-rich areas, of variable thickness, are produced during the curing (Figure 3). The latter, will play a critical role differentiating the interlaminar crack propagation from the intralaminar one and is further explained in this article. Although it can be modified to reproduce fiber clusters and diameter variation, the RSA is not efficient to generate microstructures containing fiber clusters or resin-rich zones.

Regarding the NNA, it can capture fiber distribution patterns present in the real microstructure and fibers of different diameters. However, it requires more complex iterations and much more statistical information than the RSA, namely fiber diameter distribution and distances to the first and second neighboring fibers. This information must be extracted from the micrographs [30], which is not a problem when dealing with medium size microstructures, like the ones used to predict ply's transverse strength [5, 32]. However, when the studied microstructure is large and includes large resin-rich areas, the performance of the algorithm is quite poor. The convergence of the algorithm becomes slow and the final volume fraction remains close to the so called "jam limit" ($V_f = 54\%$). Most likely, the poor performance of the algorithm is associated to the dispersion of the statistics obtained from such "layered" microstructures (Figure 3).

Both the RSA and NNA are included in the Virtual Ply Property Predictor (VIPER) Python-based software [33] together with a more powerful approach based on Potential Minimization Algorithm (PMA) [34]. This new algorithm can generate very large microstructures with high fiber volume fractions for circular or non-circular fibers, a few thousand times faster than the RSA for equivalent microstructures. However, when using the PMA, the microstructure generation is controlled by a single statistical descriptor, V_f , and, hence, it has the same problem as the RSA, namely, the fiber dispersion should be homogeneous. The adopted solution, consists of generating large homogeneous microstructures, using the PMA, with a high fiber volume fraction ($61\% < V_f < 66\%$), corresponding to the denser regions in the ply, and then creating resin rich zones by removing fibers until the RVE's volume fraction corresponds to the global one measured in the real micrograph ($V_f \approx 54\%$). **This framework was proven to be the most robust taking advantage of the efficiency of PMA algorithm and also allowing for generation of microstructures with variations of the local V_f .**

Once the periodic microstructure has been generated using the described approach, the 2D FE model is created using the commercial software ABAQUS [35] by means of an in-house developed Python script. As a result, two $1600 \times 200 \mu\text{m}^2$ RVEs were generated, corresponding to the microstructure "seen" by the interlaminar and the intralaminar crack front (section 4.2). The periodic RVE's were then discretized using isoparametric 2D plane strain wedge and brick FE for fibers and matrix with full integration at Gauss points (CPE3 and CPE4). The fiber/matrix interface debonding was modeled by inserting cohesive elements (COH2D4) around each fiber. Typically, each RVE contained approximately $\approx 1,200,000$ elements representing a discretization fine enough to capture the large stress gradients between neighboring fibers. Node positions on opposite edges of the RVE's are identical to apply periodic boundary conditions [29]. Simulations were carried out with Abaqus/Standard within the framework of the finite strain theory. Similar approach is adopted by several authors [6, 31, 11], resulting in a realistic crack propagation.

The RVE's were initially subjected to a homogeneous temperature drop of $\Delta T = -58^\circ\text{C}$, from the curing (78°C) to room temperature (20°C), hence generating realistic residual stress states in the material before mechanical loading. **This step is included to increase the accuracy of the model, even-though the temperature gradient is not significant compared to other**

high curing temperature systems. In a second step, homogeneous transverse tension stress state was introduced by applying the appropriate displacements to the master nodes linked with the periodic boundary conditions [5].

The same procedure was repeated with RVEs of smaller size ($360 \times 360 \mu\text{m}^2$) with variable microstructure to define the parameters affecting the formation of the crack front and corresponding fiber bundles.

3.2. Mechanical behavior of the RVE constituents

The carbon fibers were modeled as linear elastic and transversely isotropic solids defining five independent elastic constants (E_{f1} , E_{f2} , ν_{f12} , G_{f12} , G_{f23}) and two different thermal expansion coefficients (α_{f1} , α_{f2}) (-1- is fiber's direction). This set of elastic constants for the carbon fibers was obtained using the Chamis rule of mixtures [36] and the elastic properties of the SE-70 were taken from [19]. The two thermal expansion coefficients are those reported in [33] for a similar material.

The polymer matrix was simulated as an isotropic linear elastic solid with E_m and ν_m as its elastic modulus and Poisson ratio, respectively. The model allows the resin to undergo plastic deformation with the possibility of damage (microcracking). This coupled damage-plasticity model, available in ABAQUS [35], is a modification of the Drucker-Prager plasticity yield surface [37] enriched with a damage variable to capture the quasi-brittle behavior of the polymer under dominant tensile loads. The constitutive equation is based on the yield function proposed by [38] and includes modifications proposed by [39] to account for different strength evolution under tension and compression loads. Details of this constitutive model and of the numerical implementation can be found in [6, 11, 35].

The polymer's elastic modulus, tensile and compressive strengths (E_m , σ_{myt} and σ_{myc} , respectively) are provided by the supplier. The remaining properties, namely, fracture energy, Poisson ratio, coefficient of thermal expansion (CTE) and internal friction angle ($\mathcal{G}_m$, ν_m , α_m and β , respectively) were taken from literature [31] for a similar resin. These properties represent typical values for an aeronautical grade epoxy resin.

