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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

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reduction during conceptual and detailed design and operation.

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

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

2.2. Inter-laminar material properties

2.2.1. Inter-laminar shear strength

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

2.2.2. Inter-laminar fracture toughness

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

2.3. Discussion

It is apparent that the ply uniformity resulting from ply thinness has a strong beneficial effect on the longitudinal compressive strength and ILSS of UD composites. The enhanced longitudinal compressive strength can be considered one of the main contributors to the improved compressive plain strength of multi-directional laminates with thin plies [6, 14, 15]. However, the enhanced ILSS measured from UD coupons is not expected to have a major role on the ILSS of multi-directional laminates, since the effect of ply thickness on the inter-laminar stress distribution [7] is considered the dominant effect on the improved inter-laminar shear performance of multi-directional thin-ply laminates, overwhelming the micro-structural effects observed on UD laminates. For example, Kalfon-Cohen et al. [16] tested blocked (*thick baseline*) and dispersed (*thin baseline*) laminates manufactured from the same thin-ply composite system, obtaining higher ILSS on the *thin baseline*. Since the same thin-ply composite system was used in both laminates, no micro-structural effects should be expected, 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 strain-to-failure and the toughness (measured as the area under the experimentally obtained stress-strain curves) of angle-ply laminates is higher when plies of the same orientation are dispersed through the laminate instead of blocked together due to a delay in matrix cracking that caused a strong change in the nature and sequence of the failure mechanisms (Fig. 6). More recently, Fuller and Wisnom [82] showed that matrix cracking parallel to the fibre direction in angle-ply laminates can be completely suppressed using a spread-tow thin-ply composite material, leading to a brittle type of net-section fibre-dominated failure mode (Fig. 6 – right).

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

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

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

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

To clarify the nature of this size effect, Amacher et al. [4] tested thick-ply laminates produced from blocks of ten 30 g/m^2 thin plies and from individual 300 g/m^2 plies, and no substantial difference was found between these two laminates, demonstrating that the observed size effect was not related to changes in the material properties or micro-structure, but to the deterministic *in-situ* effect.

3.2.3. In-situ in-plane shear strength

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

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

3.2.4. In-situ transverse shear and transverse compressive strengths

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

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

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

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

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

3.3. Delamination

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

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

The study of the effect of ply thickness on delamination onset and growth dates back to the early 1980's [25, 26], when relevant analysis programmes showed that delaminations form at a lower nominal applied strain in laminates scaled by increasing the ply thickness (or the number of plies blocked together). Similar results were reported more recently [27]. Similarly, matrix cracking-induced delamination (MCID) initiates at lower applied strains when the ply thickness increases, findings supported by other authors [85, 105] and by analysis methods [106]. Due to lower *in-situ* strengths, early ply cracking occurs in the transverse plies (see Fig. 14), causing local stress concentrations at the interfaces with the surrounding plies that induce premature delamination onset. In addition, higher energy release rates (ERRs) at the interfaces between thick plies lead to early delamination onset and growth at those interfaces. The combination of these two effects contribute to the marked reduction of the delamination onset strain shown in Fig. 14 as the number of ply repetitions increases. Fig. 14 also shows that this early delamination onset leads to premature laminate failure, caused by the local stress concentrations at the load carrying plies induced by the early matrix cracks and delaminations. As the transverse plies become 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 laminate 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 increase of fracture toughness is due to the development of a large damage process zone containing split cracking and localised delaminations that modifies the stress field in the vicinity of the notch tip, as detailed earlier by Li et al. [33]. Intra-laminar fracture toughness scaling [39] allows accounting for the additional stress relaxation that results from the development of a larger damage process zone. However, it is not clear whether such virtual scaling of the fracture toughness is applicable to more complex loading scenarios, where the sequence and extent of the blunting mechanisms can be different.

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

Size effect law. As an alternative to CT- and CC-type test specimens, a methodology based on the *size effect law* [135] can be used to measure the intra-laminar fracture toughness and the $\mathcal{R}$-curve of composite laminates in mode I tension [116, 136] and compression [115], as well as in mode II [117]. According to this methodology, the $\mathcal{R}$-curve can be measured taking into account that, for different characteristic sizes (w_n), assuming that the size effect law $\bar{\sigma}^{\infty} = \bar{\sigma}^{\infty}(w)$ is known, the driving force curves $\mathcal{G}_I$ at the ultimate remote stresses ($\bar{\sigma}^{\infty}(w_n)$) are tangent to the $\mathcal{R}$-curve (Fig. 18). In other words, the $\mathcal{R}$-curve can be determined as the envelope of the crack driving force curves at the ultimate remote stresses (e.g. Ref. [137]). The size effect law can be easily obtained from *geometrically* *similar specimens*[1] of different sizes with a *positive geometry*[2] [115–117].

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

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

CNT specimens can also be used to measure the mode I tensile fracture 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].

[2] In 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.

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

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

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

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

3.5. Size effects

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

3.5.1. Size effects in smooth coupons

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

3.5.2. Hole size effects in tension

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

constant width-to-hole diameter ratio[4]) [48–50].

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

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

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

[4] It 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].

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

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

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

3.5.3. Hole size effects in compression

For the hole size effect in compression, three distinct failure mechanisms, all possibly leading to catastrophic failure, have been observed depending on size and lay-up, namely longitudinal compressive failure induced by shear-driven fibre fracture or by fibre kinking, push-out between plies and delamination [50]. Laminates with sufficiently thin plies exhibit a more brittle failure mode, characterised by a straight fracture across the laminate. Since sub-critical damage mechanisms, such as delamination and fibre splitting, are inhibited by the constraining effect of the thinner sub-laminates, the stress in the 0° plies increases to a level high enough to trigger longitudinal compressive failure. Laminates with thick plies (or thick ply blocks) show fracture of the angle plies, fibre-matrix splitting and local delamination or push-out between plies before 0° ply failure (Fig. 21). This local damage leads to a stress redistribution around the hole, reducing the stress concentration and delaying ultimate laminate failure to a higher applied stress. For this reason, the compressive notched strengths of laminates with blocked plies tend to be higher than the notched strengths of laminates with thin plies [15, 40, 41, 46, 47].

