**QUEEN'S
UNIVERSITY
BELFAST**

Thin-ply polymer composite materials: A review

Arteiro, A., Furtado, C., Catalanotti, G., Linde, P., & Camanho, P. P. (2020). Thin-ply polymer composite materials: A review. *Composites Part A: Applied Science and Manufacturing*, 132, [105777].
<https://doi.org/10.1016/j.compositesa.2020.105777>

Published in:
Composites Part A: Applied Science and Manufacturing

Document Version:
Peer reviewed version

Queen's University Belfast - Research Portal:
[Link to publication record in Queen's University Belfast Research Portal](#)

Publisher rights
Copyright 2020 Elsevier.
This manuscript is distributed under a Creative Commons Attribution-NonCommercial-NoDerivs License (<https://creativecommons.org/licenses/by-nc-nd/4.0/>), which permits distribution and reproduction for non-commercial purposes, provided the author and source are cited.

General rights
Copyright for the publications made accessible via the Queen's University Belfast Research Portal is retained by the author(s) and / or other copyright owners and it is a condition of accessing these publications that users recognise and abide by the legal requirements associated with these rights.

Take down policy
The Research Portal is Queen's institutional repository that provides access to Queen's research output. Every effort has been made to ensure that content in the Research Portal does not infringe any person's rights, or applicable UK laws. If you discover content in the Research Portal that you believe breaches copyright or violates any law, please contact openaccess@qub.ac.uk.

Thin-ply polymer composite materials: a review

A. Arteiro^{a,*}, C. Furtado^{a,b}, G. Catalanotti^c, P. Linde^d, P.P. Camanho^{a,b}

^aDEMec, Faculdade de Engenharia, Universidade do Porto, Rua Dr. Roberto Frias, s/n,
4200-465 Porto, Portugal

^bINEGI, Instituto de Ciência e Inovação em Engenharia Mecânica e Engenharia Industrial,
Rua Dr. Roberto Frias, 400, 4200-465 Porto, Portugal

^cAdvanced Composites Research Group (ACRG), School of Mechanical and Aerospace
Engineering, Queen's University Belfast, Belfast BT9 5AH, UK

^dAIRBUS Operations GmbH, Kreetzlag 10, 21129 Hamburg, Germany

Abstract

The introduction of the *spread-tow thin-ply technology* enabled the development of composite plies as thin as 0.020 mm. The availability of composite plies with a broader thickness range makes the understanding of the effects of ply thickness more pertinent than ever, therefore, a comprehensive literature review is presented in this paper. The micro-structural effects of ply thickness and ply uniformity on the mechanical response of unidirectional laminae is described. Then, the effect of ply thickness scaling on several aspects of the mechanical response of composite laminates is reviewed. Finally, the current state-of-art and recent developments in manufacturing, design and application of thin plies on novel engineered composite laminates are presented. This review demonstrates that thin plies not only bring improvements to the plain strengths and design flexibility of composite laminates, but can also enhance the performance of primary structural applications, namely those driven by residual strength and damage tolerance requirements. This can be achieved by either combining thin plies with existing material technologies, or through novel design principles. Moreover, it is shown that thin plies provide increased flexibility for multifunctional optimisation and for adoption of more efficient manufacturing technologies, with great potential gains in terms of weight savings and cost

*Corresponding author. Tel.: +351 220414049. Fax: +351 225081440
Email address: aarteiro@fe.up.pt (A. Arteiro)

reduction during conceptual and detailed design and operation.

Keywords: A. Polymer-matrix composites (PMCs), B. Mechanical properties, E. Tow spreading

1. Introduction

Advanced *tow spreading* techniques are currently used to continuously and stably open thick fibre tows and obtain uniform plies significantly thinner than conventional plies [1–3]. A careful control of the tension in the tows (or yarns) allows to cost-effectively produce flat and straight plies with a dry ply thickness as low as 0.020 mm without damaging the filament fibres. The availability of such *thin plies*, with a ply thickness equal to or below 0.100 mm, opens a broad range of new possibilities in composites design, but also new challenges in understanding how ply thickness will affect the mechanical response of composite laminates.

With the aim of understanding the main effects of ply thickness in advanced laminate design, a comprehensive literature review is presented in this paper. Firstly, the micro-structural effects arising from ply thickness on the mechanical response of unidirectional (UD) laminae are discussed. Then, the effect of ply thickness scaling on the mechanical response of multi-directional composite laminates is thoroughly described. Finally, the current state-of-art and recent developments on manufacturing and design of laminates with thin plies and on the application of thin plies in novel engineered composite laminates are presented.

2. Ply thinness and ply uniformity: effect on UD composites

2.1. Intra-laminar material properties

The increased uniformity of spread tows when compared with standard grade materials (Fig. 1) has motivated the study of the effect of ply thickness on the mechanical response of UD composites. The experimental characterisation

25 at the lamina level performed by Amacher et al. [4] show that ply thickness
 26 has no significant effect on the elastic and strength properties of the UD lam-
 27 inae, except for longitudinal compression. In this case, thin-ply UD specimens
 28 show substantially higher strength (up to 20% higher on average [4]). This en-
 29 hanced compressive strength is attributed to the more uniform micro-structure
 30 of spread-tow thin plies. Optical micrographs (Fig. 1) show that the micro-
 31 structure of laminates made of thicker plies are fairly inhomogeneous, with large
 32 variations of the local fibre volume fraction and resin rich regions throughout
 33 the composite. This heterogeneous micro-structure causes early fibre micro-
 34 instabilities that conduct to premature compressive failure of UD laminae [4].
 35 As the ply thickness decreases, a better uniformity of the micro-structure is
 36 observed, consequently delaying fibre micro-instabilities.

37 When used in the weaving process of woven fabrics, spread tows not only
 38 contribute to an improved uniformity of the micro-structure [4], but also to
 39 a better uniformity of the woven architecture [5]. Due to the reduced fibre
 40 waviness and crimp angles, micro- and meso-instabilities in the fibre direction
 41 can be effectively delayed. Longitudinal compressive strength improvements of
 42 up to 18% have been reported for UD and multi-directional plain weave [spread-](#)
 43 [tow fabric](#) laminates just by reducing the tow areal weight [6].

44 *2.2. Inter-laminar material properties*

45 *2.2.1. Inter-laminar shear strength*

46 Micro-structural uniformity has also potential effects on the inter-laminar
 47 properties of composites. Not only the inter-laminar shear strength (ILSS)
 48 of UD laminates increases (in the range of 30%), but also the coefficients of
 49 variation of the ILSS data decrease with decreasing ply thickness [7]. Both
 50 effects were attributed to the more uniform micro-structure and smaller inter-
 51 laminar resin-rich regions of thin-ply laminates (Fig. 1) [4, 7].

52 2.2.2. Inter-laminar fracture toughness

53 The inter-laminar fracture of UD composites is often characterised by an ap-
54 parent increase of the fracture toughness with crack extension, leading to a crack
55 resistance curve (or $\mathcal{R}$ -curve) type of response [8–11]. This can be attributed,
56 among other phenomena, to fibre bridging [10, 11]. The extremely uniform
57 micro-structure of very thin plies means that the potential for large-scale fibre
58 bridging in UD composites is reduced when compared with more heterogeneous
59 grades — see Fig. 2. Consequently, a flatter $\mathcal{R}$ -curve, characterised by a lower
60 steady-state value of the fracture toughness, can be expected for crack prop-
61 agation parallel to the fibres. These observations have been corroborated by
62 experiments carried out by Frossard et al. [12] for mode I inter-laminar frac-
63 ture. Conversely, the mode II fracture toughness is practically insensitive to ply
64 thickness [13], attributed to the minor role of dissipative mechanisms such as
65 fibre bridging in the mode II component of the fracture toughness.

66 2.3. Discussion

67 It is apparent that the ply uniformity resulting from ply thinness has a
68 strong beneficial effect on the longitudinal compressive strength and ILSS of
69 UD composites. The enhanced longitudinal compressive strength can be con-
70 sidered one of the main contributors to the improved compressive plain strength
71 of multi-directional laminates with thin plies [6, 14, 15]. However, the enhanced
72 ILSS measured from UD coupons is not expected to have a major role on the
73 ILSS of multi-directional laminates, since the effect of ply thickness on the
74 inter-laminar stress distribution [7] is considered the dominant effect on the im-
75 proved inter-laminar shear performance of multi-directional thin-ply laminates,
76 overwhelming the micro-structural effects observed on UD laminates. For exam-
77 ple, Kalfon-Cohen et al. [16] tested blocked (*thick baseline*) and dispersed (*thin*
78 *baseline*) laminates manufactured from the same thin-ply composite system,
79 obtaining higher ILSS on the *thin baseline*. Since the same thin-ply composite
80 system was used in both laminates, no micro-structural effects should be ex-
81 pected, and the higher ILSS solely attributed to the effect of ply thickness on

the inter-laminar stress distribution. Similar trends were obtained by Huang et al. [7] on multi-directional laminates manufactured from plies with different grades.

The same applies to the inter-laminar fracture toughness. Although bridging effects play a minor role on the inter-laminar fracture behaviour of UD thin-ply composites, leading to lower steady-state values of the fracture toughness when compared with conventional composites, in multi-directional laminates bridging phenomena is expected to be governed not by the uniformity of the micro-structure, but by the relative ply orientations and local stress state at the crack tip [17, 18] due to the fact that delaminations tend to propagate between plies with dissimilar orientation.

3. Ply thickness scaling and its effect on composite laminates

3.1. Motivation

In the past, several studies have addressed the effect of stacking sequence and ply thickness scaling on several aspects of the mechanical response of composite laminates [19–65]. These studies, which involved detailed experimental campaigns and/or analysis methods, have contributed to the understanding of the effect of ply thickness scaling on the nature of the damage mechanisms that cause laminate failure. However, the introduction of spread-tow thin plies makes this analysis more pertinent than ever, since the majority of the studies carried out in the past addressed ply thickness effects by means of ply scaling restricted to a minimum nominal ply thickness of 0.125 mm, considerably above what can be obtained nowadays using tow spreading.

3.2. Matrix cracking and in-situ effect

3.2.1. Constraining effects and in-situ strengths

When embedded in a multi-directional laminate, the laminae whose fibre orientation is perpendicular to the loading direction generally develop matrix cracks at strains lower than the ultimate failure strain of the laminate (Fig. 3)

[66, 67]. These matrix cracks are responsible for the deterioration of the mechanical performance of the laminate, originating other damage modes (e.g. delamination), and creating pathways for chemicals and other substances [68].

Nevertheless, it is observed that the development of these matrix cracks (which define the actual strength of the transverse plies) occurs at applied stresses greater than the transverse strength measured in UD coupons; in fact, the actual strengths of transverse plies are not only higher than those measured in UD coupons, but they reportedly increase with decreasing ply thickness [22, 23, 67, 69]. This is a deterministic size effect that occurs at the meso-scale, known as the *in-situ* effect. The neighbouring plies cause a constraining effect on the embedded ply, reducing the available elastic energy and delaying damage propagation in the matrix [4, 70–73]. Therefore, the ply strengths cannot be treated as intrinsic lamina properties [22, 74], but as *in-situ* properties that depend on the material and geometry of the laminate [23, 69, 75].

Several experimental studies in the literature show that the transverse tensile strength (Y_T^{is}) and the in-plane shear strength (S_L^{is}) of an embedded ply depend on the ply thickness [19–22, 24, 67, 74], on the orientation (or stiffness) of the adjacent plies [22, 74] (see Fig. 4), and on the ply location in the laminate [40, 76] (see Fig. 5). Indirect observations also show that the transverse shear strength (S_T^{is}) [16, 77–80] and the transverse compressive strength (Y_C^{is}) [77, 78, 81] are *in-situ* properties too.

3.2.2. *In-situ transverse tensile strength*

From the late 1970s [19–21] to the late 1990s [22–24], experimental studies demonstrated that, when subjected to tensile loads, the transverse stress at the onset of matrix cracking in the 90° laminae of glass and carbon fibre-reinforced polymer laminates changes with the thickness of the 90° laminae and orientation of the adjacent plies. Experimental observations (e.g. Fig. 4) show that the transverse stress at the onset of matrix cracking increases with decreasing ply thickness and increasing stiffness of the adjacent laminae.

In the early 1980s, Herakovich [76] showed that the ultimate stress, the

140 strain-to-failure and the toughness (measured as the area under the experimen-
 141 tally obtained stress-strain curves) of angle-ply laminates is higher when plies
 142 of the same orientation are dispersed through the laminate instead of blocked
 143 together due to a delay in matrix cracking that caused a strong change in the
 144 nature and sequence of the failure mechanisms (Fig. 6). More recently, Fuller
 145 and Wisnom [82] showed that matrix cracking parallel to the fibre direction
 146 in angle-ply laminates can be completely suppressed using a spread-tow thin-
 147 ply composite material, leading to a brittle type of net-section fibre-dominated
 148 failure mode (Fig. 6 – right).

149 Similar experimental observations reported severe degradation of the elastic
 150 properties with increasing transverse ply thickness in cross-ply laminates [70].
 151 Thinner transverse plies are less susceptible to transverse crack initiation or
 152 propagation in the width direction, with edge cracks occurring at higher strain
 153 levels than in thicker 90° plies of quasi-isotropic and cross-ply laminates [83].
 154 In addition, the stress concentrations on the 0° plies arising from transverse
 155 cracking in thick 90° ply blocks can have a measurable effect on the strength of
 156 the 0° ligaments and induce premature laminate failure too [44].

157 To predict these effects, progressive damage modelling formulated taking
 158 into account the *in-situ* effect revealed high efficiency, accurately capturing the
 159 reduction in the maximum failure load due to ply clustering in open-hole angle-
 160 ply and cross-ply laminates under tensile loading [84].

161 More recently, an increasing number of studies have demonstrated that the
 162 applied stress corresponding to the occurrence of the first transverse crack in-
 163 creases with decreasing ply thickness [34, 40, 68, 85–89], and that a reduction
 164 of the stiffness of the adjacent plies promotes transverse cracking [68, 85]. The
 165 first studies addressing transverse cracking in spread-tow thin plies appeared in
 166 the early 2000s. Uniaxial tensile quasi-static tests on un-notched and open-hole
 167 multi-directional laminates [4, 6, 28, 30, 66, 80, 90–94] showed that first-ply fail-
 168 ure (FPF) is delayed nearly up to ultimate failure, as observed by free-edge ob-
 169 servations, scanning electron microscope (SEM) visualisation, acoustic emission
 170 and/or digital image correlation (e.g. Fig. 7), and often supported by analysis

171 [73, 91, 92]. In addition, much lower crack-opening displacements have been
 172 reported for transverse cracks in thin plies, which eventually never penetrate
 173 completely through the thickness [66]. In the case of quasi-isotropic laminates,
 174 this provides a safe space that is independent of the loading direction [95], since
 175 laminate failure will be fibre-dominated.

