Mechanical performance of NCF composites

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Abstract

Composite materials are today used in the aerospace and marine industries due to their excellent strength and stiffness to weight ratio. In most marine and aerospace applications traditional composites (pre-impregnated tapes) are still the main choice for design of load bearing structures. However, due to their high material costs and sensitivity to out-of-plane loads, the use of composites manufactured from pre-impregnated tapes has so far been limited. Non-crimp fabric (NCF) based composites address both these concerns, without any significant drop in in-plane performance. In the presented thesis, different aspects of the mechanical performance of NCF composites are investigated.

NCF cross-ply composites response to tensile load is investigated in both Paper A and D. In Paper A, intralaminar cracks in the 90° fibre bundle layers and their effect on laminate mechanical properties has been monitored. Occurrence of ‘novel’ type of cracks propagating in the load direction (longitudinal cracks) is explained by a thorough FE analysis using an RVE approach, revealing stress concentrations caused by 0° fibre bundle waviness. Effects of damage on mechanical properties are modelled using modified micro mechanical models developed for analysis of conventional laminated composites. The analysis reveals mechanical degradation to be ruled by the crack opening displacement (COD).

However, it is shown in Paper D that NCF cross-ply composites response to tensile loading show large effect of the fabric layer stacking sequence: much larger elastic modulus reduction was observed in [0/90/0/90]S than in [90/0/90/0]S case. Since transverse cracks in 90°-bundles may give modulus decrease about 5%, the observed 40% stiffness reduction in the [0/90/0/90]S case is attributed to failure and delamination of bundles oriented in the direction of the applied load.

Failure initiation under compressive loading in NCF composites containing bundles with out-of-plane orientation imperfections were analyzed in paper B using FEM in plane stress and linear elastic formulation. Failure initiation strain was determined comparing failure functions corresponding to two alternative failure mechanisms: a) plastic micro buckling in bundle due to mixed compressive and shear load; b) plastic matrix yielding according to von Mises criterion. It was found that for all parameter and boundary condition combinations the strain at failure initiation decreased several times when the imperfection angle in the longitudinal bundle is changed between 0 and 14°.
A power law is developed in Paper C which predicts the average strain in bundles as a function of applied strain to a RVE of the NCF composite. The introduced H-matrix which establishes the relationship between strains in meso-element and RVE strains is used to calculate the “effective stiffness” of the bundle. This “effective stiffness” is the main element in simple but exact expressions derived to calculate the stiffness matrix of NCF composites.

The most important geometrical parameters which control mechanical properties of NCF composites are identified in paper E. The identification is based on experimental observations and available theoretical findings. Characteristics of the internal structure of NCF composites are analyzed in context of their significance for in-plane elastic and failure properties. A methodology for determination of most typical geometrical parameters of composites using optical observations of cross-sections of manufactured laminates is described. The methodology is applied to characterize cross-ply and quasi-isotropic composite laminates.
Preface

The work presented in this thesis has been carried out at the Division of Polymer Engineering at Luleå University of Technology during a period from January 2002 to November 2005.

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Of course, not only the supervision was important during the completion of the thesis, therefore I also want to express my appreciation towards all my present and former colleagues at the division of Polymer engineering for the nice working environment which has made my time at the division most pleasant.

I would also especially like to thank Peter L. with whom I shared the office during the work with this thesis, his support and encouragement throughout our time at the university has been most valuable.

Finally, to my family, thanks for all your support during my years of studying.

Luleå, November 2005

David Mattsson
List of publications


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1. Introduction

By definition, a material which has two or more distinct phases or constituents can be regarded as a composite material [1]. However, it is only when the constituent phases have significantly different physical properties and thus the composite properties are noticeable different from the constituent properties that we have come to recognize these materials as composites. For example, metals have different phases and polymers have impurities and still these materials are not considered as composite materials. In the case of metals, the two phases are rather similar to each other in physical properties and therefore the material can be considered as homogenous. However, in terms of structure, the metal alloys are a good example of particulate composites. The same discussion are also valid for polymers where fillers or additives are used for cost purposes which normally not change the properties to any significant extent, hence these materials can also be considered as homogenous. Composite materials are normally classified into two main types, particle and fiber reinforced composites. This classification is based on the geometry of the reinforcement where the particle is approximately equiaxed and the fiber is characterized by its length being much larger than its cross-sectional dimensions.

1.1 Fiber reinforced polymer composites

The fiber reinforced composite materials can be divided into two main types which uses either continuous or discontinuous fibers as its reinforcement. By definition the length of short fibers affect the properties of the composite whereas in the case of long fibers it could be assumed that the load is directly applied to the fibers and hence the fibers in the direction of load are the principle load carrying constituents. To form the composite, the fibers are embedded in a matrix material which binds the fibers together, transfer load to the fibers and protect them from environmental attack and damage due to handling. Reinforcement in form of fibers can be arranged in a number of ways both in the form of bundles or homogenously dispersed fibers depending on the application of the final composite. Examples of different types of long fiber reinforcements are visible in Fig 1.
Traditionally, 2D-fiber reinforced polymer composites (Fig 1a) have been used in the industry for many years for reasons such as their excellent stiffness to weight ratio and resistance to corrosion. Typical Properties of conventional structural materials and bidirectional (cross-ply) 2D-fiber reinforced composites are presented in Table 1.

Table 1. Properties of conventional structural materials and bidirectional (cross-ply) fiber composites [1].

<table>
<thead>
<tr>
<th>Material</th>
<th>Fiber volume fraction ($V_f$) (%)</th>
<th>Tensile modulus (E) (GPa)</th>
<th>Tensile strength, ($\sigma_u$) (GPa)</th>
<th>Density ($\rho$) (g/cm³)</th>
<th>Specific modulus ($E/\rho$) ($10^6$ Nm/kg)</th>
<th>Specific strength ($\sigma_u/\rho$) ($10^6$ Nm/kg)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mild Steel</td>
<td></td>
<td>210</td>
<td>0.45-0.83</td>
<td>7.8</td>
<td>26.9</td>
<td>0.058-0.106</td>
</tr>
<tr>
<td>Aluminium</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2024-T4</td>
<td></td>
<td>73</td>
<td>0.41</td>
<td>2.7</td>
<td>27.0</td>
<td>0.152</td>
</tr>
<tr>
<td>6061-T6</td>
<td></td>
<td>69</td>
<td>0.26</td>
<td>2.7</td>
<td>25.5</td>
<td>0.096</td>
</tr>
<tr>
<td>E-glass/EP</td>
<td></td>
<td>57</td>
<td>21.5</td>
<td>1.97</td>
<td>10.9</td>
<td>0.26</td>
</tr>
<tr>
<td>Kevlar49/EP</td>
<td></td>
<td>60</td>
<td>40</td>
<td>1.40</td>
<td>29.0</td>
<td>0.46</td>
</tr>
<tr>
<td>CF/EP</td>
<td></td>
<td>58</td>
<td>83</td>
<td>1.54</td>
<td>53.5</td>
<td>0.24</td>
</tr>
<tr>
<td>Boron/EP</td>
<td></td>
<td>60</td>
<td>106</td>
<td>2.00</td>
<td>53.0</td>
<td>0.19</td>
</tr>
</tbody>
</table>

The marine industry [2] was one of the initial users of 2D composite materials with around 60 years of experience in this field whereas the aircraft industries have used polymer based composites for over 30 years [3-4]. Other examples of areas were large amounts of composite materials is used are the automobile industry [5] as well as in bridges [6] which more recently started to use these materials.

The 2D laminates give excellent in-plane mechanical properties in terms of stiffness and strength due to well aligned fibers with high volume fraction in the composite. However, the in-plane orientation of the reinforcement also results in a material which is rather sensitive to delamination cracking under an impact loading due to their poor interlaminar fracture toughness. As a consequence, the post-impact in-plane mechanical properties can be severely degraded [7]. Another drawback of 2D laminates are the high costs associated with the manufacturing technique in which the fibers usually are pre-impregnated with resin and assembled in the desired lay-up before consolidation into the final composite. The relatively high manufacturing cost originate from both high storing costs (pre-pregs require low temperature to prevent curing) and high labour cost. This type of composites will in the following sections be referred to as UDPT composites.
The drawbacks found in UDPT laminates can be avoided by using some other type of reinforcement such as woven blankets (Fig. 1b). This reinforcement type is produced by the weaving of fiber bundles into a blanket which creates reinforcement with bidirectional orientation. The main driving force for the development of woven reinforcement is the reduced manufacturing costs and its enhanced fracture toughness compared to UDPT composites. The reduction in manufacturing cost is related to a decrease in both labour and material cost.

However, since waviness in the out-of plane direction is introduced in the fiber bundles in the manufacturing stage, the in-plane material properties of woven composites become inferior to the properties found in UDPT composites. The drawbacks associated with both the UDPT’s and the woven composites calls for new types of reinforcements which combine the excellent in-plane properties of UDPT composites with the relatively good fracture toughness properties and lower manufacturing costs found in woven composites. One relatively new type of reinforcement, which has the possibility to combine the benefits found in both types of reinforcements, is the Non-Crimp Fabric (NCF).

As can be seen in Fig. 1, the internal structure of these three reinforcement types differs quite significantly. Fig. 1a, shows an example of the internal structure of a cross-ply UDPT laminate which has an internal structure consisting of homogenously dispersed fibers in layers with a certain orientation. Fig. 1b, shows an example of the internal structure of a woven reinforcement. As can be seen in the figure, the bundles out-of plane orientation introduced in the interlaced regions between two bundles with different orientation are dependent on the weaving pattern. A number of
blankets with various orientations are normally stacked together in a mould to meet the design criteria in terms of mechanical properties before consolidation into the final composite by a resin infusion technique.

A schematic figure over the internal structure of a NCF reinforcement is displayed in Fig. 1c and it can be seen that this material represents a combination of the two reinforcement types previously described. In this case, the fibers in each layer are gathered in bundles with a well defined geometry in a similar fashion as for the woven composites. However, in this case the bundles are kept together by a knitting yarn instead of weaving the bundles together as described for the woven composites which theoretically should produce a fabric without out-of-plane waviness in the fiber bundles. This manufacturing technique creates a dry fabric which is easily formed into complicated shapes before it is consolidated into the final composite.

1.1.1 Modelling of elastic properties in composite materials

Due to its complicated internal structure, predictions of the elastic properties of composite materials are normally performed on different length scales such as the micro- meso and macro scale. In this case the microscale concerns the properties of a UD composite on fiber and matrix level. Mesoscale concerns in the case of woven and NCF composites the bundles and the resin rich regions in a layer whereas the macroscale considers the properties of the final composite. As mentioned above, these three different types of reinforcement show large differences in the internal structure on the mesoscale. However, the fiber bundles in woven and NCF composites consist of a large amount of fibers which are homogenously dispersed in a similar way as in the layers of UDPT composites. Therefore the bundle can be considered as a UD composite and the elastic properties of the bundles can be determined using the same models as used for predicting the properties of a layer in UDPT composites. An example of a simple and common model used for this purpose is the rule of mixtures which determines the in-plane mechanical properties of the bundle. This model predicts the longitudinal modulus with rather high accuracy since the model in this case is based on physically reasonable assumptions. However, in the case of transverse and shear modulus the model is based on the assumption of constant stress in the fibers and the matrix region (constant stress model). Since this assumption is incorrect, the predicted transverse and shear modulus are lower than the real values. Another and more accurate model used for the determination of transverse and shear properties in a UD composite are the Halpin-Tsai
equations [9]. These are simple and empirical expressions based on numerical results from a finite difference analysis. The model uses a fitting parameter which needs to be determined from a comparison of the model with an experiment or numerical calculation such as FEM. Another example of a more sophisticated model which can be used to predict the properties of a fiber bundle are the Composite Cylinder Assemblage model (CCA) developed by Z. Hashin [10, 11]. This model determines the elastic constants of the UD composite using the elastic properties of the fiber and the matrix together with the volume fraction of fibers in the considered UD composite. In this model the energy theorem of classical elasticity is used to obtain bounds on the elastic properties of the UD composite. There are also several other models available in the literature which predict the elastic constants of a UD composite and the above discussion is meant as a brief overview over some of the most common models.

The differences between these reinforcement types in internal structure on the mesoscale imply that different models have to be used to predict the elastic properties of the composite on the macroscale. Several models have been suggested in the literature for the UDPT and woven composites whereas only a few models so far have been employed for NCF composites. The classical laminate theory (CLT) is an example of a model used for UDPT laminates, which predicts the in-plane elastic properties with rather high accuracy. The elastic properties of woven composites are due to their complicated internal heterogeneous structure more difficult to predict. However, it is shown in [12] that the KKM model developed by Karayaka and Kurath gives predictions of the 3D elastic properties accurate enough for practical applications. This model is based on iso-strain assumption in the in-plane direction and iso-stress assumption in the out-of-plane direction. The internal structure of NCF composites implies that models similar to the ones used for UDPT’s and woven composites also could be used in this case. Paper C of this thesis also shows that in fact a number of models can be used to predict the elastic properties of NCF composites with high accuracy assuming that the fiber bundles are well aligned with no waviness on the mesoscale. In addition to the models mentioned above, numerical methods such as the finite element method can of course also be used to determine both the elastic properties of UD composites as well as the elastic properties of the composites on the macroscale independent on reinforcement type.
2. Non-Crimp Fabric composites

Non-Crimp Fabric composites can trace their early steps to the marine industry and it was first manufactured in 1983 in terms of a +45° ply knitted together with a -45° ply to form a double bias fabric [13]. As mentioned earlier, one of the main reasons for the industry to use NCF composites instead of UDPT laminates are the economical benefits. Bibo et al [14] studied a rectangular plate (300*600 mm) with the objective of determining the cost savings when switching from UDPT to NCF composites. For UDPT constructions, labour costs can account for as much as 50% of the final cost for the part in which 25% are material cost. The major part of the cost for labour consists in progressively debulking and laying down the material, which can be reduced to large extent by using NCF and a resin infusion technique. Since one layer in the NCF composite corresponds to approximately 7 pre-preg layers, deposition rates are much higher in the NCF case. In the studied case, 35% cost reduction was achieved by switching from UDPT/autoclave to NCF/resin infusion.

2.1 Manufacturing technique

The manufacturing technique used for producing NCF composites originate from the textile industry, were highly automated techniques such as stitching, weaving, braiding and knitting has been used for some time [15]. Especially knitting is very well suited for rapid and cost effective manufacturing and the existing machines used in the textile industry have been successfully adapted to handle high performance fibers such as glass, carbon, aramid and even ceramic fibers.

NCF composites are manufactured in two main steps; manufacturing of the preforms and moulding of the preform into the final composite. The preforms are made from layers of non-crimped fiber tows stacked in the desired orientation and bounded together by a warp-knitting procedure in which the binder yarns are inserted through the thickness of the perform. A schematic figure of the production of NCF preforms is presented in Fig 2. Typically this production technique is used to produce bi- tri and quadaxial fabrics of carbon or glass fiber using polyester or aramid warp knitting yarns [14, 16-18]. In order to minimize damage and fibre tow crimp, the amount of binder yarn is usually kept small but sufficient to hold the bundles in the layers during handling of the fabric. Several preforms are normally stacked together creating a dry fabric thick enough for structural applications.
This manufacturing technique creates dry preforms which easily can be shaped into a desired structure before consolidation into a composite by a suitable moulding technique such as resin transfer moulding (RTM) or resin film infusion (RFI). The manufacturing technique is attractive due to small amounts of material wastage, reduced production time and virtually unlimited shelf life which results in lower production costs [15]. Manufacturing parameters associated to both the preform and composite manufacturing are important for the final performance of the composite. Important parameters for the preform manufacturing are, besides fiber used in the tows, the inter-tow gap, stitch tension, stitch type, stitch density and stitch material. Important parameters, besides the matrix properties, associated with the composite manufacturing process are different for each manufacturing technique. For example, during compression inside the mould in the RTM process, the compaction forces which compresses the fibre stacks might create nesting and crimp as well as distortion of the cross-section shape of the fiber tows larger than in the case of the RFI or RIFT process which uses a vacuum bag. Similar to the UDPT laminates, the fibers in NCF composites should theoretically be straight with only small scale waviness on the micro-scale. However, this is not always the case and certain waviness is normally introduced by the manufacturing procedure at the mesoscale.

2.2 NCF composites and industrial use

NCF composites have due to their excellent performance and relatively low cost become an attractive alternative for aerospace, marine and even
automotive applications. For example, glass fiber NCF composites are widely used by the marine industry in USA to produce recreational ships (motor boats and yachts). Another example from the marine industry is the large, 73 m long, stealth ships (Visby corvette) used by the Swedish navy visible in Fig. 3. In this case the hull is produced using a sandwich structure in which the outer layers are manufactured from carbon fiber NCF composites.

Figure 3. The Visby corvette, (Picture from Kockum AB).

NCF composites have also been used to enhance the reinforcement of steel and concrete structures. In 1998, stitch bonded fabrics were for example used to fabricate composite jackets for the wrapping of more than 3800 columns of the yolo Causeway in Sacramento, USA [13].

2.3 Micro- and meso-structure in NCF composites

As described in previous sections, NCF composites are an inherently multi-scale material since the layers due to the knitting procedure are divided into fiber bundles. On the microscale each bundle can be described as a UD composite with a certain fiber content and the homogenized bundle properties may be calculated using micromechanics expressions for long fiber composites. On the mesoscale the bundle can be considered as
homogeneous transversely isotropic material surrounded by matrix and other bundles of the same or different orientation. An important mesoscale characteristic of the NCF composite is the bundle content in the composite. Micrographs presented in Fig. 4 demonstrate the hierarchical structure of NCF composites.

Figure 4. Hierarchical structure of the NCF composites.

The geometrical shape of the bundles (cross-section and axial alignment) is complex and depends on bundle orientation in the blanket, surface compression during production, resin pockets etc. The described mesoscale configuration determines the NCF composite properties on macroscale and is typically used in simulations of macro behaviour [19-21]. These studies demonstrated the importance of geometrical parameters of the micro/meso structure of the material on different mechanical properties of NCF composites.

Also waviness is dependent on the manufacturing parameters and a characterization of the waviness is essential in order to predict the mechanical properties of NCF composites with sufficient accuracy. Miller [22] showed that the waviness of the 0°-bundles on the mesoscale was mainly determined by the adjacent layers of different orientation. For biaxial laminates this means that during manufacturing the 0°-bundles are nesting into the matrix channels between each two adjacent 90°-bundles as
schematically shown in Fig. 5. Since the matrix regions between the 90°-bundles are periodically distributed, the waviness of the 0°-bundles was approximated by a periodic wave.

![Schematic picture of the out-of-plane misalignment of the 0°-bundle, caused by the periodicity in the 90°-bundle structure.](image)

Figure 5. Schematic picture of the out-of-plane misalignment of the 0°-bundle, caused by the periodicity in the 90°-bundle structure.

However, the wavelength is a good characteristic of the misalignment of the 0°-bundles only if some periodicity can be observed in the laminates. If the waviness is not periodic, another parameter of imperfection should be used instead of wavelength and amplitude, for example, the maximum inclination angle.

As can be seen in Fig. 6, the mesoscale waviness in NCF composites are in some cases induced by the stitching tread in the sense that the tread separates the 0°-bundle from the adjacent bundles in the area closest to the tread (at a distance of approximately 2 bundle thicknesses). This phenomenon can introduce waviness both in the in-plane and out-of-plane direction. The case with the waviness introduced in the in-plane direction can be seen in detail in Fig. 6 b) as varying orientation of the fibers in the 0°-bundle.

However, in many cases there is no influence of stitching tread on the waviness of the bundles. Therefore, at present stage of understanding the role of stitches on the waviness is unclear and therefore it is difficult to establish a measure of the waviness induced by the stitching thread.
2.4 Mechanical properties of NCF composites

The mechanical behaviour of NCF composites differs from the behaviour of UDPT’s and woven composites. The difference between the materials can be addressed to its differences in the mesoscale structure in terms of bundle arrangement of the fibers as well as the knitting of the bundles in the out-of-plane direction. As a result of these differences in internal structure, NCF composites have been reported to show higher out-of-plane fracture toughness and damage tolerance compared with UDPT laminates [23-26]. A number of studies have also been performed exclusively on the effect of stitching or knitting of composites in the out-of-plane direction. As an
example, a FE study on the effect of translaminar reinforcement (TLR) of composites was performed by Dickinson et al [27]. In this study, the authors found that adding a few percent TLR, had a small effect on the in-plane stiffness and a large effect on the out-of-plane stiffness, where the stitched laminate had up to 60% higher stiffness. This result suggests that TLR is improving the composites mechanical properties without any apparent drawbacks. However, Mouritz et al [7] concluded by reviewing the available literature that contradictory results were obtained by different authors concerning the effect of stitching on mechanical properties of composites. Some studies reveal that the in-plane properties are unaffected or improved by the stitching whereas some studies revealed the opposite. It was also found that predicting the effect on in-plane properties from stitches are complicated and that the properties depend on the type of fibre, resin, lay-up etc. used in the composite as well as the stitching conditions (type of thread, stitch pattern, stitch density, stitch tension etc.).

2.4.1 Tensile strength

Generally NCF composites are considered to have lower tensile strength compared to equivalent UDPT composites. However, the strength degradation varies substantially in different studies. For example the reduction in tensile strength reached only 1% in the study performed by Godbehere et al [28] whereas Bibo et al [14] performed a study were around 35% decrease in tensile strength was found for carbon fiber NCF composites. In the study performed by Bibo et al [14] the differences in volume fraction between the UDPT and NCF composites were only 6%. Hence, in this case the drop in tensile strength can only partially be attributed to the difference in volume fraction of fibers. The degradation in tensile strength of NCF composites compared to UDPT composites can be attributed to the damage and misalignment induced in the fibers by the knitting process. By reviewing the literature it also seems that the failure mechanisms due to tensile loading differ between NCF’s and UDPT’s. For example, Bibo et al [14] reported a very well defined fracture surface in a triaxial NCF carbon composite whereas the fracture of UDPT laminates is associated to multiple-ply-delamination damage. Another study of the failure mechanisms in NCF composites were performed by Wang [29]. In this study the author concluded that failure due to tensile loading occurred in glass fiber NCF’s due to delaminations which were concentrated within a short section of the specimen.
2.4.2 Compressive strength

Compression strength of NCF composites is also generally considered lower than for UDPT composites [14, 29, 30]. However, it is also obvious from the literature that the compression strength compared to UDPT’s vary within a broad range. In the study performed by Godbehere et al [28], the authors found that the compressive strength even increased with 2 and 10% for triaxial carbon fiber NCF’s with the 0°-bundles in spread and stacked styles respectively. Failure mode due to compressive loads is in the literature reported as similar for NCF’s and UDPT’s, where failure is initiated by delamination between plies with different orientation [15, 29, 32]. However, FE analysis performed in Paper B of this thesis demonstrated that plastic microbuckling in the bundle due to mixed compressive and shear loads was the initial failure mode in NCF’s with misaligned fiber bundles (out-of plane waviness). For instance, a bundle misalignment angle of 14 degrees, depending on the extent of support from adjacent bundles, may reduce the failure initiation strain by 50%. It was also shown in Paper B that the average fiber content alone does not govern the kink-band formation strain. At fixed average fiber content and mesoscale waviness, the composite with more evenly distributed fibers (lower fiber content in the bundle and larger bundle content in the composite) showed larger resistance to compressive failure.

The effect of longitudinal fiber bundle waviness on the compressive strength of NCF composites was also theoretically analyzed in [19] using FE in generalized plane strain formulation. It means that the heterogeneous bundle structure was kept only in the plane containing the direction of loading and the thickness (out-of-plane) direction, whereas in the width direction the composite is homogeneous. Starting with the experimental nonlinear shear response of the resin, the nonlinear bundle properties were calculated using micromechanics relationships. Thus, the specimen collapse due to decreasing shear resistance was the only compressive failure mode in this study.

The geometry and boundary conditions used in the FE simulations indicate that the model is a repeating element in both longitudinal and thickness direction which implies that the failure initiation stress in the model is equal to the stress to failure of the NCF plate. This study was performed with the objective to obtain insight into the mechanisms leading to failure and to provide guidelines for the fabric structure to optimize the compressive strength of the composite. It was found that the stitch tension should be kept low so that the bundles may spread and thereby delay the mesobuckling of the tows which are leading to failure.
2.4.3 Interlaminar shear strength

Reviewing the literature, it is clear that the available data on interlaminar shear strength (ILSS) of NCF composites compared to UDPT composites is inconsistent. The most widely used test to establish the ILSS of composites are the short-beam-test performed as a flexural test with a short span. Godbehere et al [28] reported a degradation of ILSS performance compared to UDPT composites of 22 and 17% for carbon/epoxy NCF composites for spread and stacked 0° tows respectively. Even stronger degradation of ILSS was found by Backhouse [31] where an average of 30% reduction was measured for different carbon NCF styles.

As a good example of the inconsistence of results in these properties, Bibo et al [14] reported an increase in ILSS of 22% for triaxial carbon fiber NCF’s compared to UDPT’s. The authors found the results surprising and they speculate that the resin have degraded in the UDPT laminates. However, the properties of the resin coincide with the resin properties in the study performed by Godbehere et al [28] and Backhouse [31].

Wang et al [32] also performed investigations of the ILSS of NCF composites and examination of the failed specimens generally showed a mix-mode type of failure. The failure mode was compressive failure on the top of the specimen below the loading point, tensile failure at the tensile side, in-plane shear failure and delaminations. Recently, the ILSS of quadriaxial, biaxial, bidirectional and unidirectional carbon fiber/epoxy NCF composites was investigated by Truong et al [33]. In this study it was shown that the ILSS of the unidirectional laminates is up to twice as high as for the other laminates with varying orientation of the plies. In the quadriaxial and biaxial laminates, the fibers with off-axis orientation may easily initiate cracks in the out-of plane direction (through the ply thickness) which resulted in a lower ultimate stress compared to the UD laminates.