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

3.5.4. Effect of laminate thickness

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

Thickness scaling of un-notched laminates using sub-laminate repetitions leads to a slight strength increase (Fig. 19). Matrix cracking starts in the outer plies [69], which represent a smaller portion of the thicker laminates with 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 faces [4, 31, 58, 59] (due to very high compressive and tensile stresses) and large delaminations at the mid-surface [4, 32, 57, 59, 60, 151, 152] (where the highest shear stresses occur) appear in these laminates, an observation that was also confirmed numerically [153]. Deeper indentation footprints can also be expected due to the high local compressive stresses beneath the impactor, which can reduce the threshold for *barely visible impact damage* and lead to higher residual strengths at the threshold of detectability.

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

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

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

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

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

Improvements of the damage tolerance of laminates with thin plies can also be achieved through hybrid solutions [58] combining thick or standard plies with thin plies [42, 43], and through proper, more flexible laminate design [60]. On one hand, ply thickness hybridisation can alleviate the amount of fibre failure with increased delamination damage, improving damage tolerance [58]. In addition, the introduction of 0°-ply blocking, namely at the outer layers of the laminate, can improve the post-impact response, attributed to an increase of the bending stiffness of the surface sub-laminates [55]. On the other hand, motivated by the unsymmetrical nature of the damage modes induced by LVI in the laminate thickness direction, unsymmetrical designs have been proposed [60], enlarging the design space for optimal hybrid solutions combining thick or standard plies with thin plies. Appropriate unsymmetrical designs allow controlling 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 failure modes, namely bearing, net-tension, shear-out, cleavage and tear-out when subjected to in-plane loading (Fig. 27) [164], or pull-through when subjected to out-of-plane loading [165].

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

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

3.9. Environmental exposure

In a multi-directional laminate, the different ply orientations lead to a mismatch of the coefficients of thermal and moisture expansion. Moreover, temperature and moisture are known to aggressively affect the polymer matrix, degrading the mechanical properties of composite laminates [4]. Hence, the temperature variation during the curing process and the in-service environment can lead to early matrix cracking and inter-laminar damage growth. However, 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].

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

4. Opportunities in manufacturing and design

4.1. Ply uniformity

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

In addition, this unique uniformity leads to excellent surface appearance that may help reducing part production costs in terms of surface finishing and painting (Fig. 29). The improved fibre distribution also favours wider inter-fibre spacing, which allows easier flow of the matrix material, either in liquid moulding techniques employing thermosetting resins, or during impregnation by thermoplastic matrices [174], leading to fewer weak zones, i.e., thin plies lead to lower void content and smaller resin-rich areas [89, 174]. However, this wider inter-fibre spacing leads to lower fibre volume fractions in spread-tow yarns [62]. In addition, the number of resin-rich interfaces between layers is higher in thin-ply laminates (since they have more layers for the same thickness), increasing the relative amount of resin in the laminate. Consequently, the fibre volume fraction of thin-ply laminates is usually lower than the fibre volume fraction of conventional composite laminates [4]. However, this difference is typically small, with no implications reported so far [4].

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

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

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

4.2. Laminate design, homogenisation and hybridisation

With thinner plies, for the same laminate thickness, more plies can be accommodated. On one hand, the design space widens, due to higher flexibility in accommodating plies of different orientation. On the other hand, laminate homogenisation by sub-laminate repetition is easier to achieve for the typical laminate thickness currently used in advanced structural design. In this case, mid-plane symmetry becomes irrelevant[5] due to negligible thermal warpage [176, 177]. This can be beneficial, for example, in the design of stringers for aero-structures, where criteria to avoid warpage are required. With homogenised laminates, optimisation may rely only on topological optimisation, since warpage is negligible. This also enables simpler lay-up processes, since reversing the stacking sequence at the mid-plane is no longer needed. Continuous stacking becomes possible, and ply drops no longer need to be symmetric, reducing processing time, cost, waste and stacking errors. Besides, by avoiding the symmetry rule, the laminate design space widens further and optimisation for out-of-plane loading and damage tolerance, where non-symmetric laminates may perform better [60], becomes easier [178].

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

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

[5] A 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 superior 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.

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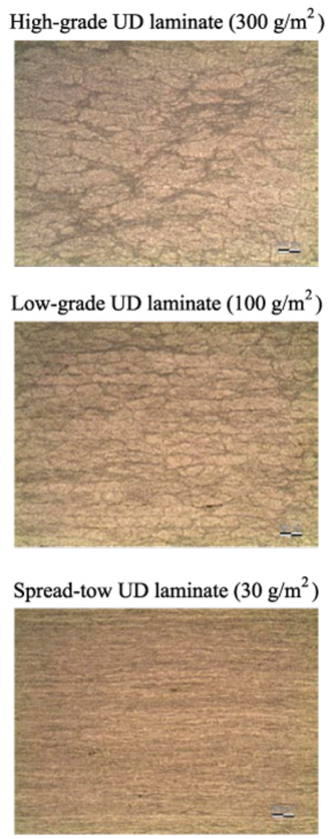


Figure 1: Optical micrographs of carbon fibre/epoxy composites of different grades (after Amacher et al. [193]).

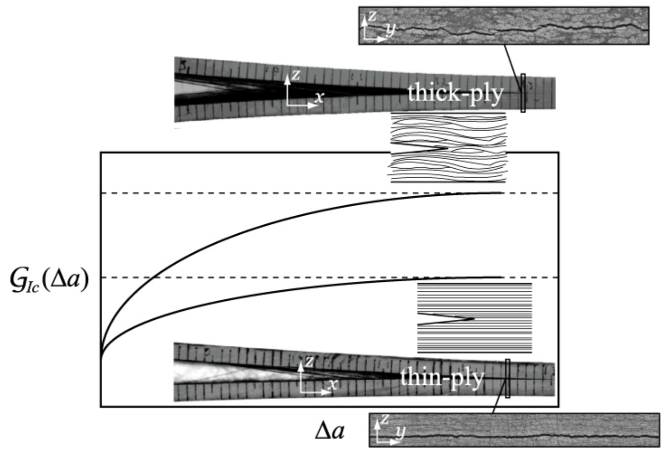


Figure 2: Effect of the uniformity of spread-tow laminae on the crack resistance curve associated with mode I inter-laminar fracture of UD polymer composites (after Frossard et al. [12]).