Fiber-matrix debonding is modeled by means of a cohesive constitutive equation relating the displacement jump across the interface to the traction vector acting on it for cracking

under the full range of mode-mixities as in [40]. The initial response of the cohesive interaction is assumed to be linear elastic governed by a contact penalty stiffness K . Such numerical parameter should be large enough to ensure displacement continuity in the absence of interface damage while avoiding convergence difficulties due to ill-conditioned stiffness matrix. The onset of debonding is controlled by a quadratic interaction criterion depending on the fiber/matrix interface strengths (normal - σ_N^0 , shear transversal - τ_T^0 , and shear longitudinal - τ_L^0), and stresses (normal - σ_N , shear transversal - τ_T , and shear longitudinal - τ_L), as

$$\left(\frac{\langle\sigma_N\rangle}{\sigma_N^0}\right)^2 + \left(\frac{\tau_T}{\tau_T^0}\right)^2 + \left(\frac{\tau_L}{\tau_L^0}\right)^2 = 1 \quad (1)$$

wherein only positive normal tractions affect the criterion (i.e. $\langle\cdot\rangle$ operator). Once debonding is initiated, the cohesive tractions transferred through the interface decrease linearly to zero by means of a single normalized scalar damage variable. The energy-based Benzeggagh-Kenane (BK) propagation criterion [41] is used to account for the fracture energy dependence on the mode mixity as

$$\mathcal{G}_c = \mathcal{G}_{Ic} + (\mathcal{G}_{IIc} - \mathcal{G}_{Ic}) \left(\frac{\mathcal{G}_{shear}}{\mathcal{G}_{Ic} + \mathcal{G}_{shear}} \right)^m \quad (2)$$

A value of $\sigma_N^0 = 42$ MPa has been assumed for the mode I strength, based on the SE-70 transverse tensile yield stress provided in [19]. The mode II and III strengths were assumed to be equal to 3/2 of that value ($\tau_T^0 = \tau_L^0 = 3/2\sigma_N^0$), as proposed by [42]. Finally, the interface fracture energies for mode I, II and III ($\mathcal{G}_{Ic}$, $\mathcal{G}_{IIc}$ and $\mathcal{G}_{IIIc}$, respectively) were set using micromechanical assumptions. The value of $\mathcal{G}_{Ic}$ was estimated using a linear decay in the cohesive law, with the given σ_N^0 and a crack opening in the order of 1 μm , supported by micromechanical observations in [6]. Regarding, the value of $\mathcal{G}_{IIc}$, it is assumed equal to the resin fracture energy, in order to take into account the matrix crashing and friction dissipation during mode II [11]. The implemented $\mathcal{G}_{Ic}$ and $\mathcal{G}_{IIc}$ values are in line with those reported in the literature [32, 43]. Finally the exponent of the energy-based Benzeggagh-Kenane (BK) damage propagation criterion is set to $m = 1.2$, according to the value found in the literature for a similar material [33]. The constituent material properties used in the present work are all shown in Table 1.

3.3. Homogenized model

Apart from the simulations using the RVE, a second two-dimensional plane strain FE model was used in this work. The aim of this model was to explain the effect of ply and resin layer thickness in the development of the intralaminar crack along the cross section. To this end, the composite was modeled as homogenized, linear elastic and transversely isotropic material. It is therefore described using five independent elastic constants, E_1 , $E_2 = E_3$, $\nu_{12} = \nu_{13}$, $G_{12} = G_{13}$ and G_{23} reported in [19] (-1- is fiber's direction). To reproduce the crack initiation and evolution, a set of cohesive planes is introduced in the model (Figure 4). Basically, there is a large homogenized layer of thickness t_{hg} , corresponding to several composite plies and containing a single fracture plane aimed to reproduce the main crack (located in the laminate symmetry plane). The homogenized properties in this region correspond to the global fiber volume fraction, $V_f = 54\%$, measured in the micrograph analysis. Then, a unique layer-section (the region of interest), containing several fracture planes distributed every $5\mu\text{m}$ was consider in the center of the layer-section. The homogenized properties in this section, fiber-rich layer (thickness, t_{fr}), correspond to a local $V_f = 60\%$, measured from the analysis of micrographs. The same cohesive properties used to model the fiber/matrix interface are used here for simplicity. This cohesive law can be easily modified to match the composite (homogenized) behavior obtained from the RVE model. Both region were separated by a resin-rich layer, t_{rr} , aimed to reproduce the transition between two composite plies. The resin properties are the same ones used in the RVE model, whereas the elastic constants for the homogenized material have been obtained using the Chamis rule of mixtures with the appropriate volume fraction [36] and the fiber properties reported in Table 1.