176 To clarify the nature of this size effect, Amacher et al. [4] tested thick-ply
 177 laminates produced from blocks of ten 30 g/m² thin plies and from individual
 178 300 g/m² plies, and no substantial difference was found between these two lam-
 179 inates, demonstrating that the observed size effect was not related to changes
 180 in the material properties or micro-structure, but to the deterministic *in-situ*
 181 effect.

182 3.2.3. *In-situ in-plane shear strength*

183 Even though most of the studies addressing ply *in-situ* effects have been
 184 focused on matrix cracking in tension, this deterministic size effect was also
 185 observed under other loading scenarios. For example, experiments suggest a
 186 significant ply thickness effect on the in-plane shear strength of embedded plies
 187 [74, 96, 97].

188 The relation between the *in-situ* in-plane shear strength and the ply thick-
 189 ness can be taken into account using fracture mechanics models and the intra-
 190 laminar mode II fracture toughness [23, 69]. It is noted, though, that linear
 191 elastic fracture mechanics (LEFM) alone is not able to accurately predict the
 192 *in-situ* in-plane shear strength [69]; in this case, the nonlinear shear response
 193 typically observed in fibre-reinforced polymers (FRPs) must be included in the
 194 fracture mechanics model (Fig. 8).

195 3.2.4. *In-situ transverse shear and transverse compressive strengths*

196 Experimental studies have also shown that a substantial reduction in longi-
 197 tudinal compressive and transverse shear strengths can result from transverse
 198 cracking (see Ref. [22] for earlier references). Therefore, an *in-situ* effect on other
 199 matrix-dominated failure mechanisms is expected to exist, namely on transverse

200 shear cracking, wedge transverse compressive fracture, and fibre kinking.

201 The application of three-dimensional (3D) phenomenological failure criteria
202 has been proposed to estimate and take into account the *in-situ* effect on the
203 transverse shear and transverse compressive strengths of embedded plies [77, 78].
204 According to these models, not only Y_T^{is} and S_L^{is} are assumed *in-situ* properties
205 (calculated using e.g. the models proposed by Camanho et al. [69]), but also
206 Y_C^{is} and S_T^{is} (Fig. 9).

207 In multi-directional laminates, transverse shear cracks typically appear as
208 inclined mode I transverse shear cusps (Fig. 10), which deflect into delaminations
209 [79]. Similarly to constrained transverse plies loaded in tension, the constraining
210 effect of the stiffer adjacent plies reduces the available elastic energy within the
211 ply, delaying transverse shear crack growth in the matrix [4, 70–72]. Although
212 there is no direct experimental evidence of an *in-situ* effect in transverse shear,
213 analysis models can predict this deterministic size effect [77–79].

214 Similarly, there is no direct experimental evidence of an *in-situ* effect in com-
215 pression. However, recent studies, which addressed directly or indirectly the
216 effect of ply thickness on the compressive properties of multi-directional lami-
217 nates, indicate that ply thickness affects the compressive response of polymer
218 composites [40, 41, 98]. These observations were later supported by a detailed
219 representation of the mechanics of transverse compressive failure and associ-
220 ated *in-situ* effect using a 3D computational micro-mechanics framework [81].
221 A clear *in-situ* effect in transverse compression was identified (see Figs. 11–12).

222 Finally, it is noted that analysis models [77, 78, 99] also predict a positive
223 contribution of the *in-situ* effect to the resistance to fibre-dominated compres-
224 sive failure mechanisms (Fig. 13). Assuming that fibre kinking is triggered by
225 localised matrix failure in the vicinity of misaligned fibres, higher matrix *in-situ*
226 strengths will lead to an enhanced failure behaviour under combined longitudi-
227 nal compressive and transverse stress states, which can also be related to the
228 improved compressive response of laminates with thin plies [40, 41, 98].

229 3.3. Delamination

230 Delamination between plies is a common damage mechanism in composite
 231 laminates due to their relatively weak inter-laminar properties [25, 49]. The
 232 structural integrity of composite structures is particularly sensitive to this failure
 233 mechanism because, typically, no visual defect is visible from the surface of the
 234 laminates, making delamination difficult to detect during visual inspection.

235 Delaminations can have several sources, including manufacturing defects
 236 (due to incomplete curing, introduction of foreign particles, or from transverse
 237 cracks caused by residual stresses) [25], impact damage [51, 54, 100–102], early
 238 matrix cracking [26, 30, 103], or geometric discontinuities such as free edges,
 239 curved sections, sudden changes of cross section, ply drops, stiffener termina-
 240 tions and flanges, or open holes, which cause high inter-laminar stress concen-
 241 trations [25, 29, 30, 104].

242 The study of the effect of ply thickness on delamination onset and growth
 243 dates back to the early 1980’s [25, 26], when relevant analysis programmes
 244 showed that delaminations form at a lower nominal applied strain in laminates
 245 scaled by increasing the ply thickness (or the number of plies blocked together).
 246 Similar results were reported more recently [27]. Similarly, matrix cracking-
 247 induced delamination (MCID) initiates at lower applied strains when the ply
 248 thickness increases, findings supported by other authors [85, 105] and by anal-
 249 ysis methods [106]. Due to lower *in-situ* strengths, early ply cracking occurs in
 250 the transverse plies (see Fig. 14), causing local stress concentrations at the inter-
 251 faces with the surrounding plies that induce premature delamination onset. In
 252 addition, higher energy release rates (ERRs) at the interfaces between thick plies
 253 lead to early delamination onset and growth at those interfaces. The combina-
 254 tion of these two effects contribute to the marked reduction of the delamination
 255 onset strain shown in Fig. 14 as the number of ply repetitions increases. Fig. 14
 256 also shows that this early delamination onset leads to premature laminate fail-
 257 ure, caused by the local stress concentrations at the load carrying plies induced
 258 by the early matrix cracks and delaminations. As the transverse plies become
 259 thinner, matrix cracking is delayed, forcing delamination to start from the free

edges. The predominant type of inter-laminar damage changes from MCID to FED (free-edge delamination), delaying laminate failure (see Fig. 14).

More recently, other studies have demonstrated, through direct or indirect experimental observations and analysis methods, the benefits of ply thinness in suppressing the onset and propagation of delamination in coupons [4, 7, 16, 28–30, 80, 95, 107, 108] and structural details [109]. It is interesting to note that, based on numerical studies, independently of the ultimate failure mode, neglecting delamination in the analysis of conventional laminates leads to inaccurate predictions of the ultimate strengths [39]. However, given the ability of thin-ply laminates to suppress delamination, complex nonlinear analysis models that simulate inter-laminar damage onset and growth are no longer necessary for accurate prediction of the failure behaviour of this class of composite materials [110, 111].

Finally, it should be noted that, due to the low out-of-plane strength of composite laminates, their use in applications with important through-thickness stresses, such as those observed in stiffened panels (in particular when subjected to bending loads), ply drops (e.g. from laminate tapering) or L-shaped structures (typical of box structures, such as wings or wind turbine blades), is generally restricted. In fact, delamination failure caused by through-thickness, out-of-plane reactions is one of the main obstacles to the replacement of current metallic curved sections by composite ones in advanced lightweight construction. The ability of thin plies to delay, or even suppress [delamination caused by free edge effects](#) without the need of special resins, interleaving solutions or through-thickness reinforcements makes laminates with thin plies [also](#) attractive for this type of applications.

3.4. Fibre failure modes

3.4.1. Intra-laminar longitudinal strengths

Intra-laminar fibre failure is typically characterised by fibre fracture, fibre-matrix debonding and subsequent pull-out in tension, or by shear-driven fibre failure and fibre kinking in compression, often leading to ultimate, catastrophic

laminar failure. It is therefore important to understand how ply thickness can affect intra-laminar failure involving fibre fracture. This has special significance since the strength and the energy released by fibre-dominated failure modes are much larger than those involving matrix-dominated failure modes, either intra-laminar matrix cracking (Sect. 3.2) or delamination (Sect. 3.3).

In tension, a ply thickness effect on the longitudinal ply strengths is not expected [42, 43]. However, the laminate strength is strongly influenced by the thickness of the transverse and off-axis plies due to early matrix cracking (Sect. 3.2) and delamination (Sect. 3.3), as shown in Fig. 14. In fact, the sub-critical damage suppression capability of thin plies allows obtaining strains-to-failure approaching the maximum elongation of the reinforcing fibres [112, 113], thus exploiting the full load carrying capacity of the composite system. In compression, an effect of the matrix *in-situ* strengths is also believed to improve the overall response under combined longitudinal compressive and transverse stress states (see Fig. 13).

3.4.2. Intra-laminar longitudinal fracture toughness

The determination of the intra-laminar fracture toughness associated with fibre failure modes is currently extremely important in the assessment of the damage tolerance of composite structures and their behaviour during damage propagation [114–117], as well as in the definition of the softening laws used in recent computational analysis models [118–121] and in establishing the energy equilibrium equations used in closed-form solutions [122] that predict the ultimate strength of composite laminates. Because lay-up, geometry and size can affect the measured ERR, care must be taken when determining the fracture toughness associated with this failure mode.

It is also worth noting that, even though the fracture toughness can be characterised by a single parameter, it may be represented more accurately in the form of an $\mathcal{R}$ -curve, relating the change in the critical ERR with crack growth [115–117, 123–125], or in the form of a cohesive law, relating the cohesive stresses inside the fracture process zone with the cohesive crack opening [126, 127]. This

is because intra-laminar damage propagation involving fibre fracture is characterised by growing resistance before the damage process zone is completely developed, namely due to (i) load redistribution resulting from micro-cracking, splitting and/or delamination, which relieve the stress concentration and delays fracture to higher applied loads [122, 128–131], and (ii) bridging by the intact fibres of the plies that are adjacent to the principal load-carrying plies with broken fibres [129, 130].

The studies focused on compressive failure are considerably less comprehensive than those on tension [125], even though the intra-laminar fracture toughness, $\mathcal{R}$ -curve, or cohesive law associated with the propagation of a kink band are very important in the analysis of the damage tolerance of FRP structures.

Compact Tension (CT) and Compact Compression (CC). The CT configuration (Fig. 15a) is perhaps the most widely used specimen configuration for intra-laminar fracture toughness measurement of composites [125]. For characterisation of the compressive fracture toughness, a configuration similar to the CT test specimen has also been used by reversal of the loading configuration, known as the CC test specimen (Fig. 15b) [123, 125].

CT-type tests are usually devised such that extraneous damage modes are minimised and a brittle type of crack progression occurs from the pre-machined notch to measure the laminate or the ply intra-laminar fracture toughness. However, whereas dispersed-ply laminates exhibit stable crack growth through the whole thickness, blocked-ply laminates are characterised by a large amount of splitting and delamination [33].

In laminates with dispersed plies, the local stress at the notch tip is sufficiently high to allow through-thickness crack progression across the width of the specimen due to the absence of blunting effects as a result of reduced sub-critical damage. The constraining effect imposed by the off-axis plies on dispersed 0° plies prevents the occurrence of split cracking (in-plane shear matrix cracks, Sect. 3.2.3), resulting in the desired brittle type of failure mode. This split cracking suppression is specially strong in thin-ply laminates (Fig. 16).

Laminates with blocked plies, however, are characterised by a larger damage process zone due to the development of splitting and delamination (Fig. 17a) [34–36], which can be attributed to the lower constraint [44] and higher ERR at the interfaces adjacent to the 0° ply blocks. Damage often spreads from the pre-notch without clear through-thickness fracture propagation [132]. This leads to a significant reduction of the stress concentration at the notch tip that ultimately conducts to an apparent tougher laminate. This apparent increase of the intra-laminar fracture toughness for fibre tensile failure modes has been also attributed to an increase in fibre pull-out in the thicker 0° layers (Fig. 17b) [37].

Based on these results, a ply thickness effect on the energy released during intra-laminar fibre tensile failure has been documented [13, 37, 133], suggesting that the intra-laminar fracture toughness associated with mode I crack propagation perpendicularly to the fibre direction is not simply a material property, but rather an *in situ* property [37, 114], in spite of contradictory results showing no notable difference in the intra-laminar fracture toughness of quasi-isotropic laminates with dispersed or blocked plies [33, 38]. In fact, the occurrence or extent of other failure mechanisms and their effect on the measured intra-laminar fibre fracture toughness are often not discussed in sufficient detail [13, 37, 38, 133].

Based on the assumption of an *in situ* intra-laminar fibre fracture toughness, Chen et al. [39] employed a cohesive zone model accounting for a thickness-dependent mode I intra-laminar fracture toughness perpendicular to the fibre direction in the prediction of the tensile strength of open-hole laminates. A simple linear scaling of the fracture toughness with the thickness of the 0° ply block was applied. Chen et al. [39] concluded that accounting for this dependency is necessary to predict ply thickness effects in multi-directional laminates. However, the effect of other damage mechanisms such as splitting or delamination was not detailed.

Because split cracking and local delamination in the vicinity of the crack tip blunts the stress concentration, crack restraining in thick 0° plies cannot be attributed solely to an intrinsically higher fracture resistance, as the apparent

381 increase of fracture toughness is due to the development of a large damage
 382 process zone containing split cracking and localised delaminations that modifies
 383 the stress field in the vicinity of the notch tip, as detailed earlier by Li et al.
 384 [33]. Intra-laminar fracture toughness scaling [39] allows accounting for the
 385 additional stress relaxation that results from the development of a larger damage
 386 process zone. However, it is not clear whether such virtual scaling of the fracture
 387 toughness is applicable to more complex loading scenarios, where the sequence
 388 and extent of the blunting mechanisms can be different.

389 By suppressing split cracking (e.g. Fig. 16a), stress redistribution ahead
 390 of the crack tip is precluded in laminates with thin plies, leading to a lower
 391 apparent fracture toughness. This effect was confirmed by Bullegas et al. [134],
 392 by showing that the intra-laminar fibre fracture toughness of thin-ply laminates
 393 can be increased several fold when split cracking is favoured through properly
 394 placed micro-cut patterns ahead of the crack front, promoting bundle pull-out
 395 during crack propagation (Fig. 17b).

396 *Size effect law.* As an alternative to CT- and CC-type test specimens, a method-
 397 ology based on the *size effect law* [135] can be used to measure the intra-laminar
 398 fracture toughness and the $\mathcal{R}$ -curve of composite laminates in mode I tension
 399 [116, 136] and compression [115], as well as in mode II [117]. According to this
 400 methodology, the $\mathcal{R}$ -curve can be measured taking into account that, for differ-
 401 ent characteristic sizes (w_n), assuming that the size effect law $\bar{\sigma}^\infty = \bar{\sigma}^\infty(w)$ is
 402 known, the driving force curves $\mathcal{G}_I$ at the ultimate remote stresses ($\bar{\sigma}^\infty(w_n)$) are
 403 tangent to the $\mathcal{R}$ -curve (Fig. 18). In other words, the $\mathcal{R}$ -curve can be determined
 404 as the envelope of the crack driving force curves at the ultimate remote stresses
 405 (e.g. Ref. [137]). The size effect law can be easily obtained from *geometrically*

406 *similar specimens*¹ of different sizes with a *positive geometry*² [115–117].