Drapier et al [20] used a representative volume element (RVE) and FE calculations for the study of the interlaminar shear behavior of NCF composites as a function of the geometrical features of a RVE. The authors found that the ILS behavior improved for increasing resin modulus and yield stress and that the ILSS decreased with increasing thickness of the resin layer in between the 0 and 90°-layer. As a manufacturing guideline for improved ILSS in NCF composites, the authors recommended a lower stitch tension which would give a more homogenous meso-structure and smaller resin rich areas in the layers.
2.4.4 Flexural strength

The flexural strength of NCF composites is not well documented in the literature compared to some of the other mechanical properties which makes it difficult to draw any valid conclusions concerning this property. However, Bibo et al [14] found that the flexural strength of NCF composites is 15% lower compared to corresponding UDPT composites which might indicate that NCF composites have inferior flexural properties compared to UDPT composites.

Failure modes due to flexure loading was found by Wang et al [32] to consist of brittle failure on the tensile side of the specimen were outer ply delamination preceded fiber failure. In this case, the authors tested both bi-, tri-, and quadriaxial glass fiber NCF composites. A more complex failure sequence was found by Kang et al [34] when testing Kevlar NCF composites. In this study, multiple cracking appeared due to a shear dominated and mixed shear/compression failure mode.

2.4.5 Elastic modulus

The stiffness of NCF composites is strongly dependent on the contribution of fibers. The main parameter characterizing fiber contribution is the average fiber content of a certain orientation in the composite. Knowing the constituent density, the fiber content may be obtained from the area weight data for the fabric and the measured weight of the composite. More detailed mesoscale information would include the volume content of bundles with a certain orientation, the fiber volume fraction inside a bundle of each orientation, the shape of the cross-section of bundles etc.

The stiffness of NCF composites measured for compressive loads are in most cases slightly lower compared to UDPT composites [14, 30, 31]. However, these results are usually explained by differences in the volume fraction of fibers in the composites and therefore it is difficult to make any conclusions from these particular measurements. However, in the study performed by Dexter et al [30], the authors reported a 15% decrease in modulus for a certain type of epoxy system and fiber blanket which could not be explained by differences in volumes fraction of fibers alone.

Godbehere et al [28] on the other hand, reported an increase in modulus of 6% for two styles of carbon triaxial NCF compared to UDPT composites. However, this result is not likely to be representative for the modulus of NCF composites since the two different fabrics had large differences in waviness of the 0°-tows. In this study, the authors also compared the tensile stiffness of two NCF configurations, stacked and spread 0° tows with
equivalent UDPT laminates. The authors found that the stiffness was 3% lower for the spread 0° tows whereas the stiffness was 10% lower for the case of stacked 0° tows compared to UDPT’s. This result indicates that the stiffness degradation is governed by the waviness of the bundles since the 0° tows were considerably wavier in the stacked case. Bibo et al [14] compared CLT predictions of the stiffness for both UDPT and NCF composites with experimentally measured values. The authors found that CLT predicted the stiffness of UDPT laminates with very good correlation whereas the NCF composites, experimentally, exhibits a 10% reduction compared to the modulus predicted by CLT.

The effect of local waviness of bundles on the NCF composites longitudinal modulus has also been described by an analytical model by Edgren et al [35]. In this study a knock-down factor was introduced to calculate the effective longitudinal stiffness of an imperfect ply with bundle waviness which is introduced in laminate theory. This effect explains the significantly lower measured elastic modulus of the NCF composite comparing with CLT or FEM predictions (with straight 0°-bundles). The main mesoscale parameters of the local bundle waviness used in the stiffness model are the wavelength of the imperfection and its amplitude suggested by Miller [22] which implies a simplified representation of the imperfection as a periodic phenomenon.

It may be concluded that if the imperfect bundle alignment, which is described as waviness, and its effect on NCF composite stiffness is analyzed, a rather detailed mesoscale information is required: fiber content in the layer and in the bundle, the relative content of bundles and resin in the layer, the shape of the bundle and the resin region between bundles etc. Flexural stiffness is not so well studied in the literature and therefore it is difficult derive general trends for this property. Nevertheless, flexural stiffness is usually predicted with rather high accuracy using CLT from constituent properties and lay-up.

2.4.6 Damage accumulation in tensile loading

Since NCF composites have an inherently heterogeneous structure of fiber bundles and matrix areas, the mechanisms controlling damage evolution and failure are in most cases more complex than for a traditional UDPT composite which complicates the analysis of these materials. The damage phenomena in NCF composites should be analyzed in comparison with the very well characterized damage mechanisms in UDPT composites revealing similarities and significant differences. Whereas similar damage behavior can be analyzed by improved “old” approaches, new phenomena such as
failure of bundles in the loading direction require development of novel models. Due to the geometrical complexity of NCF composites on the mesoscale, the potential to develop analytical models with sufficient accuracy is limited and combined numerical-analytical approaches have to be preferred.

The first mode of damage observed during tensile loading of UDPT cross-ply composites is formation of multiple intrabundle cracks in layers with off-axis orientation with respect to the load. Truong et al. [33] showed that intra-laminar cracks also develop in NCF cross-ply composites in fiber bundles oriented transverse to the main loading direction. This study also showed by x-ray and c-scan analysis that the damage occurs periodically. Hence there is a relation between the stitching and damage pattern. In Paper A of this thesis it is shown that the effect on mechanical properties from damage in bundles transverse to the loading direction could be modelled using modified micro mechanical models developed for analysis of conventional UDPT laminates.

Another example where stiffness reduction due to tensile loading are modelled were presented by Varna et al [36]. In this study the authors developed a unified approach to analyze the reduction in elastic properties of the composite in result of damage development for cases when the constituent with the distributed damage is continuous at least in one direction. As a case study, a NCF composite with broken bundles in the loading direction is considered. This study resulted in 10% stiffness reduction when 20 of a total of 56 0°-bundles had failed without following delamination from the surrounding material.

However, the inelastic behaviour of NCF composites due to damage evolution is still not thoroughly investigated in the literature [15] and requires additional work which would lead to deeper understanding of the failure mechanisms present in these materials.

2.4.7 Impact resistance

The impact resistance and damage tolerance has been studied to rather great extent in the literature since improvements is expected compared to UDPT’s due to the through-thickness reinforcement in NCF’s provided by the knitting tread. One example is Bibo et al [37] which have studied the impact resistance in NCF composites with different processing routes and compared these results to UDPT composites. The NCF laminates were produced by three different manufacturing techniques; pre-pregging the fabrics followed by auto-clave curing, by RFI and RTM. Their result showed that pre-pregged NCF composites and pre-preg tape based
composites with the same lay-up had similar behavior under impact. It was also found that NCF composites manufactured by resin infusion techniques (RFI and RTM) showed improvements in impact resistance with lower level of damage for a given impact energy.

The effect of stitch density on impact performance has been studied by Kang et al [34]. In this study the authors used a multiaxial Kevlar/vinylester NCF fabric held together by Kevlar and nylon knitting yarns and compared the results to a woven Kevlar composite. The authors observed that the Kevlar-knitted NCF composites absorbed less energy and less out-of-plane deformation after impact compared to the woven composites. Unfortunately, the lower amount of absorbed energy could not be translated into reduced damage extent.

2.4.8 Compression after impact

A number of investigations are performed were the authors have studied the behavior of NCF composites in compression after impact (CAI) and the general opinion is that NCF composites offer definitive advantageous compared to equivalent UDPT composites. For example, Dexter et al [30] analyzed the CAI properties of quadriaxial carbon NCF composites manufactured from different resin infusion techniques (RFI and RTM). In this study the obtained CAI strength for NCF composites was between 45 and 80% higher than for the UDPT composites. Another study was performed by Bibo et al [37] were the results were less promising. The authors showed that the CAI strength of NCF composites is similar as the CAI strength found in UDPT composites. In this study NCF composites with different production techniques were used and the highest CAI strength was obtained for a NCF composite manufactured by RFI.

3. Objective

Research concerning NCF composites has mostly been conducted in the field of experimental characterization of mechanical properties with or without damage. However, the mechanisms responsible for the mechanical performance as well as the degradation of mechanical properties in NCF composites have so far received little attention in the literature.

The objective with the presented thesis is to investigate some of the mechanisms controlling the mechanical performance and degradation of properties in NCF composites for load cases relative to the marine and the aerospace industry both by analytical models and experimental techniques.
4. Summary of papers

**In paper A**, NCF cross-ply laminates have been tested in tension and intralaminar cracks caused in the 90° fibre bundle layers and their effect on laminate mechanical properties has been monitored. The occurrence of ‘novel’ type of cracks propagating in the load direction (longitudinal cracks) is explained by a thorough FE analysis using an RVE approach, revealing stress concentrations caused by 0° fibre bundle waviness. Effects of damage on mechanical properties are modelled using modified micro mechanical models developed for analysis of conventional laminated composites. The analysis reveals mechanical degradation to be ruled by the crack opening displacement (COD). However, unlike traditional composites, transverse cracks do not generally extend through the entire thickness of the 90° layer, but are rather contained in single fibre bundles, limiting the COD.

**In paper B**, two failure mechanisms have been recognized as possible reasons of compressive failure in NCF composites. One of them is related to large out-of-plane stresses (normal and shear stresses) in the matrix region between imperfect bundles. They could lead to interbundle splitting which may be analyzed using a Von Mises type of criteria for matrix failure. The second mechanism is related to kink-band formation in the longitudinal bundle. To understand the kink-band formation specifics in NCF composites it has to be realized that the NCF composite has imperfections in form of waviness on two scales. On the microscale the bundle has an inherent waviness, \( \overline{\Phi} \) exactly of the same nature as UDPT composites. Hence, the same compressive failure criterion can be applied for NCF composites as for kink band formation in UDPT composites. Due to the in-plane and out-of-plane waviness of the bundle on mesoscale the bundle is in its local axes subjected to compressive as well as shear stress. The combined action of these two stress components \( (\sigma_L \text{ and } \tau_{LT}) \) facilitates the kink-band formation in the bundle (UD composite) according to [38]

\[
\sigma_L = \frac{k^* - \tau_{LT}}{\Phi + \gamma}
\]

In Eq. (1) \( k^* \) is the shear strength of the bundle material and \( \gamma \) is the yield strain in shear. On the mesoscale the combined stress failure criterion (1) has the form
Here $\sigma_{cu}$ is the compressive strength in L-direction of the bundle material. The shear stress component in Eq. (2) is caused by the bundle mesoscale misalignment only. FE analysis performed in this study demonstrated that the compressive failure strain due to kink band formation in the imperfect bundle depends on many mesoscale parameters, without a doubt the misalignment angle being the most crucial one. For instance, a bundle misalignment angle of 14 degrees, depending on the extent of support from adjacent bundles, may reduce the failure strain by 50%. It was also shown that the average fiber content alone does not govern the kink-band formation strain. At fixed average fiber content and mesoscale waviness, the composite with more evenly distributed fibers (lower fiber content in the bundle and larger bundle content in the composite) has larger resistance to compressive failure.

In paper C, FE analysis showed that the effect of mesoscale details on the NCF laminate stiffness may be in many cases neglected and a “smearing out“ routine leads to high accuracy of the determined stiffness matrix. The most suitable homogenization routine is based on the average fiber content in a layer consisting of fibers with a given orientation. The bundle structure in the layer is replaced by a homogenized layer and the elastic properties of the layer are calculated using Hashin’s composite cylinder assemblage model [10, 11]. As the next step the classical laminate theory is used to calculate the in-plane stiffness matrix. This approach was compared with 3-D FE modelling. The results even for NCF laminates with large matrix regions between bundles are within 1% accuracy which exceeds the typical precision in tests. This laminate analogy gives slightly lower accuracy for the shear modulus, which may be an indication that the bundle mesostructure is of more importance for this property. However, the error is still negligible (2% for cross-ply type of NCF composite). The above conclusions are valid if the thickness, $h_\phi$ of a layer with bundle orientation $\phi$ is constant and the layer (bundles) does not experience waviness along the fiber direction. Furthermore, transverse strain in bundles governs transverse cracking in NCF composites. FE analysis shows that this strain may be significantly lower than the applied macroscopic strain component in the same direction, which proves that the mesostructure is important in this case. This feature is important for damage evolution modelling. The iso-
strain assumption which in different combinations is widely used in stiffness models is inadequate because the strain in different meso-elements (bundles of different orientation and matrix regions) is assumed the same. Analysing by FEM the importance of media surrounding the bundle on average transverse strain it was found that an increasing ratio of the bundles transverse stiffness to the matrix stiffness leads to a decrease of the strain in the bundle. An increase of the stiffness in the same direction in adjacent layers leads to an increase of the transverse strain in the bundle. Higher bundle volume fraction in the layer leads to larger transverse strain in the bundle. These trends are described by a power law and used to predict the average strain in bundles. A calculated H-matrix which establishes the relationship between strains in meso-element and RVE strains is used to calculate the “effective stiffness” of the bundle. This “effective stiffness” is the main element in simple but exact expressions derived to calculate the stiffness matrix of NCF composites. Considering a 3-D FE model as the reference, it was found that all homogenization methods used in this study have sufficient accuracy for stiffness calculations, but only the presented method gives reliable predictions of strains in bundles.

In paper D, experimental results are presented which show that the reduction of elastic modulus in NCF cross-ply laminates due to increasing tensile loading in some cases is much higher than in laminates containing only transverse cracks. Moreover, reduction of stiffness exceeds the worst case scenario predicted by simple ply-discount model (complete failure of 90°-bundles). This indicates that in fact more complicated damage modes are present in the NCF laminate than just transverse cracking or local delaminations caused by cracks. Breaks of longitudinally oriented imperfect fiber bundles (usually in locations with certain waviness) were revealed using optical microscopy and attributed to the observed stiffness changes. The crack plane is usually normal to the longitudinal bundle axis. The energetically preferable locations of broken fiber bundles and the effect of bundle breaks and bundle delaminations on stiffness, considering the latter as a secondary damage mechanism, were analyzed using FEM. Two possible reasons for the higher stiffness reduction in [0/90/0/90]s NCF composite compared to [90/0/90/0]s case were suggested: a) If two imperfect 0°-bundle layers are separated by a 90°-bundle layer their resistance to failure is lower than when they are situated next to each other; b) the effect of each surface 0°-bundle break on the composite stiffness is larger (due to less constraint from the surrounding material the opening of surface bundle breaks is much larger).
In paper E, the most important parameters which control mechanical properties of NCF composites are identified and characterized. The identification is based on experimental observations and available theoretical findings. Characteristics of the internal structure of NCF composites are analyzed in context of their significance for in-plane elastic and failure properties. A methodology for determination of most typical geometrical parameters of composites using optical observations of cross-sections of manufactured laminates is described. The methodology is described for systematic characterization of the NCF composite using CF/EP cross-ply and quasi-isotropic composites as examples. This paper is focused on structural parameters and therefore it does not analyze the role of fiber and matrix properties which is, certainly, of great importance. Also the stitching effect on the material performance is not discussed. It is mainly because the used optical inspection tools for analysis of different cross-sections in the composite are not the most convenient tool for such investigation. More convenient is to characterize stitches on the NCF fabric before impregnation. However, then the information regarding the disturbances introduced during manufacturing is not available.
5. References


Paper A
Formation of damage and its effects on non-crimp fabric reinforced composites loaded in tension

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Abstract

Non-crimp fabric (NCF) composites, manufactured by resin infusion techniques are one of the most promising next generation composite materials. They offer large potential for application in primary structures as they give excellent performance at low production costs. However, before NCF composites will be efficiently used in design, detailed understanding of governing micro mechanisms must be accumulated and described by predictive models. In the present study, NCF cross-ply laminates have been tested in tension. Intralaminar cracks caused in the 90\textdegree/C14 fibre bundle layers and their effect on laminate mechanical properties have been monitored. Occurrence of ’novel’ type of cracks propagating in the load direction (longitudinal cracks) is explained by a thorough FE analysis using an Representative Volume Element (RVE) approach, revealing stress concentrations caused by 0\textdegree fibre bundle waviness. Effects of damage on mechanical properties are modelled using modified micro mechanical models developed for analysis of conventional laminated composites. The analysis reveals mechanical degradation to be ruled by the crack opening displacement (COD). However, unlike traditional composites, transverse cracks do not generally extend through the entire thickness of the 90\textdegree layer, but are rather contained in single fibre bundles, limiting the COD.

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1. Introduction

Damage in composite laminates subjected to mechanical loads is known to consist of intralaminar cracks lying along fibres in the plies, interlaminar cracks formed by local separation of plies, and fibre failures [1]. Studies of initiation of these cracks, their growth characteristics and the effects they collectively produce on the globally measured laminate properties belong to an active field of study called damage mechanics, playing a central role in the assessment of durability and damage tolerance of composite structures.

Composite structures can undergo multiple micro-cracking before losing ability to carry the design loads. In the sub-critical stage, however due to the surface displacements of the ply cracks, the deformation response can change significantly from the virgin state. These surface displacements are affected by the stiffness and geometry of the neighbouring plies (constraint effect) and therefore do not occur freely. The effect of damage, given by stiffness-damage relationships, has been a subject of extensive research for laminated composites in recent years. One type of damage mechanics approach, characterised as micromechanics, has mostly treated intralaminar (transverse) cracks in cross-ply laminates [2–5] relating the ply level properties and ply thickness to the laminate properties for given crack spacing. A recent work in micromechanics [6] has analysed transverse cracks in 90\textdegree plies in symmetric laminates containing plies inclined to the axial (loading) direction. Another type of approach taken is the continuum damage mechanics approach [7–11]. In the continuum damage mechanics approach damage is represented by internal state variables and relationships between measures of damage and the overall laminate properties are generated. This approach allows, in principle, to treat intralaminar cracking in off-axis plies of any orientation. However, certain material coefficients...
are usually to be determined experimentally for each laminate configuration considered. This limitation has recently been eased significantly by a combined micro-mechanics and continuum damage mechanics approach called synergistic damage mechanics [12].

Composites offer considerable potential to reduce structural weight. Nevertheless, due to their high material costs and sensitivity to out-of-plane loads (eg impact), the use of composites manufactured from prepreged tapes (prepregs) in civil aircraft primary structures has so far been limited. Non-crimp fabric (NCF) based composites address both these concerns, without any significant drop in in-plane performance. This has resulted in a strong interest among the European aircraft manufacturers to introduce NCF based composites in primary structures. However, to allow use of NCF composites in primary structures reliable damage tolerance models, to be used already at the design stage, must be at hand.

Any reliable damage tolerance model for composite structures must involve criteria based on failure mechanisms. Obviously, the mechanisms depend on the loads operating on the structure. For heterogeneous material, such as composites, structural behaviour is also dependent on the material structure on the micro-scale. As described above, models for prediction of effects of damage growth in composite materials based on prepregs are well established. For NCF based composites this is not the case. NCFs are textile preforms with multiaxial layers of fibre bundles stitched together in directions desired for structural design. During composite manufacture, preforms are stacked in a mould and infiltrated by a thermoset resin to form the composite. The architecture of the NCFs implies that these materials are heterogeneous not only at the micro-scale but also on the meso-scale (fibre bundles). In analogy with prepreg composites, reliable damage tolerance models of NCF composite structures must consider effects of the material heterogeneity on both these scales.

In the literature, most studies on NCF based composites are concerned with their manufacture, testing and applications. A few, however, address modelling issues. Particularly relevant are two papers by Drapier and Wisnom [13,14]. These papers describe a two-dimensional finite element model that can be replicated through the thickness of a bi-axial non-crimp fabric as a repeating unit cell. The models were validated by associated experimental work. The results presented in these papers imply that, on the one hand, physically sound failure criteria must consider features on the micro- and meso-scales. On the other hand, modelling of elastic properties requires homogenisation of elastic properties of the fibre bundle layers. These models must consider both out-of-plane and in-plane waviness of the fibre bundles as well as effects of resin rich areas, fibre bundle geometry and fibre volume fractions. A limited number of models to predict mechanical properties of NCF composites have been proposed in the literature [15,16]. Benard et al. [15] considered a hybrid carbon/glass fibre epoxy fabric of bidirectional non-interlocking warp and fill tows, and modelled the behaviour of this material using micro-mechanical equations in conjunction with the laminated plate theory. Classical thin laminate theory was used by Bibo et al. [16] to predict the “elastic” macro-mechanical behaviour of unidirectional and non-crimp fabrics, and the failure mechanisms were found related to crimp in the tows, although with slight differences in tension and compression. Cox and Dakhilah [17] developed a knock-down factor approach for woven composites using laminate theory considering the misalignment angle due to out-of-plane waviness.

The aim of the current study is two-fold. Firstly, the stiffness–damage relationship for a cross-ply NCF composite is to be monitored experimentally and modelled. For this purpose, analytical micro-mechanics models used for analysis of conventional laminated composites are to be modified and tensile tests are to be performed. Secondly, RVE based models of the composite mesostructure implemented in FEM are to be developed. RVE analysis is to provide information on stress and strain concentrations caused by the material heterogeneity and, hence, allow for explanations to damage mechanisms observed in tests.

2. Experimental

2.1. Materials

2.1.1. Manufacture

The laminate used in this study was manufactured from the NCF 0°/90° LT450-C10-R2VE manufactured by Devold AMT. This fabric is based on the Toray T700SC carbon fibre where the two fibre bundle layers are stitched together by polyester threads. The Vinyl-DION 9500-501, supplied by Reichhold, was used as matrix. The laminate was manufactured by vacuum infusion of the resin into the stacked NCF-reinforcement. The cross-ply laminate has the lay-up [0/90]s. Thus, each laminate was built up from two NCF blankets. After manufacturing, the laminate was post cured at 70 °C for 4 h.

The test specimens were cut from the 1 mm thick laminate, the edges of the specimens were ground and polished to achieve smooth surfaces. The final polishing was performed using a diamond suspension with 8 and 1 μm particles for the specimens aimed for stiffness and crack density characterisation respectively. The final width of the specimens, after grinding and polishing, was approximately 22 mm for the tensile stiffness specimens and 10 mm for the crack density specimens. The
length of the stiffness specimens was 230 mm whereas that of the crack density specimens was 100 mm. The dimensions of the specimens used for crack density monitoring were chosen so that the specimens fit both macro- and mini-testing machines and allowed introduced strain in the MINIMAT tester to be high enough to observe the damage.

2.1.2. NCF architectural characteristics

To allow for modelling of the NCF composite the architectural features must be characterised. For this purpose an investigation of the bundle shapes and geometry, etc. on the meso-scale of the material was performed. The results of the investigation provide the necessary and correct input data for the analytical and numerical analyses described in the theoretical section. The term “non-crimp” implies that the fibres in each lamina are straight and have no waviness perpendicular to the general fibre direction. However, it has been shown that this is not the case [13,14]. A micrograph of the laminate presented in Fig. 1 reveals that there is certain amount of crimp present in the outer 0° layers. The meso-scale structure of the laminate was characterised by means of image analysis of laminate micrographs obtained by optical microscopy. The out-of-plane waviness of the 0° bundles as indicated in Fig. 1 was assumed sine-shaped. For modelling purposes, the amplitude and wavelength of the assumed sine wave must be determined from micrographs. The amplitude of the sine function describing the 0° waviness was expressed as a function of the standard deviation of the orientation (SDO) with respect to the theoretical direction. Statistics provide a tool to compute the amplitude of any function to the standard deviation of the orientation (SDO). For the assumed sine shape of the 0° tows, the amplitude \( A \) can be shown to relate to the SDO as \( A = \sqrt{2} \cdot \text{SDO} \). This approach was presented by Drapier and Wisnom [13].

The wavelength of the sine function, \( \lambda \), was calculated from the measured mean values of the gap between the 90° bundles and the width, \( w \), of the 90° bundles: \( \lambda = \text{gap} + w \). The fibre bundle amplitude, \( A \), and wavelength, \( \lambda \), were determined to be \( A = 0.059 \text{ mm} \) and \( \lambda = 2.62 \text{ mm} \) in the 0° upper layer, and \( A = 0.031 \text{ mm} \) and \( \lambda = 2.62 \text{ mm} \) in the lower layer, respectively.

2.2. Stiffness determination in tensile tests

In this section, the tensile test set-up for determination of effects of damage on the constitutive properties is described. Tensile tests with a cross-head speed of 2.1 mm/min were performed in a servo-hydraulic Instron 1272 tensile testing machine. To prevent specimen sliding in the grips a metal net was used in the gripping area. An extensometer with gauge length of 50 mm was used to measure axial strain. The specimen length outside the grips was 170 mm. Transverse strain was measured using strain gauges with 10 mm gauge length.

Elastic properties (Young’s modulus, \( E_x \), and Poisson’s ratio, \( \nu_{xy} \)) were determined in a stepwise loading–unloading sequence with increasing maximum. The elastic properties were determined in the region 0.1 to 0.3% axial strain. The initial Young’s modulus, \( E_{iso} \), and Poisson’s ratio, \( \nu_{xx} \), were calculated using the stress and strain data from the initial loading, unloading and the following reloading in this region. A linear fit to data was used. In the following steps the unloading and following reloading data were fit to obtain elastic properties affected by damage. The ratios \( E_x/E_{iso} \) and \( \nu_{xy}/\nu_{xx} \) were plotted versus the maximum strain in the loading step. The maximum strain in the first and subsequent loading steps was 0.35, 0.6, 0.9 and 1.2%, each followed by unloading to 0.1% axial strain.

2.3. Measurements of crack density

An additional test was required to monitor crack density in the cross-ply laminate. The crack density tests consisted of two parts where the first part was performed in an Instron 1272 tensile tester with a cross-head speed of 2.1 mm/min and a specimen gauge length of 70 mm. In this test, a certain damage state was introduced loading the specimen to a predetermined strain level, different for each specimen. An extensometer with gauge length of 50 mm was used to control the strain level. The second part of the test was performed in a MINIMAT miniature mechanical test machine connected to an optical microscope to observe the cracks. The MINIMAT was used to open the cracks in the damaged specimen by applying a strain of 0.1%. During this procedure no new cracks were formed. The crack density was determined over a length of 25 mm for each specimen.