4. Results and discussion

4.1. Experimental results of inter- and intralaminar mode I fracture

The mode I fracture tests results, i.e. the R-curves, of the material from batch 2 for the interlaminar and the intralaminar cases are compared in Figure 5. The data in Figure 5 show that the ERR at initiation, $J_{I,i}$, is approximately the same: $J_{I,i}^{\text{intra}} = 257 \pm 16\text{J m}^{-2}$ (intralaminar) and $J_{I,i}^{\text{inter}} = 266 \pm 7\text{J m}^{-2}$ (interlaminar). However, the maximum ERR at the steady state propagation, $J_{I,ss}$, is completely different. For the interlaminar case, $J_{I,ss}^{\text{inter}} =$

289 $480 \pm 64 \text{ J m}^{-2}$, whereas for the intralaminar one, $J_{\text{I,ss}}^{\text{intra}} = 1830 \pm 91 \text{ J m}^{-2}$, corresponding
 290 to at least 5 tested specimens. The difference between the two fracture cases is not only
 291 reflected on the $J_{\text{I,ss}}$ values but also on the shape of the R-curve, also seen for batch 1 in
 292 [18, 19] and for carbon/thermoplastic composite in [20, 25]. Whereas in the interlaminar
 293 case the R-curve shows a gradual rising part, in the intralaminar case, this rising is steeper
 294 with a rapid increase of the ERR, which is reflected in a larger extent of LSB observed
 295 macroscopically. The fact that the evaluated $J_{\text{I,i}}$ are very similar, confirms that ERR at
 296 initiation is a material property, which depends in both cases on fibers' volume fraction and
 297 the mechanical properties of the constituents.

298 Also shown in Figure 5 are the corresponding average data for batch 1, as reported in
 299 [18, 19]. It is important to notice that although the same behavior was observed ($J_{\text{I,ss}}^{\text{inter}} \ll$
 300 $J_{\text{I,ss}}^{\text{intra}}$, with $G = J$ in the case of linear response), the values of $J_{\text{I,ss}}$ and shape of the R-
 301 curves obtained in the present work are different to the ones reported in [18, 19] (Figure
 302 5), for the material from batch 1. The only difference between the aforementioned and the
 303 current investigated material is Gurit's production batch, for the same CFRP system (SE-70)
 304 with alike mechanical properties of constituents. Thorough evaluation of the micrographs in
 305 Figure 3 shows that the global fiber volume fraction is constant ($V_f \approx 54\%$), however the
 306 randomness of the microstructure (fiber clustering and resin-rich zone size and dispersion) is
 307 quite distinct between the two batches (Figure 3). Thus, these noticeable differences in the
 308 R-curve's can only be attributed to the microstructure variation between the batches.

309 To further illustrate the origin of the different behavior observed between interlaminar
 310 and intralaminar fracture resistance, a postmortem analysis of the cross section was carried
 311 out for both types of specimen. Thus, a microscopic analysis was carried out, focused in
 312 the cross-section located right ahead of the crack tip which corresponds to the region where
 313 the fiber bundles are formed. The crack patterns for both interlaminar and intralaminar are
 314 shown in Figure 6a and 6b, respectively.

315 The image in Figure 6a shows that, for the interlaminar case, the crack propagates within
 316 a ply, as in [18], and not along the "interface between two adjacent plies", as it is commonly
 317 assumed in the literature and the Standards, where the term "delamination" was devised.
 318 In fact, the crack is prone to initiate and propagate through fiber-rich regions (high local V_f)

319 belonging to the central plies of the laminates, since the stress concentration is higher and
 320 the interface prone to failure. It has also been observed that the crack-front may kink and
 321 move to a high fiber volume fraction region located in an adjacent ply (Figure 6a), creating
 322 flat elongated fiber bundles which produce LSB and a moderate increase of the ERR. This
 323 crack migration is produced due to the inevitable waviness related to the manufacturing
 324 process (impregnation, lay-up & curing kinetics) and also to the presence of resin-rich zones,
 325 namely, inter-ply or intra-ply, since crack propagation is restricted by the presence of low
 326 local V_f regions. However, the crack plane is relatively flat over the cross-section but also
 327 aligned with the laminate's plane due to the symmetry defined by the loading conditions.
 328 It is evident though that the larger the waviness, the larger the crack migration and, as a
 329 result, more intense LSB. Only an ideal laminate with uniform V_f distribution and perfectly
 330 stacked plies (i.e. infinite disturbance wavelength of ply) would have a completely flat crack
 331 profile without LSB formation and the related toughening.

332 In contrast, intralaminar fracture can be described as the opposite case with wavelength
 333 equal to the ply thickness, t_p , which means that the plies are stacked perpendicular to the
 334 crack, i.e., the main crack is normal to laminae sequence. Moreover, fiber bundle sections
 335 are no longer flat, as they were in the interlaminar case, giving rise to increase in sections's
 336 area moment of inertia and to larger bending stiffness compared to the interlaminar ones
 337 and, hence, contributing to higher increase of the ERR (Figure 6b). Analyzing the cross
 338 section of the intralaminar specimen, it can be seen how the presence of the interply resin-
 339 rich zones, created during the fabrication, act as barriers blocking the crack and forcing it to
 340 split in branches (Figure 7). This crack branching phenomenon creates large fiber bundles
 341 because the crack migration from the symmetry plane is no longer restricted by the presence
 342 of low local V_f . Therefore, the crack can move and branch along the cross-section following
 343 the non-uniformity of the local V_f . Material from batch 1 (Figures 7a and b) shown higher
 344 tendency to form such crack branches compared to the material from batch 2 (Figures 7c and
 345 d). Thus, the size of the potential bundles in the former was larger. As a result, the materials
 346 microstructure affects the fracture behavior even for the same loading case (intralaminar).
 347 This influence of the microstructure is the object of the study described in the following
 348 sections.