407 Using the size effect method, Catalanotti et al. [116] obtained the $\mathcal{R}$ -curve
408 associated with intra-laminar fracture of different carbon/epoxy systems and
409 reinforcements (including non-crimp and woven fabrics). Scaled double edge-
410 notched specimens (Fig. 18) were used. One of the laminates tested by Cata-
411 lanotti et al. [116] was an NCF thin-ply laminate, with a minimum nominal ply
412 thickness of 0.080 mm. When compared with conventional laminates (nominal
413 ply thickness of 0.125 mm), the value of the intra-laminar fracture toughness of
414 the 0° ply, estimated using the model presented by Camanho and Catalanotti
415 [140], [and the length of the fracture process zone indicates](#) that laminates with
416 thin plies are [not inherently brittle](#)³. Although sub-critical failure mechanisms
417 such as delamination and splitting were apparently absent, a crack resistance
418 behaviour was still observed, which could be linked to a crack bridging phe-
419 nomenon ahead of the crack tip [122].

420 *Notched strength.* Notched specimen configurations are often used in the design
421 process of composite structures to account for the effects of the presence of high
422 stress concentrations, originated, for instance, by discrete sources of damage
423 [141]. The Centre-Notched Tension (CNT) configuration is perhaps the most
424 widely used.

425 CNT specimens can also be used to measure the mode I tensile fracture
426 toughness of composite laminates [40, 41, 125, 140], while a similar configu-

¹*Geometrically similar specimens* have constant ratio between the characteristic dimension, e.g. the width of the specimen, and the remaining dimensions, namely the crack length and the gauge length of the specimen. Although not mandatory, the use of geometrically similar specimens in the measurement of the $\mathcal{R}$ -curve of polymer composites provides simpler and more accurate solutions [138].

²In structures with a *positive geometry*, the ERR at constant load must increase with crack extension [139]. In other words, the geometry correction factor must be an increasing function of the crack length. Nearly all classical notched fracture specimens have a positive geometry [139].

³[An inherently brittle material would exhibit a negligible fracture process zone and a low fracture toughness, which is not the case of laminates with thin plies.](#)

427 ration, the *Centre-Notched Compression* (CNC), can be used to measure the
428 compressive fracture toughness by reversal of the loading direction [40, 41, 125].
429 However, a centre-notched specimen alone only provides a single value of the
430 intra-laminar fracture toughness, which is below the steady-state value unless
431 a sufficiently large centre-notched specimen, able to develop the full fracture
432 process zone (FPZ), is tested; the characterisation of the $\mathcal{R}$ -curve is therefore
433 not possible.

434 Previous studies [40, 41, 122] indicate that the ply thickness effect on the
435 intra-laminar fracture of composite laminates is not related with an intrinsically
436 higher fracture toughness of thicker 0° plies. Instead, it is related with a higher
437 propensity to develop sub-critical damage mechanisms such as delamination
438 due to higher inter-laminar stresses at the interfaces between thicker plies, and
439 longitudinal split cracking due to lower *in situ* shear strengths that allow the
440 development of matrix cracks parallel to the fibre direction (Fig. 17a). These
441 damage mechanisms cause a stress redistribution around the notch tip that de-
442 lays intra-laminar fracture perpendicularly to the fibres, leading to an important
443 notch blunting effect [40, 142], and apparently increasing the laminate fracture
444 toughness, as explained earlier.

445 *Discussion.* Based on these observations, it seems that the development of sub-
446 critical damage mechanisms, specially longitudinal split cracking, is the main
447 reason for the differences in the measured intra-laminar fracture toughness for
448 laminates with different ply thickness. This observation motivated Furtado et al.
449 [42] and Arteiro et al. [43] to combine 0° ply blocks with dispersed transverse
450 and off-axis thin plies to successfully increase the notched strength of thin-ply
451 laminates without compromising their intrinsically high un-notched strength.

452 Finally, it is noted that the existence of different techniques to measure
453 the intra-laminar fracture toughness associated with mode I crack propagation
454 perpendicularly to the fibre direction hinders a definitive conclusion about the
455 effect of ply thickness on this property, and, therefore, further studies are still
456 required to improve the knowledge about this topic.

457 3.5. Size effects

458 The mechanical response of a composite laminate is usually dependent on
459 the coupon size (or volume), even when all other characteristics are preserved.
460 The understanding of this *size effect* is of great importance when using strength
461 data from small coupons in the design of large load-bearing structures [35, 138,
462 139, 143, 144].

463 3.5.1. Size effects in smooth coupons

464 Detailed studies show that the un-notched strengths of multi-directional
465 composite laminates with sub-laminate scaling, where the basic sub-laminate
466 is repeated as often as required to increase the laminate thickness (and the
467 thickness of each UD layer is therefore constant and equal to the thickness of
468 a single ply) is higher than laminates with ply scaling, where plies of the same
469 orientation are stacked together to increase the effective ply thickness (Fig. 19)
470 [44–47]. These studies also show that laminates scaled at the sub-laminate level
471 exhibit no load drops or visual indications of damage prior to ultimate failure,
472 which is characterised by a fairly clean break across the width in the gauge
473 section. Laminates scaled at the ply level, on the other hand, exhibit trans-
474 verse and split cracking and delamination from the off-axis plies, leading to a
475 detrimental size effect (Fig. 19).

476 3.5.2. Hole size effects in tension

477 When a notched multi-directional laminate is loaded in tension, fibre-matrix
478 splitting in the 0° plies and localised delaminations can occur at the hole edge,
479 acting as important notch blunting mechanisms that affect the stress concentra-
480 tion at the hole edge, and consequently the notched strength. But as the hole
481 size increases, it becomes more difficult for the delaminations to grow and link
482 through-the-thickness, and for splitting to occur, leading to a *hole size effect*
483 characterised by a change in strength of laminates with a centrally located open
484 hole of different sizes, but constant stress concentration across the width (i.e.

485 constant width-to-hole diameter ratio⁴) [48–50].

486 The hole size effect is generally dependent on many factors, and therefore not
487 always result in the same trends for all laminates. In particular, ply thickness
488 plays an important role on these trends, which can be explained by the role of
489 sub-critical damage in the laminate failure mode [49].

490 Laminates with sufficiently thin plies fail by fibre fracture, because the fibre
491 failure stress is reached before extensive delamination growth and fibre-matrix
492 splitting [48, 49]. Intra-laminar fracture initiates at the hole edge and prop-
493 agates across the width, sometimes restricted to the outermost sub-laminate
494 [48, 145]. This leads to a fibre failure mode, with either pull-out or brittle
495 appearance, depending on the extent of matrix damage mechanisms. As the
496 hole size increases, delamination propagation becomes more localised, reducing
497 further the blunting effect of delamination and splitting, and decreasing the
498 notched strength, giving rise to the *conventional hole size effect* (Fig. 20).

499 With sufficiently thick plies (or thick ply blocks), delamination becomes the
500 predominant failure mechanism, which easily propagates across the width un-
501 til reaching the straight free edges of the specimen. Delamination then steps
502 through the adjacent interfaces via transverse and off-axis matrix cracks, reach-
503 ing the interface with the 0° plies, and isolating them [48, 145]. This leads to
504 complete gauge section delamination of the interfaces adjacent to the 0° plies,
505 resulting in early delamination failure [48, 49]. Even though the specimens
506 may be able to sustain additional loading due to the load carrying capacity
507 of the remaining 0° ligaments, delamination propagation can be considered an
508 appropriate definition of failure because these specimens lose their structural
509 integrity upon delamination propagation [50]. By increasing the hole size, and
510 keeping constant the width-to-hole diameter ratio, it becomes harder for the
511 delaminations to propagate and join up across the wider ligaments before fi-
512 bre fracture. In these conditions, complete loss of structural integrity is de-

⁴It is important to note that when the width is kept constant for varying hole sizes, the varying stress distribution across the width may obscure the underlying size effects [50].

513 layed and the notched strength increases, resulting in an *inverse hole size effect*
514 [48, 49, 145] (Fig. 20).

515 For laminates with intermediate ply thickness, a transition of failure mecha-
516 nisms, from delamination to fibre failure, may occur as the coupon size increases.
517 Specimens with small holes may fail by delamination, and specimens with large
518 holes may fail by fibre fracture. In these cases, the competing failure mecha-
519 nisms can lead to a flat response, with an almost constant notched strength over
520 a range of hole sizes [48, 49].

521 Based on these observations, Wisnom and co-workers [48, 49] concluded that
522 the hole diameter-to-ply block thickness ratio is the critical parameter affect-
523 ing how easily delaminations propagate, and consequently how the laminate
524 strengths change with changing hole diameters. For high hole diameter-to-ply
525 block thickness ratios — the case of laminates with thin plies — delamination
526 is not likely to occur. These observations have been confirmed not only on
527 conventional laminates [40], but also on NCF thin-ply laminates [15, 41].

528 3.5.3. Hole size effects in compression

529 For the hole size effect in compression, three distinct failure mechanisms, all
530 possibly leading to catastrophic failure, have been observed depending on size
531 and lay-up, namely longitudinal compressive failure induced by shear-driven fi-
532 bre fracture or by fibre kinking, push-out between plies and delamination [50].
533 Laminates with sufficiently thin plies exhibit a more brittle failure mode, char-
534 acterised by a straight fracture across the laminate. Since sub-critical damage
535 mechanisms, such as delamination and fibre splitting, are inhibited by the con-
536 straining effect of the thinner sub-laminates, the stress in the 0° plies increases
537 to a level high enough to trigger longitudinal compressive failure. Laminates
538 with thick plies (or thick ply blocks) show fracture of the angle plies, fibre-
539 matrix splitting and local delamination or push-out between plies before 0° ply
540 failure (Fig. 21). This local damage leads to a stress redistribution around the
541 hole, reducing the stress concentration and delaying ultimate laminate failure
542 to a higher applied stress. For this reason, the compressive notched strengths

543 of laminates with blocked plies tend to be higher than the notched strengths of
544 laminates with thin plies [15, 40, 41, 46, 47].

545 It is noted, however, that results presented elsewhere [4, 6, 14, 146] show
546 that laminates with thinner plies have equal or higher compressive notched
547 strengths. On one hand, the type and extent of the damage that develops
548 around the hole can be affected by the boundary conditions used to support and
549 load the specimen in compression (e.g. out-of-plane constraining), contributing
550 to the observed differences. On the other hand, the intrinsically higher com-
551 pressive strength of spread-tow UD plies [4] and fabrics [6] promoted by their
552 superior micro- and meso-structural uniformity (Sect. 2.1) also contributes to
553 an improved compressive notched response; despite the lack of stress relaxation
554 due to the absence of sub-critical damage mechanisms, the higher compressive
555 strengths of the 0° thin plies delay laminate failure, even in the presence of
556 a notch. It is therefore clear that, to account for the effect of ply thickness
557 on the compressive strength of notched laminates, it is not only important to
558 understand meso-scale effects such as the type and extent of damage prior to ul-
559 timate failure, but also the role of extrinsic and intrinsic factors such boundary
560 conditions and micro-structural characteristics.

561 3.5.4. *Effect of laminate thickness*

562 Wisnom et al. [50], after gathering extensive work on scaling of un-notched
563 and open-hole specimens, discussed the effect of laminate thickness on the
564 strength of laminates, and how ply thickness scaling affects the observed trends.
565 Increasing the laminate thickness by ply scaling leads to a severe tensile strength
566 reduction in the un-notched and open-hole laminates, which show very similar
567 trends (Figs. 19 and 22). This is attributed to easier delamination propagation
568 as the laminate thickness increases with ply thickness.

569 Thickness scaling of un-notched laminates using sub-laminate repetitions
570 leads to a slight strength increase (Fig. 19). Matrix cracking starts in the
571 outer plies [69], which represent a smaller portion of the thicker laminates with
572 dispersed plies, leading to higher laminate un-notched strengths as the laminate

thickness increases. However, notched laminates with dispersed plies show a slight reduction in tensile strength with increasing laminate thickness, remaining constant for increasingly thicker laminates (Fig. 22). In this case, as opposed to the un-notched laminates, damage development in the outer plies (which represent a smaller portion of the thicker laminates) leads to a reduced notch blunting effect around the hole, increasing the stress concentration, leading to premature laminate failure.

In compression, un-notched laminates with sub-laminate scaling do not show significant thickness scaling effects, with a fairly constant compressive strength regardless of the specimen thickness (Fig. 19) [50, 146]. Un-notched laminates with ply scaling, on the other hand, exhibit a strength reduction with increasing laminate thickness (Fig. 19). Fibre waviness and void content [146] and a change in failure mode to delamination [50] are reportedly the main contributors to the thickness effect on the compressive failure strength of laminates with ply scaling.

Conversely, the compressive notched strength of dispersed-ply and blocked-ply laminates is fairly insensitive to laminate thickness (Fig. 23), despite a change of failure mode to delamination (with no fibre breakage) in thicker laminates [50].

3.5.5. Discussion

Currently, it is well understood that structures subjected to uniform, smooth stress distributions in uniaxial tension or uniaxial compression can benefit from constructions based on laminates with thin, dispersed plies. However, in the presence of stress concentrations, the lack of notch blunting mechanisms may lead to higher notch sensitivity, specially in tension (Sect. 3.5.2).

To ensure a damage tolerant design with thin plies, sub-critical damage should be allowed to grow near the stress concentrations, preferably without compromising the un-notched strengths of the laminate. This has been achieved by combining thin transverse and off-axis plies with thick (or intermediate) 0° plies to promote longitudinal split cracking tangent to the notch boundary, relieving the stress concentration, and thus promoting the desired notch blunting

effect [42, 43]; more interestingly, the presence of thick (or intermediate) 0° plies do not compromise the intrinsically high un-notched strengths of thin-ply laminates [42, 43].

In compression, the effect of sub-critical damage preceding ultimate failure on the laminate compressive notched strength may depend substantially on the local boundary and loading conditions; nevertheless, more recent results on spread-tow thin-ply laminates suggest that the ply uniformity and damage suppression capability of thin plies lead to a superior compressive notched strength by delaying fibre compressive failure to higher loads in spite of the local stress concentration promoted by the absence of notch blunting mechanisms [4, 6, 14].

Finally, it is important to stress that in laminates with thin plies, since delamination and matrix cracking are precluded, failure analysis becomes simpler since closed-form solutions or simple modelling strategies can be used to predict with reasonable accuracy the notched strengths and notch size effects on such laminates [15, 79, 110, 111, 122]. Conversely, in laminates with standard or thick plies, delamination and split cracking are likely to occur. Consequently, their mechanical response can only be predicted using nonlinear finite element (FE) codes [39, 147], which are usually unsuitable for preliminary design or optimisation. This is a relevant advantage of thin plies, specially in applications that require complex material scrutiny and certification procedures, for fast and optimal laminate selection, or when the time and resources to run advanced FE codes are limited.