Fig. 1. Micrograph of a cross-section of the NCF cross-ply laminate.
This procedure was repeated for each specimen in three load steps within the strain interval of approximately 0.2–1.2%. Four specimens were tested and measured crack density was plotted versus maximum strain in each load step. Both edges of the specimen were investigated and plots were based on mean values of the average number of cracks/mm.

To assure that the dimensions of the specimens used in this test did not affect the crack density measurements, tests were also performed on tensile test specimens used for characterisation of stiffness properties. In these accompanying tests, the specimens were exposed to a maximum strain of 0.8%. After unloading, the specimens were cut parallel to the loading direction in two parts and polished to allow study of the cracks in the centre of the specimen. Again, the crack density was measured on a 25 mm length of the specimens.

2.4. Experimental results

2.4.1. Characterisation of the damage state

Damage in form of cracks was observed only in layers with 90°-orientation of fibre bundles. The observed cracks were divided into four groups, depending on location and orientation with respect to the laminate mid-plane. The different crack types were named: longitudinal cracks; half cracks; whole cracks and double cracks. Specific features of these cracks are schematically illustrated in Fig. 2.

The cracks named longitudinal cracks, see Fig. 2, are novel in the sense that they never occur in traditional laminated composites (prepregs) subjected to tensile loading conditions. In this study, longitudinal cracks appeared in the 90° layer both within the bundles and at the bundle–matrix interface. Typical longitudinal cracks are shown in Fig. 3. Longitudinal cracks were most often located within bundles and only seldom found to propagate in the matrix. However, at high strains cracks were found to grow through the matrix connecting to neighbouring bundles.

Cracks oriented mainly transverse to the laminate mid-plane, only penetrating a single bundle, are named half cracks (see Fig. 2). The name is related to the feature that these cracks do not cross the whole middle layer, which in the studied case consists of two fibre bundle layers oriented perpendicular to the loading direction. Instead, half cracks are always contained within a single 90° fibre bundle, without connection to cracks present in neighbouring fibre bundles. Examples of half cracks are presented in Fig. 4.

A crack extending from one 0° layer to an other, i.e. which runs through two adjacent 90° fibre bundles and the matrix in between, is called a whole crack (see Fig. 2). In case isolated fibres or fibre clusters lie embedded in the matrix in between bundles, the crack path is usually along the fibre-matrix interface. Examples of whole cracks are presented in Fig. 5.

The fourth category of cracks defined are double cracks, see Figs. 2 and 6. This crack type is a combination of a half crack and a longitudinal crack. Double cracks have a transverse branch like a half crack and a longitudinal branch as a longitudinal crack. The longitudinal part with a varying length may be located in the central part of the bundle as well as closer to the bundle boundary. These cracks are usually forming within single bundles and are not propagating in the matrix between two fibre bundles. For high strains, however, some double cracks penetrating the matrix are also found.

After determination of the number of cracks of a certain type within a gauge length, \( L \), one can compute the corresponding crack densities as presented in Fig. 2.

\[
p_{\lambda} = \frac{N_{\lambda}}{L}, \quad p_{\text{half}} = \frac{N_{\text{half}}}{L}, \quad p_{\text{whole}} = \frac{N_{\text{whole}}}{L}, \quad p_{\text{double}} = \frac{N_{\text{double}}}{L}.\]

Fig. 2. Schematic showing the four crack types observed in NCF cross-ply laminates. \( p \) refers to crack densities of the different type of cracks.
2.4.2. Effect of damage on elastic properties

Initial elastic properties, Young’s modulus, $E_{10}$ and Poisson’s ratio, $\nu_{12}$, measured in tensile test are presented in Table 1. As shown in the table, the measured initial Young’s modulus is fairly similar for the different specimens. The measured initial Poisson’s ratio, however, has larger scatter.

Typical examples of the obtained axial stress–strain and transverse–axial strain curves are shown in Fig. 7. The hysteresis loop for unloading and reloading in the stress–strain curve is very small. This is expected for carbon fibre cross-ply type of laminate in tension. The slope change with increasing maximum strain is also moderate which implies small change in elastic modulus. The significant
unloading slope change of the transverse versus axial strain curve in Fig. 7b) indicates large Poisson’s ratio degradation due to damage during the test.

Normalised Young’s modulus and Poisson’s ratio versus axial strain are plotted in Fig. 8. As shown in the figure, the observed damage is only slightly reducing the Young’s modulus of the composite. This is expected since the laminate modulus is mainly controlled by the non-damaged 0° fibre bundle layers. For Poisson’s ratio, however, the effect of damage is strong. This effect is expected because transverse cracks (half cracks, whole cracks or double cracks) increase strain in the loading direction and simultaneously reduce the contraction in the transverse direction (transverse bundles close to the crack in 90° layer are less loaded in x-direction and hence do not contract much in y-direction). The described effects are similar to those observed in cross-ply prepreg laminates. Also, the quantitative changes are similar to those for prepregs. This is confirmed by good accuracy of simulations in Section 4.

2.4.3. Crack density evolution

As described in Section 2.3, crack densities were measured on 10 mm wide specimens. However, the degradation of mechanical properties was measured on wider specimens. To assess the effect of specimen width on cracking, comparison between crack densities in the interior of the wide, and on the edge cracks of narrow specimens was performed. As a result, no statistical difference for half cracks could be seen between the two cases, see Table 2. This result validates use of smaller dimension specimens in the damage evolution test. It

![Fig. 6. Examples of double cracks in single 90° fibre bundles of the cross-ply NCF laminate.](image)

![Fig. 7. Typical stepped tensile loading-unloading curves: (a) stress plotted versus longitudinal strain; (b) negative transverse strain plotted versus longitudinal strain.](image)

<table>
<thead>
<tr>
<th>Specimen</th>
<th>$E_{10}$ (GPa)</th>
<th>$\nu_{xy}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>55.5</td>
<td>0.066</td>
</tr>
<tr>
<td>2</td>
<td>53.8</td>
<td>0.048</td>
</tr>
<tr>
<td>3</td>
<td>56.4</td>
<td>0.087</td>
</tr>
<tr>
<td>4</td>
<td>54.1</td>
<td>0.056</td>
</tr>
<tr>
<td>5</td>
<td>55.9</td>
<td>0.043</td>
</tr>
<tr>
<td>6</td>
<td>56.3</td>
<td>0.042</td>
</tr>
<tr>
<td>Mean</td>
<td>55.3</td>
<td>0.057</td>
</tr>
<tr>
<td>Standard deviation</td>
<td>1.1</td>
<td>0.017</td>
</tr>
</tbody>
</table>
should be pointed out that no double cracks were found for the applied strain level and that longitudinal cracks were not monitored.

Measured crack density of half cracks as a function of applied strain is depicted in Fig. 9. As shown in the figure, the number of half cracks increases approximately bi-linearly with increasing strain. The crack density is defined according to Fig. 2. In the tests, cracking was initiated at approximately 0.4% strain. As also shown in the figure, measured crack density initially increases dramatically and then slows down (ideally approaching an asymptotic value at saturation). This is consistent with observations on prepreg cross-ply laminates.

It was observed that the whole and double cracks appear at slightly higher strains (~0.5%) than half cracks (see Fig. 10). The density of these cracks is significantly lower and the scatter is large. As a consequence, the effect of these cracks on laminate stiffness is expected to be limited.

3. Theoretical modelling

3.1. FE-models

Two different unit cell models (which in this case coincides with the representative volume elements, RVE) were defined from the characteristic geometry of the studied NCF laminate. Two-dimensional FE models were adopted to simulate and investigate the response of the material from tensile loading.

3.1.1. Geometry

The laminate was modelled in the plane parallel to the 0° fibre bundles and perpendicular to the 90° bundles (x,z-plane in Fig. 11). In the first model, the 90° bundles are located on top of each other, i.e. on the same x-coordinate. This represents the case where the out-of-plane waviness of the top and the bottom 0° bundles is out-of-phase with each other (see Fig. 11b). In the second model the two 90° bundles are located on different positions in the x-direction, thus the 0° bundle waviness is in-phase, see Fig. 11a. These two cases, shown in Fig. 11, represent two geometrical extremes of the studied NCF-laminate.

The interface between the two 90° bundles was located in the specimen centre. The shape of the tip area of the 90° bundles was chosen to give a total 90° bundle...
width, \( w \), that corresponds to the measured mean width \( w = \lambda - \text{gap} \). The resin layers on the surfaces of the laminate were not considered in the models.

### 3.1.2. Materials

The linear elastic material models used for the matrix and the 90° bundles were isotropic and transversely isotropic respectively. For the 0° fibre bundles a linear elastic orthotropic material model was used. The main orthotropic material axis was parallel to the 0° fibre bundle, thus following the sine shape of the corresponding 0° bundle. The fibre bundle structure of the 0° layer was simplified replacing it with an effective 0° lamina. The longitudinal elastic properties were calculated using the rule of mixtures, where the fibre volume fraction was determined from the fibre substance (weight/unit area) divided by the fibre density and lamina thickness. The transverse elastic properties were calculated using the Halpin – Tsai equations. The fibre volume fraction within the 90° bundle was determined from the 90° lamina’s overall fibre volume fraction, in the same manner as for the 0° lamina, divided by the measured 90° bundle volume fraction. The fibre volume fractions used in the analyses were: 0° lamina \( V^{0}_{f} = 48.7\% \); in-phase 90° bundle \( V^{90}_{fb} = 59.4\% \); and out-of-phase 90° bundle \( V^{90}_{kv} = 58.1\% \). The material properties for the matrix material were chosen in accordance with the information from the resin manufacturer. The properties used in the analyses are reported in Table 3. Note that the elastic properties of the polyester stitch threads are not reported since they are not considered in the models.
3.1.3. Mesh

The analyses of the RVEs were performed under the assumption of generalised plane strain conditions. The general finite element code ANSYS [18] was used for the analyses. The geometry was modelled using the element type PLANE183. This is a plane element defined by eight nodes. The geometry was meshed using mapped meshing with quadrilateral elements in the major part of the model. The areas at the 90° bundle tips were meshed with triangular elements, as depicted in Fig. 12.

3.1.4. Boundary conditions

The boundary conditions for the two models are presented schematically in Fig. 13. The translational degree of freedom in the x-direction along the right vertical boundary was set to zero. Translation in the z-direction was prevented prescribing the node at the middle of the right vertical boundary zero displacement. In the analysis, a prescribed uniform deformation of the model left edge was applied (see Fig. 13). The applied deformation corresponds to a global strain in the x-direction of 1%. This loading condition keeps the two vertical boundaries of the model parallel, thus fulfilling the demands on symmetry.

3.2. Stiffness degradation models

The stiffness reduction mechanism due to cracks is related to the relative displacements (opening and sliding) of the crack surfaces. It has been shown [19,20] that in axial loading of cross-ply laminates the average crack opening displacement (COD), normalised with respect to the far field stress in the layer and with respect to the crack size, is the entity that uniquely determines the rate of the stiffness change with increasing crack density. Closed form expressions containing the COD, crack density and elastic characteristics of material have been presented for stiffness tensor components. Thus, the stiffness is uniquely related to the damage state.

The damage state depends on the loading history (the maximum load in the previous loading ramp, assuming that fatigue and other time dependent effects may be ignored). In quasi-static tensile loading of laminated fibre reinforced composites the required relationship is density of different modes of cracks as a function of the applied maximum strain. In the presented paper, damage evolution (which from modelling point of view is the most complex problem) is not simulated. Instead, experimental crack density versus applied strain data are approximated by bi-linear and/or linear curves and used in calculations.

The normalised average COD may be related to the local stress perturbation caused by the presence of the crack. In fact, there is a unique relationship between the volume averaged axial stress perturbation in the cracked layer and the average COD. Most of the approximate analytical stiffness reduction models (shear lag, variational etc) are based on stress state calculation under simplifying assumptions. Therefore, there exist a straightforward way to use the results of these models to calculate the COD and the corresponding stiffness tensor of the damaged composite. An alternative approach has been presented in [19]. Using FEM based numerical parametric analysis it was shown that the average

![Fig. 12. Mesh of RVE models: (a) 0° bundles in-phase; 0° bundles out-of-phase.](image-url)
normalised COD is a robust parameter that depends only on the stiffness and thickness ratios of the surrounding layers with respect to corresponding properties of the cracked layer. A power law was proposed to calculate the COD.

However, the available stiffness degradation models [2,19,21] have been developed for prepreg type of materials. Therefore, homogenisation must be performed to use them for NCF composites, i.e., the bundle layered structure must be replaced by an equivalent homogeneous layer. It is important to ensure that the calculated average COD, which mostly depends on the stiffness ratio of layers, is not significantly changed by the homogenisation routine. Since the transverse modulus of a bundle and the transverse modulus of the effective layer of the same orientation are not much different (matrix gaps are small), it is reasonable to assume that the error in the calculated crack opening displacement, which governs the rate of stiffness reduction, is small. The homogenisation may be performed as follows: the average fibre volume fraction in each layer is obtained using manufacturers data regarding reinforcement area weight, fibre density and measured NCF composite layer thickness. The calculated fibre volume fraction values together with fibre and matrix elastic properties are used to calculate layer stiffness matrix applying simple micro-mechanics models (rule of mixtures, Halpin-Tsai equations). Since the fibre content in the 90°-orientation and 0°-orientation layers is different, the stiffness matrices of these layers are also different. Results are presented in Table 4.

From the description of the modelling procedure it follows that the geometrical appearance of the crack may significantly affect the COD and, hence, the crack influence on stiffness reduction. The observed damage modes in the NCF cross-ply composite described in Section 2 have different effect on NCF-laminate stiffness. For example, longitudinal cracks which have the crack plane parallel to the laminate mid-plane have no effect on laminate response to in-plane loading, except that delaminations, in general, may change the Poisson’s ratio. For this reason these cracks are not included in the following simulation of the extensional stiffness matrix. The other three damage modes (whole cracks, half cracks and double cracks) are differently reducing the stiffness due the differences in opening of these cracks in tensile loading. The whole cracks due to their larger size open more than half cracks do. The opening of the double cracks is generally larger than that of half cracks because the longitudinal part of the crack contributes to the crack opening. However, this difference may be quantitatively evaluated only if the length of the longitudinal part of these cracks is measured as a function of applied strain. These measurements were not performed in the present study.

In this paper the analysis of the stiffness reduction due to observed types of cracks will be performed in three different approximations. In all cases cracks are considered as non-interactive and hence, the effect of individual cracks may be summed. Due to lack of more detailed information, double cracks are assumed to have the same effect on stiffness as half cracks and, therefore, are not considered as a special group.

“Large crack” approximation: a large crack is considered as a crack covering the whole 90°-layer thickness, see Fig. 14. Obviously, the whole crack belongs to

![Prescribed deformation, δ](image1)

![δ](image2)

Fig. 13. Boundary and loading conditions for the two FE-models: (a) 0° bundles in-phase; (b) 0° bundles out-of-phase.

<table>
<thead>
<tr>
<th>Property</th>
<th>0-layer</th>
<th>90-layer</th>
</tr>
</thead>
<tbody>
<tr>
<td>$E_1$ (Gpa)</td>
<td>112.47</td>
<td>114.51</td>
</tr>
<tr>
<td>$E_2$ (Gpa)</td>
<td>7.30</td>
<td>7.41</td>
</tr>
<tr>
<td>$G_{12}$ (Gpa)</td>
<td>2.32</td>
<td>2.35</td>
</tr>
<tr>
<td>$G_{23}$ (Gpa)</td>
<td>0.283</td>
<td>0.281</td>
</tr>
<tr>
<td>$v_{21}$</td>
<td>0.35</td>
<td>0.35</td>
</tr>
<tr>
<td>$G_{23}$ (Gpa)</td>
<td>2.70</td>
<td>2.75</td>
</tr>
<tr>
<td>$t_s$ (mm)</td>
<td>0.236</td>
<td>0.236</td>
</tr>
<tr>
<td>$t_c/2$ (mm)</td>
<td>0.236</td>
<td>0.236</td>
</tr>
</tbody>
</table>

Table 4 Effective properties of layers in stiffness reduction modelling
this group. Regarding the half cracks in this approximation, we assume that they may be considered as a pair of two closely located half cracks (one in the upper bundle and one in the lower bundle) which together behave as one large crack. This assumption is rather rough and implies larger predicted stiffness change than the correct crack geometry would give. Hashin’s model [2] as well as the model developed in [19] will be applied. Since the half crack density in the upper and lower bundle is assumed the same, the density of large cracks in this model becomes $\rho = \rho_{wh} + \rho_0 + \rho_c$.

“Small crack” approximation: a small crack is a crack covering only a single fibre bundle (ie half of the 90° layer thickness). Half cracks obviously belong to this group. In the model, it is important to retain the feature that “small cracks” have smaller COD due to their smaller size. To use the same expressions as in “large crack” approximation we will consider these cracks in pairs, see Fig. 15. Hence, they become “large cracks” but with a rather specific shape of opening displacements. The opening is zero at the interface between two bundles. The coupled nature of the “small cracks” shown in Fig. 15 is just for the sake of illustration emphasising the difference to the “large crack”. When the average COD is calculated we consider only one “small crack” surrounded by 0° layer from one side and undamaged 90° layer from the other side.

For uniformity of treatment also whole cracks will be considered as shown in Fig. 15. The latter assumption should lead to underestimation of the COD of the whole cracks and hence to underestimation of the stiffness reduction. However, as the density of the whole cracks is small compared to the density of half cracks the inaccuracy introduced by this assumption is expected to be small. The total number of cracks in this approximation is $N_d + N_{wh} + N_c$. Using the crack density definitions in Section 2.4.1 the density of “small cracks” is $\rho = \rho_{wh} + \rho_0 + \rho_c$.

The disadvantage of this crack geometry is that stress analysis of the known analytical models is not applicable (these models assume zero shear stress on the symmetry line which is not true for the geometry in Fig. 15). Therefore the analysis will be based on model [19] only.

“Mixed crack” approximation: to obtain the resultant stiffness reduction the most adequate would be to add the stiffness reduction due to whole cracks (crack density, $\rho_{wh}$), which are represented as “large cracks”, and the stiffness reduction due to “small cracks” (half cracks with density, $\rho_c$, and whole cracks with density, $\rho_w$). The predicted stiffness curve should then lie below the “small crack” approximation and above the “large crack” approximation curves.

For completeness the basic simulation ideas and the stiffness reduction expressions for the used models are presented below. The explanation is more detailed for “large crack” approximation. Since in the “small crack” approximation the cracks are considered in pairs, as shown in Fig. 15, the same expressions (but with different COD) may be used in analysis.

3.2.1. Stiffness modelling in “Large crack” approximation

Symmetric and balanced $[\mathbf{S},90_n]_{\text{s}}$ laminate, shown in Fig. 14, contains 90° layer with cracks, that cover the whole 90° layer thickness, and two balanced sub-laminates as supporting layers. In our particular case, the sub-laminate is a 0° layer. Thus, each crack runs through both upper and lower fibre bundles. The sub-laminate in the $(x,y,z)$-system is macroscopically orthotropic. The 90° layer is transversally isotropic. In the following, upper indices ( $\uparrow$) and ($\downarrow$) denote variables in the 90° layer and sub-laminate, respectively. Variables without upper index are used to describe the laminate.

The magnitude of stiffness reduction due to cracking is governed by normalised crack opening displacement $u_{an}$ and normalised crack density $\rho_n$. These quantities are defined as follows

\[
u_{an}(z) = \frac{2 \rho_{wh}}{(\sigma_{wh}^c/E_{wh})} t_c \int_{0}^{t_c/2} u(z)dz
\]

\[
\rho_{wh} = \frac{\rho_c t_c}{2 t_c}
\]

Fig. 14. Schematic showing of the laminate model with “large cracks” covering the whole 90-layer.

Fig. 15. Schematic showing of the laminate model with couples of “small cracks” covering only one bundle.
Using the procedure described in [21] the following expressions are obtained for damaged laminate stiffness characteristics, \( E_s \) and \( \nu_{xy} \). Virgin laminate elastic properties are denoted \( E_{0} \) and \( \nu_{0} \). The obtained expressions [21] are as follows:

\[
\frac{E_s}{E_0} = 1 - Q_{09}^0 (1 - \nu_{xy0}) \frac{1}{g^2 \rho_\text{lam} \Delta t_{tr}}
\]

\[
\frac{\nu_{xy}}{\nu_{xy0}} = 1 + Q_{09}^0 (1 - \nu_{xy0}) \frac{1}{g^2 \rho_\text{lam} \Delta t_{tr}}
\]

In [21,22] it was shown that the macroscopic properties of damaged laminate can be expressed through properties of undamaged laminate and crack face opening displacement, see Eqs. (3) and (4) with coefficients given by Eqs. (6) and (7). These relationships are obtained relating the damaged laminate strains, which are equal to strains in undamaged layers, to the volume averaged strain in the damaged layer and crack opening displacement. Since in this procedure all equilibrium and displacement continuity relationships are satisfied exactly, the Eqs. (3) and (4) represent an exact relationship between global damaged laminate properties and local parameter (COD) given by solution of the local stress perturbation problem.

In Eqs. (3) and (4) \( Q_{09}^0 \) and \( Q_{09}^0 \) are 90° layer Poisson’s ratio and stiffness matrix element, respectively. Other constants involved in Eqs. (3) and (4) are calculated as follows:

\[
v_{yy0} = \frac{E_{00}}{E_{00}}
\]

\[
g_3 = \frac{t_r}{2t_s} \left( S_{11} + S_{12} + \frac{S_{12} - S_{11}}{2} \right)
\]

\[
\frac{1}{8} = \frac{E_{00} v_{xy0}}{4E_T}
\]

\[
\left[ 1 + \frac{S_{11}^0 t_r}{S_{09}^0 t_s} + \frac{S_{12}^0 t_r}{S_{09}^0 t_s} \right] ^2 \left[ \frac{S_{11}^0 t_r}{S_{09}^0 t_s} + \frac{S_{12}^0 t_r}{S_{09}^0 t_s} \right] ^2
\]

Here \( S_{11}^0, S_{12}^0, \) and \( S_{09}^0 \) are in-plane compliance elements of the 90° layer, \( S_{11}^0 = \frac{1}{E_{01}}, S_{12}^0 = \frac{1}{E_{02}}, \) and \( S_{09}^0 = -\frac{v_{01}}{E_{01}^2} \).

\[
S_{xx}^d = S_{11}^0, \quad S_{yy}^d = S_{22}^0, \quad S_{xy}^d = S_{12}^0, \quad \text{and} \quad E_s^d = E_1^0.
\]

The average stress \( \sigma_{yy}^0 \) in the 90° layer of the repeat unit depends on axial stress perturbation caused by the presence of two cracks. Without losing generality, the axial stress distribution may be written in the following form:

\[
\sigma_{yy}^0 = \sigma_{yy}^0 - \sigma_{yy}^0 \psi(x, z).
\]

where \( \sigma_{yy}^0 \) is the CLT stress in 90° layer, \( -\sigma_{yy}^0 \psi(x, z) \) is stress perturbation caused by the presence of the crack. Volume average of Eq. (10) is

\[
\sigma_{yy}^0 = \sigma_{yy}^0 - \sigma_{yy}^0 \rho_s R_0.
\]
\[ R(x) = \frac{4}{7} \int_{-S/2}^{S/2} \psi(x, z) \, dz \, dx. \] (12)

It has been demonstrated in Joffe et al. [19] that \( R(x) \) may be related to the average crack opening displacement that is normalised with respect to load

\[ R(x) = k_u w(x). \] (13)

with \( k \) given by Eq. (7).

The local stress perturbation \( R(x) \) may be calculated using any analytical model or numerical procedures. The shear lag models which are the simplest among analytical models are not considered here because the authors strongly believe that the shear lag model is oversimplified and the shear lag parameter in the model is basically a fitting parameter and a particular modification of the shear lag model is usually chosen just to get a better agreement with particular experimental data.

An analytical model based on rather accurate stress analysis but which still remains simple is the model developed by Hashin [2]. The main results are presented below.

Hashin’s model: the original model which was developed for cross-ply laminates is here modified for more general case of \([S,90_n]\) laminates. In Hashin’s model, uniform axial stress distribution across the layer thickness is assumed in sub-laminate as well as in 90°-layer. This assumption gives linear \( z \)-distributions for shear stresses and parabolic for \( z \)-axis normal stress. Assumed expressions for \( x \)-axis stress components are as follows:

\[ \sigma_{x}^{00} = \sigma_{x}^{00}(1 - \psi(x)) \quad \sigma_{x}^{z} = \sigma_{x}^{00} \frac{1}{b} \psi(x) \] (14)

where, \( b = \frac{z}{t_c} \).

Minimisation of the complementary energy of the repeating element with respect to \( \psi(x) \) leads to a forth order differential equation with constant coefficients for this function. Roots of the corresponding characteristic equation are as follows:

\[ k = \pm (a \pm \sqrt{\beta}) \]

where

\[ \alpha = q^4 \cos^2 \frac{q}{2}, \quad \beta = q^4 \sin^2 \frac{q}{2} \quad \tan \theta = \sqrt{\frac{4q}{p^2}} - 1 \] (15)

provided

\[ 4q > p^2 \quad \text{and} \quad p = \frac{C_{22} - C_{11}}{C_{22}}, \quad q = \frac{C_{00}}{C_{22}}. \] (16)

Here,

\[ C_{00} = \frac{1}{E_2} + \frac{1}{E_3} \frac{C_{22}}{b^2} \quad C_{22} = \frac{v_{23}}{E_2} \left( \frac{b^2}{3} + \frac{2}{3} \right) - \frac{v_{12}^2}{E_2}, \]
\[ C_{11} = \frac{1}{30} \left( \frac{1}{E_2} \right) \frac{b^2}{E_2}, \quad C_{22} = \frac{1}{E_2} \left( \frac{1}{4} \frac{b^2}{3} + \frac{8}{30} \right) + \frac{1}{20} \frac{b^2}{E_2}. \] (17)

Perturbation function \( R(x) \) can now be found by integrating \( \psi(x) \), see Eq. (12). By performing integration we obtain:

\[ R(x) = \frac{4q}{{a^2 + p^2}} \frac{1}{2} \beta \sinh(2\alpha z_a) \cos(2\beta z_a) \] (18)

Power law for COD: previous FE analysis [19] has demonstrated that the normalised average COD is the same for a large variety of load cases: plane strain; generalised plane strain; thermal loading; bi-axial loading; etc. Important for this crack characteristic is, that for each particular case the COD is normalised with respect to the far field transverse stress in the damaged layer which, certainly, depends on loading conditions, laminate lay-up, etc. Hence, the relationships for normalised average COD obtained below may be used in all possible thermo-mechanical loading conditions. The real CODs are obtained multiplying by the current far field stress.