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

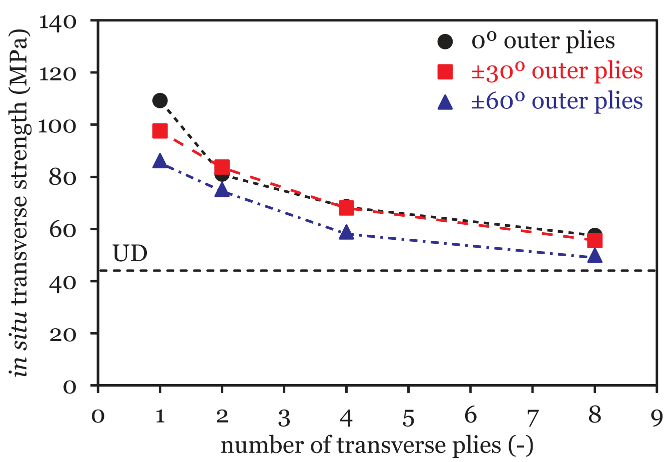


Figure 4: *In-situ* effect on the transverse tensile strength of an embedded 90° ply (after Flaggs and Kural [22]).

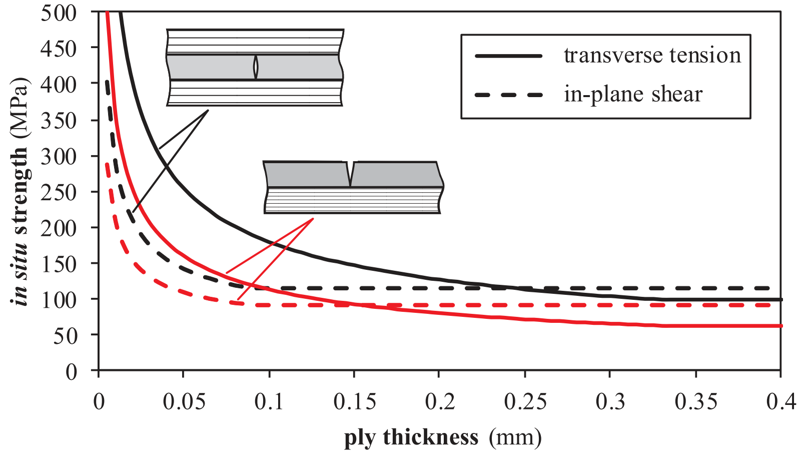


Figure 5: *In-situ* transverse tensile and in-plane shear strengths of an embedded inner or outer ply (after Camanho et al. [69]).

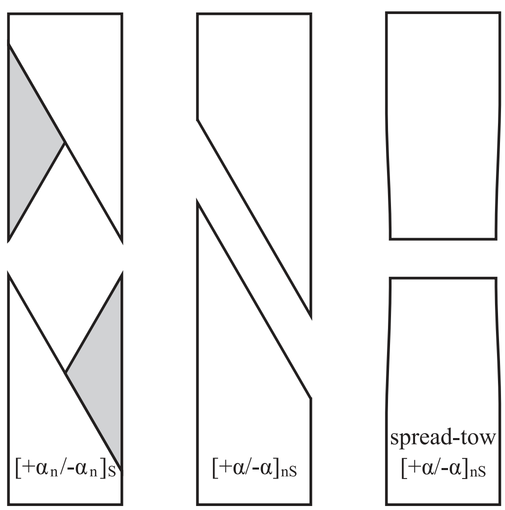


Figure 6: Failure of blocked-ply, dispersed-ply and spread-tow angle-ply laminates (after Herakovich [76] and Fuller and Wisnom [82]).

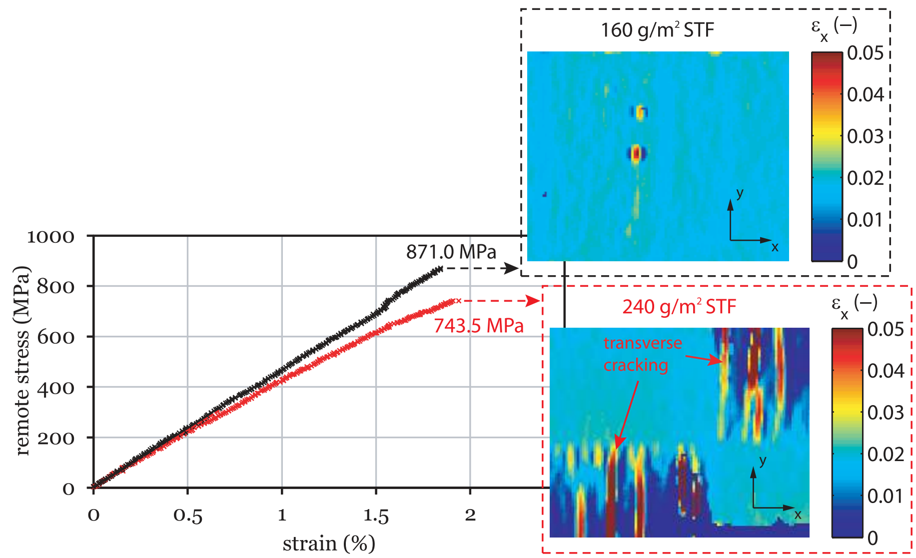


Figure 7: Remote stress-strain relations and longitudinal strain fields (ε_x) obtained by digital image correlation at the stage prior to ultimate failure in plain weave spread-tow fabrics (after Arteiro et al. [6]). STF stands for spread-tow fabric. Before ultimate failure, no transverse cracks could be observed in the 160 g/m^2 spread-tow fabrics ($\times$). However, the outer transverse yarns of the 240 g/m^2 spread-tow fabrics developed several matrix cracks that contributed to the nonlinear tensile response of this laminate ($\times$).