3.6. *Impact resistance and damage tolerance*

Load bearing structures made of composite laminates, particularly those used in aerospace applications, must have their design driven by *damage tolerance* considerations, i.e. some level of damage must be assumed to exist in the composite structure [101, 148]. The reason for this is the susceptibility of composite laminates to the introduction of visually undetectable damage caused by external sources such as out-of-plane *low-velocity impact* (LVI) events by foreign objects.

LVI events usually lead to the formation of local delaminations below and around the impact location that are particularly critical for the compressive strength of the impacted composite laminate [149]. Impact damage has, therefore, a significant effect on the residual compressive and shear strengths of composite laminates due to the appearance of instabilities and overloading of the undamaged areas, reducing the load-carrying capacity and structural integrity of composite structures [51, 53, 54, 100, 101, 149, 150].

In laminates with thick plies (or thick ply blocks), the number of interfaces available for delamination is reduced, leading to fewer, but larger delaminations when subjected to impact loading [51–56], and consequently lower delamination threshold loads and lower peak forces (Fig. 24). This is typically prejudicial for the impact response of the laminate [55].

However, when assessing the effect of ply thickness on the residual compressive strength measured in *compression after impact* (CAI) tests, the trends are not so clear [52–55, 100]. On one hand, the residual compressive strength of laminated composites with thick plies (or thick ply blocks) is expected to decrease due to larger delaminations, which reduce the laminate bending stiffness and, therefore, increase the susceptibility to local buckling. On the other hand, if the number of delamination planes increases, which is expected to happen when the number of plies in a laminate is higher [54, 55], as in laminates with thin plies, the residual compressive strength is also expected to decrease due to the formation of thinner sub-laminates that are more susceptible to local buckling. The lack of a clear trend for the effect of ply thickness on the residual compressive strength of conventional laminates is confirmed by the number of experimental studies with contradictory conclusions [51–54].

In the case of very thin plies, the overall size of delamination damage is reportedly similar to that of blocked-ply laminates [14, 28], confirmed by quasi-static indentation studies [31, 32]. However, these results contradict the typical observations on laminates with conventional dispersed and blocked plies [51–56]. Although matrix cracking and delamination onset are delayed in laminates with thin plies [57], earlier fibre breakage on the impacted and non-impacted

664 faces [4, 31, 58, 59] (due to very high compressive and tensile stresses) and
 665 large delaminations at the mid-surface [4, 32, 57, 59, 60, 151, 152] (where the
 666 highest shear stresses occur) appear in these laminates, an observation that
 667 was also confirmed numerically [153]. Deeper indentation footprints can also
 668 be expected due to the high local compressive stresses beneath the impactor,
 669 which can reduce the threshold for *barely visible impact damage* and lead to
 670 higher residual strengths at the threshold of detectability.

671 In spite of this characteristic impact damage morphology, some authors re-
 672 ported a superior CAI response in laminates with thin plies [28, 57]. Although
 673 large projected delamination areas [59], typically due to the propagation of a sin-
 674 gle or few large delaminations [57, 60], are observed, the resulting sub-laminates
 675 are substantially thicker than those originated in thick-ply laminates, leading to
 676 an improved CAI response. This superior CAI strength is also attributed to an
 677 intrinsically higher compressive strength due to an improved micro-structural
 678 uniformity (Sect. 2.1), and to a fine dispersion between plies of different fibre
 679 orientation that restrains fibre kinking [57].

680 However, more recent studies report decreasing damage tolerance with ply
 681 thinness in compression [59, 152], as well as in tension [151], attributed to ex-
 682 tensive fibre failure during the impact stage [59]; this is specially critical in thin
 683 laminates [58–60]. Moreover, since fibre breakage is one of the predominant
 684 damage mechanisms in thin laminates with thin plies, its extent also dictates
 685 the reduction of residual compressive strength with increasing impact energy
 686 [154], not just the projected damage area as in conventional laminates.

687 Based on the previous observations, it is recognised that the effect of ply
 688 thickness scaling on the impact resistance and damage tolerance of composite
 689 laminates is more complex than the effect of ply thickness scaling on the intra-
 690 and inter-laminar failure modes alone. While an improved impact resistance
 691 is usually reported for laminates with dispersed plies of intermediate thickness
 692 [4, 51–54, 59], laminates with thin plies exhibit early fibre breakage due to the
 693 delay of delamination and matrix cracking, compromising their impact response.

694 These observations also suggest that, from a damage resistance and damage

695 tolerance point of view, laminates with thin plies have to be addressed differently
696 from conventional composites [79]. The characteristic mechanical behaviour
697 under impact of laminates with thin plies must be taken into account in the
698 design process to fully exploit their special features. Since fibre fracture and
699 deeper indentation footprints appear earlier in the laminate [59, 60], which then
700 cause the delaminations inside, undetectable damage due to impact may not
701 be of concern; in other words, once damage occurs, it may be immediately
702 detectable. Therefore, instead of designing thin-ply composite structures for
703 damage tolerance, it may be necessary to design them specifically for impact
704 (fibre fracture) resistance.

705 Alternatively, thin plies can be combined with other material technologies
706 for an optimum structural response. This has been achieved through com-
707 bination of thin plies with more advanced, tougher matrix formulations and
708 interface interleaving [113, 155, 156]. Such *toughened* thin-ply composites can
709 have damage tolerances within the standard requirements of damage-tolerant
710 design, without compromising the composite strengths. This is possible due
711 to strong delamination suppression; interleaving thermoplastic veils can arrest
712 delamination onset and consequently improve the residual strength of laminates
713 with thin plies [113, 155, 156].

714 Improvements of the damage tolerance of laminates with thin plies can also
715 be achieved through hybrid solutions [58] combining thick or standard plies with
716 thin plies [42, 43], and through proper, more flexible laminate design [60]. On
717 one hand, ply thickness hybridisation can alleviate the amount of fibre failure
718 with increased delamination damage, improving damage tolerance [58]. In ad-
719 dition, the introduction of 0°-ply blocking, namely at the outer layers of the
720 laminate, can improve the post-impact response, attributed to an increase of
721 the bending stiffness of the surface sub-laminates [55]. On the other hand,
722 motivated by the unsymmetrical nature of the damage modes induced by LVI
723 in the laminate thickness direction, unsymmetrical designs have been proposed
724 [60], enlarging the design space for optimal hybrid solutions combining thick or
725 standard plies with thin plies. Appropriate unsymmetrical designs allow con-

trolling the position of the dominant delaminations caused by LVI, moving them away of the interfaces close to the mid-surface (as typically observed in laminates with thin plies [57, 60, 151]), while reducing the projected damage area and the extent of fibre breakage. This leads to the formation of a thick intact sub-laminate, improving damage tolerance.

3.7. Fatigue

Due to the ability of high performance reinforcements such as carbon fibres to transmit and disperse the high-frequency vibration loads, polymer matrix composite materials with continuous fibres are very effective when subjected to mechanical fatigue [61]. Nevertheless, cyclic loading is still a very important precursor of sub-critical damage mechanisms in composite laminates that result in the degradation of composite structures.

When subjected to cyclic loading, standard composite laminates experience damage initiation by transverse cracking in the matrix, which occurs in the very early stages of the fatigue life of the material. While not affecting the structural response and the resistance to longitudinal damage propagation, these transverse cracks induce local delaminations, which propagate under fatigue loading [61, 157, 158], causing stress redistributions that affect the residual stiffness, the residual strength and the fatigue life of the laminate [26, 49, 61, 157].

In the case of notched coupons under fatigue loading, the development of sub-critical damage is enhanced, since the relative damage growth rates will not be the same and matrix-dominated failure modes such as matrix cracking and delamination are likely to become more dominant before fibre failure (Fig. 25). In addition, damage development is more progressive and more dispersed in the 90° plies than in quasi-static loading [61].

Since decreasing ply thickness leads to the suppression of matrix-dominated damage mechanisms such as matrix cracking and delamination, which are the cause of material degradation under cyclic loading, laminates with thin plies expectedly show enhanced fatigue life behaviour. The constraining effect on

thinner transverse plies leads to limited growth of micro-cracking [4, 5, 14, 27, 28, 62, 112] and delamination [4, 28, 112] (Fig. 26), regardless of the stress level [27, 112] and sign of the stress ratio [112], up to a point where no stiffness degradation nor damage accumulation is observed [4]. Consequently, the fatigue lifetime of spread-tow woven [5, 62] and tape laminates [4, 14] is markedly superior to that of conventional laminates.

It is worth noting that the certification of primary composite structures, for example in aerospace applications, is typically based on “no growth” criteria for delaminations emanating from stress raisers such as ply drops or free edges [158]. Therefore, the intrinsic damage suppression capability and improved fatigue life of laminates with thin plies may result in benefits when it comes to the certification of airborne composite structures based on this kind of criteria.

3.8. *Bearing strength*

Virtually every large-scale composite structure contains joints. The reasons behind their use include manufacturing constraints and requirements related to accessibility to the structure, quality control, structural integrity assessment, and part replacement. Among the types of joints used in large-scale composite structures, mechanically fastened joints are particularly common, and often the only feasible solution. However, due to the stress concentrations created, mechanically fastened joints are generally the critical part of a composite structure, as they are a source of weakness and compliance [159–163]; consequently, the soundness of the joint design procedure used is reflected on the overall weight and cost of the composite structure.

The behaviour of composites in bolted joints differs considerably from that of metals. The quasi-brittle nature of composite materials requires more detailed analysis to address the different damage mechanisms occurring in the vicinity of the loaded fasteners. Therefore, when designing composite joints, several factors need to be taken into account, including the orthotropy of the laminates (which may promote higher stress concentrations) and geometrical parameters such as edge distances and hole spacings [141, 162]. These lead to different fail-

786 ure modes, namely bearing, net-tension, shear-out, cleavage and tear-out when
787 subjected to in-plane loading (Fig. 27) [164], or pull-through when subjected to
788 out-of-plane loading [165].

789 Bearing failure is one of the most common failure modes, which occurs pre-
790 dominantly when the fastener diameter is a small fraction of the plate width.
791 This is a progressive, non-catastrophic failure mode and occurs when the lami-
792 nate is no longer capable of sustaining the accumulated local damage [160, 166].
793 Before ultimate failure, damage starts as localised delaminations, followed by
794 matrix cracks through the thickness of the plies, which lead to further delam-
795 inations [167]. Bearing failure is also related to fibre kinking and fibre-matrix
796 crushing [168]. This progressive failure mode typically leads to a permanent
797 deformation of the hole [160] that can affect the integrity and functionality of
798 the structure.

799 Given the ability of laminates with thin plies to delay, or even suppress,
800 sub-critical, progressive damage mechanisms such as delamination and matrix
801 cracking, an improved performance of these materials when subjected to bear-
802 ing loads is expected. Thinner plies lead to lower intra-laminar shear stresses
803 [169], delaying the onset and propagation of damage mechanisms such as matrix
804 cracking, fibre kinking and through-thickness shear cracking [4, 41], and lead to
805 higher delamination onset loads and, consequently, smaller initial delaminated
806 areas [4, 170]. The result is a superior bearing performance of laminates with
807 thinner plies [41] at room and hot-wet conditions [4].

808 3.9. *Environmental exposure*

809 In a multi-directional laminate, the different ply orientations lead to a mis-
810 match of the coefficients of thermal and moisture expansion. Moreover, tem-
811 perature and moisture are known to aggressively affect the polymer matrix,
812 degrading the mechanical properties of composite laminates [4]. Hence, the
813 temperature variation during the curing process and the in-service environment
814 can lead to early matrix cracking and inter-laminar damage growth. However,
815 with thin plies, matrix cracking constraining (Sect. 3.2) and lower inter-laminar

stresses (Sect. 3.3) can be expected, also under thermal- and moisture-induced loadings. Therefore, the sensitivity of laminates with thinner plies to temperature and environmental effects is likely to be lower.

This feature of thin-ply laminates can be exploited to improve the resistance also under extreme environmental conditions, such as applications on low earth orbit (LEO) environments [63] and hot-wet conditions [4]. For example, thin-ply laminates show improved tensile strength either with or without LEO environment exposure, exhibiting a stable mechanical response independently of the environmental conditions [63].

3.10. Lightning strike resistance and other multifunctional properties

Low-conductive FRPs are more vulnerable to lightning-induced damage than metals. During a lightning strike to a composite structure, several direct effects can be distinguished, such as thermal effects caused by the electric arc, thermal and electrodynamic effects induced by circulation of the lightning current, and mechanical effects from air and surface shock waves. To prevent hazardous events such as catastrophic structural damage, composite structures need to be designed against these lightning direct effects [171].

When subjected to simulated lightning strikes, thin plies reportedly contribute to the reduction of lightning-induced damage, specially bulk damage, including delamination (Fig. 28) [64]. This lightning damage suppression capability results from the lower in-plane and out-of-plane electrical resistivity of laminates with thin plies [64, 172]. The applied lightning current seems to flow not only in the first, top ply, but also in the underlying plies when these are sufficiently thin, restricting more efficiently the resin vaporisation area due to lower electrical anisotropy as compared with laminates with thicker UD tapes [64]. Moreover, after being subjected to a lightning strike, laminates with thinner plies exhibit higher residual strength [65]. This favours the application of these laminates on advanced, safety-critical composite structures that may be subjected to lightning strikes [64].

845 The adoption of thin plies in laminate design also allows further improve-
846 ments of the electrical conductivity with recourse to other material technolo-
847 gies, such as carbon nano-tubes directly grown on the carbon fibres, motivated
848 by the more uniform fibre distribution and ply thinness [173]. These charac-
849 teristics can bring improvements to physical properties such as electrical and
850 thermal conductivity and facilitate the introduction of multi-functionality into
851 FRPs without compromising their intrinsically high specific elastic and strength
852 properties [173].

853 4. Opportunities in manufacturing and design

854 4.1. Ply uniformity

855 At first, it might appear that part production using laminates with thin plies
856 will result in a clear disadvantage: per laminate thickness, more plies will have
857 to be accommodated, which may result in an increased processing effort that
858 increases with the number of plies (or ply thinness) to reach a target laminate
859 thickness. Therefore, the effective use of thin plies in laminate design must rely
860 on a substantially different approach, using novel reinforcement architectures
861 enabled by the unique uniformity of thin plies. For example, thin plies reduce
862 the degree of waviness caused by the insertion of the weft threads in non-crimp
863 fabrics or the crimp angle and tow flexure in woven fabrics, improving the
864 mechanical properties of this kind of reinforcements [6, 41, 174], which become
865 sound alternatives to UD tapes. [In spite of the additional spreading step prior to](#)
866 [the stabilisation and laying or weaving steps, which contributes to an increase](#)
867 [of the relative cost of spread tows, heavier tow yarns — with up to 100k or](#)
868 [200k fibre counts — can be used to reduce the cost of the source fibre tows, as](#)
869 [opposed to the very expensive 6k or 1k yarns of conventional tows required to](#)
870 [obtain sufficiently thin tapes and woven fabrics.](#)

871 In addition, this unique uniformity leads to excellent surface appearance
872 that may help reducing part production costs in terms of surface finishing and
873 painting (Fig. 29). The improved fibre distribution also favours wider inter-

874 fibre spacing, which allows easier flow of the matrix material, either in liquid
875 moulding techniques employing thermosetting resins, or during impregnation by
876 thermoplastic matrices [174], leading to fewer weak zones, i.e., thin plies lead to
877 lower void content and smaller resin-rich areas [89, 174]. However, this wider
878 inter-fibre spacing leads to lower fibre volume fractions in spread-tow yarns [62].
879 In addition, the number of resin-rich interfaces between layers is higher in thin-
880 ply laminates (since they have more layers for the same thickness), increasing
881 the relative amount of resin in the laminate. Consequently, the fibre volume
882 fraction of thin-ply laminates is usually lower than the fibre volume fraction
883 of conventional composite laminates [4]. However, this difference is typically
884 small, with no implications reported so far [4].