Based on symmetry considerations only a quarter of the repeating unit (Fig. 14) was used in FE simulations [19]. Symmetry conditions on sides \( x = -S/2, z \in [t_c/2, t_c/2] \) and \( x = S/2, z = 0 \) were applied. Traction free conditions were used on \( z = t_c/2 \) and on the crack surface \( x = -S/2, z \in [0, t_c/2] \). Constant displacement in \( x \)-direction was applied on \( x = 0, z \in [0, t_c/2] \). To have non-interactive cracks, the normalised crack spacing \( S_b/t_c = 5 \) was used. Investigations showed [19] that CODs, normalised according to Eq. (1), are effected by the presence of neighbouring cracks only if the normalised crack density \( \rho_c > 0.25 \).

At rather high crack density, the crack interaction leads
to smaller CODs. It was also shown that COD values mainly depend on the axial stiffness ratio of 0° and 90° layers and on the laminate geometry, the dependence on other elastic constants (Poisson’s ratio, shear moduli) are negligible. Based on these studies the following power law was presented

\[ v_{\text{an}} = 1.044 + \left[ 0.89 + 0.312 \left( \frac{E_x}{E_y} - 1 \right) \right] \left( \frac{E_x}{E_y} \right)^{0.75} \]  

(19)

3.2.2. Stiffness reduction: ‘small crack’ approximation

As described above, in this case we replace all cracks (whole cracks as well as half cracks and double cracks) by a system of crack pairs passing through both fibre bundles, having zero opening displacement at the interface between bundles. In this case, the damaged laminate is characterised by damage state presented in Fig. 15.

The length of a pair of “small cracks” in Fig. 15 is \( t_x \), the same as in the whole crack case but the COD is different. Important for the “small crack” approximation is to realise that the constraints on the opening of the upper crack tip and the crack tip at the mid-plane are different. For example, considering the upper crack the top neighbouring layer is the 0° layer whereas the bottom layer is the 90° layer. Model based on crack opening displacement (COD) allow straightforward generalisation for this case.

We assume that COD in the upper part of this “small crack” is determined by the constraint from the 0° layer but the COD of the lower part of this “small crack” depends on the constraint of the 90° layer. Then the average COD is an average taken of COD in case of “small crack” constrained by 0° layers from both sides and COD of a “small crack” constrained by 90° layers from both sides. Then one may show that the normalised average COD of these cracks is

\[ v_{\text{an}} = \frac{1}{4} \left( v_{\text{an}}^0 + v_{\text{an}}^90 \right) \]  

(20)

where

\[ v_{\text{an}}^0 = 1.044 + \left[ 0.89 + 0.312 \left( \frac{E_x}{E_y} - 1 \right) \right] \left( \frac{E_x}{E_y} \right)^{0.75} \]  

(21)

and * is ‘0°’ or ‘90°’. In the analyses calculating \( v_{\text{an}} \), we use \( E_x = E_y \). Calculating \( v_{\text{an}} \) we use \( E_x = E_y^90 \).

4. Comparison between calculated results and test data

In this section modelling and experimental results are compared. In Section 4.1, the observed novel longitudinal cracks are explained using numerical results obtained by the detailed FE analysis described in Section 3.1. In Section 4.2, comparison between results from analytical stiffness degradation models, expressed in Section 3.2, and experimental results is presented.

4.1. Assessment of novel longitudinal cracks observed in tests

Results from the FE analyses of the repeating NCF micro-structure are presented in Fig. 16. Fig. 16a shows results for a NCF micro-structure where the 0° fibre bundles are in-phase, whereas Fig. 16b depicts results for a NCF with the 0° bundles out-of-phase. The figures present local stresses on a line in the centre of the upper 90° bundle layer. Stresses in the load direction (x-direction) and out-of-plane direction (z-direction) are presented at a far-field strain of 1%. As shown in the figures, stresses in the load direction exceed those out-of-plane. However, both models predict high out-of-plane stresses in the 90° bundles. For the case where the 0° bundles are out-of-phase, the peak stress in the 90° fibre bundle is the highest and reaches 57 MPa, corresponding to a out-of-plane strain of 0.66%. The peak out-of-plane stress and strain for the case with 90° bundles in-phase are 35 MPa and 0.39%, respectively. These relatively high out-of-plane strains are caused by the waviness of the 0° fibre bundles, which are forced to straighten in tension. Comparing the in-plane transverse strain (1%) with the reported out-of-plane strains in the modelled undamaged laminate it is clear that normal transverse cracks will appear first. This is consistent with experimental results (0.4% for half cracks versus 0.5% for double cracks). Nevertheless, longitudinal cracks caused by stress concentrations inherent to straightening of the 0° fibre bundles will follow.

The two geometrical extremes of the NCF composite under consideration result in different prediction of the Young’s modulus. This is consistent with what has been observed for transverse modulus predictions based on square and hexagonal unit cell representations of composite micro-structures [22]. The predicted Young’s modulus of the models with 0° bundles out-of-phase and in-phase were 57.5 and 53.6 GPa, respectively. The experimental value for the Young’s modulus was determined to 55.3 GPa. Consequently, the plausible bounds of Young’s modulus, predicted by the two models, enclose the experimentally measured value of Young’s modulus.

4.2. Degradation of elastic properties

4.2.1. Stiffness degradation in “large crack” approximation

The crack density data for whole cracks, double cracks and half cracks were reduced to a density of “large cracks”; see Fig. 14, using expression \( \rho = \rho_{\text{wh}} + \rho_{\text{d}} + \rho_{\text{h}} \). The “large crack” density versus applied strain is presented in Fig. 17.
Experimental data were fitted by a bi-linear curve. Equations of fitting curves are presented in Fig. 17. These equations were used to recalculate the experimental stiffness versus strain data to the form where stiffness is shown as a function of crack density. The theoretical stiffness reduction models (Hashin’s model and power law based model) presented in Section 3.2.1 were applied to predict the NCF laminate stiffness reduction with increasing “large crack” density. The predictions are compared with experimental stiffness versus crack density data in Fig. 18.

Obviously the “large crack” approximation is rather far from the real situation. Common for both graphs in Fig. 18 is that simulations are in a poor agreement with experimental data predicting much larger stiffness reduction than observed in tests. Hashin’s model [2] predicts larger axial modulus $E_x/E_0$ as well as Poisson’s ratio $\nu_{xy}/\nu_{0y}$ reductions than the COD based model [19]. This is explained by the used variational approach based on principle of minimum complementary energy that leads to a lower bound solution.

The results presented in Fig. 18 indicate that a pair of half cracks cannot be considered as acting like one “large crack”. Such an assumption leads to significant overestimation of stiffness reduction. Consequently, the often observed close location of two half cracks in the upper and lower bundle of the 90° layer (similar x-coordinate), does not mean that these cracks are essentially connected inside the specimen. These couples definitely do not act like a “large crack” crossing the whole 90° layer.

4.2.2. Stiffness degradation in “small crack” approximation

Crack density data for whole cracks, double cracks and for half cracks were reduced to the density of “small cracks”, see Fig. 15, using the expression $\rho = \rho_{slc} + \rho_{sc} + \rho_{lc}$. The “small crack” density versus applied strain curve which is the same as for “large crack” case is presented in Fig. 17. The difference is in the opening displacement of this crack system.

The predicted elastic modulus is compared with experimental data in Fig. 19a. Agreement is much better then using the “large crack” approximation. Similar conclusions may be drawn considering Poisson’s ratio predictions and test data versus crack density, see Fig. 19b.

The “small crack” approximation theoretically should render smaller stiffness changes than observed experimentally. However, this is not the case. Common trends considering deviation between model and test may be observed in both graphs in Fig. 19. For low crack densities the model overestimates the modulus reduction. This indicates that at this stage the cracks are not even crossing one bundle (they are not “small cracks” in the sense of definition in Section 3) or are not propagating in the fibre direction through the whole specimen. Related reason of lower stiffness reduction in test may be the elliptical shape of bundles: crack covering the
entire bundle thickness is still smaller than the 90-layer half-thickness. An obvious alternative explanation would be inaccuracy of the model. However, the only possible source of error is the value of normalised average crack opening displacement which determines the initial slope of the reduction curve. Crack interaction at higher crack density reduces the slope. The observed stiffness reduction relationships show the opposite—the slope is increasing with crack density. It may be explained only by the fact that each individual crack is evolving—changing its geometry (becoming larger)—with increasing load.

At larger crack density (corresponding to larger strain) the observed stiffness reduction becomes in a better agreement with predictions based on the “small crack” approximation and there are indications that finally experimental reduction is larger. It is not surprising because at large strains increasing number of “whole cracks”, which definitely have larger openings than the assumed in “small crack” model, were observed. Also the number of double cracks (bundle cracks with longitudinal branches) increases with strain. These cracks also have larger COD leading to additional stiffness reduction.

4.2.3. Stiffness degradation in “mixed crack” approximation

More detailed statistics regarding crack density is needed to improve predictions. In this section, whole cracks represented as “large cracks” are considered separately from half cracks and double-cracks that are represented as “small cracks”. The corresponding crack densities together with bi-linear and linear approximations are given in Fig. 20. These crack density curves were used in simulations to transform the simulated stiffness versus applied strain results to the form presented in Fig. 21a. The density of whole cracks at a certain strain level was determined from Fig. 20. This crack density was used to calculate the modulus using the “large crack” approximation. Results are presented as “wh”-curve in Fig. 21a. Similar procedure was performed for the group containing half cracks and double-cracks. This time the “small crack” approximation was used to calculate modulus. The predictions are presented as “h+d”-curve in Fig. 21a. The total of both reductions is presented as “h+d+wh”-curve. The prediction based on the assumption that all cracks belong to the “small crack” family is presented for comparison. Obviously the most detailed classification of cracks (analysis of whole cracks as “large cracks” and the rest of cracks as “small cracks”) leads to additional stiffness reduction in simulations, however, the difference is small. The explanation for that is the comparatively low density of the whole cracks.

Similar calculations regarding the Poisson’s ratio are presented in Fig. 21b. The conclusions are exactly the same as for elastic modulus predictions.

Fig. 19. Simulation performed using the “small crack” approximation. (a) Elastic modulus reduction due to cracks in transverse bundles; (b) Poisson’s ratio reduction due to cracks in transverse bundles.

Fig. 20. Density of cracks as a function of applied strain: “large cracks” schematically presented in Fig. 14; “small cracks” shown in Fig. 15. Solid lines are bi-linear and linear approximations.
5. Conclusions

The study concerns meso-mechanical modelling of a NCF cross-ply laminate subjected to tension load. In the present paper, damage caused in the 90° fibre bundles and its effect on laminate mechanical properties has been monitored. The damage caused during tension is labelled into four categories: longitudinal cracks; half cracks; whole cracks; and double cracks. Half cracks that are contained in a single 90° fibre bundle and, hence, only extend through half the 90° layer, occur first (at a strain of approximately 0.4%) and dominate over all other crack types. The other crack types initiate at slightly higher strains >0.5%. Of these, longitudinal cracks occurring by them self in a single bundle or linked to a half crack forming a double crack are never observed in traditional prepreg composites. Occurrence of this “novel” type of crack propagating in the load direction (longitudinal) is explained by a thorough FE analysis revealing stress concentrations caused by 0° fibre bundle waviness. Longitudinal cracks form at stress concentrations caused by the forced straightening of the 0° fibre bundles in tension. The effect of the fibre waviness is strongest for the analysed case where the upper and lower 0° fibre bundles are extending out-of-phase.

Experiments showed that the effect of damage on longitudinal modulus is very limited, whereas the effect on Poisson’s ratio is large, for the NCF cross-ply composite. Effects of damage on mechanical properties were modelled using conventional (Hashin [2]) and modified (power-law [19]) analytical micro-mechanics models developed for analysis of traditional laminated composites. It was shown that the mechanical degradation is ruled by the crack opening displacement (COD). Furthermore, the strong influence of half cracks on the degradation of mechanical properties may not be handled by the Hashin’s model. Models that assume transverse cracks to extend through the entire 90° layer, eg the Hashin’s model [2], overestimate the effect of damage on the mechanical properties. Modification of a model to allow for analysis of laminates containing half cracks is presented. This model takes into consideration the constraint effect of neighbouring non-cracked 90° fibre bundles. Computations considering these “small cracks” are found to successfully predict the behaviour of the NCF cross-ply laminate, with only a slight overestimation of the reductions in the mechanical properties. Theoretical considerations, however, imply the model to underestimate these reductions. This indicates that half cracks do not extend through the whole specimen width and/or the entire 90° fibre bundle layer thickness, t_c.

To further illustrate the potential of the modified model, stiffness predictions separately considering effects of experimentally determined crack densities of “small cracks” and “large cracks” are presented. Results from this analysis indicate that the occurrence of “small cracks”, ie half cracks, completely dominate the COD of the 90° layer in the NCF cross-ply laminate. This trend to form “small cracks” observed for NCF composites is beneficiary as resulting reductions in mechanical properties are less than those expected for the corresponding prepreg composite.

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References

Paper B
Compressive failure analysis of non-crimp fabric composites with large out-of-plane misalignment of fiber bundles

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Abstract

Failure initiation under compressive loading in non-crimp fabric composites containing bundles with out-of-plane orientation imperfections was analyzed using FEM in plane stress and linear elastic formulation. The bundle orientation imperfection in a composite unit was described by a sine function. Failure initiation strain was determined comparing failure functions corresponding to two alternative failure mechanisms: (a) plastic microbuckling in bundle due to mixed compressive and shear load; (b) plastic matrix yielding according to von Mises criterion. Parameters for compressive failure initiation analysis were bundle misalignment angle, fiber volume fraction inside the bundle and bundle volume fraction inside the composite unit. The support effect of the neighbouring material was analyzed varying boundary conditions and solving cases with particular configuration of surrounding material. Model prepreg tape GF/EP composite with different introduced levels of out-of-plane waviness of layers was used to validate the conclusions from parametric analysis.

Keywords: A: Layered structures; B: Elasticity; B: Mechanical properties; C: Computational modelling

1. Introduction

Non-crimp fabric (NCF) composites due to their excellent performance and relatively low manufacturing costs have become an attractive alternative for aerospace and marine applications. They have reinforcement in form of blankets each of them consisting of several layers stitched together by stitching yarns (see Fig. 1). Each layer consists of straight unidirectional fiber bundles. NCF composites are produced using techniques like Resin Transfer Molding, Resin Film Infusion etc. Additionally to substantial production cost reductions compared to prepreg tape based materials also improvements in damage tolerance and out-of-plane fracture toughness of these materials have been reported [1–3].

One of the features of the NCF composite is that it is an inherently multiscale material. On the microscale each bundle is a unidirectional (UD) fiber composite with a certain fiber content \( V_f \) and the homogenized bundle properties may be calculated using micromechanics expressions for long fiber composites. On the mesoscale bundle is considered as homogeneous transversely isotropic material surrounded by matrix and other bundles of the same or different orientation. A mesoscale characteristic of the composite is bundle content \( V_b \) in a layer. The geometrical shape of bundles (cross-section and axial alignment) is complex and depends on bundle orientation in the blanket, surface compression during production, resin pockets etc. The described mesoscale configuration determines the NCF composite properties on macroscale and is typically used in simulations of macro behavior.

To understand the NCF composite behavior in compression it is important to realize that NCF composites have imperfections in form of waviness on two scales. On microscale bundle being a UD composite has inherent fiber waviness as in any unidirectional composite (even if the bundle would have a perfect orientation). This assumption allows application of UD composite failure criteria on the bundle scale. On the mesoscale bundle orientation is not...
perfect and exhibit both in-plane and out-of-plane waviness. Thus the resulting mesostructure contains different types of imperfections bundle out-of-plane orientation imperfection being one of them. This waviness has been characterized experimentally and described by a sinusoidal shape using the standard deviation as a measure of the wave amplitude [4].

Since NCF composites in sense of bundle structure and orientation imperfections represent an intermediate configuration between prepreg tape based composite and woven composite, it is obviously necessary to identify relevant methods of compressive failure analyzes used in the latter. An interesting approach to analyze plain weave fabric composites was used in [5]. Kink band formation in bundles due to plastic microbuckling was not observed experimentally in [5]. This may be explained by large angles of bundle waviness in plain weave composites promoting shear failure. As a possible reason for initiation of failure was suggested combined effect of bending and direct axial stresses in the longitudinal bundle leading to shear failure of the specimen. In the developed compressive strength model the longitudinal bundle is consisting of a number of elements each of them being a beam on elastic foundation. The elastic foundation may be transverse bundle, matrix or a mix of them. The obtained stress distributions are used in failure analysis which is based on maximum stress criterion. Since it is not clear whether the short elements of the longitudinal bundle may be described as beams, FEM analysis for validation of presented results would be very useful. The model renders rather good strength predictions as compared to experimental but the predicted failure strain shows trends different than the experimentally observed. Simulations showed that for CF/EP composite the failure initiates in the transverse bundle if the gap versus wave length ratio is small and it initiates in the longitudinal bundle if the ratio is larger than 0.3. This conclusion is not certain because only the normal stress in the longitudinal bundle has rather smooth distribution along the fiber, whereas shear stress has a strong peak which value and also its applicability in the maximum stress criterion is questionable.

Regarding NCF composites the sequence of failure events from initiation mode to final failure and its dependence on material and geometrical parameters is still unclear. Post-failure inspection of the carbon fiber composite compression specimens [6] revealed a significant level of debris with delamination between sublamine groups. The triaxial carbon and the quasi-isotropic glass fiber NCF composites failed at rather low stress in a splaying mode. Authors correctly speculate that 'fiber waviness and the periodicity of trigonometric-like constriction in the reinforcing tows by the stitching yarn is the key factor in reducing compressive strength in the 0-direction'.

Our own studies have shown that plastic microbuckling in bundles is also a possible failure mechanism, see Fig. 2. Due to localized nature of mesoscale imperfections the final failure is not necessary immediately triggered by the microbuckling event in the most imperfect bundle. The bundle architecture of the NCF composite can help to localize the buckling region and to suppress the propagation of the plastic zone. Furthermore, the secondary damage mechanisms initiated by the primary mode lead to final failure with a very complex mix of delaminations, bundle collapsing etc, making it very difficult to identify the real sequence of events.

Theoretically the effect of longitudinal fiber bundle waviness on the compressive strength of NCF composites was analyzed in [7] using FE in generalized plane strain formulation. It means that the heterogeneous bundle mesostructure was kept only in the plane containing the direction of loading and the thickness (out-of-plane) direction, whereas in the width direction the composite is homogeneous. Starting with the experimental nonlinear shear response of the resin the non-linear bundle properties were calculated using micromechanics relationships. Thus the specimen collapse due to decreasing shear resistance was the only compressive failure mode in this study. Compressive loading of the imperfect NCF specimen until its failure in shear mode was simulated. The geometry and boundary conditions used in FE simulations indicate that the model is a repeating element in both longitudinal and thickness.
direction which implies that the failure initiation stress in the model is equal to the stress to failure of the NCF plate.

Based on our observations, see Fig. 2, and the suggested multiscale approach we consider in the presented paper two possible mechanisms for compressive failure: (a) plastic microbuckling of a bundle (considered as a UD composite) due to combined action of longitudinal compressive and shear stresses in the regions with large bundle out-of-plane misalignments and (b) matrix yielding outside the fiber bundle.

Plastic microbuckling as compressive failure mechanism for unidirectional composites was first analyzed in [8,9]. This model assumes the existence of zones of infinite width with inherent fiber misalignment (waviness). This imperfect orientation leads to plastic yielding of the matrix in this region when macroscopical axial compressive load is applied. Later this model was generalized considering combined loading case [10] when additionally to the compressive stress in the average fiber direction there is also macroscopic shear stress component acting in the plane of waviness. It was shown that the shear stress component may significantly reduce the resistance to compressive stress. As the most important in the context of the presented paper is the result that compressive stress in 0-direction is coupled with the shear stress in the compressive failure criterion. The model was developed for an infinite width of the band of imperfection and in 2-D formulation where the plane of shear stresses and the plane of fiber waviness are coinciding.

Later on results of finite element calculations were presented [11] showing that even if the initial fiber imperfection band is of finite length the microbuckle propagates outside the misalignment region. The misalignment zone was assumed elliptic with amplitude of misalignment decreasing according to cosine law. It was shown that for finite width of the misalignment zone the compressive stress at buckling lies between Rosen’s elastic microbuckling value [12] and the infinite width misalignment zone model [8,9]. Value equal to 100 fiber diameters was given as the critical misalignment zone width, where the simple criterion [8,9] is valid. Since this value approximately corresponds to the bundle thickness in NCF composites one may speculate that the compressive strength of a ‘properly’ aligned and supported bundle may be higher than for an infinite composite which may have larger zone of initial misalignment.

Since the plastic microbuckling model describes microbuckling in an infinite medium, it is unclear how factors like support of surrounding layers in the laminate and edge effects could be analyzed using this model. This problem is important also considering NCF composites on mesoscale. An interesting approach to analyze the effect of the supporting off-axis layers on the fiber buckling in the axial layer is presented in [13]. Equilibrium of a fiber in the 0-layer with sinus-shaped in-plane waviness was analyzed using beam model on elastic foundation and describing the matrix resistance to in-plane shear by a distributed moment and the effect of the restraint of non-buckling layers by a distributed load. The distributed load is calculated from out-of-plane shear stress caused by the thickness gradient of the displacement in the transverse in-plane direction. The through-the-thickness distribution is obtained choosing the shape function and calculating involved constants by minimizing the strain energy of the system. Similar model considering fiber as a beam on elastic foundation was used in [14] with a difference that the distributed moment due to in-plane shear stress was ignored but fiber volume fraction was accounted for. Also the distributed load was calculated in a different way. The described approach was further developed in [15] dividing the 0-laminae in thin fiber and matrix layers and using numerical procedures instead of a priori assumed through-the-thickness distribution as described in [14]. The method was used to study the effect of matrix properties, layer thickness and boundary conditions on the limit of stability in elastic formulation. Based on numerical studies a homogenization scheme was proposed which is in a good agreement with numerical results.

The described methods may be generalized and applied on mesoscale to analyze the buckling of an in-plane imperfect bundle in NCF composite surrounded by other bundles.

The objective of the presented paper is to study the compressive failure initiation in an imperfect element of NCF composite. FEM based parametric study is performed on the mesoscale to identify the microscale and mesoscale parameters governing failure initiation in an NCF element containing fiber bundle with a certain degree of out-of-plane waviness which is surrounded by matrix and supported by different layers. As two competitive failure mechanisms plastic microbuckling in the bundle and yielding in the matrix are considered. To verify the conclusions a model NCF composite’ with predetermined out-of-plane waviness is designed, tested in compression and analyzed theoretically.

2. Failure mechanisms and criteria for NCF composites in compression

2.1. Failure mechanisms and stress state

The difference between prepreg-tape based laminate and NCF composite is that the former has internal structure consisting of continuous fibers that are rather homogeneously dispersed, whereas NCF composite has continuous fibers combined in fiber bundles with well defined geometry (see Fig. 1a). However, each fiber bundle consists of several thousands of fibers (6k, 12k, etc., see Fig. 1b).

As described in introduction this hierarchical structure and very different length scales in NCF composite imply that failure analysis should be based on multi-scale
approach. It means that mesoscale properties of a bundle used in this paper (elastic properties and strength) are obtained using methods developed for unidirectional prepreg tape based composites. Rule of mixtures and Halpin-Tsai expressions were used to calculate elastic properties of the bundle. Plastic microbuckling phenomenon described for completeness below and observed in NCF composites by optical microscopy, shown in Fig. 2, is assumed to govern the bundle strength.

Ideally fiber bundles in NCF composites would be straight and aligned. However, in reality, mostly due to manufacturing procedure, there are imperfections within NCF composites. Most harmful imperfection for compressive strength is out-of-plane waviness of bundles (see Fig. 3 showing a cross-ply NCF composite with a large imperfection before and after compressive loading, this is view from the edge).

The out-of-plane bundle waviness on the mesoscale can be characterized by an angle which is the deviation from perfect (straight) orientation. In the wave region the longitudinal ($L$) direction of the bundle has a certain angle with respect to the macroscopic compressive load direction ($y$-direction). This results in combined compressive-shear loading in the local ($L$-$T$) axes of the bundle which facilitates kink-band formation (see Fig. 2), compressive failure mechanism observed in prepreg based composite due to plastic microbuckling.

The waved bundle within composite is surrounded by a matrix and constrained by other bundles which have ‘perfect’ orientation. As a result the stress state in the misaligned bundle is rather complex and may be analyzed only by numerical methods. The input parameters for the model are shape of bundle with a maximum angle of waviness, bundle geometry, fiber content inside the bundle $V_f$, volume content of a bundle in the layer $V_b$ and mechanical properties of fiber and matrix. The stress state in the imperfect element, containing the curved bundle and matrix, as affected by surrounding layers of different orientation is the basic information for failure analysis. Definition of adequate boundary conditions which represent different cases of the constraint of surrounding layers simplifies analysis.

In the matrix region, surrounding the imperfect bundle, a complex stress state is developed due to the bending of the bundles. Therefore, failure may initiate in the bundle as well as in the matrix (yielding) surrounding the bundle.

2.2. Plastic microbuckling

2.2.1. Uniaxial loading

The following analysis and expressions are valid for perfectly aligned bundle loaded in compression ($y$- and $L$-directions coincide in any position in the bundle). On the microscale fibers in the bundle have zones with fiber misalignment characterized by angle $\phi$. Considering the composite material as homogeneous anisotropic with rigid-perfectly plastic behavior of constituents Budiansky [8,9] identified the shear yield stress $k_y$ of the composite and the initial misalignment angle $\phi$ of fibers in the kink-band as the main factors governing the compressive strength. Definition of the initial misalignment angle $\phi$ and inclination angle $\beta$ is shown in Fig. 4 (where $\Phi$ is the composite shear yield strain, $\gamma = \Phi$). The compressive strength is the lowest if the kink band inclination angle $\beta = 0$.