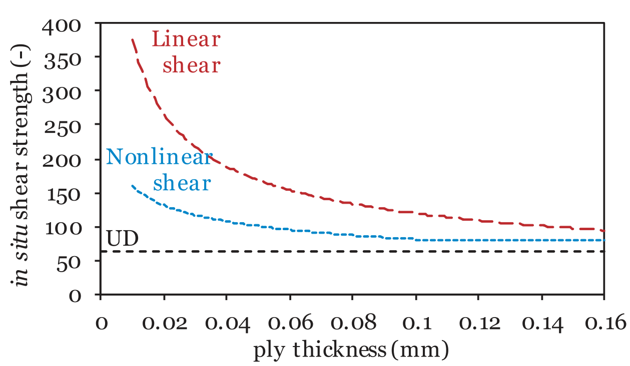


Figure 8: Linear and nonlinear predictions for the *in-situ* in-plane shear strength of a thin embedded ply (after Camanho et al. [69]).

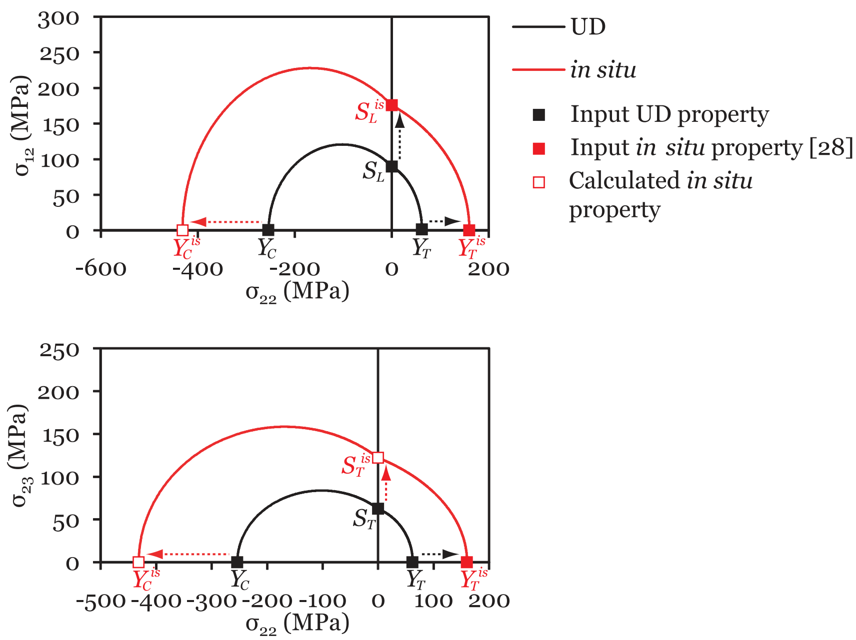


Figure 9: *In-situ* effect on the failure envelopes for matrix-dominated damage mechanisms (after Camanho et al. [78]).

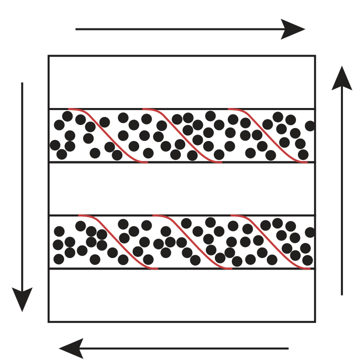


Figure 10: Transverse shear cracking in multi-directional laminates (after Olsson [79]).

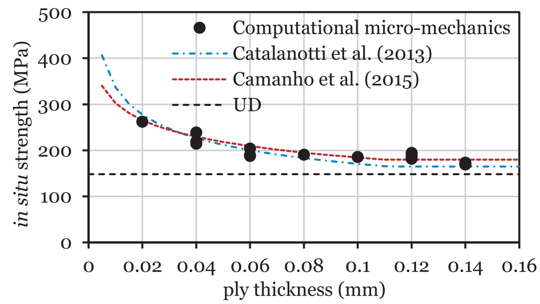


Figure 11: *In-situ* transverse compressive strength predicted by computational micro-mechanics and phenomenological failure criteria (after Arteiro et al. [81]).

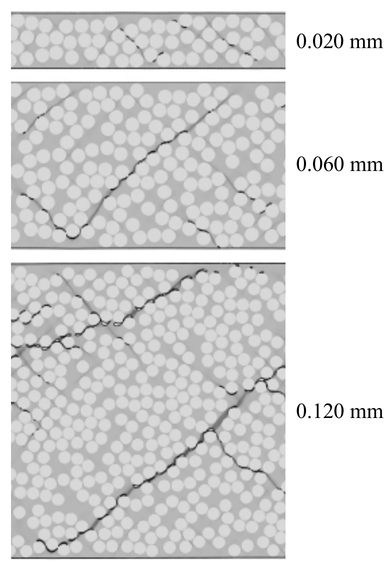


Figure 12: *In-situ* effect on wedge transverse cracking (after Arteiro et al. [81]). Representative volume elements loaded in compression by a uniaxial remote strain of 2.5% along the horizontal direction show that failure of conventional 90° plies is dominated by fibre-matrix interface cracking and 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.

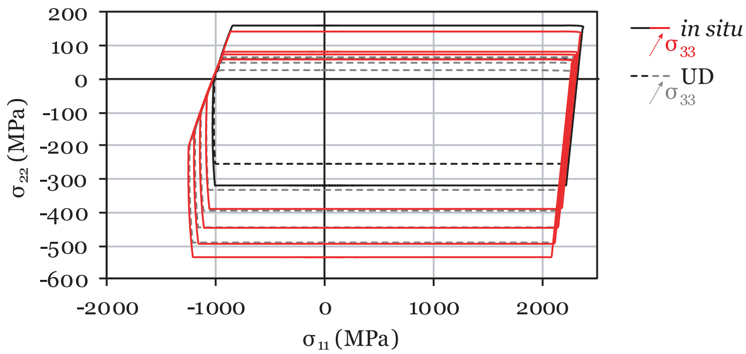


Figure 13: Longitudinal-transverse failure envelopes for different values of applied through-thickness transverse compressive stress (σ_{33}) for an embedded and a UD ply (after Arteiro et al. [99]). Black lines correspond to $\sigma_{33} = 0$, and coloured lines to non-zero σ_{33}. The same value of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