885 Interestingly, the overall fibre volume fraction of spread-tow and conventional
886 woven laminates is identical, in spite of the lower fibre volume fraction of spread-
887 tow yarns [62]. This is attributed to the resin-rich areas between the tows in
888 conventional fabrics that reduce the overall fibre volume fraction of the laminate
889 [62]. In spread-tow fabric laminates, the amount of resin between the tows is
890 very small, and, therefore, the fibre volume fraction of the fabric is similar to the
891 fibre volume fraction in the yarns [62]; conversely, in conventional fabrics, due
892 to the amount of resin between the tows, the fibre volume fraction is typically
893 lower than the fibre volume fraction of the corresponding yarns.

894 Ply thickness has also a noticeable effect on the occurrence of damage dur-
895 ing manufacturing. In spite of the mismatch of coefficients of thermal expansion
896 between plies with different fibre orientation, thinner plies suppress thermally-
897 induced damage [105, 175] as a result of the *in-situ* effect previously discussed.
898 For example, Fernberg and Joffe [94], through free-edge inspection, observed
899 very few to no transverse cracks in spread-tow fabric laminates after manufac-
900 turing, as opposed to conventional woven fabrics.

901 Another advantage of ply thinness is the design of ply drops, which become
902 much smoother (Fig. 30). Not only the appearance is improved, but also the
903 structural performance of the laminates due to lower ERR at the interfaces
904 between ply drops.

905 4.2. Laminate design, homogenisation and hybridisation

906 With thinner plies, for the same laminate thickness, more plies can be ac-
907 commodated. On one hand, the design space widens, due to higher flexibility
908 in accommodating plies of different orientation. On the other hand, laminate
909 homogenisation by sub-laminate repetition is easier to achieve for the typical
910 laminate thickness currently used in advanced structural design. In this case,
911 mid-plane symmetry becomes irrelevant⁵ due to negligible thermal warpage
912 [176, 177]. This can be beneficial, for example, in the design of stringers
913 for aero-structures, where criteria to avoid warpage are required. With ho-
914 mogenised laminates, optimisation may rely only on topological optimisation,
915 since warpage is negligible. This also enables simpler lay-up processes, since
916 reversing the stacking sequence at the mid-plane is no longer needed. Continu-
917 ous stacking becomes possible, and ply drops no longer need to be symmetric,
918 reducing processing time, cost, waste and stacking errors. Besides, by avoiding
919 the symmetry rule, the laminate design space widens further and optimisation
920 for out-of-plane loading and damage tolerance, where non-symmetric laminates
921 may perform better [60], becomes easier [178].

922 By allowing a higher number of repeated sub-laminates, the potential for
923 overall laminate thickness reduction on buckling- and damage tolerance-driven
924 laminate design also increases [179]. In addition, [minimum gauge structures](#)
925 [are expected to benefit from the introduction of thin plies due to lower discrete](#)
926 [steps in ply thickness; when the theoretical minimum gauge corresponds to](#)
927 [a non-integer number of plies, adding one thin ply will lead to thinner, and](#)
928 [therefore lighter structures than adding one conventional ply.](#)

929 [Having](#) more plies per laminate thickness [also](#) leads to the possibility of
930 using smaller relative ply orientations between adjacent plies, which further
931 improves delamination [180] and impact resistance [181]. It is also noted that

⁵A laminate is homogenised when the normalised flexural stiffness matrix $[D^*]$ approaches that of the in-plane stiffness matrix $[A^*]$, and the bending-extension coupling terms $[B^*]$ tend to zero.

sub-laminates with smaller relative ply orientations reduce the gap between FPF and LPF (last-ply failure) in such a way that matrix cracking can be completely suppressed before ultimate failure. Together with lower ERRs at the interfaces between plies due to ply thinness, shallow-angle thin-ply sub-laminates allow, for example, the exploitation of the nonlinear deformation of angle-ply laminates to generate more progressive and ductile failure modes in advanced composites [82, 182].

Thin carbon plies also allow the production, through hybridisation, of composite systems exhibiting progressive fragmentation or multiple cracking instead of sudden, catastrophic ultimate failure [183–189]. With these approaches, higher energy dissipation can be successfully obtained [184, 186–189], but usually at the compromise of ultimate plain and notched strengths [184, 186–188]. It is also worth noting the increase in fibre tensile strain-to-failure due to the hybrid effect in very thin plies [190], showing that the resistance of the carbon-fibre reinforced polymer system can be fully exploited within hybrid composites with thin plies.

4.3. Analysis

Homogenisation and the absence of sub-critical damage mechanisms such as matrix cracking (Sect. 3.2) and delamination (Sect. 3.3) before final failure makes analysis much faster and simpler. The simplest methods are based on FPF analysis, designing against the occurrence of transverse cracking. But for most conventional laminates, the FPF strain is much lower than the ultimate failure strain of the laminate, and the high specific strength of composites is therefore compromised [191]. In laminates with thin plies, the gap between FPF and LPF is reduced, and the high performance usually associated to composite laminates can be fully utilised by employing closed-form and fast solutions [15, 99, 110, 111]. This brings clear advantages to the design process, specially when involving optimisation and the generation of A- or B-basis allowables by simulation [192].

The suitability of simpler analysis models combined with the intrinsic supe-

rior strength and increased design space of laminates with thin plies can lead to better optimised laminates and higher consistency in defining safety factors, with great potential for weight savings and, consequently, cost reduction during the conceptual and detailed design stages and during operation.

5. Concluding remarks

Ply thickness can have a drastic effect on the structural performance of composite laminates. Based on experimental and numerical observations on composites with thick, intermediate, and thin plies, this review has demonstrated that the latter bring substantial improvements to the plain strengths and design flexibility of composite laminates. In addition, the use of thin plies in primary structural applications, namely those driven by residual strength and damage tolerance requirements, can enhance their mechanical and weight performance. To achieve this, thin plies can be combined with existing material technologies, for instance through ply-level hybridisation, to design laminates with lower notch sensitivity and higher damage tolerance without compromising the intrinsically high un-notched strengths [42, 43, 58]. Alternatively, novel design principles, including lower relative fibre angles and laminate asymmetry, become feasible, providing enhanced structural and damage tolerant responses [60]. Moreover, it is shown that thin plies provide increased opportunities for multifunctional optimisation due to higher design flexibility [173], and for adoption of more efficient manufacturing technologies thanks to improved multi-axial fabrics that do not compromise structural behaviour. Finally, the suitability of simpler analysis models combined with the intrinsic superior strength and increased design space of laminates with thin plies can lead to weight savings and, consequently, cost reduction during conceptual and detailed design and operation thanks to better optimised laminates and higher consistency in defining safety factors. It is concluded that thin-ply composite materials provide new perspectives for innovative structural applications.

990 **Acknowledgements**

991 This work was partially funded by AIRBUS under project 2genComp —
992 second generation Composites. The authors gratefully acknowledge the support
993 provided by AIRBUS.

994 The first and last authors would like to thank the financial support provided
995 by the Portuguese Government's FCT – Fundação para a Ciência e a Tecnologia
996 through National Funds in the scope of project MITP-TB/PFM/0005/2013.
997 The second author is also grateful for the support of FCT under the Grant
998 SFRH/BD/115859/2016.

999 The last author gratefully acknowledges the funding of Project NORTE-01-
1000 0145-FEDER-000022 – SciTech – Science and Technology for Competitive and
1001 Sustainable Industries, co-financed by Programa Operacional Regional do Norte
1002 (NORTE2020), Fundo Europeu de Desenvolvimento Regional (FEDER).

1003 The authors are also grateful to Prof. S.W. Tsai (Department of Aeronautics
1004 and Astronautics, Stanford University, CA USA) for the inspiring discussions
1005 and fruitful collaboration.

1006 **References**

- 1007 [1] Kawabe K, Matsuo T, Maekawa Z. New technology for opening various
1008 reinforcing fiber tows. J Soc Mat Sci Jpn 1998;47(7):727–34. In Japanese.
- 1009 [2] Kawabe K. New spreading technology for carbon fiber tow and its ap-
1010 plication to composite materials. Sen-i Gakkaishi 2008;64(8):262–7. In
1011 Japanese.
- 1012 [3] Quintana Casanovas J, Costa J, Mayugo JA, Galan Llongueras A. Fab-
1013 rication of hybrid thin ply tapes. IOP Conf Ser: Mater Sci Eng
1014 2018;406:012067.
- 1015 [4] Amacher R, Cugnoni J, Botsis J, Sorensen L, Smith W, Dransfeld C.
1016 Thin ply composites: Experimental characterization and modeling of size-
1017 effects. Compos Sci Technol 2014;101:121–32.

- 1018 [5] Nishikawa Y, Miki T, Okubo K, Fujii T, Kawabe K. Fatigue behaviour of
1019 plain-woven CF/epoxy composites using spread tows (effect of tow thick-
1020 ness on crack formation). Trans Jpn Soc Mech Eng A 2005;71(710):1356–
1021 61. In Japanese.
- 1022 [6] Arteiro A, Catalanotti G, Xavier J, Linde P, Camanho PP. Effect of tow
1023 thickness on the structural response of aerospace-grade spread-tow fabrics.
1024 Compos Struct 2017;179:208–23.
- 1025 [7] Huang C, He M, He Y, Xiao J, Zhang J, Ju S, et al. Exploration relation
1026 between interlaminar shear properties of thin-ply laminates under short-
1027 beam bending and meso-structures. J Compos Mater 2018;52(17):2375–
1028 86.
- 1029 [8] Suo Z, Bao G, Fan B. Delamination *R*-curve phenomena due to damage.
1030 J Mech Phys Solids 1992;40(1):1–16.
- 1031 [9] Albertsen H, Ivens J, Peters P, Wevers M, Verpoest I. Interlaminar frac-
1032 ture toughness of CFRP influenced by fibre surface treatment: Part 1.
1033 Experimental results. Compos Sci Technol 1995;54:133–45.
- 1034 [10] Sørensen BF, Jacobsen TK. Large-scale bridging in composites: R-curves
1035 and bridging laws. Compos Part A-Appl S 1998;29A:1443–51.
- 1036 [11] Airoidi A, Dávila CG. Identification of material parameters for mod-
1037 elling delamination in the presence of fibre bridging. Compos Struct
1038 2012;94:3240–9.
- 1039 [12] Frossard G, Cugnoni J, Gmür T, Botsis J. Mode I interlaminar fracture
1040 of carbon epoxy laminates: Effects of ply thickness. Compos Part A-Appl
1041 S 2016;91:1–8.
- 1042 [13] Frossard G. Fracture of thin-ply composites: effects of ply thickness.
1043 Ph.D. thesis; École Polytechnique fédérale de Lausanne; Lausanne; 2017.

- 1044 [14] Yokozeki T, Aoki Y, Ogasawara T. Experimental characteriza-
1045 tion of strength and damage resistance properties of thin-ply carbon
1046 fiber/toughened epoxy laminates. *Compos Struct* 2008;82:382–9.
- 1047 [15] Arteiro A, Catalanotti G, Xavier J, Camanho PP. Notched response
1048 of non-crimp fabric thin-ply laminates: Analysis methods. *Compos Sci*
1049 *Technol* 2013;88:165–71.
- 1050 [16] Kalfon-Cohen E, Kopp R, Furtado C, Ni X, Arteiro A, Borstnar G, et al.
1051 Synergetic effects of thin plies and aligned carbon nanotube interlaminar
1052 reinforcement in composite laminates. *Compos Sci Technol* 2018;166:160–
1053 8.
- 1054 [17] de Morais AB, de Moura MF, Marques AT, de Castro PT. Mode-I in-
1055 terlaminar fracture of carbon/epoxy cross-ply composites. *Compos Sci*
1056 *Technol* 2002;62:679–86.
- 1057 [18] Varandas L, Arteiro A, Catalanotti G, Falzon BG. Micromechanical anal-
1058 ysis of interlaminar crack propagation between angled plies in mode I
1059 tests. *Compos Struct* 2019;220:827–41.
- 1060 [19] Garrett KW, Bailey JE. Multiple transverse fracture in 90° cross-ply
1061 laminates of a glass fibre-reinforced polyester. *J Mater Sci* 1977;12:157–
1062 68.
- 1063 [20] Parvizi A, Garrett KW, Bailey JE. Constrained cracking in glass fibre-
1064 reinforced epoxy cross-ply laminates. *J Mater Sci* 1978;13:195–201.
- 1065 [21] Parvizi A, Bailey JE. On multiple transverse cracking in glass fibre epoxy
1066 cross-ply laminates. *J Mater Sci* 1978;13:2131–6.
- 1067 [22] Flaggs DL, Kural MH. Experimental determination of the in situ trans-
1068 verse lamina strength in graphite/epoxy laminates. *J Compos Mater*
1069 1982;16:103–16.

- 1070 [23] Dvorak GJ, Laws N. Analysis of progressive matrix cracking in composite
1071 laminates II. First ply failure. *J Compos Mater* 1987;21:309–29.
- 1072 [24] Boniface L, Smith PA, Bader MG. Transverse ply cracking in cross-ply
1073 CFRP laminates — Initiation or propagation controlled? *J Compos Mater*
1074 1997;31(11):1080–112.
- 1075 [25] O'Brien TK. Characterization of delamination onset and growth in a
1076 composite laminate. NASA Technical Memorandum, NASA-TM-81940;
1077 NASA Langley Research Center; Hampton, Virginia; 1981.
- 1078 [26] O'Brien TK. Analysis of local delaminations and their influence on
1079 composite laminate behavior. NASA Technical Memorandum 85728/US-
1080 AAVSCOM Technical Report 83-B-6, NASA-TM-85728; NASA Langley
1081 Research Center/U.S. Army AVSCOM; Hampton, Virginia; 1984.
- 1082 [27] Berthelot JM. Transverse cracking and delamination in cross-ply glass-
1083 fiber and carbon-fiber reinforced plastic laminates: Static and fatigue load-
1084 ing. *Appl Mech Rev* 2003;56(1):111–47.
- 1085 [28] Sihm S, Kim RY, Kawabe K, Tsai SW. Experimental studies of thin-ply
1086 laminated composites. *Compos Sci Technol* 2007;67:996–1008.
- 1087 [29] Camanho PP, Arteiro A, Turon A, Costa J, Guillaumet G. Structural
1088 integrity of thin-ply laminates. *JEC Compos Mag* 2012;71:49–50.
- 1089 [30] Guillaumet G, Turon A, Costa J, Renart J, Linde P, Mayugo JA. Damage
1090 occurrence at edges of non-crimp-fabric thin-ply laminates under off-axis
1091 uniaxial loading. *Compos Sci Technol* 2014;98:44–50.
- 1092 [31] Yokozeiki T, Kuroda A, Yoshimura A, Ogasawara T, Aoki T. Damage char-
1093 acterization in thin-ply composite laminates under out-of-plane transverse
1094 loading. *Compos Struct* 2010;93:49–57.
- 1095 [32] Wagih A, Maimí P, González EV, Blanco N, Sainz de Aja JR, de la
1096 Escalera FM, et al. Damage sequence in thin-ply composite laminates
1097 under out-of-plane loading. *Compos Part A-Appl S* 2016;87:66–77.