4.2. Computational micromechanics: prediction of fracture morphology

Figures 8a and b show the created RVEs for the investigation of the interlaminar and intralaminar crack-front creation and the corresponding results. Although the final failure of the RVE is produced, in both cases, by the development of a crack perpendicular to the loading axis, there are some differences in the failure between the interlaminar and intralaminar case. Once the interface fails, plastic deformation and damage under tension in the matrix initiates. The RVE's response becomes nonlinear and the material transverse strength, Y_T , is reached. At this point, corresponding to the maximum load, several cracks are active over the cross section. Then, the load drops due to the combined failure of interface and matrix inside the plies. Up to this stage the failure pattern is similar for both cases.

On one hand, for the interlaminar case (upper part in Figure 8), some of these multiple cracks coalesce into a main one. The RVE is practically split in two parts joined by thin matrix ligaments, where severe plastic deformation and damage are accumulated until ultimate failure (Figures 8 c and d), resulting in a crack pattern rather continuous, disrupted only in a few domains that will lead to the formation of elongated bridging bundles. On the other hand, for the intralaminar case, the initial load drop is followed by damage of the resin-rich zones located between the plies. The presence of these zones delays the crack coalescence into a single one, because the multiple cracks formed during the initial load drop are stopped by the resin-rich zones. The stress field is completely modified due to the formation of the shear bands inside the resin-rich zones, where the plastic strain accumulated and localized. Therefore, the crack kinks and produces branches that lead to the formation of new secondary cracks (Figures 8 e, f & g) able to grow independently. At this point, the crack pattern is rather irregular and contains multiple crack fronts, which constitutes a favorable scenario for fiber bundles creation. The same behavior is observed in tested specimens (Figure 7).

It is worth mentioning that there is practically no plastic deformation in the resin-rich zone for the interlaminar case. The crack is mainly located in an intraply domain in the lower ply. Nevertheless, the presence of intraply high local V_f , causes migration of the crack-front in the upper ply. In the real material, the ply waviness would promote the coalescence of both cracks (crack migration seen in Figure 6), through a resin-rich region of minimum thickness. However, the accumulation of damage in the matrix ligaments made the FE highly distorted

and, hence, the computational time increment required by the model to converge became very small (i.e. more increments and, hence, more total time are required to finish the calculation). Since the main differences during crack formation in inter- and intralaminar fracture were captured at this point, the calculations were stopped before the full coalescence of the crack to avoid excessive computation time. This problem can be partially solved by increasing the number of matrix elements. However, due to the size of the model, the computational cost of the model is prohibitive. With the number of elements implemented, each simulation needs ~ 5 days using a computer equipped with 16 Intel[®] Xeon[®] E5-2630 v3 (2.4 GHz) cores and 64 GiB of RAM memory.

4.3. Parametric study 1: the role of the microstructure

The experimental evidence (see Figure 3 and comparison of R-curves in Figure 5 with those obtained in [19, 18]) together with the RVE model pointed out the importance of the matrix-rich regions during the crack development for the intralaminar case. Therefore, to assess the importance of the microstructure, a parametric study was conducted in this section. To this end, four different RVEs were generated (Figure 9), following the same procedure used for the ones considered in section 3.1. The fiber diameter distribution, global V_f and mechanical properties of the constituents are identical too (Table 1). These models contained approximately 500,000 elements and each simulation took ~ 2 days using the aforementioned computer solver.

The only difference between the four RVEs was the size and distribution of the resin-rich zones, while the mechanical properties of constituents and cohesive interfaces are kept constant. As seen in Figure 3b, the material investigated herein (batch 2) contains thick and thin resin-rich zones in the interply domains. This is approximated by a combination of the RVEs shown in Figure 9a and b. A more complex fiber distribution was considered, shown in Figure 9c, where thick and thin resin-rich regions are present together with a non-homogeneous fiber distribution both in interply and intraply regions, corresponding to the material investigated in [18, 19] (batch 1), illustrated in Figure 3a. Finally, based on preliminary results showing the importance of the resin-rich zones in the crack evolution, a fourth type of microstructure (Figure 9d) was created. This fiber distribution maximizes the number of resin-rich zones, while keeping the global V_f constant, by using highly packed thin plies with rather forced

resin-rich zones compared to a typical thin-ply laminate microstructure [44]. This latter microstructure, as explained later, is optimized for a greater micro-crack spread that leads to higher population of thick bridging bundles resulting to intense LSB and elevated ultimate toughness.