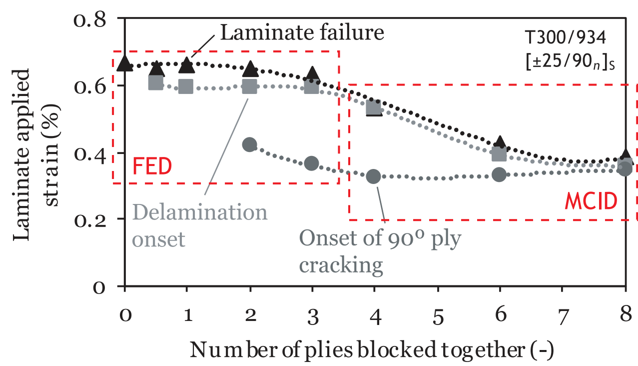
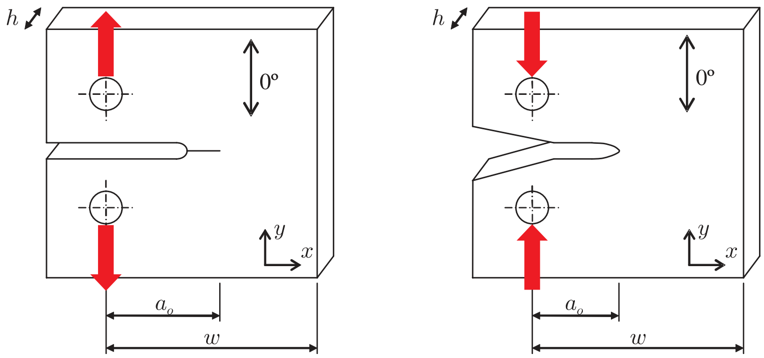


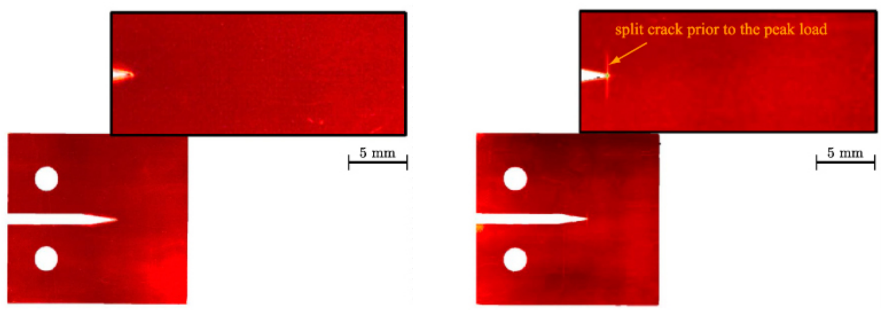
Figure 14: Laminate applied strain at the onset of 90° ply cracking, at the onset of delamination, and at laminate failure (after O'Brien [26]). In the laminates with 4 or more plies blocked together, delamination is induced by matrix cracking (MCID), whereas in the laminates with less than 3 plies blocked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



(a) CT specimen.

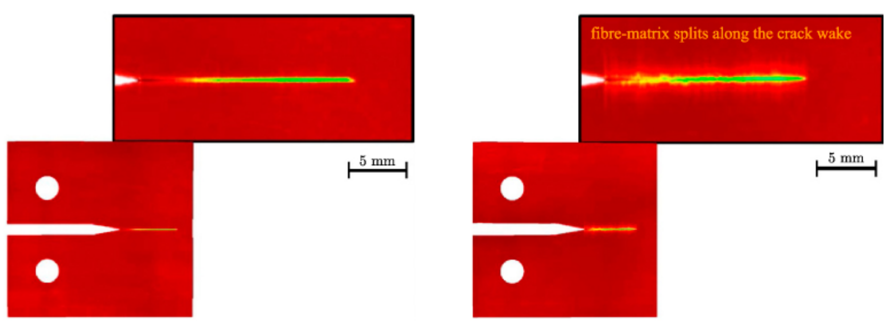
(b) CC specimen.

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



(a) 0.055 mm dispersed plies prior to the peak load.

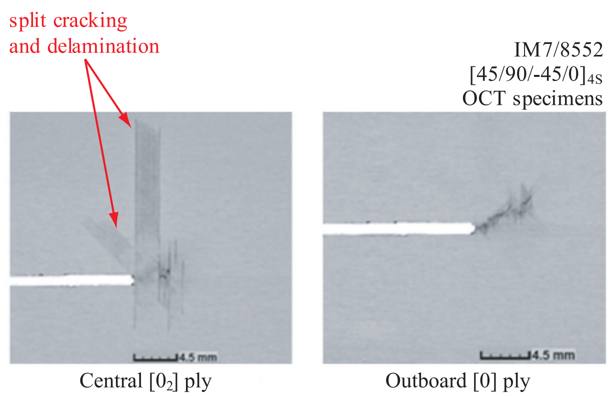
(b) 0.110 mm blocked plies prior to the peak load.



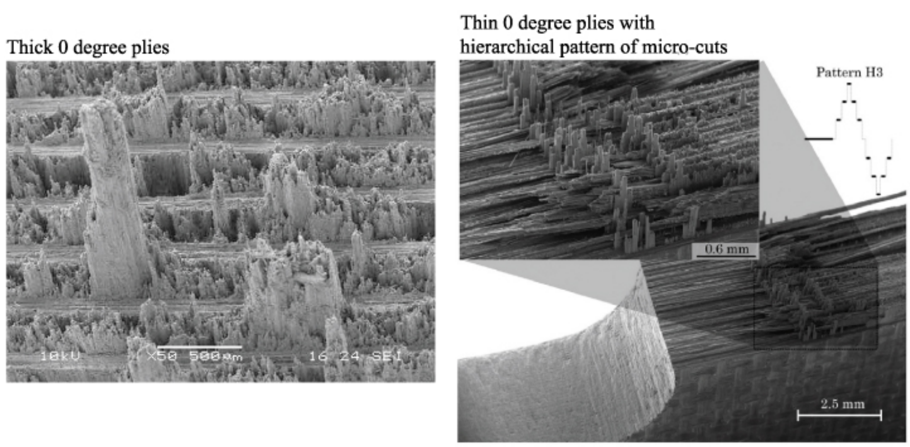
(c) 0.055 mm dispersed plies after testing.

(d) 0.110 mm blocked plies after testing.

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 hierarchical pattern of micro-cuts (after Bullegas et al. [134]).