- 1098 [33] Li X, Hallett SR, Wisnom MR, Zobeiry N, Vaziri R, Poursartip A. Ex-
1099 perimental study of damage propagation in Over-height Compact Tension
1100 tests. *Compos Part A-Appl S* 2009;40:1891–9.
- 1101 [34] O'Higgins RM, McCarthy MA, McCarthy CT. Comparison of open hole
1102 tension characteristics of high strength glass and carbon fibre-reinforced
1103 composite materials. *Compos Sci Technol* 2008;68:2770–8.
- 1104 [35] Xu X, Wisnom MR, Mahadik Y, Hallett SR. An experimental investiga-
1105 tion into size effects in quasi-isotropic carbon/epoxy laminates with sharp
1106 and blunt notches. *Compos Sci Technol* 2014;100:220–7.
- 1107 [36] Xu X, Wisnom MR, Mahadik Y, Hallett SR. Scaling of fracture response in
1108 Over-height Compact Tension tests. *Compos Part A-Appl S* 2015;69:40–8.
- 1109 [37] Laffan MJ, Pinho ST, Robinson P, Iannucci L. Measurement of the in situ
1110 ply fracture toughness associated with mode I fibre tensile failure in FRP.
1111 Part II: Size and lay-up effects. *Compos Sci Technol* 2010;70:614–21.
- 1112 [38] Teixeira RF, Pinho ST, Robinson P. Translaminar fracture toughness
1113 of CFRP: from the toughness of individual plies to the toughness of the
1114 laminates. In: *Proceedings of ECCM15. Venice; 2012*, p. 1–8.
- 1115 [39] Chen BY, Tay TE, Baiz PM, Pinho ST. Numerical analysis of size ef-
1116 fects on open-hole tensile composite laminates. *Compos Part A-Appl S*
1117 2013;47:52–62.
- 1118 [40] Erçin GH, Camanho PP, Xavier J, Catalanotti G, Mahdi S, Linde P.
1119 Size effects on the tensile and compressive failure of notched composite
1120 laminates. *Compos Struct* 2013;96:736–44.
- 1121 [41] Arteiro A, Catalanotti G, Xavier J, Camanho PP. Notched response of
1122 non-crimp fabric thin-ply laminates. *Compos Sci Technol* 2013;79:97–114.
- 1123 [42] Furtado C, Arteiro A, Catalanotti G, Xavier J, Camanho PP. Selective
1124 ply-level hybridisation for improved notched response of composite lami-
1125 nates. *Compos Struct* 2016;145:1–14.

- 1126 [43] Arteiro A, Catalanotti G, Xavier J, Linde P, Camanho PP. A strategy
1127 to improve the structural performance of non-crimp fabric thin-ply lami-
1128 nates. *Compos Struct* 2018;188:438–49.
- 1129 [44] Lavoie JA, Soutis C, Morton J. Apparent strength scaling in continuous
1130 fiber composite laminates. *Compos Sci Technol* 2000;60:283–99.
- 1131 [45] Wisnom MR, Khan B, Hallett SR. Size effects in unnotched tensile
1132 strength of unidirectional and quasi-isotropic carbon/epoxy composites.
1133 *Compos Struct* 2008;84:21–8.
- 1134 [46] Lee J, Soutis C. Measuring the notched compressive strength of composite
1135 laminates: Specimen size effects. *Compos Sci Technol* 2008;68:2359–66.
- 1136 [47] Soutis C, Lee J. Scaling effects in notched carbon fibre/epoxy composites
1137 loaded in compression. *J Mater Sci* 2008;43:6593–8.
- 1138 [48] Green BG, Wisnom MR, Hallett SR. An experimental investigation into
1139 the tensile strength scaling of notched composites. *Compos Part A-Appl*
1140 *S* 2007;38:867–78.
- 1141 [49] Wisnom MR, Hallett SR. The role of delamination in strength, failure
1142 mechanisms and hole size effect in open hole tensile tests on quasi-isotropic
1143 laminates. *Compos Part A-Appl S* 2009;40:335–42.
- 1144 [50] Wisnom MR, Hallett SR, Soutis C. Scaling effects in notched composites.
1145 *J Compos Mater* 2010;44(10):195–210.
- 1146 [51] Hitchen SA, Kemp RMJ. The effect of stacking sequence on impact dam-
1147 age in a carbon fibre/epoxy composite. *Composites* 1995;26:207–14.
- 1148 [52] Fuoss E, Straznicky PV, Poon C. Effects of stacking sequence on the
1149 impact resistance in composite laminates — Part 1: parametric study.
1150 *Compos Struct* 1998;41:67–77.
- 1151 [53] de Freitas M, Reis L. Failure mechanisms on composite specimens sub-
1152 jected to compression after impact. *Compos Struct* 1998;42:365–73.

- 1153 [54] González EV, Maimí P, Camanho PP, Lopes CS, Blanco N. Effects of
1154 ply clustering in laminated composite plates under low-velocity impact
1155 loading. *Compos Sci Technol* 2011;71:805–17.
- 1156 [55] Sebaey TA, González EV, Lopes CS, Blanco N, Costa J. Damage resis-
1157 tance and damage tolerance of dispersed CFRP laminates: Effect of ply
1158 clustering. *Compos Struct* 2013;106:96–103.
- 1159 [56] García-Moreno I, Caminero MA, Rodríguez GP. Effects of laminate thick-
1160 ness and ply-stacking sequence on the low velocity impact resistance of
1161 carbon fiber-reinforced laminates. *Materiales Compuestos* 2017;1(1):35–
1162 44.
- 1163 [57] Saito H, Morita M, Kawabe K, Kanesaki M, Takeuchi H, Tanaka M, et al.
1164 Effect of ply-thickness on impact damage morphology in CFRP laminates.
1165 *J Reinf Plast Comp* 2011;30:1097–106.
- 1166 [58] Sasikumar A, Trias D, Costa J, Blanco N, Orr J, Linde P. Effect of
1167 ply thickness and ply level hybridization on the compression after impact
1168 strength of thin laminates. *Compos Part A-Appl S* 2019;121:232–43.
- 1169 [59] Sasikumar A, Trias D, Costa J, Blanco N, Orr J, Linde P. Impact and
1170 compression after impact response in thin laminates of spread-tow woven
1171 and non-crimp fabrics. *Compos Struct* 2019;215:432–45.
- 1172 [60] Sasikumar A, Trias D, Costa J, Singery V, Linde P. Mitigating the weak
1173 impact response of thin-ply based thin laminates through an unsymmetri-
1174 cal laminate design incorporating intermediate grade plies. *Compos Struct*
1175 2019;220:93–104.
- 1176 [61] Nixon-Pearson OJ, Hallett SR, Withers PJ, Rouse J. Damage develop-
1177 ment in open-hole composite specimens in fatigue. Part 1: Experimental
1178 investigation. *Compos Struct* 2013;106:882–9.

- 1179 [62] Nishikawa Y, Okubo K, Fujii T, Kawabe K. Fatigue crack constraint
1180 in plain-woven CFRP using newly-developed spread tows. *Int J Fatigue*
1181 2006;28:1248–53.
- 1182 [63] Moon JB, Kim MG, Kim CG, Bhowmik S. Improvement of tensile prop-
1183 erties of CFRP composites under LEO space environment by applying
1184 MWNTs and thin-ply. *Compos Part A-Appl S* 2011;42:694–701.
- 1185 [64] Yamashita S, Sonehara T, Takahashi J, Kawabe K, Murakami T. Ef-
1186 fect of thin-ply on damage behaviour of continuous and discontinuous
1187 carbon fibre reinforced thermoplastics subjected to simulated lightning
1188 strike. *Compos Part A-Appl S* 2017;95:132–40.
- 1189 [65] Yamashita S, Hirano Y, Sonehara T, Takahashi J, Kawabe K, Murakami
1190 T. Residual mechanical properties of carbon fibre reinforced thermoplas-
1191 tics with thin-ply prepreg after simulated lightning strike. *Compos Part*
1192 *A-Appl S* 2017;101:185–94.
- 1193 [66] Saito H, Takeuchi H, Kimpara I. Experimental evaluation of the damage
1194 growth restraining in 90° layer of thin-ply CFRP cross-ply laminates. *Adv*
1195 *Compos Mater* 2012;21:57–66.
- 1196 [67] Sebaey TA, Costa J, Maimí P, Batista Y, Blanco N, Mayugo JA. Mea-
1197 surement of the *in situ* transverse tensile strength of composite plies by
1198 means of the real time monitoring of microcracking. *Compos Part B-Eng*
1199 2014;65:40–6.
- 1200 [68] Yokozeiki T, Aoki T, Ogasawara T, Ishikawa T. Effects of layup angle and
1201 ply thickness on matrix crack interaction in contiguous plies of composite
1202 laminates. *Compos Part A-Appl S* 2005;36:1229–35.
- 1203 [69] Camanho PP, Dávila CG, Pinho ST, Iannucci L, Robinson P. Prediction
1204 of in situ strengths and matrix cracking in composites under transverse
1205 tension and in-plane shear. *Compos Part A-Appl S* 2006;37:165–76.

- 1206 [70] Adolfsson E, Gudmundson P. Matrix crack initiation and progression in
1207 composite laminates subjected to bending and extension. *Int J Solids*
1208 *Struct* 1999;36:3131–69.
- 1209 [71] Andersons J, Joffe R, Spārniņš E, Rubenis O. Progressive cracking
1210 mastercurves of the transverse ply in a laminate. *Polym Composite*
1211 2009;30(8):1175–82.
- 1212 [72] Saito H, Takeuchi H, Kimpara I. A study of crack suppression mechanism
1213 of thin-ply carbon-fiber-reinforced polymer laminate with mesoscopic nu-
1214 merical simulation. *J Compos Mater* 2014;48(17):2085–96.
- 1215 [73] Arteiro A, Catalanotti G, Melro AR, Linde P, Camanho PP. Micro-
1216 mechanical analysis of the *in situ* effect in polymer composite laminates.
1217 *Compos Struct* 2014;116:827–40.
- 1218 [74] Chang FK, Chen MH. The in situ ply shear strength distribution in
1219 graphite/epoxy laminated composites. *J Compos Mater* 1987;21:708–33.
- 1220 [75] Catalanotti G. Prediction of *in situ* strengths in composites: Some con-
1221 siderations. *Compos Struct* 2019;207:889–93.
- 1222 [76] Herakovich CT. Influence of layer thickness on the strength of angle-ply
1223 laminates. *J Compos Mater* 1982;16:216–27.
- 1224 [77] Catalanotti G, Camanho PP, Marques AT. Three-dimensional failure
1225 criteria for fiber-reinforced laminates. *Compos Struct* 2013;95:63–79.
- 1226 [78] Camanho PP, Arteiro A, Catalanotti G, Melro AR, Vogler M. Three-
1227 dimensional invariant-based failure criteria for transversely isotropic fibre-
1228 reinforced composites. In: Camanho PP, Hallett SR, editors. *Numerical*
1229 *Modelling of Failure in Advanced Composite Materials*; chap. 5. Cam-
1230 bridge: Woodhead Publishing; 2015, p. 111–50.
- 1231 [79] Olsson R. Analytical prediction of damage due to large mass impact on
1232 thin ply composites. *Compos Part A-Appl S* 2015;72:184–91.

- 1233 [80] Bhudolia SK, Joshi SC, Boon YD. Experimental and microscopic investi-
1234 gation on mechanical performance of textile spread-tow thin ply compos-
1235 ites. *Fiber Polym* 2019;20(5):1036–45.
- 1236 [81] Arteiro A, Catalanotti G, Melro AR, Linde P, Camanho PP. Micro-
1237 mechanical analysis of the effect of ply thickness on the transverse
1238 compressive strength of polymer composites. *Compos Part A-Appl S*
1239 2015;79:127–37.
- 1240 [82] Fuller J, Wisnom MR. Pseudo-ductility and damage suppression in thin
1241 ply CFRP angle-ply laminates. *Compos Part A-Appl S* 2015;69:64–71.
- 1242 [83] Yokozeki T, Aoki T, Ishikawa T. Transverse crack propagation in the
1243 specimen width direction of CFRP laminates under static tensile loadings.
1244 *J Compos Mater* 2002;36(17):2085–99.
- 1245 [84] Chang KY, Liu S, Chang FK. Damage tolerance of laminated composites
1246 containing an open hole subjected to tensile loadings. *J Compos Mater*
1247 1991;25:274–301.
- 1248 [85] Huchette C. Sur la complémentarité des approches expérimentales et
1249 numériques pour la modélisation des mécanismes d’endommagement des
1250 composites stratifiés. Ph.D. thesis; Université Paris 6; Paris; 2005. In
1251 French.
- 1252 [86] Pierron F, Green B, Wisnom MR. Full-field assessment of the damage pro-
1253 cess of laminated composite open-hole tensile specimens. Part I: Method-
1254 ology. *Compos Part A-Appl S* 2007;38:2307–20.
- 1255 [87] Pierron F, Green B, Wisnom MR, Hallett SR. Full-field assessment of the
1256 damage process of laminated composite open-hole tensile specimens. Part
1257 II: Experimental results. *Compos Part A-Appl S* 2007;38:2321–32.
- 1258 [88] Laurin F, Carrere N, Huchette C, Maire JF. A multiscale hybrid approach
1259 for damage and final failure predictions of composite structures. *J Compos*
1260 *Mater* 2013;47(20–21):2713–47.