The following expression for compressive strength $\sigma_{cu}$ of UD composite was obtained

$$\sigma_{cu} = \frac{k_y}{\Phi + \gamma}$$  \hspace{1cm} (1)

In (1) if $\beta = 0$, $k_y$ is equal to composite shear strength ($k_y = k$). It should be noted that the stress level $\sigma_{cu}$ is not related to fiber failure. Fibers would still hold an increasing load, but they lose stability due to matrix yielding, thus fiber properties are not fully utilized in this failure mode. It is
assumed in following calculations that the composite shear strength \( k^* \), shear strain \( \gamma \) and the initial misalignment angle of fibers are independent on fiber volume fraction in the bundle. Since the real effect of \( V_f \) on these parameters is unclear such assumption can be justified. In order to more accurately account for fiber content influence on these parameters, additional experimental characterization is needed. According to this model the compressive strength of an aligned bundle does not depend on the fiber properties or fiber volume fraction. The composite yield stress \( k_s \) is often set equal to the matrix yield stress. This result follows from the rather rough constant shear stress model.

Direct experimental evidence to support (1) was obtained by Piggott and Hariss [16] who varied the matrix yield stress and showed that compressive stress of the composite is proportional to it as long as the failure mode is microbuckling. When matrix yield stress was very high fibers collapsed prior to microbuckling. More evidence was obtained in tests at varying temperatures by Barker and Balasundaram [17], with increasing temperature the yield strength of the matrix decreased and failure mode changed from fiber collapse to microbuckling.

2.3.2. Combined shear-compressive loading

If unidirectional laminate is subjected to combined remote axial and shear loading the necessary compressive stress for kink-band formation, \( \sigma_c \) depends on the level of the remote shear stress \( \tau_{Lx} \). This result was obtained in [10] and is expressed through following modification of (1)

\[
\sigma_c = k^* - \tau_{Lx} \frac{1}{\alpha + \gamma}
\]  

(2)

As it was mentioned before, in case of \( y \)-axis compressive loading of NCF composite with misaligned bundle the stress state within a bundle depends on many factors and becomes rather complex. Obviously in \( L-T \) system of coordinates even in simplest cases shear stress \( \tau_{Lx} \) is present additionally to compressive stress component \( \sigma_c \) and Eq. (2) should be used to determine the stress state at microbuckle initiation.

2.3.3. Mesoscale failure criteria

In order to perform failure analysis on mesoscale we need to define failure criteria using quantities defined on this scale. As stated in Section 2.1, due to complex internal structure of the imperfect unit of the NCF composite failure can initiate not only in the misaligned bundle but in the surrounding media as well. Therefore, failure criteria should be defined for each region within the analyzed unit.

2.3.3.1. Plastic microbuckling criterion for bundle

The stress state in the imperfect unit subjected to uniform \( y \)-axis compressive strain depends on many factors. The out-of-plane bundle misalignment is analyzed in this paper using plane stress FEM and the mesoscale local stresses in an arbitrary location in the bundle are \( \sigma_c \), \( \tau_{Lx} \), and \( \sigma_T \), \( T \)-direction here and in following is the out-of-plane direction. The criterion for microbuckle initiation given by Eq. (2) may be rearranged using Eq. (1) in a different form containing only terms characterizing mesoscale properties

\[
\sigma_c = k^* - \tau_{Lx} \frac{1}{\alpha + \gamma} = k^* - \frac{\tau_{Lx}}{k_s} \frac{\alpha}{\alpha + \gamma} = k^* - \frac{\tau_{Lx}}{k_s} \sigma_{cu} - \frac{\tau_{Lx}}{k_s} \sigma_{cu}
\]  

(3)

Thus the criterion which the local stresses \( \sigma_c \), \( \tau_{Lx} \) have to satisfy to initiate microbuckling is

\[
\frac{\sigma_c}{\sigma_{cu}} + \frac{\tau_{Lx}}{k^*} = 1
\]  

(4)

It is noteworthy that the failure criterion (4) describes the interaction between compressive and shear stress in a linear form. The absolute values should be used in this criterion. In extreme cases when only shear stress or only compressive stress is acting we obtain the usual maximum stress criterion. One can see that if the loading is predominantly in shear the material resistance will be reduced if additionally compressive stress is applied. Similar conclusion is obtained if the loading is predominantly compressive. Eq. (4) is valid if the microscale fiber waviness plane coincides with plane where the shear stress is acting. In current case before otherwise is proven we assume that the microscale fiber misalignment in the bundle is 3-D random phenomenon and has statistically the same value in in-plane and out-of-plane directions.

The failure criterion (4) may be rewritten in form

\[
F_b(\sigma_c, \tau_{Lx}) = 1
\]  

(5)

where the failure function \( F_b \) is defined as

\[
F_b = \frac{\sigma_c}{\sigma_{cu}} + \frac{\tau_{Lx}}{k^*}
\]  

(6)

Here \( \sigma_{cu} \) is compressive strength in \( L \)-direction and \( k^* \) is shear strength of the bundle material. If for a given local stress combination in a certain position this function becomes equal or larger than 1 bundle plastic microbuckling takes place.

2.3.3.2. Matrix failure

The constraint due to bundle deformation and bending can lead to large stress concentrations in the surrounding matrix. High stress level may cause matrix yielding. For matrix yielding we apply the common von Mises criterion:

\[
\sigma_{EQV} = \sigma^\text{yield}
\]  

(7)

Here the equivalent stress is defined as

\[
\sigma_{EQV} = (\sigma_c^2 + \sigma_T^2 + \sigma_L^2 + 3\sigma_{Lx}^2)^{1/2}
\]  

(8)

Expression is simplified since the stress state is calculated using plane stress assumption.
Introducing yield function

\[ F_Y = \frac{\sigma_{11,y}}{\sigma_{11,\text{yield}}} \]  

(9)

the matrix failure criterion may be rewritten as

\[ F_Y = 1 \]  

(10)

3. Finite element analysis

The stress analysis was performed on the model in plane stress formulation by means of finite element method (FEM) using commercial package ANSYS 6.1. Standard 4-node elements PLANE42 were used, number of elements across the bundle was 10 and the same amount on each side of the bundle, which gave 30 elements across the whole unit. Number of elements in the longitudinal direction was chosen so, that element shape was approximately square. In all simulations compression is applied on the unit along the \( y \)-direction (vertical direction in this paper as shown in Fig. 5) by prescribing compressive strain of 1%.

3.1. Definition of the imperfect unit

We consider an imperfect unit in the NCF composite. This unit contains a bundle with an out-of-plane waviness described by sine function. In Fig. 5, \( y \) is the direction of the applied compressive displacement and \( x \) is the out-of-plane coordinate. The bundle structure in \( z \)-direction is ignored and plane stress analysis is used. This simplification is equal to the assumption that the bundles in units behind the curved bundle (in \( z \)-direction) have the same imperfection as the considered bundle and do not exhibit additional support to the bundle in the imperfect unit. Furthermore, these units are even smeared out in \( z \)-direction building a unit with imperfect layer.

![Fig. 5. Schematic showing of an imperfect unit used in calculations.](image)

Denoting the length of the unit in the \( y \)-direction by \( L \) the imperfect bundle has a shape which mid-plane is given by equation

\[ x = a \sin \frac{2\pi(y - L/4)}{L} \]  

(11)

Here \( a \) is the wave amplitude. The maximum angle (tangent at \( y = L/4 \)) is used to characterize the bundle waviness

\[ \theta \approx \tan \theta = \frac{a}{2L} \]  

(12)

From here we obtain the relationship between the amplitude and the maximum angle

\[ a = \frac{2L}{2\pi} \]  

(13)

In following height of the unit is assumed equal to one, \( L = 1 \).

Bundle misalignment angles 14°, 7° and 3° are considered in the following analysis.

The bundle size in the \( x \)-direction is \( h_b \) and the size of the unit in the same direction is \( h \).

Then the volume fraction of the bundle in the unit can be expressed as

\[ V_b = h_b/h \]  

(14)

3.2. Constituent properties for mesoscale analysis

Matrix is assumed to be isotropic with

\[ E_m = 3 \text{ GPa} \quad \nu_m = 0.4 \]

Properties of the transversely isotropic T700 fibers are

\[ E_1 = 256 \text{ GPa} \quad E_2 = 25.5 \text{ GPa} \quad G_{12} = 22.06 \text{ GPa} \quad G_{23} = 9.25 \text{ GPa} \quad \nu_{12} = 0.289 \quad \nu_{23} = 0.38 \]

Bundle properties (UD composite properties) have been calculated using the constituent properties given above and fiber volume fraction in the bundle \( V_f \) using simple engineering expressions (rule of mixtures and Halpin-Tsai expressions).

Results presented in this paper are for bundle volume fractions \( V_f = 0.5; 0.7 \) and 0.8.

Following values for bundle strength (estimated from well-aligned UD laminate tests) and matrix yielding (experimentally obtained) are used in failure initiation analysis:

\[ \sigma_{cu} = 1000 \text{ MPa} \quad k^* = 60 \text{ MPa} \quad \sigma^\text{yield(tension)} = 90 \text{ MPa} \quad \sigma^\text{yield(compression)} = 120 \text{ MPa} \]

Analyzing the constraint effect of surrounding layers on failure initiation we will use 3-D elastic properties of (45, -45)s laminate which are calculated using 3-D laminate
Here direction 1 coincides with y and direction 3 with x.

3.3. Boundary conditions

Boundary conditions on the vertical boundaries of the imperfect unit containing bundles with waviness, see Fig. 5, were formulated to reflect the real typical conditions in a NCF laminate containing several layers of fabric. The idea is illustrated in Fig. 6 for a case of periodic out-of-phase arrangement of units. Certainly the out-of-phase or in-phase arrangements of imperfect units are just two extreme examples. These units in the out-of-phase case are arranged in a row and number of those units is large enough to simulate symmetry condition in the middle of laminate.

Three different cases of boundary conditions are shown in a schematic drawing in Fig. 6. Most prone to buckling and with the lowest resistance to compressive load is a unit which is not supported by other units. In this case the boundaries are free (see Fig. 6a), this case is noted as F–F. A row of in-phase imperfect units would have similar behaviour.

Similarly, units belonging to the middle block of the ‘units-in-row’ model (see Fig. 6c) we model constraining both sides of the unit by symmetry conditions (notation S–S).

For the case of the surface unit of the ‘units-in-row’ model (see Fig. 6b) we assume that one edge is free and another is constrained by symmetry condition (noted as F–S). Boundary conditions with assigned notations are summarized in Table 1. The adequacy of these simplified boundary conditions was accessed comparing with results when the whole row of units is modelled. Particular cases when the boundaries of the imperfect unit are supported by perfectly oriented 0–90-bundles or sub-laminates were also modelled.

3.4. Methodology for failure initiation analysis

Due to considerable number of parameters to vary (boundary conditions, waviness angle, bundle properties, volume content of bundles in the unit) it is important to identify areas with stresses that are critical in terms of failure. In order to demonstrate how these areas were identified and how the stress distributions were used we consider in detail a unit with F-S boundary conditions and 14° of bundle waviness. The bundle material is No. 1 in Table 2 and the volume content of bundle \( V_f = 0.7 \). Similar analysis was performed for all boundary conditions but only one case is presented due to limited space.

The distribution of shear stress \( \tau_{12} \) in the unit is presented in Fig. 7. These results show that the highest shear stress values can be found in sections were the misalignment angle has maximum value. It should be noted that although different by sign, absolute value of the shear stress is symmetric with respect to the middle of the unit. Therefore, both areas were the misalignment angle is highest will contribute equally to the failure function. This means that one important location to analyse stresses is across the unit at the level of maximum misalignment angle (quarter length of the unit).

The distribution of transverse stress \( \sigma_T \) is given in Fig. 8. These data point out that the critical stress values are in the middle of the unit and thus, another essential location to examine stresses is defined.

### Table 1

Boundary conditions used for stress analysis

<table>
<thead>
<tr>
<th>Notation</th>
<th>Left boundary</th>
<th>Right boundary</th>
<th>Drawing</th>
</tr>
</thead>
<tbody>
<tr>
<td>F–S</td>
<td>Free</td>
<td>Symmetry</td>
<td>Fig. 6b</td>
</tr>
<tr>
<td>F–F</td>
<td>Free</td>
<td>Free</td>
<td>Fig. 6a</td>
</tr>
<tr>
<td>S–S</td>
<td>Symmetry</td>
<td>Symmetry</td>
<td>Fig. 6c</td>
</tr>
</tbody>
</table>

### Table 2

Bundle properties

<table>
<thead>
<tr>
<th>Property/ material</th>
<th>Nr 1: ( V_f = 50% )</th>
<th>Nr 2: ( V_f = 60% )</th>
<th>Nr 3: ( V_f = 52.5% )</th>
</tr>
</thead>
<tbody>
<tr>
<td>( E_1 ) (GPa)</td>
<td>129.9</td>
<td>155.3</td>
<td>136.2</td>
</tr>
<tr>
<td>( E_2 ) (GPa)</td>
<td>11.4</td>
<td>15.4</td>
<td>12.3</td>
</tr>
<tr>
<td>( \nu_{12} )</td>
<td>0.345</td>
<td>0.333</td>
<td>0.342</td>
</tr>
<tr>
<td>( G_{12} ) (GPa)</td>
<td>3.11</td>
<td>3.63</td>
<td>3.02</td>
</tr>
<tr>
<td>( \nu_{23} )</td>
<td>0.45</td>
<td>0.45</td>
<td>0.45</td>
</tr>
<tr>
<td>( G_{23} ) (GPa)</td>
<td>3.93</td>
<td>5.31</td>
<td>4.23</td>
</tr>
</tbody>
</table>
The distribution of longitudinal stress $\sigma_L$ is given in Fig. 9. Similarly to transverse stress, critical stress values are in the middle of the unit.

This overview of complete stress distribution allowed determining key locations for stresses and thus amount of analysed data can be drastically reduced.

In following analysis only stresses on the lines across the unit thickness ($x$-direction in Fig. 5) are considered to calculate failure function in the bundle $F_b$, see Eq. (6), and the failure function in the matrix $F_Y$, see Eq. (9).

Example of the failure function $F_b$ distribution is presented in Fig. 10 (only values in the bundle are shown). From these results follows that failure will occur at the level of maximum bundle misalignment angle ($Y=0.25$) where failure function $F_b$ reaches values of 2.75. It can also be concluded that the failure function has a maximum in the middle part of the bundle at the level $Y=0.25$ whereas at level $Y=0.5$, $F_b$ reaches a peek closer to the right boundary between bundle and matrix. This peek is due to bending forces which increases the compressive stress in this area.

In order to calculate failure function in the matrix, von Mises stresses are analysed, example of distribution of these stresses is presented in Fig. 11. Here maximum is in the middle of the unit.

Example of the failure function $F_Y$ distribution is presented in Fig. 12 (only values in the matrix, outside the bundle, are shown). These results clearly show that failure function $F_Y$ has its maximum of 1.60 in the middle of unit.
and, more important, it is lower than failure function in the bundle \( F_b \). It should be noted that this conclusion is valid for all analyzed cases and not only for this particular example. This means that failure always occurred in the bundle and not in the matrix.

4. Results of parametric study of compressive failure initiation

Since the performed calculations are based on linear elastic material behavior in the small deformation region, we can not simulate the final compressive failure of the NCF laminate. When we in following say ‘failure’ we mean failure initiation in the imperfect unit. It is very possible that in many cases the failure initiation in the imperfect unit will not be followed by collapse of the whole laminate containing also perfect units which can sustain higher load. The parametric study was performed in order to investigate influence of the following factors: constraint effects of surrounding material, fiber volume fraction \( V_f \) within bundle, fiber bundle volume fraction \( V_b \) in the composite unit and angle of misalignment.

Maximum value of failure function was calculated according to procedure described in previous section. However, for comparison between different cases it seems more sensible to present strain at failure rather than value of failure function. Therefore, additional procedure of failure function and strain scaling was performed before final result was obtained.

The scaling is possible because of the abovementioned linear material model and small strain assumption in these simulations. Thus all stresses can be simply scaled down by scaling down applied strain. The scaling procedure is the following:

1. arbitrary compressive strain \( \varepsilon_{\text{appl}} \) is applied (for simplicity 1% was used);
2. maximum value of the failure function \( F_b \) is obtained and scaled down (or up) to 1. Scaling coefficient is calculated as \( SC = 1/F_b \);
3. applied strain is scaled down (or up) according to scaling coefficient \( \varepsilon_{\text{fail}} = \varepsilon_{\text{appl}} SC \).

The strain \( \varepsilon_{\text{fail}} \) obtained by this method is the failure initiation strain in the imperfect unit under compression.

Since both failure functions \( F_b \) and \( F_y \) were always compared and \( F_y \) was always lower than \( F_b \), we conclude that in all studied cases plastic microbuckling was the mode of failure.

It should also be noted that in the analysis the bundle compressive strength \( \sigma_{cu} \) was assumed constant, independent on the fiber volume fraction in the bundle.

![Fig. 9. Longitudinal stress \( \sigma_L \) distribution (in local axes) in the unit.](image)

![Fig. 10. Distribution of failure function \( F_b \) (b.c. \( F-S \), misalignment angle of 14°).](image)
This assumption is consistent with the plastic microbuckling mechanism of compressive failure which is controlled by composite shear strength and fiber misalignment angle. These two parameters have a weak and unknown dependence on fiber volume fraction. However, assuming bundle strength independent of $V_f$ and recognizing that the bundle elastic modulus has a rule of mixtures dependence on fiber content, we obtain that the failure strain of a perfect unit decreases with an increasing fiber content in the bundle. For this reason the results of parametric analysis are in several cases expressed in terms of normalized strain, where normalization is performed with respect to strain of failure of a corresponding perfectly aligned unit.

4.1. Influence of bundle misalignment angle

In order to investigate influence of the bundle misalignment waviness, three angles were chosen $3^\circ$, $7^\circ$, $14^\circ$. Calculations were performed for different combinations of fiber volume fractions $V_f$ and bundle volume fractions $V_b$. Results are presented in Figs. 13–15. Strain at failure for laminate with waved bundle is normalized with respect to strain at failure of laminate containing perfectly aligned bundles. Strain at failure for perfectly aligned bundle is 0.77% for $V_f = 0.5$ and 0.64% for $V_f = 0.6$.

These results indicate that with increasing misalignment angle strain at failure is decreasing. This of course is somewhat expected from information available on the subject. However, these data give additional qualitative and quantitative information, such as: case when both edges of the unit with waved bundle are free ($F$–$F$) is the worst case scenario, compare to $F$–$S$ or $S$–$S$ cases. At certain misalignment (about $6^\circ$) strain at failure becomes insensitive to the misalignment angle for the case with $F$–$F$ boundary conditions. It is the opposite for units with ($F$–$S$) and ($S$–$S$) boundary conditions: the strength is rather insensitive at small misalignments and drops much faster when the angle is increasing.

Another interesting feature is that at larger misalignment angles all three lines corresponding to different boundary conditions start to converge indicating that shear stress component is dominating the bundle failure.

Increasing $V_f$ (compare Figs. 14 and 15) has a very marginal effect on the presented curves. Increasing $V_b$ (see Figs. 13 and 14) leads to larger reduction of the failure strain with the misalignment angle in case of ($F$–$S$) and ($S$–$S$) boundary conditions.

4.2. Influence of fiber volume fraction within the bundle

Calculation was performed for different values of fiber volume fraction $V_f$ within the bundle, for laminate unit with bundle volume content of 0.7 and waved bundle with misalignment angle of $7^\circ$. These results are presented in Fig. 16.
Fig. 16 (a) shows the effect on the failure strain in a normalized form. Obviously there is no noticeable increase or decrease. In contrast we see in Fig. 16 (b) that increasing fiber content reduces the failure strain in a non-normalization form. This indicates that the failure strain reduction with $V_f$ in Fig. 16(b) is caused by the decrease of the failure strain of a unit with perfectly aligned bundle and is not related to imperfect alignment. Strain at failure is decreasing with increasing fiber volume fraction within the bundle, independent of boundary conditions. Moreover, the rate of this decrease is approximately the same for all cases (lines are roughly parallel).

4.3. Influence of bundle volume fraction

The effect of bundle volume fraction $V_b$ on failure strain when all other parameters are kept unchanged is rather complex. Considering data for bundle misalignment angle $7^\circ$ in Figs. 13 and 14 and recalculating them to real strain we see that for all boundary conditions the failure strain decreases with $V_b$ change from 0.5 to 0.7. Failure strains are given in Fig. 17(a). The same effect is observed also at other misalignment angles. Results in Fig. 17(b) are for the same misalignment angle $7^\circ$ for fiber volume content within the bundle of 0.525 and for two rather high values of bundle volume fraction $V_c$. Apparently in this region there is no substantial effect of bundle content on strain at failure for ($F$–$F$) boundary conditions and only minor increase of strain at failure for $F$–$S$ boundary conditions. So, there appears to be a minimum in the failure strain versus bundle content curve.

4.4. Influence of total fiber volume fraction

Results from analysing influence of $V_b$ and $V_f$ does not really give an answer how the total fiber volume fraction affects strain at failure due to some contradiction in trends for $V_f$ and $V_b$. Indeed, with increase of $V_f$ failure strain decreases but with increase of $V_b$ there is first a decrease and then a slight increase in failure strain level. It is necessary to look at total fiber volume fraction which is obtained as a product of $V_b$ and $V_f$ to find out whether the total fiber volume fraction alone determines the failure strain and how much it depends on fiber dispersion parameters $V_f$ and $V_b$. These data are presented in Fig. 18. Generally speaking strain at failure indeed slightly decreases with increasing total fiber volume fraction. However, focusing on fiber volume fraction of 0.42 in this figure we see two values of strain corresponding to the same total fiber content: the lower failure strain value corresponds to case $V_b=0.7$, $V_f=0.5$ and the upper to case $V_b=0.8$, $V_f=0.525$. This result proves that fiber bundle structure has larger effect than the average (total) fiber content. Several practical
recommendations can be made based on this result. Composite with imperfect bundles where fibers are better dispersed can sustain higher compressive stresses. Laminate with larger fiber bundles and lower fiber content in bundle is more favourable in terms of compressive strength. This conclusion correlates very well with previous statements about prepreg laminates having higher compressive strength than NCF based composites.

4.5. Constraint effect of neighbouring layers

Some constraint effects expressed through three types of boundary conditions were already discussed above. But as it was mentioned earlier, boundary conditions $F-F$, $F-S$ and $S-S$ correspond to rather extreme cases ($F-F$) or very specific situations ($S-S$). More likely an imperfect unit with a waved bundle will be surrounded either by straight, perfectly aligned bundles or by other type of material. In order to investigate these, more realistic, constraint...
scenarios and to understand which of the three previously used boundary conditions may be applicable to describe them with sufficient accuracy, calculations were performed for configurations where the imperfect unit is supported by layers of different type of materials. Thickness of each support layer in $x$-direction was equal to the imperfect unit thickness. Summary of all these cases is given in Table 3.

When analyzing supporting layers which include $(G_{45})$ configurations these two layers were considered as a sublaminate with properties calculated using 3-D laminate theory, see Table 2. Results of calculations are presented in Fig. 19. These compressive failure initiation strain data actually validate our previous statements: unit without any support on the sides fails at lowest strain and the higher strain is sustained by the unit with symmetry conditions on both sides. Additional layers on the sides of the imperfect unit improve performance and show approximately the same failure strain level as $F-S$ boundary conditions. Softer material seems to be less supportive and does not contribute to performance as much as stiffer ones. It is shown by slightly lower strain at failure for the case when resin layer is used as surrounding media, compare to composite layers. However, when stiffer material is used as support it does not matter any more if it is UD ply, off-axes layer (angle-ply laminate) or cross-ply layers. In all those cases strain at failure is equal. This indicates that at certain level there is no performance gain for the imperfect unit with addition of more stiff materials as a surrounding media.

5. Model experiment

Parametric study clearly showed that the initial failure mode in imperfect units with misaligned bundles is plastic microbuckling in mixed compressive and shear loading. Increasing the misalignment angle from zero to $14^\circ$, the shear part in the failure function is increasing and finally the shear term is dominating the failure. The strain for failure initiation is decreasing with increasing bundle misalignment. To validate previous conclusions model experiment was designed. Compressive testing was performed on prepreg glass fiber/epoxy laminates with introduced out-of-plane waviness. Waviness was introduced by putting prepreg inserts of $0$-orientation between two other $0$-layers as schematically shown in Fig. 20. The insert thickness was equal to $6t_0$, where $t_0$ is prepreg layer thickness. Thus the total thickness of the imperfect unit was $16t_0$. It was supported by UD 0-laminate of thickness $7t_0$ from each side. Due to this arrangement resin pockets (regions with very low fiber content) were observed in several regions as shown in Fig. 20. Although distance $d$ was chosen to obtain predetermined misalignment angles ($\Phi=5^\circ$, $10^\circ$, $15^\circ$), values of those angles slightly varied from designed ones due to manufacturing procedure. Therefore, misalignment angle was measured from actual samples as the mean value of all angles within the distance $d$.

![Fig. 19. Effect of boundary conditions on strain at failure in compression.](image1.png)

![Table 3](image2.png)

<table>
<thead>
<tr>
<th>Notation</th>
<th>Left side of the unit</th>
<th>Right side of the unit</th>
</tr>
</thead>
<tbody>
<tr>
<td>R-R</td>
<td>Resin layer</td>
<td>Resin layer</td>
</tr>
<tr>
<td>UD-UD</td>
<td>Two 0-layers</td>
<td>Two 0-layers</td>
</tr>
<tr>
<td>CR-CR</td>
<td>0/90 layers</td>
<td>90/0 layers</td>
</tr>
<tr>
<td>AU-AU</td>
<td>0/±45 layers</td>
<td>0/±45 layer</td>
</tr>
<tr>
<td>AP-AP</td>
<td>Two ±45 layers</td>
<td>Two ±45 layers</td>
</tr>
</tbody>
</table>

![Fig. 20. Schematic picture of the laminate used in model experiment.](image3.png)
Compression tests were performed on 80 mm long specimens with end tabs (working zone 20 mm) by using in-house built test fixture which utilizes shear and end load transfer. Tests were performed in displacement controlled mode with loading rate of 1 mm/min. More detailed information on test procedures can be found in [18].