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

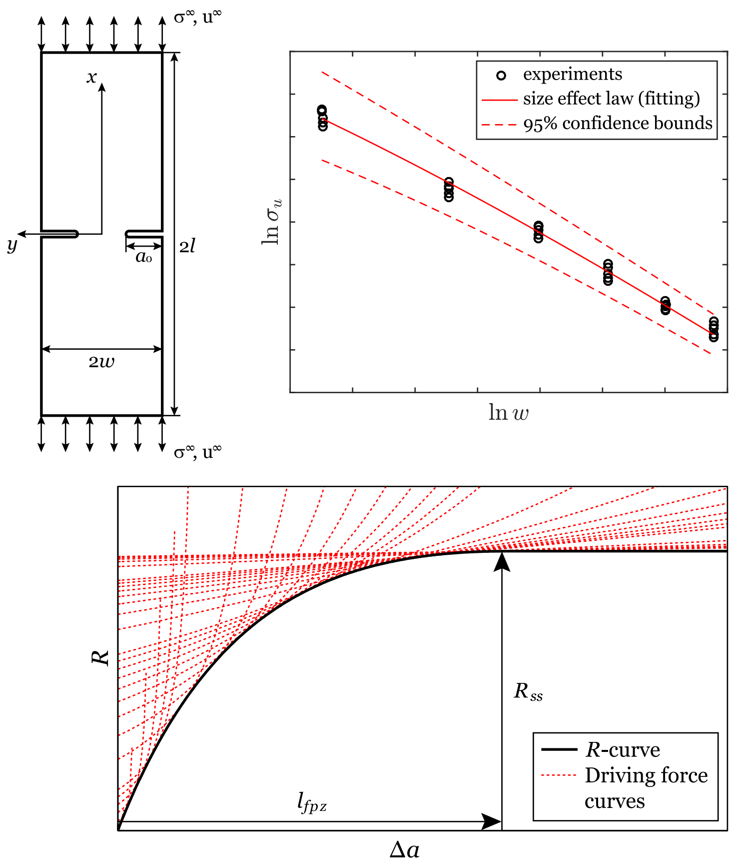


Figure 18: Size effect law fitting and $\mathcal{R}$-curve determined as the envelope of driving force curves (after Catalanotti et al. [116]).

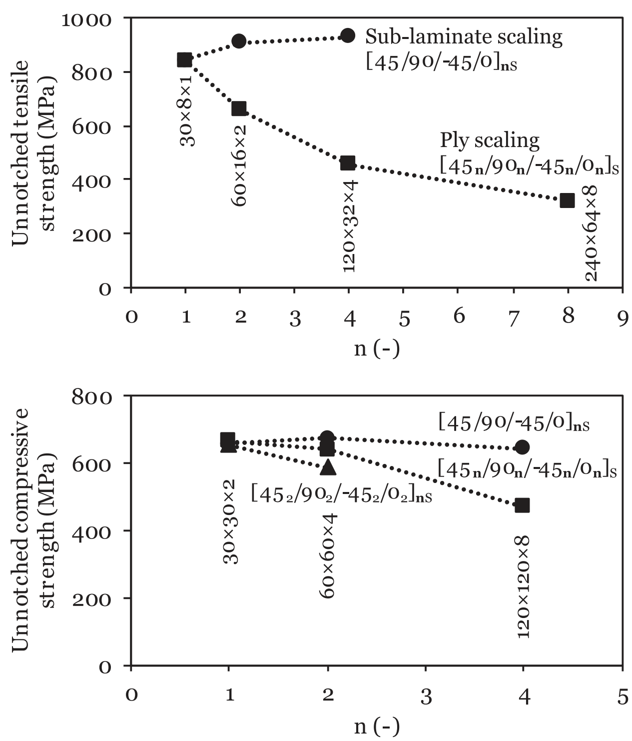


Figure 19: Size effects in un-notched IM7/8552 carbon/epoxy coupons loaded in tension (after Wisnom et al. [45, 50]) and in compression (after Lee and Soutis [46]). Specimen dimensions are given as: gauge length (mm) × width (mm) × thickness (mm).

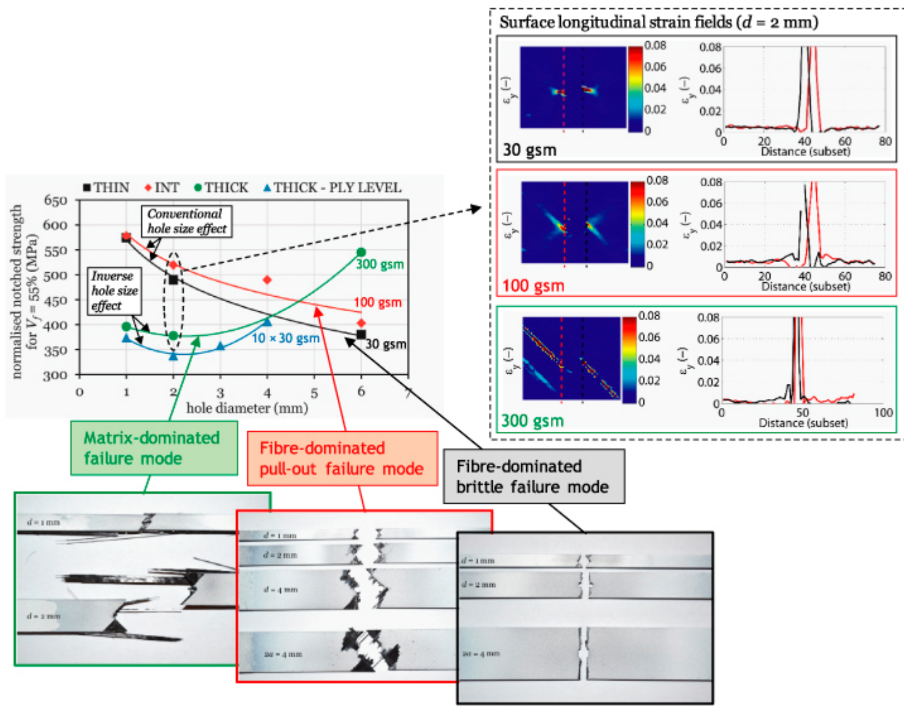


Figure 20: Size effects in notched M40JB/ThinPreg™ 80EP carbon/epoxy coupons loaded in tension, representative failed specimens, and surface longitudinal strain fields obtained with digital image correlation from specimens with a hole diameter of 2 mm (after Furtado et al. [42]).