- [89] Ho YC, Sasayama H, Yanagimoto J. Mechanical properties and drawing process of multilayer carbon-fiber-reinforced plastic sheets with various prepreg thicknesses. *Adv Mech Engng* 2017;9(3):1–12.
- [90] Sasayama H, Kawabe K, Tomoda S, Ohsawa I, Kageyama K, Ogata N. Effect of lamina thickness on first ply failure in multidirectionally laminated composites. *J Jpn Soc Compos Mater* 2004;30(4):142–8. In Japanese.
- [91] Takagi K, Nakatani H, Suga K, Koike A, Ogiwara S. Effect of ply thickness on mechanical property of CFRP symmetric angle-ply laminates. In: *Proceedings of ICCM-18*. Jeju Island; 2011, p. 1–4.
- [92] Ogiwara S, Nakatani H. Effect of ply thickness on mechanical properties in CFRP angle-ply laminates. In: *Proceedings of ECCM15*. Venice; 2012, p. 1–6.
- [93] Kohler S, Cugnoni J, Amacher R, Botsis J. Transverse cracking in the bulk and at the free edge of thin-ply composites: Experiments and multiscale modelling. *Compos Part A-Appl S* 2019;124:105468.
- [94] Fernberg P, Joffe R. Thin-ply effects on long-term thermal stability of high temperature polyimide composites. In: *Proceedings of ICCM-22*. Melbourne; 2019, p. 1–2.
- [95] Guillaumet G. Improved delamination resistance of thin-ply based laminates: an experimental and numerical study. Ph.D. thesis; Universitat de Girona; Girona; 2016.
- [96] Chang FK, Scott RA, Springer GS. The effect of laminate configuration on characteristic lengths and rail shear strength. *J Compos Mater* 1984;18:290–6.
- [97] Liu H, Falzon BG, Catalanotti G. Studies on the effects of lamina thickness and orientation on the shear response of composite laminates. In: *Proceedings of ICCM-21*. Xi'an; 2017, p. 1–10.

- 1288 [98] Kawabe K, Sasayama H, Kageyama K, Ogata N. Effect of ply thickness
1289 on compressive properties in multidirectionally laminated composites. J
1290 Jpn Soc Compos Mater 2008;34:173–81. In Japanese.
- 1291 [99] Arteiro A, Catalanotti G, Reinoso J, Linde P, Camanho PP. Simulation
1292 of the mechanical response of thin-ply composites: From computational
1293 micro-mechanics to structural analysis. Arch Computat Methods Eng;
1294 2018. <https://doi.org/10.1007/s11831-018-9291-2>.
- 1295 [100] Reis L, de Freitas M. Damage growth analysis of low velocity impacted
1296 composite panels. Compos Struct 1997;38(1–4):509–15.
- 1297 [101] Aoki Y, Suemasu H, Ishikawa T. Damage propagation in CFRP laminates
1298 subjected to low velocity impact and static indentation. Adv Compos
1299 Mater 2007;16:45–61.
- 1300 [102] Lopes CS, Seresta O, Coquet Y, Gürdal Z, Camanho PP, Thuis B. Low-
1301 velocity impact damage on dispersed stacking sequence laminates. Part I:
1302 Experiments. Compos Sci Technol 2009;69:926–36.
- 1303 [103] París F, Blázquez A, McCartney LN, Barroso A. Characterization and
1304 evolution of matrix and interface related damage in $[0/90]_S$ laminates
1305 under tension. Part II: Experimental evidence. Compos Sci Technol
1306 2010;70:1176–83.
- 1307 [104] Dávila CG, Camanho PP. Analysis of the effects of residual strains and
1308 defects on skin/stiffener debonding using decohesion elements. Tech. Rep.
1309 AIAA Paper 2003-1465; NASA Langley Research Center; Hampton, Vir-
1310 ginia; 2003.
- 1311 [105] Lavoie JA, Adolfsson E. Stitch cracks in constraint plies adjacent to a
1312 cracked ply. J Compos Mater 2001;35(23):2077–97.
- 1313 [106] Zubillaga L, Turon A, Maimí P, Costa J, Mahdi S, Linde P. An energy
1314 based failure criterion for matrix crack induced delamination in laminated
1315 composite structures. Compos Struct 2014;112:339–44.

- [107] Hajikazemi M, Van Paepegem W. Variational analysis of free-edge stress and displacement fields in general un-symmetric and thin-ply laminates under in-plane, bending and thermal loading. *Compos Part A-Appl S* 2018;113:220–32.
- [108] García-Rodríguez SM, Costa J, Singery V, Sasikumar A. A 3D tomographic investigation to elucidate how thin-ply laminates improve the interlaminar shear strength and the effect of toughening the interfaces with veils. In: *Proceedings of the 10th European Solid Mechanics Conference*. Bologna; 2018, p. 1.
- [109] Arca MA. Strengthening of L-shaped composite laminates using carbon nanotube reinforcement and thin ply non-crimp fabrics. Master’s thesis; Graduate School of Natural and Applied Sciences, Middle East Technical University; Ankara; 2014.
- [110] Furtado C, Arteiro A, Bessa MA, Wardle BL, Camanho PP. Prediction of size effects in open-hole laminates using only the Young’s modulus, the strength, and the $\mathcal{R}$ -curve of the 0° ply. *Compos Part A-Appl S* 2017;101:306–17.
- [111] Reinoso J, Arteiro A, Paggi M, Camanho PP. Strength prediction of notched thin ply laminates using finite fracture mechanics and the phase field approach. *Compos Sci Technol* 2017;150:205–16.
- [112] Kötter B, Polyak D, Körbelin J, Fiedler B. Influence of ply thickness on failure initiation, propagation and mechanical properties in CFRP laminates. In: *Proceedings of ICFC 7*. Vicenza; 2018, p. 1–10.
- [113] Cugnoni J, Amacher R, Kohler S, Brunner J, Kramer E, Dransfeld C, et al. Towards aerospace grade thin-ply composites: Effect of ply thickness, fibre, matrix and interlayer toughening on strength and damage tolerance. *Compos Sci Technol* 2018;168:467–77.

- [114] Laffan MJ, Pinho ST, Robinson P, Iannucci L. Measurement of the in situ ply fracture toughness associated with mode I fibre tensile failure in FRP. Part I: Data reduction. *Compos Sci Technol* 2010;70:606–13.
- [115] Catalanotti G, Xavier J, Camanho PP. Measurement of the compressive crack resistance curve of composites using the size effect law. *Compos Part A-Appl S* 2014;56:300–7.
- [116] Catalanotti G, Arteiro A, Hayati M, Camanho PP. Determination of the mode I crack resistance curve of polymer composites using the size-effect law. *Engng Fract Mech* 2014;118:49–65.
- [117] Catalanotti G, Xavier J. Measurement of the mode II intralaminar fracture toughness and R-curve of polymer composites using a modified Iosipescu specimen and the size effect law. *Engng Fract Mech* 2015;138:202–14.
- [118] Maimí P, Camanho PP, Mayugo JA, Dávila CG. A continuum damage model for composite laminates: Part I — Constitutive model. *Mech Mater* 2007;39:897–908.
- [119] Maimí P, Camanho PP, Mayugo JA, Dávila CG. A continuum damage model for composite laminates: Part II — Computational implementation and validation. *Mech Mater* 2007;39:909–19.
- [120] Dávila CG, Rose CA, Camanho PP. A procedure for superposing linear cohesive laws to represent multiple damage mechanisms in the fracture of composites. *Int J Fract* 2009;158:211–23.
- [121] Camanho PP, Bessa MA, Catalanotti G, Vogler M, Rolfes R. Modeling the inelastic deformation and fracture of polymer composites — Part II: Smeared crack model. *Mech Mater* 2013;59:36–49.
- [122] Arteiro A, Catalanotti G, Xavier J, Camanho PP. Large damage capability of non-crimp fabric thin-ply laminates. *Compos Part A-Appl S* 2014;63:110–22.

- 1371 [123] Pinho ST, Robinson P, Iannucci L. Fracture toughness of the tensile and
1372 compressive fiber failure modes in laminated composites. *Compos Sci*
1373 *Technol* 2006;66:2069–79.
- 1374 [124] Catalanotti G, Camanho PP, Xavier J, Dávila CG, Marques AT. Measure-
1375 ment of resistance curves in the longitudinal failure of composites using
1376 digital image correlation. *Compos Sci Technol* 2010;70:1986–93.
- 1377 [125] Laffan MJ, Pinho ST, Robinson P, McMillan AJ. Translaminar fracture
1378 toughness testing of composites: A review. *Polym Test* 2012;31:481–9.
- 1379 [126] Ortega A, Maimí P, González EV, Trias D. Characterization of the
1380 translaminar fracture Cohesive Law. *Compos Part A-Appl S* 2016;91:501–
1381 9.
- 1382 [127] Ortega A, Maimí P, González EV, Trias D. Specimen geometry and speci-
1383 men size dependence of the $\mathcal{R}$ -curve and the size effect law from a cohesive
1384 model point of view. *Compos Part A-Appl S* 2017;205:239–54.
- 1385 [128] Walker TH, Avery WB, Ilcewicz LB, Poe, Jr. CC, Harris CE. Tension
1386 fracture of laminates for transport fuselage. Part I: material screening.
1387 In: Soderquist JR, Neri LM, Bohon HL, editors. *Proceedings of the 9th*
1388 *DoD/NASA/FAA conference on fibrous composites in structural design;*
1389 *vol. II. Lake Tahoe, Nevada; 1991, p. 747–87.*
- 1390 [129] Coats TW, Harris CE. A progressive damage methodology for residual
1391 strength predictions of notched composite panels. *Tech. Rep. NASA/TM-*
1392 *1998-207646; NASA Langley Research Center; Hampton, Virginia; 1998.*
- 1393 [130] Coats TW, Harris CE. A progressive damage methodology for resid-
1394 ual strength predictions of notched composite panels. *J Compos Mater*
1395 1999;33(23):2193–224.
- 1396 [131] Caminero MA, Lopez-Pedrosa M, Pinna C, Soutis C. Damage moni-
1397 toring and analysis of composite laminates with an open hole and adhe-

1398 sively bonded repairs using digital image correlation. *Compos Part B-Eng*
1399 2013;53:76–91.

1400 [132] Bergan AC, Dávila CG, Leone FA, Awerbuch J, Tan TM. Mode I cohesive
1401 law characterization of through-crack propagation in a multidirectional
1402 laminate. Tech. Rep. NF1676L-18222; NASA Langley Research Center;
1403 Hampton, Virginia; 2014.

1404 [133] Fernandes MFC. Translaminar fracture of thin-ply composite laminates.
1405 Master’s thesis; Faculdade de Engenharia, Universidade do Porto; Porto;
1406 2014.

1407 [134] Bullegas G, Pinho ST, Pimenta S. Engineering the translaminar fracture
1408 behaviour of thin-ply composites. *Compos Sci Technol* 2016;131:110–22.

1409 [135] Bažant ZP, Planas J. Fracture and Size Effect in Concrete and Other
1410 Quasibrittle Materials. CRC Press LLC; 1997.

1411 [136] Bažant ZP, Daniel IM, Li Z. Size effect and fracture characteristics of
1412 composite laminates. *Mater Struct* 1996;118:317–24.

1413 [137] Bažant ZP, Kim JK, Pfeiffer PA. Nonlinear fracture properties from size
1414 effect tests. *J Struct Eng-ASCE* 1986;112:289–307.

1415 [138] Bažant ZP. Size effect on structural strength: a review. *Arch Appl Mech*
1416 1999;69:703–25.

1417 [139] Bažant ZP, Yavari A. Is the cause of size effect on structural strength
1418 fractal or energetic-statistical? *Engng Fract Mech* 2005;72:1–31.

1419 [140] Camanho PP, Catalanotti G. On the relation between the mode I fracture
1420 toughness of a composite laminate and that of a 0° ply: Analytical model
1421 and experimental validation. *Engng Fract Mech* 2011;78:2535–46.

1422 [141] CMH-17-3G . Composite Materials Handbook; vol. 3 of 6: Polymer Matrix
1423 Composites Materials Usage, Design and Analysis. SAE International;
1424 2012.

- [142] Iarve EV, Mollenhauer D, Kim R. Theoretical and experimental investigation of stress redistribution in open hole composite laminates due to damage accumulation. *Compos Part A-Appl S* 2005;36:163–71.
- [143] Wisnom MR. Size effects in the testing of fibre-composite materials. *Compos Sci Technol* 1999;59:1937–57.
- [144] Bažant ZP. Size effect. *Int J Solids Struct* 2000;37:69–80.
- [145] Hallett SR, Green BG, Jiang WG, Wisnom MR. An experimental and numerical investigation into the damage mechanisms in notched composites. *Compos Part A-Appl S* 2009;40:613–24.
- [146] Lee J, Soutis C. Thickness effect on the compressive strength of T800/924C carbon fibre-epoxy laminates. *Compos Part A-Appl S* 2005;36:213–27.
- [147] Camanho PP, Maimí P, Dávila CG. Prediction of size effects in notched laminates using continuum damage mechanics. *Compos Sci Technol* 2007;67:2715–27.
- [148] Chang JB, Goyal VK, Klug JC, Rome JI. Composite structures damage tolerance analysis methodologies. Tech. Rep. NASA/CR-2012-217347; NASA Langley Research Center; Hampton, Virginia; 2012.
- [149] de Moura MFSF, Gonçalves JPM, Marques AT, de Castro PMST. Modelling compression failure after low velocity impact on laminated composites using interface elements. *J Compos Mater* 1997;31(15):1462–79.
- [150] Richardson MOW, Wisheart MJ. Review of low-velocity impact properties of composite materials. *Compos Part A-Appl S* 1996;27A:1123–31.
- [151] Olsson R, Ahlqvist F, André A, Hellström P, Alvarez E, González EV, et al. Testing and modelling of tension after impact of a thin ply textile composite. In: *Proceedings of ECCM17*. Munich; 2016, p. 1–8.

- 1451 [152] García-Rodríguez SM, Costa J, Bardera A, Singery V, Trias D. A 3D
1452 tomographic investigation to elucidate the low-velocity impact resistance,
1453 tolerance and damage sequence of thin non-crimp fabric laminates: effect
1454 of ply-thickness. *Compos Part A-Appl S* 2018;113:53–65.
- 1455 [153] Soto A, González EV, Maimí P, Martín de la Escalera F, Sainz de Aja JR,
1456 Alvarez E. Low velocity impact and compression after impact simulation
1457 of thin ply laminates. *Compos Part A-Appl S* 2018;109:413–27.
- 1458 [154] Yamada K, Yamamoto S, Kanesaki M, Nishikawa M, Matsuda N, Kawabe
1459 K, et al. Effect of ply thickness on impact damage modes of thin-ply CFRP
1460 laminates. In: *Proceedings of ICCM-22*. Melbourne; 2019, p. 1–12.
- 1461 [155] García-Rodríguez SM, Bardera A, Sasikumar A, Ruiz M, Singery V,
1462 Boada I, et al. Effect of ply-thickness on the impact response of inter-
1463 leaved thin-ply non-crimp-fabric laminates: X-ray tomography investiga-
1464 tion. *Materiales Compuestos* 2017;1(1):57–64.
- 1465 [156] García-Rodríguez SM, Costa J, Singery V, Boada I, Mayugo JA. The effect
1466 interleaving has on thin-ply non-crimp fabric laminate impact response:
1467 X-ray tomography investigation. *Compos Part A-Appl S* 2018;107:409–20.
- 1468 [157] Naghipour P, Bartsch M, Voggenreiter H. Simulation and experimen-
1469 tal validation of mixed mode delamination in multidirectional CF/PEEK
1470 laminates under fatigue loading. *Int J Solids Struct* 2011;48:1070–81.
- 1471 [158] Allegri G, Wisnom MR. A non-linear damage evolution model for mode
1472 II fatigue delamination onset and growth. *Int J Fatigue* 2012;43:226–34.
- 1473 [159] Chen WH, Lee SS, Yeh JT. Three-dimensional contact stress analysis of
1474 a composite laminate with bolted joint. *Compos Struct* 1995;30:287–97.
- 1475 [160] Camanho PP, Matthews FL. Stress analysis and strength prediction of
1476 mechanically fastened joints in FRP: a review. *Compos Part A-Appl S*
1477 1997;28A:529–47.