For the interlaminar loading, all RVEs respond in a similar manner (Figure 10). Upon loading, several small cracks are initiated at different fiber-rich zones, which eventually interact and coalesce to form the main crack. Upon continued loading, the secondary cracks (previously formed around the main one) stop growing, in the transverse direction. The main crack propagates within one of the plies and not at the resin-rich zone, while the plastic deformation is concentrated in the matrix ligaments formed after interface failure (Figure 10). The crack pattern shows formation of elongated bundles, that are particularly remarkable for the last two cases of batch 1 with the highly ERR dissipative real (recorded experimentally) and the optimized microstructures (shown in Figures 10c and d). Evidently, the ply migration mechanism observed in the experiments is reproduced here. In the real microstructure of batch 1, the heterogeneity created by the resin-rich zones inside the plies contributes to the crack migration, producing the characteristic elongated bundles. On the other hand, in the optimized case, due to the small thickness of the plies, a small waviness can produce the crack migration to the adjacent ply.

For the intralaminar fracture, the situation is even more interesting. Unlike the interlaminar case, where the fracture pattern is similar for the four RVEs (apart from the ply migration observed for batch 1 and optimized microstructures), for the intralaminar fracture each microstructure exhibits a different behavior. The RVE containing thick interply resin regions (Figure 11a) can kink the crack in a remarkable way, because the thick resin-rich zones are able to concentrate an important amount of plastic deformation. The presence of the thick resin layers can produce secondary cracks around the main one (lower & upper parts of the RVE, remote from the symmetry plane), potentially leading to fiber bundle formation. The opposite case is represented by the RVE containing thin resin layer (Figure 11b), where the fracture type is quite akin to the interlaminar one with barely no secondary cracks formed.

For the case of the higher energy dissipation in a real microstructure (batch 1 in Figure

11c), it can be seen how the resin-rich zones located inside the ply, act as nucleation points producing secondary cracks. After interface failure and multiple crack formation in the fiber-rich zones, the resin-rich regions start carrying an important amount of stress and, hence, undergo extensive plastic deformation. The secondary cracks can nucleate easier thanks to the presence of multiple resin-rich zones, which allows the growth of multiple cracks simultaneously. As a result, the crack pattern is quite irregular, setting the foundations for the creation of fiber bundles with large section. This crack multiplication and spread effects are even clearer for the optimized microstructure (Figure 11d). In this case multiple cracks are formed inside the thin plies prior to the activation of the resin rich zones. When the cracks tend to coalesce, multiple small bundles can form.

Overall, these results show that the resin rich zones between and within the plies (Figure 3a) lead to an increased number of nucleation sites. These results suggest that a greater resistance to fracture of specimens from batch 1 as compared to specimens from batch 2 is achieved for a given crack increment Δa (Figure 5). This effect may be enhanced with evolution of bundles' cross-section during the crack propagation, but such investigation is out of the scope of this study.

When comparing the stress-strain curves, a very similar material transverse response is obtained for interlaminar and intralaminar cases, as well as for the four RVEs (Figure 12a and b). This is because the failure process is triggered by the interface decohesion. However, it is worth mentioning that some negligible difference in RVE's stiffness (isostress vs. isostrain) is observed, while the interface failure starts approximately at the same transverse strain level for both types of failure.

In general, the average energy dissipated due to matrix plasticity is slightly larger for intra- than interlaminar ($< 5\%$), whereas for the energy dissipated due to interface damage the trend is opposite. Thus, the differences between the microstructures, in terms of total energy dissipation (interface damage and matrix plasticity), are negligible (Figures 12c and 12d).

Interface and matrix tensile strengths are similar for the Gurit SE-70, which means that the matrix undergoes plastic deformation right after debonding. Under this condition, the average amount of plastic dissipated energy is more than double the energy due to interface

damage for all microstructures and load types (50 μJ vs. 20 μJ for plastic and interface damage dissipation, respectively). Whereas for the interlaminar case the main plastic dissipation takes place in the matrix ligaments created after the microstructure collapse (peak load), for the intralaminar case, plastic deformation is mainly localized in the resin rich regions between the composite plies.

In additional simulations using a stronger matrix ($2 \times \sigma_{\text{myt}}$), the onset of plasticity was delayed, allowing more interface debonding prior the peak load. This increase of matrix tensile strength is accompanied by an increase of the composite's transverse strength, as compared to the "standard" Gurit SE-70. For this "stronger-matrix" case, both interface damage and plastic dissipated energy become closer for the interlaminar case. Nevertheless, for this stronger matrix and the intralaminar case, the amount of plastic dissipated energy for the optimized (55 μJ) and the thick (49 μJ) cases was larger than the one obtained for the thin (27 μJ) and batch 1 (35 μJ) microstructures for the same level of strain ($\varepsilon = 2.5\%$). In summary, early activation of plasticity, allows more energy dissipation, sacrificing transverse strength, compared to a material with a stronger matrix for the same level of strain.

In general, a larger influence of the microstructure, in terms of plastic dissipated energy, is expected for the intralaminar loading case due to the presence of resin-rich zones in the investigated RVEs (specially in the optimized and the thick resin-rich zones microstructures). Their presence allows accumulation of more plastic strain after the collapse of the microstructure (peak load) and before the development of critical damage in the matrix, due to microcracking.