Results of the compression tests show that measured compressive strength indeed decreases with increasing misalignment angle (see Fig. 21). It should be noted that the presented stress is not the failure initiation stress. Samples were loaded until complete failure occurred, i.e. test was stopped only when load supported by specimen considerably dropped, almost to the zero level. The reduction of compressive strength is very substantial: it changes from approximately 750 MPa for 'perfectly' aligned laminate down to approximately 400 MPa for a misalignment angle of 15°.

Characteristic pictures of the specimens after the failure are presented in Fig. 22. The pictures for samples with introduced misalignment (Φ = 5°, 10°, 15°) suggest that failure mode is not necessarily in compression, it might be either splitting in the matrix region or plastic microbuckling and it is impossible to identify which is the first mode and the sequence of failure effects.

In order to separate and identify the initial failure mode, numerical analysis was performed on a simple model by means of FEM using ANSYS 6.1. Symmetric lay-up of the laminate in the model experiment allows simulating the whole area by using only quarter of the specimen and employing symmetry conditions. Standard PLANE42 4-node elements were used in calculations with plane stress formulation of the model. Sinus shape of the waviness was assumed. The average angle of misalignment was defined as \( \tan \Phi = 3v/d \). Schematic picture of the model used in FEM analysis is shown in Fig. 23. Composite properties used in simulation are given in Table 4, resin modulus \( E_z \) = 3.5 GPa and Poisson’s ratio \( v \) = 0.37 were also used. Similarly as for parametric analysis, 1% of strain was applied and scaling procedure was utilized to calculate strain at initiation of failure. Naturally, failure criteria for the curved bundle and matrix also were the same as in parametric study. However, for the laminate surrounding the curved bundle simple maximum stress criterion was employed.

In order to localize position at which failure will be initiated, distribution of the failure function is analysed in the whole model. In FEM calculations, the compressive...
and shear strength was set to 930 and 65 MPa, respectively. Plot of the failure function for bundle with 5° misalignment is presented in Fig. 24 as an example.

In all cases there is a local concentration of the failure function at the points where three different materials come in contact. These concentrations should be attributed to the general limitation of the FEM codes to deal with stress singularities and our meshing procedure at these points which does not insure smooth change of the slope of the 0-layer (in the curved bundle) from zero to nonzero. Recalculating large y-axis stress (horizontal direction in Fig. 20) to local axes with almost zero orientation may lead to numerical instability resulting in relatively larger shear stresses there. Another argument of importance is that the real geometry at these three material contact points is different than in model: the real geometry is without corners and without real contact; resin pockets still contain fibers, etc. Calculations performed with meshes of different sizes of the failure function $F_b$ close to the points of stress concentrations (see Fig. 25) show apparent differences in maximum values obtained using different size of elements. These data indicate singularity and prove that high values of failure function in the vicinity of border point between different materials are indeed artefacts related to assumed geometry and should be discarded in analysis.

Apart from these two high concentration points there is a large region where the failure function has large values. It is part where the fibre bundle has the largest waviness (see Fig. 24). We conclude that the initial failure always occurs in the curved bundle in the abovementioned region. The plot of the failure function $F_b$ within the misaligned bundle along its middle-line in Fig. 26 shows that the largest value is at the location were misalignment angle is highest. This value increases with increase of the misalignment angle. The scaling coefficient was used for the calculation of the compressive failure initiation strain in the model composite.

Experimental results now can be compared with prediction of failure strain ($\varepsilon_{f,0}$) obtained from FEM calculations. Both, experimental and numerical results are presented in Table 5. There are two sets of data obtained from experiment. One is strain when final failure occurred, it is denoted by $\varepsilon_{f,exp}$. Comparison shows that the experimental failure strain $\varepsilon_{f,exp}$ is higher than the numerically obtained values. This difference has its origin in differences in experimental procedure and numerical analysis: the experiment was not stopped at the stage of initiation of failure but rather at the complete failure of samples which is most likely splitting in the matrix region. One more reason may be that we use too low UD compressive strength in

### Table 4

<table>
<thead>
<tr>
<th>Property</th>
<th>$E_1$ (GPa)</th>
<th>$E_2$ (GPa)</th>
<th>$\nu_{12}$</th>
<th>$\nu_{23}$</th>
<th>$G_{12}$ (GPa)</th>
<th>$G_{13}$ (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$E_1$</td>
<td>42.5</td>
<td>13.7</td>
<td>0.291</td>
<td>0.42</td>
<td>5.83</td>
<td>4.82</td>
</tr>
</tbody>
</table>

Fig. 24. Failure function for the misalignment angle of 5°.

Fig. 25. Dependence of failure function maximum (near stress concentration) on element size.

Fig. 26. Plot of failure function within the curved bundle for different misalignment angles, along the bundle middle-line.
Failure mode Microbuckling Microbuckling Microbuckling

on this curve, it is denoted by failure strain with fiber content for perfectly aligned bundle. However, this change is determined by the reduction of the yield strength with increase of fiber content in the NCF composite layer $V_f$, from 50 to 60% at fixed volume content of the bundle.

Another experimental strain values presented in Table 5 are obtained from stress–strain curve at the point of ‘knee’ on this curve, it is denoted by $\varepsilon_f^0$. This ‘knee’ is thought to be first damage event that takes place (kink band formation and buckling). Thus, values of $\varepsilon_f^0$ should be compared with results of simulation, since both of those values indicated failure initiation. However, in order to verify these rather uncertain ‘knee’ strain data, test should be stopped as soon as first failure occurs and specimens must be examined under microscope.

Nevertheless, main trend shown by this comparison is the same as from previous parametric investigation, failure strain decreases with increasing misalignment angle. Moreover, ratio between strains for different misalignment angles ($\varepsilon_f^5 : \varepsilon_f^10 : \varepsilon_f^15$) is in reasonable agreement for experimental values for $\varepsilon_f^0$, and simulation. From experiment we obtain $1:0.67:0.65$ as compared with $1:0.55:0.36$ in simulation.

### 6. Summary and conclusions

The failure initiation analysis under compressive loading in non-crimp-fabric composites using FEM in plane stress and linear elastic formulation was aimed to study the effect of bundle out-of-plane orientation imperfection (described by a sine function), fiber volume fraction inside the bundle and bundle volume fraction inside the composite unit. The support effect of the neighbouring material was analyzed varying boundary conditions and solving cases with particular lay-up of surrounding material.

Failure initiation strain was determined comparing failure functions corresponding to two alternative failure mechanisms: (a) plastic microbuckling in bundle due to mixed compressive and shear load; (b) plastic matrix yielding according to von Mises criterion.

It was found that for all parameter and boundary condition combinations the strain at failure initiation decreases several times with change of the imperfection angle between 0° and 14°.

The increase of fiber volume fraction in the bundle $V_f$ from 50 to 60% at fixed volume content of the bundle $V_b$ led to about 15% reduction of failure initiation strain. However, this change is determined by the reduction of the failure strain with fiber content for perfectly aligned bundle and the failure initiation strain in normalized form does not depend on $V_f$. The bundle volume fraction increase at fixed $V_f$ has a complex effect on failure: initiation strain is significantly decreasing in $V_f$ region between 50 and 70% but for higher values of $V_f$ the strain is stabilized and even slightly increasing. It was found that the total (average) fiber content in the NCF composite layer $V_f/V_b$ alone cannot explain the strain at failure initiation because much larger failure initiation strain was found at the same total volume fraction but for better dispersed fibers (larger $V_b$).

Analysing the support of surrounding media to failure initiation it was found that the result is rather insensitive to what type of material (90-45-0-layers) is surrounding the imperfect unit. The results are best described by idealized $(F-S)$ boundary conditions which are between $(F-F)$ and $(S-S)$ cases but closer to the latter.

Experiments on designed model composite with introduced out-of-plane layer misalignment confirmed the above conclusions and were in reasonable agreement with simulation results for the model configuration.

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### References

Paper C
Average strain in fiber bundles and its effect on NCF composite stiffness

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Abstract

Transverse strain in bundles governs transverse cracking in NCF composites. FE analysis shows that this strain may be significantly lower than the applied macroscopic strain component in the same direction. This feature is important for damage evolution modelling. The iso-strain assumption which in different combinations is widely used in stiffness models is inadequate because the strain in different meso-elements (bundles of different orientation and matrix regions) is assumed the same. Analysing by FEM the importance of media surrounding the bundle on average transverse strain it was found that an increasing ratio of the bundle transverse stiffness to the matrix stiffness leads to a decrease of the strain in the bundle. An increase of the stiffness in the same direction in adjacent layers leads to an increase of the transverse strain in the bundle. Higher bundle volume fraction in the layer leads to larger transverse strain in the bundle. These trends are described by a power law and used to predict the average strain in bundles. The calculated H-matrix which establishes the relationship between strains in meso-element and RVE strains is used to calculate the “effective stiffness” of the bundle. This “effective stiffness” is the main element in simple but exact expressions derived to calculate the stiffness matrix of NCF composites. Considering a 3-D FE model as the reference, it was found that all homogenization methods used in this study have sufficient accuracy for stiffness calculations, but only the presented method gives reliable predictions of strains in bundles.

Keywords: Non-Crimp fabric composites, micromechanics, constitutive modelling, FEM.
1 Introduction

Non-Crimp fabric (NCF) composites are manufactured from textile preforms consisting of layers of fiber bundles with a certain orientation assembled by warp-knitted threads [1]. This production technique allows for substantial reductions in production costs compared to pre-preg tape based materials. In addition to economical benefits, authors have reported improvements in damage tolerance as well as out-of-plane fracture toughness [2-4]. In the manufacturing process, the reinforcement (preform) is infiltrated by a resin, creating a NCF composite which is heterogeneous not only on micro-scale as for pre-preg based composites but also on meso-scale. The heterogeneities described imply that a multi-scale homogenization of the composite is desirable in order to perform a simplified stress and strain analysis of the NCF composite.

The architecture of NCFs described above shows similarities to woven fabric composites and therefore methods and theoretical models used for woven fabric composites [5] are relevant also for NCF composites. Ishikawa and Chou [6, 7] proposed two models for woven composites named the mosaic and fiber undulation model. These models used the idea of dividing the representative volume element (RVE) into infinitesimal strips where classical lamination theory (CLT) can be used to calculate the properties of the strip. The mosaic model disregarded the waviness of the yarn which is always present in woven fabrics whereas the fiber undulation model included the waviness. Two-scale homogenization of the RVE is performed, first creating a homogenized material in each bundle of the composite (micro-scale homogenization). The constitutive equation of the RVE can then be derived by dividing the RVE into elements where each element in the composite has a stiffness matrix determined by its meso-elements (bundle and matrix regions). Similar methods with different complexity to obtain the stiffness matrix were used in [8-10].

A limited number of analytical models deal with mechanical properties of NCF composites. Wang et al. [11] and Bibo et al. [12], given the similarity of NCF and prepreg material in terms of virtually straight in-plane yarns, used CLT in order to predict the stiffness of NCF composites. Chun et al. [13] studied the mechanical properties of multi-axial warp knitted fabric composites using iso-strain assumption in the meso-elements. The stitching tread was included in the analysis. Robitaille et al. [14] presented a set of equations accounting for the geometry of NCF composites assembled by a warp knitting procedure which was previously developed for woven and braided fabrics. This set of equations predicts the properties of the NCF
laminate using the manufacturing parameters as input. The method used by Klopp et al [15] is based on a reduced volume fraction of fiber and matrix due to the distortion of fiber and matrix created by the stitching tread. The reduced volume fraction was then used together with CLT to predict the mechanical properties of the laminate.

Theoretically, NCF composites consist of straight fiber bundles where the size of each bundle is determined by the stitching procedure. However, due to the complex manufacturing technique, NCF composites have both in- and out-of-plane waviness of the bundles. Recently, Edgren et al [16] developed a model which accounts for the NCF composites stiffness reduction due to out-of-plane waviness of the tows. The reduction is described by a knock-down factor to longitudinal modulus, derived using Timoshenko beam theory.

Very little attention has been paid to micro-damage development in NCF composites, where transverse intra-bundle cracking is the first mode [17]. The evaluation of the accompanied stiffness reduction using model developed for laminates was performed in [17]. However, damage evolution analysis requires detailed knowledge of the strain in the bundle which depends on the meso-geometry and on the elastic properties of meso-elements which is not available at present. The average bundle strain also determines the contribution of the bundle to the stiffness of the composite.

The objective of the present study is to develop a reliable and accurate model to determine a) average strains in fiber bundles and other meso-elements; b) to calculate NCF composite stiffness with a high accuracy.

3-D FE results will be used to evaluate the models. Simple analytical models based on “smearing-out” techniques or on iso-strain assumptions for all meso-elements or for meso-elements belonging to one “super-element” will also be inspected. Exact closed form expressions for NCF composite stiffness elements will be presented containing as an input the “effective stiffness” of the bundle which is uniquely related to the average strain in the bundle as compared to the RVE strain. Parametric studies of the average bundle strain are performed, analysing the importance of the media surrounding the bundle; the stiffness ratio between the meso-elements in the same layer and in the neighbouring layers and the volume content of the bundle in the layer. This study has resulted in a power law expression for the transverse strain in the bundle.

The main focus in this paper is on normal strains and normal stiffness elements and only a limited amount of results for shear properties is presented.
2 Models representing NCF composites

2.1 Structure of the NCF composite

NCF composites are heterogeneous materials with a layered structure; see an example in Fig. 1, each layer consisting of theoretically straight fiber bundles with distinct orientation and matrix cylinders parallel to them with a complex cross-section. In following sections, these constituents of a layer are called “meso-elements”. Each bundle is characterized by the fiber volume fraction $V_f$. The average volume fraction of fibers in a layer $V_f^a$ is defined by $V_f^a = V_f V_b$, were $V_b$ is the volume fraction of bundles in the layer.

Figure 1. 3-D model of a NCF composite with lay-up [0, 90].

An accurate prediction of the NCF composite stiffness and determination of strain in bundles needed for damage analysis is a complex task. Since the heterogeneities of a bundle and of a layer have different length scales the analysis may be significantly simplified performing homogenization separately on both length scales:

- Micro-scale homogenization is over the fiber and matrix micro structure within the bundle. It is performed using the Composite Cylinder Assemblage model (CCA model) developed by Z. Hashin [18, 19], which is one of the most accurate for UD composite stiffness. The
result is bundle mechanical properties described as homogeneous material.
- Homogenization over the meso structure of the RVE which consists of meso-elements like bundles of different orientation and matrix regions. Macroscopic constitutive properties of the NCF composite are obtained.

The representations of the NCF composite analyzed in this paper are described below.

### 2.1.1 Laminate approximation of the NCF composite

In this model the micro- and meso-structure is smeared-out and the effective properties of a layer are calculated using the average fiber content of the layer $\nu_f$ in the CCA model. This implies that the non-uniformity of the distribution of fibers in the layers (bundle structure) only has a secondary effect and we are replacing the heterogeneous structure by a homogeneous material as shown in Fig. 2 a). The obtained “effective laminate” is a RVE which is treated by CLT. Due to iso-strain assumption in CLT the calculated strains in all layers will be equal to the macroscopically applied strain. It is not possible to calculate the real strains in the meso-elements.

![Figure 2](image.png)

Figure 2. Representation of NCF composite layer: a) homogenization in a layer; b) simplification of the meso-scale geometry

### 2.2 NCF composite models with bundle structure

#### 2.2.1 3-D model of the RVE

In order to simplify the analysis, the real meso-geometry (“quasi-elliptical” shape of the bundle) is replaced by a rectangular bundle containing the same
$V_f$ and bundle content $V_b$, see Fig. 2b. Certainly this procedure effects the stress distribution close to the corners and interfaces but it has less significance when average strain in a bundle and composite stiffness are analyzed. After this simplification the NCF composite in Fig. 1 can then be divided in meso-elements as shown in Fig. 3a). Observing the composite from the top we can recognize 9 “super-elements” (see Fig. 3c)) each of them consisting of 4 meso-elements according to Fig 3 b).

Figure 3. The schematics of $[0,90]_s$ NCF composite meso-scale structure: a) simplified RVE; b) a “super-element”; c) 2-D model consisting of 9 “super-elements”.
Model with some homogenized layers

In this slightly simplified model some layers of the NCF composite (usually layers which are not in the focus of the investigation) are replaced by a homogeneous material as described in Section 2.1. The bundle structure of layers subjected to detailed analysis is retained. As an example, the 0°-layer of the [0, 90]s NCF composite may be homogenized retaining the bundle structure in the 90°-layer only. It is expected to introduce a negligible inaccuracy when stiffness in 0°-direction and 90°-bundle strains are analyzed. In this example the NCF composite may be divided into three “super-elements” containing 4 meso-elements each. The analysis becomes two-dimensional.

3 Stiffness of NCF composites

Meso-scale models differ by the way in which the (average) strain in the k-th meso-element is related to the strain applied to the RVE. The relationship may be written in the local axes of the meso-element as

$$\varepsilon^0_k = [H]_k [\varepsilon]_{\text{RVE}}$$

and the $H$-matrix depends on the RVE meso-structure and constituent properties.

3.1 Exact expressions for NCF composite in-plane stiffness

In this section we will derive an exact expression for stiffness matrix $[\mathbf{Q}_{\text{RVE}}]$ which defines the stress-strain relationship of the NCF composite (Voigt notation)

$$\{\sigma\}_{\text{RVE}} = [\mathbf{Q}_{\text{RVE}}] \{\varepsilon\}_{\text{RVE}} \quad (1)$$

Several types of meso-elements (matrix regions of different geometries, bundles of several orientations etc) may be recognized in the RVE. The k-th meso-element is characterized by its volume fraction $V_k$ in the RVE, geometry, material stiffness matrix $[\mathbf{Q}_k]$ in symmetry axes and orientation angle $\theta_k$. They define the stress-strain relationship for a meso-element in the global axes

$$\{\sigma\}_k = [\mathbf{Q}_k] \{\varepsilon\}_k \quad (2)$$
The line above the stiffness matrix $[\mathbf{D}_k]$, and other variables related to the meso-element indicates that the expressions are written in global coordinates. We introduce volume averaged stresses and strains in the following way

$$
\{\sigma\}^a = \frac{1}{V} \int \{\sigma\} dV \quad \{\varepsilon\}^a = \frac{1}{V} \int \{\varepsilon\} dV
$$

(3)

Superscript $a$ denotes volume average. The averaged stress-strain relationship of the $k$-th meso-element has the same form as in an arbitrary point

$$
\{\sigma\}^a_k = [\mathbf{D}_k] \{\varepsilon\}^a_k
$$

(4)

It has been shown using the divergence theorem [20] that the volume averaged stress over the whole RVE is equal to the boundary averaged stress $\Sigma_{ij}^{RVE}$ defined below.

$$
\Sigma_{ij}^{RVE} = \frac{1}{V_{RVE}} \int_{S_{RVE}} \sigma_{ik} n_k x_j dS
$$

(5)

Here $S_{RVE}$ is the external boundary of the RVE. It can be shown that $\Sigma_{ij}^{RVE}$ is equal to the average stress applied to the RVE boundary $\{\sigma\}_{RVE}$. One can show [20] that boundary averaged strain introduced as

$$
\varepsilon_{ij}^{ba} = \frac{1}{V_{S_x}} \frac{1}{2} \int_{S_x} \left( u_i n_j + u_j n_i \right) dS
$$

(6)

is equal to the volume averaged strain. In this case, $S_E$ is the external boundary of the volume. Hence, we can write for the $k$-th meso-element

$$
\{\varepsilon\}^a_k = \{\varepsilon\}_{k}^{ba}
$$

(7)

The boundary averaged strain is a meso-scale strain in the meso-element defined by the deformation of its outer boundaries and it is easier to determine using FE than the volume average. If the meso-element is a layer in a laminate this strain is equal to the strain applied to the laminate (RVE
strain). In a general case the strain defined by displacements at the meso-element boundary is not equal to the RVE strain but it is still uniquely related to it. We can write this linear relationship in the local axes of the meso-element as

\[
\{\epsilon\}^a_{ba} = [H]_{kk} \{\tilde{\epsilon}\}_{RVE}
\]  

(8)

The unknown matrix \(H_k\) depends on RVE mesoscale morphology and constituent properties. In Eq. (8) \(\{\tilde{\epsilon}\}_{RVE}\) is the RVE strain transformed to the local axes of the meso-element. Expressing \(\{\tilde{\epsilon}\}_{RVE}\) in Eq. (8) through the RVE strain in the global axes and transforming Eq. (8) to the global system using strain transformation expressions we obtain

\[
\{\epsilon\}_{RVE} = [T] \{\tilde{\epsilon}\}_{RVE} \quad \Rightarrow \quad \{\epsilon\}^a_{ba} = [T] [H]_{kk} [T]^T \{\epsilon\}_{RVE}
\]  

(9)

Here \([T]\) is the stress transformation matrix from global to local axes used in CLT. We start the RVE stiffness derivation by expressing the stress vector applied to the RVE, \(\{\sigma\}_{RVE}\) through the average stress in the RVE and expressing the volume integral as a sum of volume integrals over meso-elements.

\[
\{\sigma\}_{RVE} = \{\sigma\}^a_{RVE} = \sum_k V_k \{\sigma\}^a_{kk}
\]

(10)

Here \(V_k\) is the volume fraction of the meso-element in the RVE. The average stress in the meso-element in Eq. (10) may be replaced by Eq. (4) in which the volume average strain in meso-element is expressed through Eq. (7). Using Eq. (9) we finally obtain

\[
\{\sigma\}_{RVE} = \sum_k V_k \{\tilde{\sigma}\}_{kk} \{T\} \{H\} \{T\}^T \{\epsilon\}_{RVE}
\]

(11)

From Eq. (11) follows the expression for stiffness matrix of NCF composite

\[
[Q_{0}]_{RVE} = \sum_k V_k \{\tilde{\sigma}\}_{kk} \{T\} \{H\} \{T\}^T
\]

(12)
Equation (12) can be rewritten as

\[
[\mathcal{Q}_{0}]_{RVE} = \sum_{k} V_k \left[ T \right]_{k}^T \mathcal{Q}_k \left[ H \right]_{k} \left( T \right)_{k}^{-1} \tag{13}
\]

We introduce the “effective stiffness” of the sub-element as

\[
[\mathcal{Q}_H]_k = [\mathcal{Q}]_k [H]_k \tag{14}
\]

Using it in Eq. (12) we see that the RVE stiffness may be calculated using the rule of mixtures

\[
[\mathcal{Q}_{0,RVE}] = \sum_{k} V_k [\mathcal{Q}_H]_k \tag{15}
\]

Obviously, an accurate determination of \([H]_k\) is a key issue for strain calculation in the meso-element and for NCF composite stiffness. The determination of \([H]_k\) can be both approximate analytical and numerical and some examples of simplified approaches are described in Section 3.3. However, the most efficient in our opinion is the approach presented in Section 4 where the expression for \([H]_k\) is developed based on results of FE parametric analysis.

3.2 Form of the \([H]_k\) matrix for cylindrical meso-elements

The \([H]_k\) -matrix defines the relationship between k-th meso-element boundary averaged strains and the strains applied to the RVE. In the local system of coordinates coupled with the material symmetry or with the local geometry Eq. (8) has the form

\[
\begin{bmatrix}
\varepsilon_{L}^b \\
\varepsilon_{T} \\
\gamma_{LT}^b \\
\end{bmatrix}
= 
\begin{bmatrix}
H_{11} & H_{12} & 0 \\
H_{21} & H_{22} & 0 \\
0 & 0 & H_{66} \\
\end{bmatrix}_k
\begin{bmatrix}
\varepsilon_{L} \\
\varepsilon_{T} \\
\gamma_{LT}^b \\
\end{bmatrix}_{RVE} \tag{16}
\]

The exact dependence of \(H_{ij}\) on geometrical and stiffness parameters comes from solution of a complex 3-D problem. We will show here that in
case of cylindrical meso-elements the number of unknown elements in this matrix can be reduced and they may be easily determined.

**Laminate.** In case of laminate the meso-element is a unidirectional layer with a certain orientation. In this case the strains applied to the meso-element boundary are equal to the strain applied to the RVE (isostrain assumption in the CLT). Obviously the \( [H] \) -matrix has the form

\[
[H] = \begin{bmatrix}
1 & 0 & 0 \\
0 & 1 & 0 \\
0 & 0 & 1 \\
\end{bmatrix}
\]  

(17)

**Cylindrical meso-elements: Fiber bundles, matrix regions parallel to them.** We start the analysis assuming that only one global strain component is applied to the RVE and that the element has a form of a cylinder (with an axis in the L-direction) crossing the whole RVE.

1. Applying only \( \varepsilon_T \) to the RVE, the average strain in the L-direction in the element is equal to this strain component applied to the RVE (it is zero). Hence, \( H_{12} = 0 \). The average strain in the T-direction in the meso-element is a complex function of all parameters and is unknown at the present stage. We denote it \( H_T \).

2. If only \( \varepsilon_L \) is applied to the RVE, the strain in the meso-element in the L-direction is equal to the applied strain, \( H_{11} = 1 \). Examples are longitudinal strain in the bundle and the same strain component in the cylindrical matrix region located between bundles and with axis parallel to them. Even if the RVE transverse strain is zero, the transverse strain in the bundle is nonzero because of different Poisson’s ratios of the bundle and the matrix. It would be zero only if the matrix is not present in the RVE, which means that the Poisson’s contraction strain of the bundle would be compensated by external transverse strain \( \nu_{12} \varepsilon_L^{RVE} \) (\( \nu_{12} \) is the Poisson’s ratio of the meso-element). In the presence of matrix region this compensating strain is \( H_T \nu_{12} \varepsilon_L^{RVE} \) and the difference with the above builds the transverse strain experienced by the meso-element.