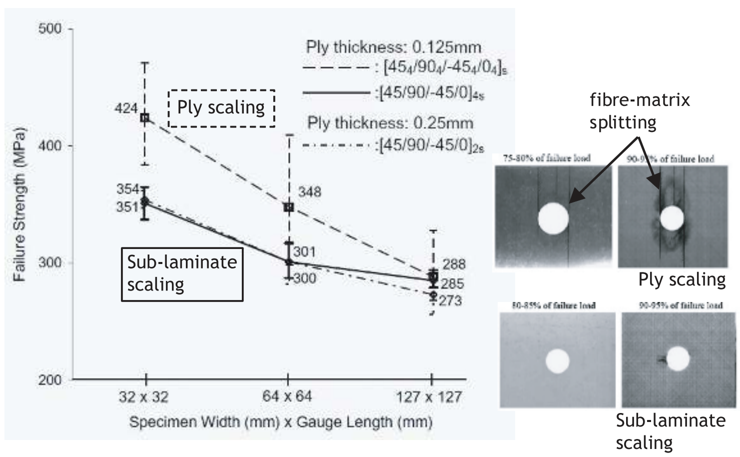


Figure 21: Size effects in notched IM7/8552 carbon/epoxy coupons loaded in compression and damage development in representative open-hole compression specimens (after Lee and Soutis [46]).

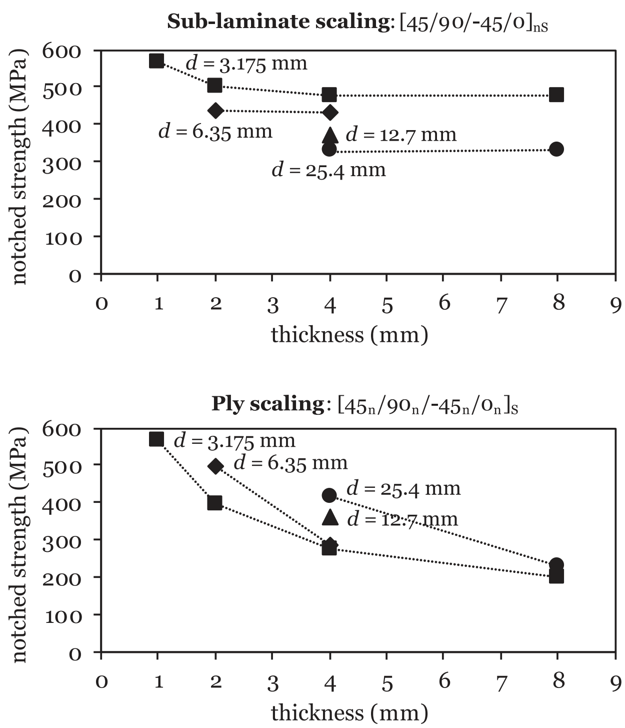


Figure 22: Effect of laminate thickness on the notched tensile strengths of IM7/8552 carbon/epoxy coupons with different hole diameters d (after Wisnom et al. [50]).

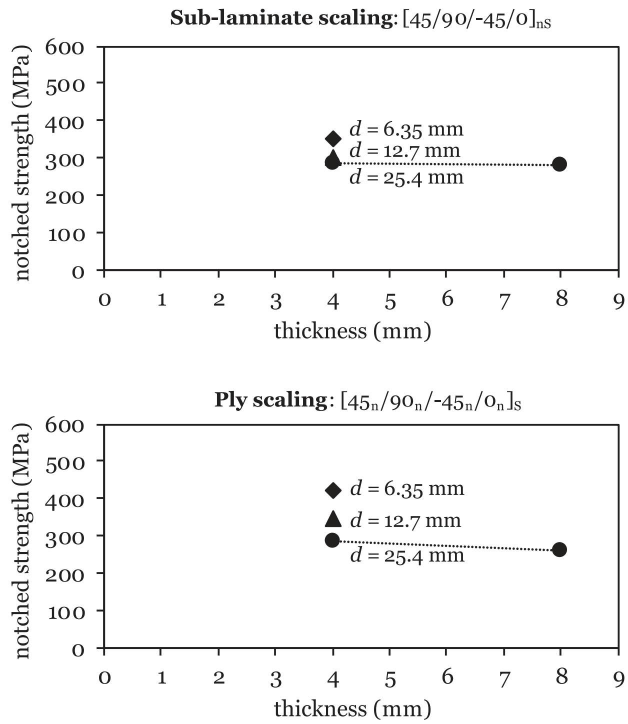


Figure 23: Effect of laminate thickness on the notched compressive strengths of IM7/8552 carbon/epoxy coupons with different hole diameters d exhibiting valid failure mode (after Wisnom et al. [50]).

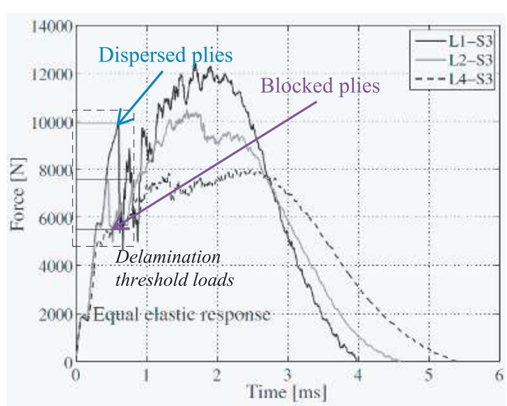


Figure 24: Impact response of laminates with different degrees of ply blocking (after González et al. [54]).