Nevertheless, these differences in terms of dissipated energy play a secondary role in the definition of the microcrack pattern. The different fiber distributions are sufficient to produce different fracture patterns between inter- and intralaminar loading, which leads to different intensity of LSB.

4.4. Parametric study 2: ply thickness and resin-rich zone

As already pointed out in the previous sections, the presence of the resin-rich zones plays an important role in the formation of fiber bundles, especially for the intralaminar case. To understand better these underlying mechanisms, a complementary set of parametric studies was carried out using the FE model described in section 3.3. This model, although simpler

than the RVE one, provides a phenomenological description of the process using much less computational resources. Indeed, the homogenized model has around 120,000 elements and each simulation took a couple of hours using the aforementioned computer solver.

In the first series, the influence of the resin rich area separating the composite plies is studied. To this end, five different thickness (i.e. 5, 10, 15, 20 and 40 μm) of this resin rich zone, t_{rr} , were simulated keeping a baseline fiber-rich layer thickness of $t_{\text{fr}} = 180 \mu\text{m}$. The equivalent values for the physically tested material are: $t_{\text{rr}} = 20 \mu\text{m}$ ($V_{\text{f,rr}} = 0\%$) and $t_{\text{fr}} = 180 \mu\text{m}$ ($V_{\text{f,fr}} = 60\%$) resulting in the nominal $t_{\text{p}} = 200 \mu\text{m}$ ($V_{\text{f}} \approx 54\%$). The results are shown in Figure 13. As it can be observed, the thicker the resin layer, the wider the span between microcracks and, hence, the size of the potential bundle. A thin resin-rich layer is not able to kink the main crack thus the propagation is almost straight through the composite ply and the potential bundle has a relatively flattened ellipsoidal form (Figure 13a). Increasing the thickness of the resin layer, the span between microcracks increases and so does the microcrack's deflection inside the ply. As a result, the potential bundle section tends to a circular morphology. The extreme situation is represented by the $t_{\text{rr}} = 40 \mu\text{m}$ case, where the ratio between the resin rich zone and the ply thickness is large enough ($\approx 45\%$) to allow the propagation of dominant, highly deflected cracks through the ply. These results point out to two main remarks. First, the main crack deflection span is controlled by the thickness of the resin-rich zone. Second, the larger the ratio between the resin-rich zone and the ply thickness, $r = t_{\text{rr}}/t_{\text{fr}}$, the larger the number of microcracks developing through the entire ply

The second parametric series was aimed to explain the importance of the ply thickness in the intralaminar crack morphology. To this end, five different fiber-rich layer thicknesses ($t_{\text{fr}}/8$, $t_{\text{fr}}/4$, $t_{\text{fr}}/2$, t_{fr} and $2t_{\text{fr}}$, where $t_{\text{fr}} = 180 \mu\text{m}$) were simulated keeping a baseline resin rich zone thickness of $t_{\text{rr}} = 20 \mu\text{m}$. The results are shown in Figure 14. As it can be observed, the thinner the composite ply, the larger the diffusion (i.e. spread) of microcracks. Right after the resin-rich layer, the crack band span (in -y-) is almost the same for all cases since it is controlled by the thickness of the resin-rich zone (as already demonstrated), which is constant. However, at ultimate failure ($\sigma_{yy} = 0$) for the smallest r -ratio, one crack dominates, while when r is close to 1 (Figure 14a), all microcracks reach the maximum extend (i.e. one ply).

However, it shows a large amount of microcracks which can be associated to the presence of multiple small fiber bundles. Increasing the ply thickness (i.e reducing r), the size of the cracked region decreases and so does the extension of microcracks (Figure 14b). A further increase of t_{fr} leads to an optimum situation (i.e. bundles with larger section, moment of inertia and bending stiffness will be formed). Moreover, the span of the cracked zone is maximized and the presence of remote microcracks is diminished (Figure 14c and d). If the ply thickness is further augmented (between t_{fr} and $2t_{fr}$), the influence of the resin-rich zone is lost. Therefore, a single dominant crack is formed in the middle of the ply. This situation corresponds to a perfectly homogeneous microstructure similar to what the crack sees in the interlaminar failure mode.

Finally, the vertical stress distribution along the symmetry plane is analyzed for three representative cases ($t_{fr}/8$, $t_{fr}/2$ and $2t_{fr}$). The results are shown in Figure 15. As it can be observed, the stress profile is strongly influenced by the ply thickness. For the thin ply configuration, multiple peaks are developed inside the cracked region (Figure 15a) near the microcracks. This irregular stress distribution is lost increasing t_{fr} (Figure 15b), as it is reflected by two peaks due to the two main cracks that define a large potential bundle. For the thickest ply, the stress distribution shows no peaks inside the cracked region (Figure 15c), corresponding to the development of a single crack.