The above analysis leads to the following H-matrix and the “effective stiffness matrix \( Q_H \).
\[
[H] = \begin{bmatrix}
1 & 0 & 0 \\
(H_T - 1)\nu_{12} & H_T & 0 \\
0 & 0 & H_{LT}
\end{bmatrix}
\]  
(18)

\[
[Q_H] = \begin{bmatrix}
Q_{11} + Q_{12}\nu_{12}(H_T - 1) & Q_{12}H_T & 0 \\
Q_{12}H_T & Q_{22}H_T & 0 \\
0 & 0 & Q_{66}H_{LT}
\end{bmatrix}
\]  
(19)

As an example for application of Eq. (11) we consider the upper half (due to symmetry) of the cross-ply NCF composite shown in Fig. 3 and focus on \( Q_{xx}^{RVE} \) and \( Q_{xy}^{RVE} \). For simplicity we consider a special case when the mesostructure of the 90°- and 0°-layer is the same, \( V_b = V_{90} = V_{b0} \). From Eq. (15) neglecting higher order terms with respect to Poisson’s ratio we obtain

\[
Q_{xx}^{RVE} = V_1 Q_{22}^b H_T^b + V_1 Q_{11}^b + V_2 Q_{11}^m + V_2 Q_{11}^m H_T^m
\]
\[
Q_{xy}^{RVE} = 2V_1 Q_{12}^b H_T^b + 2V_2 Q_{12}^m H_T^m
\]  
(20)

Indexes \( b \) and \( m \) denote bundle and matrix. Here \( V_1 \) and \( V_2 \) is the volume fraction of a bundle and matrix region of one given orientation in the RVE, respectively. In the considered case \( V_1 = \frac{1}{2} V_b \). The unknown \( H_T^b \) and \( H_T^m \) are related

\[
H_T^m = \frac{1 - L_{90n} H_T^b}{L_{Mn}}
\]  
(21)

Here \( L_{90n} \) and \( L_{Mn} \) is the size in the x-direction of the 90°-bundle and the matrix region, normalized with respect to \( L_M + L_{90} \), see Fig. 4.

3.3 Simplified models based on assumed strains in meso-elements

A. Iso-strain approach. In this simple approach which is often applied to woven composites, the in-plane strain components of all meso-elements are assumed to be equal and, hence, identical to the applied strain. This approximation may lead to rather good results if the composite contains
many layers randomly shifted in horizontal directions [21, 22]. This implies that $[H]_k$ in Eq. (13) are an identity matrix and the expression for NCF composite stiffness turns to $[Q]_{RVE} = \sum_{k=1}^{K} V_k [\bar{Q}]_k$ which is just a rule of mixtures.

B. Through-the-thickness homogenization and partial iso-strain model

In this approach the RVE is considered as a 2-D structure of “super-elements” as shown in Fig. 3. Each “super-element” contains meso-elements stacked in a certain sequence in the thickness direction. Homogenization through the thickness is performed in each “super-element” to obtain its stiffness matrix, assuming that in-plane strains in all meso-elements in the “super-element” are equal (iso-strain condition). This assumption means that CLT can be used to calculate the stiffness of a “super-element”. In result the NCF composite is reduced to a 2-D structure divided into “super-elements” according to Fig. 3c and each “super-element” is a homogeneous material. In order to use Eq. (15) for NCF composite stiffness we have to find the $[H]_k$-matrix. In other words, we need to determine the strain in each “super-element”. It can be done a) numerically using FEM in 2-D formulation referred as “2D FE”; b) using approximate analytical solution called “partial iso-strain model”. In the partial iso-strain model the 2-D structure is divided in “rows” and “columns” which we call “vectors”. The strain transverse to the “vector” is assumed the same in all “super-elements” belonging to this “vector” but it is different than in other “vectors”. The magnitude of the strain depends on the resultant stiffness of the row and is a result of solution. This means that straight lines in the 2-D structure remain straight after deformation. Additionally conditions are: a) the total deformation in a certain direction is equal to the sum of transverse deformations of “vectors”; b) global equilibrium is satisfied in any direction and in any cross-section.

4 Effect of meso-parameters on average transverse strain in bundles

4.1 Material properties

The parametric study was performed varying elastic properties and meso-structure to cover typical values for both carbon and glass fiber composites. Hashin’s Composite Assemblage model was used to calculate the material
properties of homogenized bundles and layers for these materials. The volume fraction of fibers within a bundle \( V_f \) was set to 0,55. The calculated material properties for carbon and glass fiber laminates can be found in Table 1. In the glass fiber case homogenized layers were not considered.

Table 1. Material properties used in the numerical analysis (GPa).

<table>
<thead>
<tr>
<th></th>
<th>Carbon fiber</th>
<th>Glass fiber</th>
<th>Matrix</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fiber bundle</td>
<td>( V_f = 0.55 )</td>
<td>( V_f = 0.55 )</td>
<td>4.2 FE modelling</td>
</tr>
<tr>
<td>( V_{fib} )</td>
<td>( V_{fib} = 0.275 )</td>
<td>( V_{fib} = 0.44 )</td>
<td></td>
</tr>
<tr>
<td>( V_{hom} )</td>
<td>( V_{hom} = 0.5 )</td>
<td>( V_{hom} = 0.8 )</td>
<td></td>
</tr>
</tbody>
</table>

| \( E_L \)      | 230          | 20           | 13 |
| \( E_T \)      | 30           | 30,36        | 30 |
| \( G_{LT} \)   | 20           | 12           | 12 |
| \( G_{TT} \)   | 12           | 20,77        | 12 |
| \( v_{LT} \)   | 0,25         | 0,269        | 0,25 |
| \( v_{TT} \)   | 0,25         | 0,407        | 0,25 |

**4.2 FE modelling**

Parametric studies with focus on the transverse strain in the 90º-bundle were performed using a 2-D FE model. A schematic picture of the model is presented in Fig. 4. In all cases, the thickness ratio of the layers was the same (\( t_0 = t_{90} = 1 \)). The commercial FE program Ansys 7.1 with Plane82 elements was used for the numerical analysis. Plane82 is a two dimensional four node structural solid element with three degrees of freedom at each node. The analysis was performed using plane stress formulation. Boundary conditions are shown in Fig. 4. Applied strain in x-direction, \( \varepsilon_{xRVE} \) was 1% and the ratio \( (L_{90} + L_{M})/t_{90} = 10 \) was constant. The upper surface was free or with coupled nodes. The number of elements varied between 500-2040 depending on the size of the matrix region. Average strain in the 90º-bundle \( \varepsilon_{x90} \) was calculated as \( \varepsilon_{x90} = \Delta L_{90}/L_{90} \), where \( \Delta L_{90} \) is the average x-displacement of the nodes at the interface between 90º-bundle and matrix.
Figure 4. 2-D FE model used for the parametric study of interface distortion between 90°-bundle and matrix.

A comparison was made between the free upper boundary condition and the case where the nodes at the upper surface are coupled, having the same displacement in z-direction. This boundary condition simulates periodic and symmetric lay-up of RVE's in the thickness direction. Both Carbon and Glass fiber composites with different geometrical parameters were analyzed. As can be seen in Table 2, the strain in the 90°-bundle is increasing when additional constraint in form of coupling in z-direction is used. The coupling of the nodes prevents the laminate from local bending and leads to smaller distortion of the interface. However, the effect on the average strain $\varepsilon_{z}^{90}$ is rather small and is not influencing the conclusions in the following section to any significant extent. Therefore, the model without coupling will in the following be employed.

Table 2. $[0\ 90]_s$, CF=Carbon fiber, GF=Glass fiber.

<table>
<thead>
<tr>
<th>Mat</th>
<th>L₀₀</th>
<th>Lₘ</th>
<th>$E₀$ (GPa)</th>
<th>$E₉₀$ (GPa)</th>
<th>$\varepsilon_{z}^{90}$ (free nodes)</th>
<th>$\varepsilon_{z}^{90}$ (coupled nodes)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CF</td>
<td>8</td>
<td>2</td>
<td>128,1</td>
<td>10,36</td>
<td>0,92</td>
<td>0,927</td>
</tr>
<tr>
<td></td>
<td>5</td>
<td>5</td>
<td>128,1</td>
<td>10,36</td>
<td>0,866</td>
<td>0,874</td>
</tr>
<tr>
<td>GF</td>
<td>8</td>
<td>2</td>
<td>43,4</td>
<td>12,61</td>
<td>0,882</td>
<td>0,894</td>
</tr>
<tr>
<td></td>
<td>5</td>
<td>5</td>
<td>43,4</td>
<td>12,61</td>
<td>0,797</td>
<td>0,813</td>
</tr>
</tbody>
</table>

$E₀$, $E₉₀$ and $Eₘ$ are the modulus of the 0°-, 90°- bundles and matrix region respectively. A simple analytical model was also employed which predicts the 90°-bundle strain for different thickness of the layers ($t₉₀$) of the RVE. This model uses the 90°C-bundle strain for the case when
(L_{90}+L_{M})/t_{90}=10 \) (reference case) calculated with FEM, and the difference in thickness of the layers between the reference case and the studied RVE. The length of the 90°-bundle and matrix region \((L_{90} \text{ and } L_{M})\) are constant.

\[
\varepsilon_{x90}^{ba}\left|_{(L_{90}+L_{M})/t_{90}=\varphi} = \varepsilon_{xRVE} - \left(\varepsilon_{xRVE} - \varepsilon_{x90}^{ba}\right)\left|_{(L_{90}+L_{M})/t_{90}=10}\right) \left(\frac{10}{\varphi}\right)\right. \quad (22)
\]

Here \(\varepsilon_{xRVE} = 1\%\) for all cases. As can be seen in Table 3, the strain in the 90°-bundle is predicted with rather high accuracy in the carbon fiber case for both \((L_{90}+L_{M})/t_{90}=8\) and \((L_{90}+L_{M})/t_{90}=15\). However, for glass fibers, the accuracy is not as good due to lower stiffness of the layers.

Table 3. \([0\ 90]_s\), CF=Carbon fiber, GF=Glass fiber. Effect of layer thickness on 90°-bundle strain.

<table>
<thead>
<tr>
<th>Mat</th>
<th>L_{90}</th>
<th>L_{M}</th>
<th>E_{90} \text{ (GPa)}</th>
<th>E_{90} \text{ (GPa)}</th>
<th>\varepsilon_{x90}^{ba} \text{ (free nodes)}</th>
<th>\varepsilon_{x90}^{ba} \text{ (free nodes)}</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>(L_{90}+L_{M})/t_{90}=8\ FEM</td>
<td>(L_{90}+L_{M})/t_{90}=10\ FEM</td>
</tr>
<tr>
<td>CF</td>
<td>5</td>
<td>5</td>
<td>128.1</td>
<td>10.36</td>
<td>0.839</td>
<td>0.833</td>
</tr>
<tr>
<td>GF</td>
<td>5</td>
<td>5</td>
<td>43.4</td>
<td>12.61</td>
<td>0.769</td>
<td>0.746</td>
</tr>
</tbody>
</table>

4.3 Average strain in the bundle: power law

It was found that Poisson’s ratio and shear moduli of constituents have negligible effect on results and the main parameters affecting the average transverse strain in the fiber bundle are

A) Volume fraction of the bundle in the layer characterized by the parameter \(\frac{V_{90}}{1-V_{90}} = \frac{L_{90}}{L_{M}}\)

B) Stiffness ratio of the constraint layer and the transverse bundle \(\frac{E_0}{E_{90}}\)

C) Matrix and 90°-bundle stiffness ratio \(\frac{E_M}{E_{90}}\)

The influence of these parameters on 90°-bundle transverse strain was analysed and a superposition of these separate contributions is described in a simple model. From Fig. 5 a) follows that when the volume fraction of 90°-
bundle increases, the strain in the 90°-bundle $\varepsilon_{90}^{ba}$ also increases and asymptotically approaches to the value of the applied strain to the RVE. Presenting $\varepsilon_{xRVE} - \varepsilon_{x90}^{ba}$ as a function of $\frac{V_{90}}{1-V_{90}}$ in logarithmic axes, see Fig. 5 b), the strain dependence on volume fraction of 90°-bundle approximately forms a straight line. We conclude that the $\varepsilon_{x90}^{ba}$ dependence on bundle volume fraction in the layer can be described by a power law

$$\frac{\varepsilon_{x90}^{ba}}{\varepsilon_{xRVE}} = 1 + B \left( \frac{V_{90}}{1-V_{90}} \right)^{k_1}$$

(23)

where $k_1$ is the slope of the curve in Fig. 5 b). Presenting in logarithmic axes the bundle strain as a function of $\frac{E_0}{E_{90}}$, see Fig. 5 c) we obtain similar conclusion: the dependence is linear when using a constant $A_2$ according to figure 5 c) and may be described by a power law. Finally, the dependence on the matrix stiffness, $\frac{E_M}{E_{90}}$ presented in Fig. 5 d) is also linear when using the constant $A_2$. Assuming superposition of all three contributions leads to the following expression

$$\frac{\varepsilon_{x90}^{ba}}{\varepsilon_{xRVE}} = 0.95 - \left[ B_1 \left( \frac{V_{90}}{1-V_{90}} \right)^{k_1} + B_2 \left( \frac{E_0}{E_{90}} \right)^{k_2} - B_3 \left( \frac{E_M}{E_{90}} \right)^{k_3} \right]$$

(24)

Here $B_1 = 0.153$, $B_2 = 0.115$, $B_3 = 0.175$, $k_1 = -0.433$, $k_2 = -0.342$, $k_3 = 0.628$. Parameters $B_i$ were found as the best fit to all available data. Since according to Eq. (8) $H_{T}^{b}$ in Eq. (18) is defined as

$$H_{T}^{b} = \frac{\varepsilon_{x90}^{ba}}{\varepsilon_{xRVE}}$$

(25)

one can see that Eq. (24) defines the correction factor $H_{T}^{b}$ necessary for RVE stiffness prediction.
5 Stiffness predictions for NCF composites

5.1 Details of the FE analysis

Stiffness of the RVE of a cross-ply NCF composite with lay-up [0, 90]s, containing 9 “super-elements” as shown in Fig. 3, was calculated using the homogenization models described in Section 2 and 3 and compared with direct 3-D FE results. In section 5 and 6 the thickness to length ratio of the RVE was constant and set to \((L_{90}+L_{M})/t_{90}=10\). Solid185 which are an eight node structural solid elements with three degrees of freedom at each node were used in all 3-D FE models. The number of elements for the 3-D FE calculation was 3200 for normal loading and 1600 for shear loading. Plane82 elements were used in the 2-D plane stress case. The number of elements was 256. An example of the 3-D FE model with bundle structure in both 0° and 90°- layers is presented in Fig. 6.
Since the RVE is symmetric, only the upper half of the RVE was modelled. Boundary conditions were as follows:

a) In **normal loading**, the nodes at the two related surfaces \( (x=0 \) and \( x=L_x \) or \( y=0 \) and \( y=L_y \)), were coupled so that two corresponding nodes experienced the same tangential displacement. Nodes at the top surface were coupled, forcing the nodes to have the same displacement in z-direction. Strains \( \varepsilon_{xRVE} = 1\% \) and \( \varepsilon_{yRVE} = 0 \) were applied.

b) In **shear loading**, the same coupling conditions as already described for normal loading were applied. However, additional boundary conditions are also used in form of coupling of nodes at related surfaces, so that two corresponding nodes experienced the same displacement in the out-of-plane direction \( (z\)-direction). In this case a tangential displacement of 0.25% of the side length of the RVE was applied to each side surface resulting in a total of 1% shear strain.

### 5.2 Effect of the RVE surface conditions

The effect of boundary conditions at the top surface on the stiffness was analyzed. The coupled node boundary condition (the same z-displacement for all nodes) which simulates a periodic lay-up in the thickness direction was compared with free surface condition. The results presented in Table 4 for CF cross-ply composite \( (V_b=0.5) \) show that the model becomes stiffer if the nodes are coupled on the top-surface. In the following sections the coupling on the top surface will be used for all FE models concerning stiffness predictions.
Table 4. Stiffness and bundle strain of \([0,90]_s\) CF NCF composite with bundle structure, \(V_b=0.5\).

<table>
<thead>
<tr>
<th></th>
<th>3-D FEM, free top surfaces nodes</th>
<th>3-D FEM, coupled top surface nodes</th>
</tr>
</thead>
<tbody>
<tr>
<td>(Q_{11}) (GPa)</td>
<td>36.405</td>
<td>36.555</td>
</tr>
<tr>
<td>(Q_{12}) (GPa)</td>
<td>2.008</td>
<td>2.106</td>
</tr>
<tr>
<td>(\varepsilon_{x1,3,7,9})</td>
<td>0.877</td>
<td>0.902</td>
</tr>
<tr>
<td>(\varepsilon_{x2,8})</td>
<td>1.123</td>
<td>1.098</td>
</tr>
<tr>
<td>(\varepsilon_{x4,6})</td>
<td>0.796</td>
<td>0.848</td>
</tr>
<tr>
<td>(\varepsilon_{x5})</td>
<td>1.204</td>
<td>1.153</td>
</tr>
</tbody>
</table>

In Table 4, \(\varepsilon_{x1-9}\) are average strains in the “super-elements” according to Fig. 3a.

5.3 Brief characterization of the used stiffness models

In order to validate the homogenization methods described in Section 2 and 3, parametric studies, analysing the stiffness matrix of NCF laminates were performed for both carbon and glass fiber composites. In this case only in-plane elements of the stiffness matrices are considered. In the case of glass fiber, only a limited number of cases are studied to validate the conclusions obtained for the carbon fiber laminates. Stiffness analysis was performed using 7 different methods:

1. **3D FE**: 3-D FE model with bundle structure in both 0\(^\circ\) and 90\(^\circ\)-layers. This model is considered as the reference model.
2. **3D FE Hom0°**: 3-D FE model were the 0\(^\circ\)-layer bundle structure is replaced by a UD layer according to section 2.1. The bundle structure is kept in the 90\(^\circ\)-layer only.
3. **CLT**: Homogenization over both 0\(^\circ\) and 90\(^\circ\)-layer. Each layer with a bundle structure is replaced by an effective UD layer and CLT is used for NCF composite stiffness.
4. **Eq 13**: Analytical solution which uses 3-D FE or power law expressions for average transverse strains in bundles. This model is the main result of this paper.
5. **Iso-strain**: This method assumes the same strain in all meso-elements. The strain is equal to the applied strain.
6. Partial iso-strain: The problem is reduced to 2-D problem for a domain consisting of NCF “super-elements”. This method keeps the strain of all meso-elements in a given NCF “super-element” equal. The solution of the 2-D problem may be found using an analytical model in section 3.3, which assumes that straight lines defining “super-element” rows and columns in the NCF composite remain straight after deformation.

7. 2D FE: The problem is reduced to the 2-D problem exactly as in the partial iso-strain model. The difference is that the 2-D problem is solved using FEM.

5.4 Comparison of models

From Tables 5 and 6, it is obvious that all homogenization techniques predict the stiffness of the laminate with accuracy sufficient for practical needs. The iso-strain model gives the highest stiffness of all models. This is due to the very rigid constraint that all meso-elements have the same strain as the RVE strain. Partial iso-strain model on the other hand, allows the different “super-elements” to have a unique strain in each super-element belonging to the same row or column and the model becomes less stiff compared to the iso-strain model. Nevertheless, the Partial-iso-strain model is still stiffer than the 2D FE model, partially due to the assumption that each super-element after deformation of the laminate retains its rectangular shape. Homogenization over only the 0º-layer gives higher stiffness than homogenization over all layers (CLT) and the stiffness is generally lower than for the reference model (3D FE model with bundle structure). The analytical solution “Eq 13” gives the results closest to the reference model. However, the accuracy of this model depends on the precision in the determination of the average strain in the 90º-bundle used as input in the model. This model predicts the stiffness closest to the reference model when the strain in the 90º-bundle comes from 3D FE calculations. Comparing carbon and glass fibers, we see that CF composites show less “dispersion” in results than GF composites regarding the different homogenization models. This is not surprising since the deformation of the NCF laminate is mostly controlled by longitudinal properties of bundles, which means that higher stiffness carbon fiber bundles are controlling the stiffness of the laminate to a higher degree than bundles with lower stiffness.
Table 5. Stiffness of a $[0, 90]_s$ CF NCF composite according to different homogenization models.

<table>
<thead>
<tr>
<th>Stiffness (GPa)</th>
<th>3D FE</th>
<th>CLT</th>
<th>Iso-strain</th>
<th>Partial Iso-strain</th>
<th>2D FE</th>
<th>Eq 13 $H_T^b$ from Power law $H_T^b$ from 3D FE</th>
</tr>
</thead>
<tbody>
<tr>
<td>$V_b=0.5$</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$Q_{11}$</td>
<td>36.555</td>
<td>36.553</td>
<td>36.355</td>
<td>36.847</td>
<td>36.775</td>
<td>36.646</td>
</tr>
<tr>
<td>$Q_{12}$</td>
<td>2.106</td>
<td>2.046</td>
<td>1.989</td>
<td>2.154</td>
<td>2.126</td>
<td>2.099</td>
</tr>
<tr>
<td>$Q_{66}$</td>
<td>2.115</td>
<td>2.095</td>
<td>2.090</td>
<td>2.469</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>$V_b=0.8$</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$Q_{11}$</td>
<td>56.299</td>
<td>56.297</td>
<td>56.190</td>
<td>56.522</td>
<td>56.491</td>
<td>56.461</td>
</tr>
<tr>
<td>$Q_{12}$</td>
<td>2.505</td>
<td>2.473</td>
<td>2.441</td>
<td>2.546</td>
<td>2.534</td>
<td>2.528</td>
</tr>
<tr>
<td>$Q_{66}$</td>
<td>2.911</td>
<td>2.895</td>
<td>2.890</td>
<td>3.183</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 6. Stiffness of a $[0, 90]_s$ GF NCF composite according to different homogenization models.

<table>
<thead>
<tr>
<th>Stiffness (GPa)</th>
<th>3D-FE</th>
<th>CLT</th>
<th>Iso-strain</th>
<th>Partial Iso-strain</th>
<th>Eq 13 $H_T^b$ from Power law</th>
</tr>
</thead>
<tbody>
<tr>
<td>$V_b=0.5$</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>$H_T^b$ from Power law</td>
</tr>
<tr>
<td>$Q_{11}$</td>
<td>15.78</td>
<td>15.466</td>
<td>16.356</td>
<td>16.039</td>
<td>15.914</td>
</tr>
<tr>
<td>$Q_{12}$</td>
<td>2.372</td>
<td>2.181</td>
<td>2.556</td>
<td>2.413</td>
<td>2.345</td>
</tr>
<tr>
<td>$Q_{66}$</td>
<td>2.177</td>
<td>2.144</td>
<td>2.591</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>$V_b=0.8$</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>$H_T^b$ from Power law</td>
</tr>
<tr>
<td>$Q_{11}$</td>
<td>23.27</td>
<td>23.074</td>
<td>23.74</td>
<td>23.58</td>
<td>23.273</td>
</tr>
<tr>
<td>$Q_{12}$</td>
<td>3.034</td>
<td>2.92</td>
<td>3.19</td>
<td>3.12</td>
<td>2.968</td>
</tr>
<tr>
<td>$Q_{66}$</td>
<td>3.06</td>
<td>3.019</td>
<td>3.377</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

6 Average transverse strain in bundles

6.1 Average strain in super-elements

Damage analysis in bundle composites, which will be addressed in our future work, requires detailed information about the strain in meso-elements. According to Table 4, the boundary conditions on the upper surface (free or coupled) affect the average strain and the strain variation is smaller if the top surface is coupled. In the following FE results the coupling condition is used. Results in Table 4 show rather large differences in the average strain in different “super-elements” leading to failure of all attempts to model damage evolution based on the RVE strain only as it is done in the iso-strain model.

The two methods employing “super-elements”, with stiffness calculated using iso-strain assumption for all meso-elements belonging to the “super-element”, show more adequate results presented in Table 7. The analytical
Partial Iso-Strain model gives even closer results to the reference model than the 2D FE model. This implies that the 2-D problem is solved with sufficient accuracy and the cause of differences with the reference model is in the iso-strain assumption used for meso-elements in the “super-element”. The numbering of the “super-elements” presented in Table 7 is defined in Fig. 3.

<table>
<thead>
<tr>
<th>Material</th>
<th>$V_b$</th>
<th>Super-element</th>
<th>3D-FE</th>
<th>2D FE</th>
<th>Partial Iso-strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>CF</td>
<td>0.5</td>
<td>1,3,7,9</td>
<td>0.902</td>
<td>0.970</td>
<td>0.957</td>
</tr>
<tr>
<td></td>
<td>0.8</td>
<td>1,3,7,9</td>
<td>0.948</td>
<td>0.989</td>
<td>0.988</td>
</tr>
<tr>
<td>GF</td>
<td>0.5</td>
<td>1,3,7,9</td>
<td>0.860</td>
<td>-</td>
<td>0.865</td>
</tr>
<tr>
<td></td>
<td>0.8</td>
<td>1,3,7,9</td>
<td>0.926</td>
<td>-</td>
<td>0.958</td>
</tr>
</tbody>
</table>

Due to lower difference in stiffness between the “super-elements” the Partial Iso-Strain model for Glass-fiber composites is in a better agreement with the reference model than for Carbon-fiber.

6.2 Average strain in meso-elements

More detailed analysis of 3D FE results reveals, see Table 8 for the “super-elements” defined in Fig. 3, that the average strain in the $0^\circ$-bundle differs from the average strain in the $90^\circ$-bundle. Obviously the iso-strain hypothesis for a NCF “super-element” is not accurate enough.