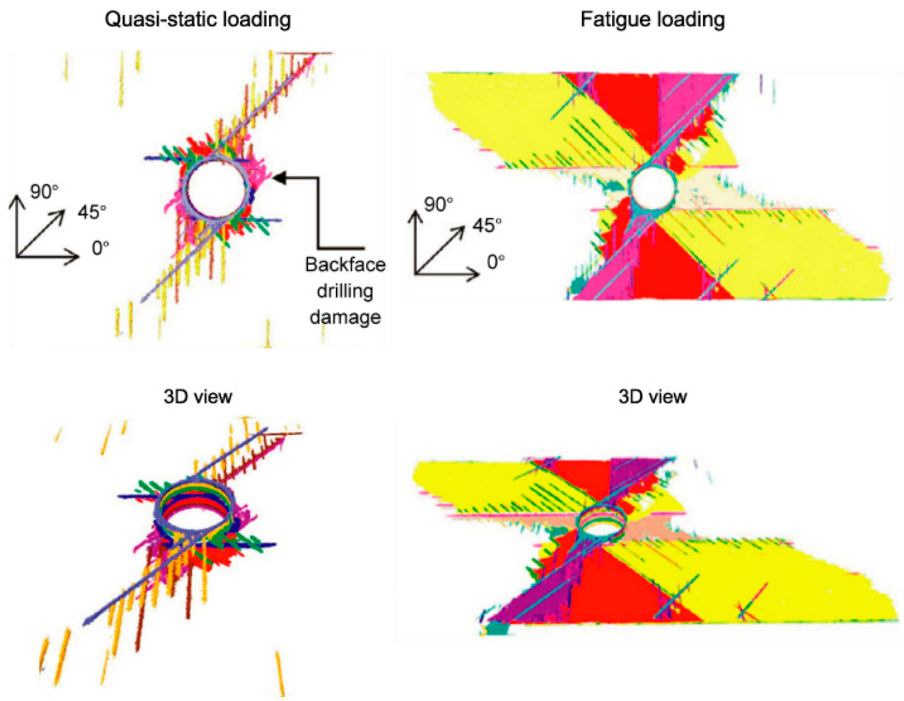


Figure 25: X-ray computed tomography images of interrupted quasi-static and fatigue open-hole tension tests (after Nixon-Pearson et al. [61]).

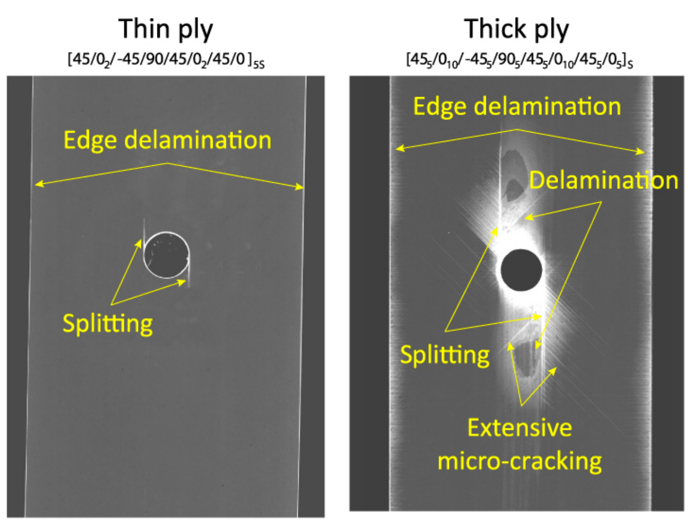


Figure 26: X-ray pictures of specimens subjected to tensile fatigue loading at room temperature, after 73,000 cycles, with a ply thickness of 0.04 mm and a total laminate thickness of 3.2 mm (after Sihn et al. [28]).

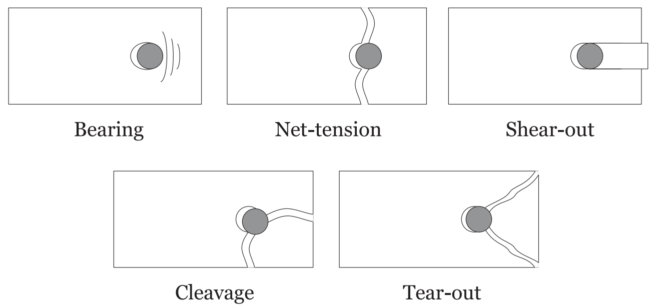


Figure 27: Joint failure modes (after ASTM D5961/D5961M – 13 [164]).

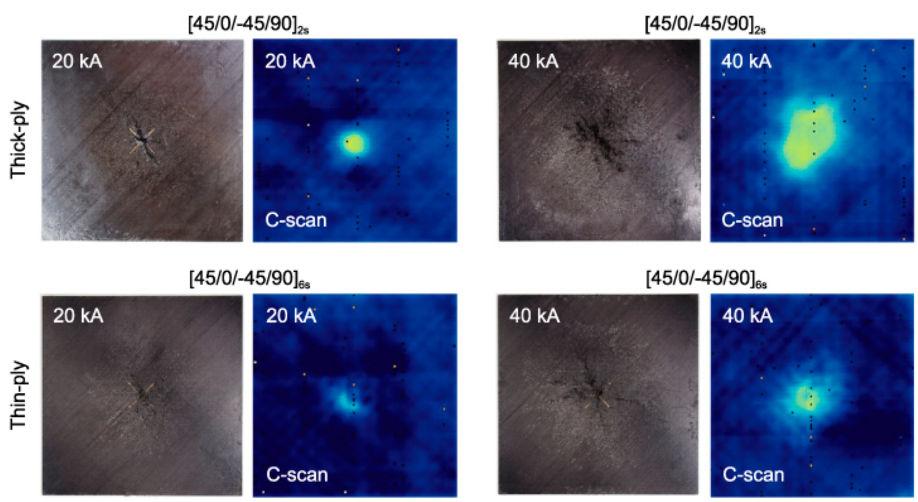


Figure 28: Visual and ultrasonic C-scan images of carbon fibre-reinforced thermoplastic laminates subjected to lightning strikes (after Yamashita et al. [64]).

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

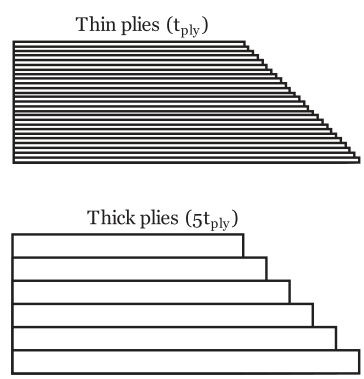


Figure 30: Illustration of ply drops with thin and thick plies.