The homogenized model presented here, although simplistic, provides some interesting insights. By varying the r -ratio between the resin-rich zone and the ply thickness, the stress distribution inside the ply during fracture can be altered. Very small values of r , lead to a ply-through crack development that can be associated to small potential fiber bundles. Increasing r , larger and more rounded potential bundles are created. However, there is an optimum range of $r \approx 0.1 - 0.2$, which means that, after a certain point, a further increase will lead to a single crack, hence, no potential bridging bundles at all. These conclusions define an interesting framework for future investigations, in microstructural optimization to understand and control LSB.

5. Concluding remarks

The initial formation of fiber bridging under interlaminar and intralaminar mode I failure has been studied by means of computational micromechanics and compared to experimental evidence. The intense increase of the ERR, associated to the presence of large fiber bundles is linked to the fiber dispersion distribution. Indeed, the differences found for the interlaminar and intralaminar cases are attributed to the different interactions between microstructure and microcrack as a result of the presence of resin rich zones within and between the plies. Experiments and RVE FE modeling demonstrate that the crack will initiate from the weak fiber-matrix interface. For the interlaminar fracture it will follow the path defined by fiber-rich interply regions and the waviness of the plies, even though the term delamination is commonly used to describe interlaminar fracture. The random crack-front path will lead to creation of LSB, thus, more pronounced fiber clustering and randomly distributed resin-rich zones will favor LSB formation.

The different crack-front patterns, for interlaminar and intralaminar, are revealed by the study of material's cross-section at the crack tip obtained postmortem from tested DCB specimens. The single fiber and flat bundles observed in the interlaminar case are associated to the ply waviness. On the contrary, the more rounded bundles seen for the intralaminar case are attributed to the presence of the resin-rich zones located between the composite plies. In addition, the presence of resin-rich "defects" inside a ply, play an important role in the development of LSB, acting as nucleation points for the creation of new potential fiber bundles.

This influence of the microstructure and the presence of the resin-rich zones are reproduced using a 2D RVE model. With the properties of the material constituents as an input, the developed models can reproduce the experimentally observed ply cracking behavior and failure mechanisms and correlate the effect of the microstructure with the more intense toughening due to LSB for batch 1. This approach allowed also the definition of microstructures that favor LSB. Nevertheless, the detailed RVE approach had a high computational cost.

A simplified model is also proposed to study the influence of microstructural parameters like fiber- & resin-rich zone thicknesses. The thickness of the resin-rich zone, t_{rr} , between the plies is directly related to the size of the potential fiber bundles. On the other hand, the

fiber-rich layer thickness, t_{fr} , is related to the microcrack spreading phenomenon responsible for the bundle size increase. The influence of both, represented by their ratio, $r = t_{\text{rr}}/t_{\text{fr}}$, arises as an important micromechanical parameter controlling the crack front morphology. A small r -ratio would provide a wide cracked region containing a large amount of microcracks, whereas a large one would suppress the influence of the resin-rich zones and reduce the formation of potential fiber bundles. Values of 10 – 20% appear to be optimum.

In summary, this paper points out the importance of the microstructure in the development of the initial stages of LSB. The methodology described here reproduces the composite behavior based on reliable properties of the micro-constituents and can be used to replace the physical experiments, at least for material screening purposes. These virtual tests with full control of the composite microstructure and constituent properties, allow microstructural optimization to be performed in the future. Moreover, the computational approach allows the simulation of complex stress states, not possible to be evaluated experimentally. The importance of micromechanical parameters, such as fiber dispersion, fiber- and resin-rich zone thicknesses is pointed out, opening a vast field for future research opportunities aimed to predict and control the development of toughening mechanisms like LSB.

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Figure 1: Schematics of specimen showing fiber orientation within the laminates and loading direction. Interlaminar plate (a) and specimen (b) Intralaminar plate (c) and specimen (d). The y - and z - axis represent the ply stacking directions followed during fabrication for inter and intralaminar plates.

Figure 2: Idealization of the stress state (a) at the limit ahead of the crack tip of the composite, used to define the 2D computational approach for (b) inter- and (c) intralaminar fracture. Nominal ply thickness, $t_p = 200 \mu\text{m}$.

Figure 3: Cross-section images obtained with digital microscope for two different batches of Gurit SE-70, (a) material used in [18, 19] (batch 1) and (b) material manufactured for the present work (batch 2). Although the global fiber volume fraction is the same ($V_f = 54\%$), the different fiber dispersion is evident between the two microstructures. Insert in (b) indicates an Image J reconstruction used to calculate the local fiber volume fraction.

Figure 4: Homogenized model representing cross-section of investigated composite material. The main intralaminar crack, which goes through several plies of homogenized material, is represented using a single row of cohesive elements, whereas micro-cracking that leads to formation of potential fiber bundles is modeled using a set of distributed cohesive planes placed inside a fiber-rich layer (the region of interest is indicated in blue). The model dimensions are $540 \times 540 \mu\text{m}^2$. Nominal ply thickness, $t_p = t_{fr} + t_{rr} = 200 \mu\text{m}$.

Figure 5: Gurit SE-70 (batch 2) mode I R-curves, representing the Energy Release Rate (ERR), measured using the J-integral, against the crack advance, Δa , obtained from the interlaminar and intralaminar DCB tests (section 2.2). Average values for the ERR [18, 19], corresponding to batch 1 of Gurit SE-70 material, are also included for the sake of comparison (black and red squares). Stars indicate beginning of Steady State.