- 1478 [161] Camanho PP, Tavares CML, de Oliveira R, Marques AT, Ferreira AJM.
1479 Increasing the efficiency of composite single-shear lap joints using bonded
1480 inserts. *Compos Part B-Eng* 2005;36:372–83.
- 1481 [162] Camanho PP, Lambert M. A design methodology for mechanically fas-
1482 tened joints in laminated composite materials. *Compos Sci Technol*
1483 2006;66:3004–20.
- 1484 [163] Catalanotti G, Camanho PP. A semi-analytical method to predict net-
1485 tension failure of mechanically fastened joints in composite laminates.
1486 *Compos Sci Technol* 2013;76:69–76.
- 1487 [164] Standard test method for bearing response of polymer matrix compos-
1488 ite laminates, ASTM D5961/D5961M – 13. ASTM International; West
1489 Conshohocken, PA, USA; 2013.
- 1490 [165] Catalanotti G, Camanho PP, Ghys P, Marques AT. Experimental and nu-
1491 merical study of fastener pull-through failure in GFRP laminates. *Compos*
1492 *Struct* 2011;94:239–45.
- 1493 [166] Cooper C, Turvey GJ. Effects of joint geometry and bolt torque on the
1494 structural performance of single bolt tension joints in pultruded GRP
1495 sheet material. *Compos Struct* 1995;32:217–26.
- 1496 [167] Camanho PP. Application of numerical methods to the strength predic-
1497 tion of mechanically fastened joints in composite laminates. Ph.D. thesis;
1498 Centre for Composite Materials, Imperial College of Science, Technology
1499 and Medicine; London; 1999.
- 1500 [168] Chang FK, Chang KY. Post-failure analysis of bolted composite joints in
1501 tension or shear-out mode failure. *J Compos Mater* 1987;21:809–33.
- 1502 [169] Garbo SP, Ogonowski JM. Effect of variances and manufacturing toler-
1503 ances on the design strength and life of mechanically fastened composite
1504 joints. Volume 1 - Methodology development and data evaluation. Fi-
1505 nal report AFWAL-TR-81-3041; Flight Dynamics Laboratory, Air Force

- 1506 Wright Aeronautical Laboratories; Air Force Systems Command, Wright-
1507 Patterson Air Force Base, Ohio; 1981.
- 1508 [170] Camanho PP, Matthews FL. Delamination onset prediction in me-
1509 chanically fastened joints in composite laminates. *J Compos Mater*
1510 1999;33(10):906–27.
- 1511 [171] Feraboli P, Miller M. Damage resistance and tolerance of carbon/epoxy
1512 composite coupons subjected to simulated lightning strike. *Compos Part*
1513 *A-Appl S* 2009;40:954–67.
- 1514 [172] Wu H, Li S, Zhang Z, Tong L. Electrical responses of thin-ply composites.
1515 In: *Proceedings of ICCM-22*. Melbourne; 2019, p. 1–8.
- 1516 [173] Russello M, Diamanti EK, Catalanotti G, Ohlsson F, Hawkins SC, Fal-
1517 zon BG. Enhancing the electrical conductivity of carbon fibre thin-ply
1518 laminates with directly grown aligned carbon nanotubes. *Compos Struct*
1519 2018;206:272–8.
- 1520 [174] El-Dessouky HM. Spread tow technology for ultra lightweight CFRP com-
1521 posites: potential and possibilities. In: Bafekrpour E, editor. *Advanced*
1522 *Composite Materials: Properties and Applications*; chap. 6. Warsaw: De
1523 Gruyter Open; 2017, p. 323–48.
- 1524 [175] Mouane K. Polyimide thin-ply composite. Master’s thesis; Luleå Univer-
1525 sity of Technology, Department of Engineering Sciences and Mathematics;
1526 2018.
- 1527 [176] Tsai SW. *Strength and Life of Composites*. Composites Design Group,
1528 Department of Aeronautics and Astronautics, Stanford University; 2008.
- 1529 [177] Tsai SW, Melo JDD. An invariant-based theory of composites. *Compos*
1530 *Sci Technol* 2014;100:237–43.
- 1531 [178] Baker N, Butler R, York CB. Damage tolerance of fully orthotropic lam-
1532 inates in compression. *Compos Sci Technol* 2012;72:1083–9.

- 1533 [179] Cadena DSA. Improvement in damage tolerance and buckling behaviour
1534 of a composite fuselage panel using thin plies. Master's thesis; Politecnico
1535 de Milano, Facoltà di Ingegneria Industriale; Milan; 2014.
- 1536 [180] Tao J, Sun CT. Influence of ply orientation on delamination in composite
1537 laminates. *J Compos Mater* 1998;32(21):1933–47.
- 1538 [181] Mencattelli L, Pinho ST. Realising bio-inspired impact damage-tolerant
1539 thin-ply CFRP Bouligand structures via promoting diffused sub-critical
1540 helicoidal damage. *Compos Sci Technol* 2019;182:107684.
- 1541 [182] Fuller J, Wisnom MR. Ductility and pseudo-ductility of thin ply angle-ply
1542 CFRP laminates under quasi-static cyclic loading. *Compos Part A-Appl*
1543 *S* 2018;107:31–8.
- 1544 [183] Czél G, Wisnom MR. Demonstration of pseudo-ductility in high per-
1545 formance glass/epoxy composites by hybridisation with thin-ply carbon
1546 prepreg. *Compos Part A-Appl S* 2013;52:23–30.
- 1547 [184] Czél G, Rév T, Jalalvand M, Fotouhi M, Wisnom MR. Demonstration of
1548 pseudo-ductility in quasi-isotropic laminates comprising thin-ply UD car-
1549 bon/epoxy hybrid sub-laminates. In: *Proceedings of ECCM17*. Munich;
1550 2016, p. 1–8.
- 1551 [185] Czél G, Jalalvand M, Wisnom MR, Czigány T. Design and characterisa-
1552 tion of high performance, pseudo-ductile all-carbon/epoxy unidirectional
1553 hybrid composites. *Compos Part B-Eng* 2017;111:348–56.
- 1554 [186] Fotouhi M, Jalalvand M, Wisnom MR. High performance quasi-isotropic
1555 thin-ply carbon/glass hybrid composites with pseudo-ductile behaviour in
1556 all fibre orientations. *Compos Sci Technol* 2017;152:101–10.
- 1557 [187] Czél G, Rev T, Jalalvand M, Fotouhi M, Longana ML, Nixon-Pearson OJ,
1558 et al. An experimental investigation into the tensile strength scaling of
1559 notched composites. *Compos Part A-Appl S* 2018;104:151–64.

- 1560 [188] Fotouhi M, Jalalvand M, Wisnom MR. Notch insensitive orientation-
1561 dispersed pseudo-ductile thin-ply carbon/glass hybrid laminates. *Compos*
1562 *Part A-Appl S* 2018;110:29–44.
- 1563 [189] Cugnoni J, Frossard G, Amacher R, Botsis J. Translaminar fracture of
1564 regular and hybrid thin ply composites: experimental characterization
1565 and modeling. In: *Proceedings of ECCM18*. Athens; 2018, p. 1–8.
- 1566 [190] Wisnom MR, Czél G, Swolfs Y, Jalalvand M, Gorbatiikh L, Verpoest I.
1567 Hybrid effects in thin ply carbon/glass unidirectional laminates: Accu-
1568 rate experimental determination and prediction. *Compos Part A-Appl S*
1569 2016;88:131–9.
- 1570 [191] Ogiwara S, Takeda N. Interaction between transverse cracks and delami-
1571 nation during damage progress in CFRP cross-ply laminates. *Compos Sci*
1572 *Technol* 1995;54:395–404.
- 1573 [192] Vallmajó O, Cózar IR, Furtado C, Tavares R, Arteiro A, Turon A, et al.
1574 Virtual calculation of the B-value allowables of notched composite lami-
1575 nates. *Compos Struct* 2019;212:11–21.
- 1576 [193] Amacher R, Cugnoni J, Botsis J. Thin ply composites: experimental
1577 characterization and modeling. In: *Proceedings of ICCM-19*. Montréal;
1578 2013, p. 1–13.
- 1579 [194] Teixeira RF, Pinho ST, Robinson P. Thickness-dependence of the
1580 translaminar fracture toughness: Experimental study using thin-ply com-
1581 posites. *Compos Part A-Appl S* 2016;90:33–44.

High-grade UD laminate (300 g/m²)

Low-grade UD laminate (100 g/m²)

Spread-tow UD laminate (30 g/m²)

Figure 1: Optical micrographs of carbon fibre/epoxy composites of different grades (after Amacher et al. [193]).

Figure 2: Effect of the uniformity of spread-tow laminae on the crack resistance curve associated with mode I inter-laminar fracture of UD polymer composites (after Frossard et al. [12]).

Figure 3: Transverse matrix cracks in a cross-ply laminate (after Sebaey et al. [67]).

Figure 4: *In-situ* effect on the transverse tensile strength of an embedded 90° ply (after Flaggs and Kural [22]).

Figure 5: *In-situ* transverse tensile and in-plane shear strengths of an embedded inner or outer ply (after Camanho et al. [69]).

Figure 6: Failure of blocked-ply, dispersed-ply and spread-tow angle-ply laminates (after Herakovich [76] and Fuller and Wisnom [82]).

Figure 7: Remote stress-strain relations and longitudinal strain fields (ϵ_x) obtained by digital image correlation at the stage prior to ultimate failure in plain weave [spread-tow fabrics](#) (after Arteiro et al. [6]). *STF* stands for [spread-tow fabric](#). Before ultimate failure, no transverse cracks could be observed in the 160 g/m² [spread-tow fabrics](#) (×). However, the outer transverse yarns of the 240 g/m² [spread-tow fabrics](#) developed several matrix cracks that contributed to the nonlinear tensile response of this laminate (×).

Figure 8: Linear and nonlinear predictions for the *in-situ* in-plane shear strength of a thin embedded ply (after Camanho et al. [69]).

Figure 9: *In-situ* effect on the failure envelopes for matrix-dominated damage mechanisms (after Camanho et al. [78]).

Figure 10: Transverse shear cracking in multi-directional laminates (after Olsson [79]).

Figure 11: *In-situ* transverse compressive strength predicted by computational micro-mechanics and phenomenological failure criteria (after Arteiro et al. [81]).

Figure 12: *In-situ* effect on wedge transverse cracking (after Arteiro et al. [81]). Representative volume elements loaded in compression by a uniaxial remote strain of 2.5% along the horizontal direction show that failure of conventional 90° plies is dominated by fibre-matrix interface cracking and large localised plastic deformation in a plane not aligned with the loading direction. Thin plies, on the other hand, show a dispersed damage mechanism, with reduced transverse cracking at the same applied remote strain.

Figure 13: Longitudinal-transverse failure envelopes for different values of applied through-thickness transverse compressive stress (σ_{33}) for an embedded and a UD ply (after Arteiro et al. [99]). Black lines correspond to $\sigma_{33} = 0$, and coloured lines to non-zero σ_{33} . The same value of σ_{33} is applied for envelopes in the same position with respect to the reference black lines ($\sigma_{33} = 0$).

Figure 14: Laminate applied strain at the onset of 90° ply cracking, at the onset of delamination, and at laminate failure (after O'Brien [26]). In the laminates with 4 or more plies blocked together, delamination is induced by matrix cracking (MCID), whereas in the laminates with less than 3 plies blocked together, delamination starts from the free edges (FED).

Figure 15: CT and CC coupon configurations (after Pinho et al. [123]).

Figure 16: X-ray imaging of $[90_2/(90/0)_{16}/90]_s$ (a and c) and $[90_2/(90_2/0_2)_8/90]_s$ (b and d) thin-ply CT specimens (after Teixeira et al. [194]). Split cracking is observed in laminates with intermediate ply thickness (b and d), but it is completely suppressed in laminates with thin plies (a and c).

(a) Split cracking and delamination at the notch tip on the central block of two 0° plies and on a single 0° ply of an over-height CT test specimen (after Xu et al. [36]).

(b) SEM micrograph of fracture surfaces of a thick 0° ply (after Laffan et al. [37]) and a thin 0° ply with an hieraphical pattern of micro-cuts (after Bullegas et al. [134]).

Figure 17: Post-mortem observations of intra-laminar fibre fracture.

Figure 18: Size effect law fitting and $\mathcal{R}$ -curve determined as the envelope of driving force curves (after Catalanotti et al. [116]).

Figure 19: Size effects in un-notched IM7/8552 carbon/epoxy coupons loaded in tension (after Wisnom et al. [45, 50]) and in compression (after Lee and Soutis [46]). Specimen dimensions are given as: gauge length (mm) × width (mm) × thickness (mm).

Figure 20: Size effects in notched M40JB/ThinPreg™ 80EP carbon/epoxy coupons loaded in tension, representative failed specimens, and surface longitudinal strain fields obtained with digital image correlation from specimens with a hole diameter of 2 mm (after Furtado et al. [42]).

Figure 21: Size effects in notched IM7/8552 carbon/epoxy coupons loaded in compression and damage development in representative open-hole compression specimens (after Lee and Soutis [46]).

Figure 22: Effect of laminate thickness on the notched tensile strengths of IM7/8552 carbon/epoxy coupons with different hole diameters d (after Wisnom et al. [50]).

Figure 23: Effect of laminate thickness on the notched compressive strengths of IM7/8552 carbon/epoxy coupons with different hole diameters d exhibiting valid failure mode (after Wisnom et al. [50]).

Figure 24: Impact response of laminates with different degrees of ply blocking (after González et al. [54]).

Figure 25: X-ray computed tomography images of interrupted quasi-static and fatigue open-hole tension tests (after Nixon-Pearson et al. [61]).

Figure 26: X-ray pictures of specimens subjected to tensile fatigue loading at room temperature, after 73,000 cycles, with a ply thickness of 0.04 mm and a total laminate thickness of 3.2 mm (after Sih et al. [28]).

Figure 27: Joint failure modes (after ASTM D5961/D5961M – 13 [164]).

Figure 28: Visual and ultrasonic C-scan images of carbon fibre-reinforced thermoplastic laminates subjected to lightning strikes (after Yamashita et al. [64]).

Figure 29: Excellent surface appearance of thin-ply non-crimp fabrics. Courtesy of Chomarar
©Laurent Becot Ruiz.

Figure 30: Illustration of ply drops with thin and thick plies.