Figure 6: Images of Gurit's SE-70 cross-sections before the crack tip showing (a) interlaminar ($z=2\text{ cm}$) and (b) intralaminar ($z=5\text{ cm}$) fracture patterns in specimens from batch 2. For the interlaminar, the crack develops within a single ply of composite material until it moves to the adjacent one due to the presence of high local V_f regions and the inevitable ply waviness created during the fabrication process.. The intralaminar shows a smaller amount of irregular fiber bundles, but much larger than the ones found in [19] for the batch 1 material.

Figure 7: Image of Gurit's SE-70 cross-section showing the intralaminar fracture pattern at the crack tip: (a) and (b), specimens from batch 1; (c) and (d), specimens from batch 2. The main crack front kinks, due to the presence of resin-rich zones, producing branches (a) and (b). Since the embedding resin partially penetrated the specimens' cross section during sample preparation, cracks are highlighted for the sake of clarity.

Figure 8: RVEs showing the crack pattern (interfaces where the fracture criterion is met are highlighted in red) for (a) interlaminar and (b) intralaminar fracture. Crack development within the ply for the interlaminar case (c) and (d). Accumulated plastic strain (PEEQT) in the resin-rich zones and crack branching is evident for the intralaminar fracture (e), (f) and (g). The RVE dimensions are $1600 \times 200 \mu\text{m}^2$.

Figure 9: Set of RVEs used to evaluate the influence of microstructure for interlaminar and intralaminar crack formation. Layered microstructures (a) and (b), representative regions batch 2 with thick and thin resin-rich zone, respectively; (c) Representative region of material from batch 1 [18, 19]; (d) optimized layered microstructure. The four microstructures have the same global fiber volume fraction, $V_f = 54\%$. The RVE dimensions are $360 \times 360 \mu\text{m}^2$.

Figure 10: Accumulated plastic strain (PEEQT) for interlaminar failure of representative regions of tested material from batch 2 (a) and (b); representative region of material from batch 1 [18, 19] (c); optimized layered microstructure (d). Load is applied along the horizontal direction ($-x$).

Figure 11: Accumulated plastic strain (PEEQT) for intralaminar failure of representative regions of tested material from batch 2 (a) and (b); representative region of material from batch 1 [18, 19] (c); optimized layered microstructure (d). The presence of secondary cracks (branches) is highlighted. Load is applied along the vertical direction ($-y$).

Figure 12: Graphs showing stress-strain curves and the evolution of dissipated energy of RVEs for intralaminar and interlaminar fracture of the representative regions of material from batch 2, batch 1 [18, 19] and optimized microstructure. Stress-strain curves: (a) interlaminar and (b) intralaminar; Energy dissipated due to plastic deformation (ALLPD) of the matrix and interface failure (ALLDMD); (c) interlaminar and (d) intralaminar.

Figure 13: Parametric study showing the influence of resin-rich zone thickness (illustrated only in (e)) in the development of intralaminar crack. Crack pattern and transverse (vertical) stress contour in the region of interest (indicated in Figure 4) are illustrated, for increasing resin rich rich-zone thicknesses. Load is applied along the vertical direction ($-y$).

Figure 14: Parametric study showing the influence of ply thickness in the development of intralaminar crack. Crack pattern and transverse (vertical) stress contour in the region of interest (indicated in Figure 4) are illustrated, for increasing ply thicknesses. Load is applied along the vertical direction ($-y$). Note that the microcrack span is constant since it is controlled by the constant t_{tr} .

Figure 15: Crack pattern and transverse (vertical) stress contour in the region of interest (indicated in Figure 4) and stress profile along the symmetry plane for fiber-rich layer: (a) thin ($t_{fr}/8$), (b) standard ($t_{fr}/2$) and (c) thick ($2t_{fr}$).

Table 1: Properties of the SE-70 material constituents used in the FE simulations. 1: Inverse homogenization. 2: Manufacturer & literature. 3: Derived herein & literature.

SE-70 carbon fiber properties ¹						
E_{f1} (GPa)	E_{f2} (GPa)	G_{f12} (GPa)	G_{f23} (GPa)	ν_{f12}	α_{f1} (K ⁻¹)	α_{f2} (K ⁻¹)
217	13.07	11.43	5.94	0.33	-0.9e-6	7.2e-6
SE-70 epoxy matrix properties ²						
E_m (GPa)	ν_m	σ_{myt} (MPa)	β	σ_{myc} (MPa)	$\mathcal{G}_m$ (J/m ²)	α_m (K ⁻¹)
3.6	0.35	54	29	142	100	52e-6
SE-70 fibre/matrix interface properties ³						
K (MPa/m)	σ_N (MPa)	τ_T (MPa)	τ_L (MPa)	$\mathcal{G}_{Ic}$ (J/m ²)	$\mathcal{G}_{IIc}$ (J/m ²)	$\mathcal{G}_{IIIc}$ (J/m ²)
100e9	42	63	63	2	100	100