<table>
<thead>
<tr>
<th>Material</th>
<th>$V_b$</th>
<th>Super-element</th>
<th>3D FE, bundle structure</th>
<th>Power law</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>Upper meso-element</td>
<td>Lower meso-element</td>
</tr>
<tr>
<td>CF</td>
<td>0.5</td>
<td>1,3,7,9</td>
<td>0.987</td>
<td>0.826</td>
</tr>
<tr>
<td></td>
<td>0.8</td>
<td>1,3,7,9</td>
<td>0.995</td>
<td>0.907</td>
</tr>
<tr>
<td>GF</td>
<td>0.5</td>
<td>1,3,7,9</td>
<td>0.964</td>
<td>0.762</td>
</tr>
<tr>
<td></td>
<td>0.8</td>
<td>1,3,7,9</td>
<td>0.986</td>
<td>0.870</td>
</tr>
</tbody>
</table>

The “Upper meso-element” and “Lower meso-element” refer to two meso-elements of a “super-element” above the RVE’s symmetry plane. For example in “super-element” 1, the “Lower meso-element” is a $90^\circ$-bundle whereas in “super-element 2” it is a matrix-region. The average strain in the
90°-bundle, which is the most probable locus of transverse failure, is lower than for the “super-element” in average.

Strain in the matrix region of “super-element” 2 is much larger than in the 0°-bundle of the same “super-element”.

7 Conclusions

The NCF composite stiffness and average bundle strain was analyzed using approximate methods with different level of complexity: a) smearing out the bundle structure in a layer, replacing the NCF composite by a laminate and using CLT; b) assuming iso-strain for all meso-elements; c) using assumption that the composite consists of “super-elements” and the stiffness matrix of it can be obtained assuming iso-strain assumption for constituents, thus reducing the inherently 3-D stress problem to a 2-D problem. It was found that for stiffness calculations all homogenization methods used in this study have sufficient accuracy considering a 3D FE model as the reference case. Damage evolution analysis in bundle composites, which will be addressed in our future work, requires detailed information about the transverse strain in bundles which governs transverse cracking in NCF composites.

FE analysis showed that the transverse strain in a bundle may be significantly lower than the strain in the adjacent longitudinal bundle or in the “super-element” in average. The iso-strain approach is inadequate because in this model the strain in each meso-element is the same. Models using homogenization over the thickness and which approximately describe the strain distribution in the RVE give better results.

A more accurate model to predict the average transverse strain in the bundle was developed, analysing by FEM, the importance of media surrounding the 90°-bundle in cross-ply composite. This study showed that an increase of the 90°-bundle to matrix stiffness ratio leads to a decrease of the average strain in the 90°-bundle. An increase in the 0°-bundle stiffness leads to an increase of the average strain in the 90°-bundle. Higher 90°-bundle volume fraction in the layer results in larger average strain in the bundle. These trends are described by a power law and used to predict the average strain in a 90°-bundle.

The H-matrix, introduced to describe the relationship between the average bundle strains and strains applied to the RVE, is used to calculate the “effective stiffness” of the bundle. The “effective stiffness” is the main parameter in simple but exact expressions which were derived to calculate the stiffness matrix of NCF composites.
8 References

Paper D
Damage in NCF composites under tension

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Abstract

Non-crimp fabric (NCF) cross-ply composites response to tensile loading is investigated showing large effect of the fabric layer stacking sequence: much larger elastic modulus reduction was observed in [0/90/0/90]S than in [90/0/90/0]S case. Since transverse cracks in 90°-bundles may give modulus decrease about 5%, the observed 40% stiffness reduction is attributed to failure and delamination of bundles oriented in the direction of the applied load. Analysis of micrographs shows extensive delaminations and 0°-bundle breaks. FE calculations showed that failure of 0°-bundles at the surface is energetically more favorable. However, the fracture resistance of surface bundles is higher due to smaller bundle waviness and the density of bundle cracks on the surface was not larger than inside.

Two possible reasons for the higher stiffness reduction in the [0/90/0/90]S NCF composite were suggested: a) If two imperfect 0°-bundle layers are separated by a 90°-bundle layer their resistance to failure is lower than when they are situated next to each other; b) the effect of each surface 0°-bundle break on the composite stiffness is larger (due to less constraint from the surrounding material the opening of surface bundle breaks is much larger).

Keywords: Non-crimp fabric composites, Damage accumulation, Stiffness, FEM, Bundle failure

*corresponding author
1. Introduction

Composite materials are today in widespread use in the aerospace and marine industries due to their excellent strength and stiffness to weight ratio. In most marine and aerospace applications traditional composites (pre-impregnated tapes) are still the main choice for design of load bearing structures. However, the extensive use of composite materials in both primary and secondary constructions calls for new production techniques which render lower manufacturing costs while maintaining the same excellent mechanical properties as traditional composite materials. One promising manufacturing technique which can meet these requirements involves the textile process which is used to produce non-crimp fabric (NCF) composites as a relatively new class of reinforcement suitable for high performance structures. NCF composites are manufactured from preforms (blankets) with multiple layers of straight fiber bundles with different orientations stitched together by a warp knitting procedure [1]. This manufacturing procedure creates a material which is heterogeneous not only at the microscale but also at the mesoscale due to the bundle structure in the layers.

Non-crimp fabric (NCF) composites combine fairly good mechanical properties with low production costs, high deposition rates and virtually unlimited shelf life. Additionally to these benefits, NCF composites have also been reported to show higher out-of-plane fracture toughness and damage tolerance [2-5] as compared with prepreg laminates. The Visby corvette manufactured by the Karlskrona shipyard (a 72 meter long stealth ship) is one example where the marine industry successfully used NCF based composites also in load bearing structures. Currently there are attempts to evaluate the possibility to use NCF composites in more hi-tech applications, namely in the aerospace industry, where good mechanical performance and damage tolerance is crucial.

However, before NCF composites can be used in these areas, more research must be conducted on the subject of damage evolution and accumulation as well as its influence on mechanical properties of these composites. The logical start for such research would be characterization of properties degradation during the exploitation (mechanical loading) of NCF composites and identification of micro-damage mechanisms in these materials responsible for observed changes. Since NCF composites have an inherently heterogeneous structure of fiber bundles and matrix areas, the mechanisms controlling damage evolution and failure are in most cases more complex than for a traditional composite which complicates the
analysis of these materials. The damage phenomena in NCF composites have to be analyzed in comparison with the very well characterized damage mechanisms in laminated composites revealing similarities and significant differences. Whereas similar damage behavior can be analyzed by improved “old” approaches, the new phenomena require development of novel models. Due to geometrical complexity of NCF composites on the meso-scale, the potential to develop analytical models with sufficient accuracy is limited and combined numerical-analytical approaches have to be preferred. FE simulations provide the possibility for a thorough investigation of stresses and strains in the heterogeneous structure found in NCF composites and there are several publications on this subject. For example in Wisnom et al. [6, 7] the compressive strength and intra-laminar shear behaviour of NCF composites was investigated. The objective of these studies was to provide manufacturing guidelines concerning the fabric structure in order to optimize the properties of the composite. The authors used a 2D-trough the thickness repeating cell in which the material heterogeneous structure is modeled in detail. The two competing mechanisms (kink-band formation in fiber bundles with imperfect orientation and inter-bundle matrix yielding) during compressive loading of unidirectional and multi-directional NCF plates were analyzed using FEM in [8].

It is well known that the most common mode of damage causing reduction of stiffness in cross-ply prepreg laminates is transverse cracks in 90°-layers. Properties degradation in such materials is very well studied and there are a number of accurate models available [9-11] which can predict reduction of elastic properties as a function of transverse crack density. Obviously, intra-laminar cracks also develop in NCF cross-ply composites in fiber bundles oriented transverse to the main loading direction, resulting in stiffness reduction of approximately the same magnitude as in traditional composites. It has been shown [12] that the reduction in elastic properties in carbon fiber NCF cross-ply composites may reach 4% if the only damage in the cross-ply composite are transverse cracks in the fiber bundles oriented transverse to applied load. The stiffness of the damaged NCF composite was described with sufficient accuracy by models previously developed for prepreg laminates and adapted to the NCF case [12]. However, the development of damage in bundles is much more difficult to model: the strain in bundles is not uniformly distributed and usually even the average strain in bundle is smaller than the applied strain [13].

The inelastic behaviour of NCF composites due to damage evolution is still not thoroughly investigated [14] and requires additional work which would lead to deeper understanding of failure mechanisms present in NCF composites. This forms the objectives of our study. We hope that the
The presented paper will provide a deeper insight into micro-damage modes evolving during tensile loading and their effect on degradation of NCF composite elastic properties. We present experimental results which show that the reduction of elastic modulus in NCF cross-ply laminates due to increasing tensile loading in some cases is much higher than in laminates containing only transverse cracks. Moreover, reduction of stiffness exceeds the worst case scenario predicted by simple ply-discount model (complete failure of 90°-bundles). This indicates that in fact more complicated damage modes are present in the NCF laminate than just transverse cracking or local delaminations caused by cracks. Breaks of longitudinally oriented imperfect fiber bundles (usually in locations with certain waviness) were revealed using optical microscopy and attributed to the observed stiffness changes. The energetically preferable locations of broken fiber bundles and the effect of bundle breaks and bundle delaminations on stiffness, considering the latter as a secondary damage mechanism, were analyzed using FEM.

2 Experimental

2.1 Geometrical features of Non-crimp fabric composites

The difference between prepreg-tape based laminates and NCF composites is that the former has an internal structure consisting of continuous fibers that are rather homogeneously dispersed over the whole layer; whereas NCF composite has continuous fibers combined in fiber bundles with well defined geometry. A schematic picture of non-crimp fabric is shown in Fig. 1a and a micrograph of an NCF cross-ply laminate is presented in Fig. 1b. However, each fiber bundle consists of several thousands of rather homogeneously distributed fibers (see Fig. 1b, lower part) and thus, the bundle is similar to prepreg tape based UD composite in terms of microstructure and elastic properties. Such hierarchical structure and very different length scales (fiber scale and bundle scale) in NCF composite imply that analysis of the NCF material should be based on a multi-scale approach [13].
2.2 Materials and experimental procedure

Experiments were performed on cross-ply laminates produced using carbon fiber based NCF reinforcement and epoxi resin. The reinforcement was in a form of blankets consisting of 2 plies manufactured with an orientation of ± 45° and a surface density of 534 g/m². The composite was manufactured using the resin infusion technique (flexible tooling) where a total of 4 fabric blankets were used, thus obtaining an 8 ply NCF composite with a volume fraction of fibers corresponding to approximately 60%. Depending on cutting direction, 2 different lay-ups of laminates were obtained; lay-up A: [0/90/0/90]S and lay-up B: [90/0/90/0]S. Thus, for lay-up A, outer layers were oriented in the loading direction and for lay-up B, the outer layers were orientated transverse to the load. Schematic pictures of both lay-ups are shown in Fig. 2.
Tensile specimens were cut and polished to reach dimensions according to the European Standard EN2597(B) resulting in a specimen length and width of 250 and 25 mm respectively. The specimen edges were polished to relatively high extent by using diamond pastes and fine polishing emulsion with a particle size of 0.5 μm in the final polishing sequence. All samples were fitted with end tabs (composite with woven glass fibre fabric and epoxy resin) which were glued to the specimens by Araldite 2011 2-component epoxy adhesive. Mechanical tests were carried out according to the European standard EN2597 which is a displacement controlled tensile test with a loading rate of 0.01 mm/s. Tests were performed on a Dartec hydraulic machine with a load capability of 250 KN. During the test, load and longitudinal strains were recorded. Measurements of longitudinal strain were performed using an extensometer with a gauge length of 50 mm.

Evaluation of damage accumulation and resulting degradation of elastic properties were performed using a stepwise loading sequence with increasing load level in each step. After each loading step, the specimen was unloaded to a strain level of 0.05% allowing for the measurement of the Young’s modulus within the strain interval 0.05-0.25%. In this region, the material shows linear behaviour and does not undergo any additional damage. Young’s modulus was measured from the mean value of one unloading and reloading sequence in the above region and obtained from linear fitting of the stress-strain curve. A total of 6 specimens of each lay-up were used in the experiments.

2.3 Macro-properties

Experimental results for normalized elastic modulus as a function of applied strain for NCF carbon fiber based cross-ply laminates (both lay-up A and B) are presented in Fig. 3. Normalization is done with respect to the modulus for undamaged laminate. It should be noted from Fig. 3 that the scatter is significantly larger for lay-up A than for lay-up B.
Figure 3. Degradation of elastic modulus as a function of applied strain for carbon fiber based NCF cross-ply laminates with diverse lay-up.

The results in Fig. 3 show that the Young’s modulus starts to decrease at approximately 0.65% strain for both lay-up A and B. However, at higher strain levels the behavior of the laminates differs significantly. In the case of lay-up B, the maximum decrease in modulus is within 5% which correlates very well with the modulus reduction found in cross-ply pre-preg carbon fiber based laminates where the dominating damage mechanisms is intra-laminar (transverse) cracks in the 90°-layers [9].

When a prepreg cross-ply laminate is subjected to tensile load, as shown in Fig. 4a, multiple transverse cracking takes place (Fig. 4b) which can be accompanied by local (at the crack tip) delaminations on the interface between 0° and 90°-layers. This damage is usually quantified by a crack density (crack per unit length) and changes of elastic properties are presented either as a function of this damage parameter or as a function of applied load if the number of cracks is unknown. Usually there is very little or no damage at all in the 0°-layers.
Figure 4. Cross-ply laminate (a) subjected to uniaxial loading and (b) resulting damage.

A large number of models with different sophistication have been developed to deal with prediction of degradation of elastic properties in cross-ply laminates at known level of damage, see for details [11]. Analytical expressions for modulus degradation contain thermo-elastic ply properties, laminate lay-up and crack density [11]. For example, such expression can in generalized form be written as:

\[ \frac{E_s}{E_{s0}} = \frac{1}{1 + a \rho u(l_0)} \]  

(1)

where \( E_s \) and \( E_{s0} \) is the elastic modulus of the laminate with damage and without damage respectively, \( a \) is a constant containing elastic properties and geometry of 0° and 90°-layers, \( \rho \) represents the crack density, \( u(l_0) \) is average opening displacement of the crack (COD) normalized with respect to the load and \( l_0 \) is the normalized half-distance between cracks. Equation (1) is rather general and can actually be used not only for cross-ply laminates but also for more general symmetric laminates with cracks in 90°-layers. Moreover, the COD can be obtained by using different models (for example Shear Lag, Hashin’s, FEM etc.), see [9-11] for more details.

However, modeling of damage evolution (crack density as a function of strain) is a much more difficult task and the evolution is usually determined from experiments. Example [9] of experimentally determined damage evolution for carbon fiber/epoxy prepreg laminates and predictions of the degradation of elastic modulus as a function of crack density are shown in Fig. 5.
A lower bound estimation for degradation of elastic modulus of cross-ply laminate due to cracks in the 90°-layer can be obtained using the “ply discount” model which assumes that the elastic modulus of the damaged 90°-layer is zero (may be visualized as a case of infinite crack density). Since “rule of mixtures” is very accurate for cross-ply laminate we can use it to determine the initial modulus:

\[ E_{x0} = \frac{t_0E^0 + t_90E^{90}}{t_0 + t_90} \]  \hspace{1cm} (2)

where \( E^0 \) and \( E^{90} \) is the axial modulus of 0° and 90°-layer respectively; \( t_0 \) and \( t_90 \) is thickness of 0° and 90°-layers in the laminate respectively. When the 90°-layer is completely destroyed, \( E^{90}=0 \), and equation (2) gives the modulus of the damaged cross-ply laminate:

\[ E_x = \frac{t_0E^0}{t_0 + t_90} \]  \hspace{1cm} (3)

The ply discount solution for the CF/EP cross-ply laminate is also shown in Fig. 5b.

In NCF composite case, the maximum possible modulus reduction due to transverse cracks in bundles can be estimated using laminate analogy of the NCF composite: the bundle structure of a layer is smeared out and replaced by a layer. The elastic properties of a layer are determined by the Composite Cylinder Assemblage model (CCA model) [15, 16] using the average fiber
content in the composite and the properties of the fiber and matrix as an input. The accuracy of the “smearing –out” procedure for stiffness calculation has been proven in [13]. The calculated layer properties are given in Table 1 for $V_f^{layer} = 60\%$. Using them in (2) we obtain a Young’s modulus of 76.48 GPa for an undamaged laminate. Assuming that the whole layer with 90°-bundles is completely destroyed or delaminated from the rest of the material and applying equation (3) we obtain a Young’s modulus of 71.72 GPa as the lower bound for the damaged laminate. Thus, in worst case scenario, elastic modulus of such NCF laminate would decrease by approximately 6%.

However, in the case of lay-up A, the reduction in modulus was as high as 40% which is much higher than the above estimation for carbon fiber based cross-ply laminates with transverse cracks only. This means that the extensive reduction in modulus found in experiments for laminates with lay-up A can only be explained by some additional and more complex damage mechanism than transverse cracks in the 90°-layer. Damage MUST BE also in bundles with 0°-orientation which is confirmed by the fractography analysis performed in Section 2.4.

2.4 Damage modes

Micrographs of a carbon fiber reinforced NCF cross-ply laminate with lay-up A after tensile loading to different strain levels are shown in Fig. 6-8. Obviously the meso-structure is not perfect and the theoretically aligned 0°-bundles have local waviness. Fig. 6 reveals that there are only a few transverse cracks in the 90°-bundles at 0.66% strain, this is consistent with the results obtained for degradation of elastic modulus in Fig. 3.

Figure 6. Type A cross-ply laminate with lay-up [0/90/0/90]$_s$ at a strain level of ~0.66%.
These cracks are more often located in the middle 90°-layer which has double thickness. Cracks are often crossing the entire cross-section of these layers (both 90°-bundles). It appears that the crack on the left side has partially penetrated the neighbouring 0°-bundle below it. This effect has been observed also in laminates and the cause of it is a very large stress concentration in fibers close to the tip of the transverse crack [17].

At higher strain levels of approximately 0.87% shown in Fig. 7, an extensive damage is developing in the specimen. In this case not only multitude of transverse cracks in 90°-bundles occurs but also vast delaminations between 0° and 90°-layers as well as fracture of 0° bundles takes place (see also Fig. 8 for details). The actual sequence of events is still not clear. One possible scenario includes 0°-bundle break as the first event followed by delamination due to dynamic effects and shear stresses. It may be complemented by intense transverse cracking. Another possibility is that delamination is initiated by 90°-bundle cracks and/or out-of-plane stresses due to local waviness of the 0°-bundle. Once the 0°-bundles are separated from other neighbouring bundles, their out-of-plane movement is no longer constrained. Thus, if the 0°-bundle is wavy (as it is shown in Fig. 6-8) it is simply straightening out with reduced contribution to the axial stiffness of the laminate or it is breaking due to bending stresses.

Figure 7. Type A cross-ply laminate with lay-up [0/90/0/90]s at a strain level of ~0.87%.
Figure 8. Type A cross-ply laminate with lay-up [0/90/0/90]_S at a strain level of ~0.87%. Close-up from Fig. 7 (within rectangle), showing extensive delaminations and failure of the 0°-bundle.

Micrograph of a type B cross-ply laminate is presented in Fig. 9 for comparison. Although the strain level at which this Fig. 9 is produced is rather high (0.91%), only a few transverse cracks in the 90°-bundles are visible and no delaminations or breaks in the 0°-bundles are present in the laminate. The smaller amount of transverse cracks in this case may be related to the geometrical differences: in laminates with lay-up A there are two transverse layers together but in laminates with lay-up B only one. Cracking in thick layer (through two bundles) requires less energy than in thin layer (only one bundle). Micrographs (Fig. 6-9) validate the assumptions stated in Section 2.3, that the extreme decrease of elastic modulus in the laminate with lay-up A is caused by additional failure events related to 0°-bundles (not only transverse cracks in the 90°-bundles). However, modeling of the observed damage to reveal its effect on the stiffness reduction in the composite is required in order to further validate the mechanisms suggested in this section.

Figure 9. Type B cross-ply laminate with lay-up [90/0/90/0]_S at a strain level of ~0.91%. 
3 Modeling

The aim of modeling was:

a) to determine the location of the $0^\circ$-bundle in the NCF composite most prone to break. This would give an explanation why we have the very large stiffness reduction attributed to $0^\circ$-bundle breaks in type A but not in type B NCF laminates.

b) to analyze the effect of the delamination length of the broken $0^\circ$-bundle as well as the effect of the number of broken bundles (without delamination) on the stiffness of the composite.

Modeling of the observed damage phenomena was performed by FEM.

3.1 Theoretical prerequisites

The energetically most favorable location of the bundle break can be analyzed comparing the energy release rate due to bundle break formation for bundles at different locations. Virtual crack closure technique, which involves the assumption that the energy needed to close a crack is the same as the energy released when the crack forms, can be used. The crack closure energy for the failure of a $0^\circ$-bundle was calculated according to the procedure described in [5] were the work needed to increase the original crack length, $a$, by a small incremental length $\Delta a$, will be equal to the energy required for the prolonged crack length, $a + \Delta a$ to return to its original length. The work $W$ can then be expressed as:

$$ W = \frac{1}{2} \sum_{i=1}^{N} \sigma_i^b \int_{A_i} u_i \, dA $$

(4)

were $\sigma_i^b$ is the stress in the loading direction for a finite element with index $i$, situated in the considered cross-section of the bundle before failure, $u_i$ is the crack opening displacement (COD) of the crack surface for an finite element with index $i$ after failure. In this case $u_i$ were determined using nodal displacement and the element shape functions. $N$ is the total number of finite elements in the bundle at the crack surface.

The stiffness of the damaged NCF composite may be calculated as follows. First the total reaction force on all nodes in the loading direction is determined at the surface where displacements in longitudinal direction are
applied. From the total reaction force, the stress in the loading direction is calculated dividing by the total cross-sectional area of the surface. Application of Hooke’s law for this uniaxial loading case (dividing the stress by the applied strain which is calculated from the displacement) gives the Young’s modulus of the NCF composite.

3.2 Details of the used FE model

The FE models were chosen equal to the representative volume element (RVE) of the investigated NCF material. The specimen dimensions and manufacturing parameters used in the experiments were used in the FE model: 12 bundles (in average) oriented in the loading direction within each 0°-layer. It resulted in the following bundle dimensions; \( t_B = 0.3\ mm \) and \( w_B = 25/12 = 2.08\ mm \). In this case \( t_B \) and \( w_B \) are the thickness and width of the bundle respectively. A schematic picture of the models employed in the parametric studies (Case A-D), different location of the broken 0°-bundle and the boundary conditions used in the simulation are presented in Fig. 10. The broken bundle is represented by the black area.

Figure 10. Schematic figure of the models used in the FE calculations with the dimensions; \( w_{RVE} = 25\ mm \), \( l_{RVE} = 25\ mm \) and \( t_L = 0.3\ mm \); a) Case A; \([90, 0, 90, 0]\); b) Case B; \([90, 0, 0, 90]\); c) Case C; \([0, 90, 0, 90]\); d) Case D; \([90, 0, 90, 0]\).

Symmetry conditions were applied to surface \( S_3 \) and \( S_6 \), whereas 1 % strain in x-direction was applied to surface \( S_5 \). Surfaces \( S_1 \), \( S_2 \) and \( S_4 \) are modeled without constraints. The break of one longitudinal bundle is
simulated by releasing the symmetry condition on the bundle at the crack surface (black area) separately from the rest of the surface $S_6$. Delamination of a broken bundle was represented by setting the elastic properties to zero in the part of the bundle which was delaminated. This procedure neglects Poisson's effects and details of the stress transfer at the delamination crack tip which would lead to a somewhat stiffer behavior of the model. However, since the delamination length of the longitudinal bundle is a rather uncertain parameter, the difference can be neglected. The delamination length was varied in 12 equally long steps starting from perfectly bonded (no delamination) to a totally debonded bundle. The numerical analysis was performed using the commercial FE-package Ansys 9.0. The FE models were meshed using the 3-D element solid185 with quadratic shape. The element Solid185 is defined by eight nodes having three degrees of freedom at each node.

In the analysis concerning delaminated longitudinal bundles, the mesh size in the bundle corresponded to 6 elements in width direction and 2 elements in the out-of plane direction resulting in a total of 13104 elements in the entire model. A similar mesh was used analyzing the strain energy release rate due to bundle break by crack closure technique. However, in this case, the mesh size in the bundle corresponded to 4 elements in width direction and 2 elements in the out-of plane direction which resulted in a total of 6750 elements in the entire model. The models were meshed using increasing element size in the loading direction (x-direction) with the smallest elements situated at the surface were bundle failure was simulated. The model is rather rough but keeping the mesh constant it was good enough to reveal the qualitative differences between cases.

The bundle structure in each layer of the NCF composites was kept in the FE calculations. However, the inter-bundle matrix regions were not included in the analysis and each bundle is in a direct contact with the adjacent bundle. The bundle properties were assumed equal to the average layer properties considering it as a homogenous material with retained average volume fraction of fibers in the layer. The effective elastic properties of one homogenous layer were calculated using the Composite Cylinder Assemblage model (CCA model) developed by Z. Hashin [15, 16]. The input data used in this model are the elastic properties of fiber and matrix and the volume fraction of fibers. The obtained properties of the NCF layer used in the FE calculations can be seen in Table 1.
Table 1. Material properties used in calculations.

<table>
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<tr>
<td>$G_{23}$</td>
<td>8.33</td>
<td>3.35</td>
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4 Simulation results and discussion

4.1 Energy release due to longitudinal bundle break

The energy released when a $0^\circ$-bundle breaks was investigated in this study with the objective of establish the location of the $0^\circ$-bundle most prone to fail. It was assumed that the bundle break is not followed by delamination at the interface and the released energy was calculated using (4) as the work to close the crack.

The results in Fig. 11 are presented in the form of Normalized crack closure energy. The normalization was done with respect to the crack closure energy for the broken $0^\circ$-bundle with location according to case A.

![Normalized crack closure energy](image)

Figure 11. Normalized crack closure energy for bundles with different locations in the NCF composite.

Analyzing the crack closure energy obtained for one $0^\circ$-bundle failure, it is obvious from Fig. 11 that cracks corresponding to case C (broken surface bundles) experience higher energy release than in the other cases. The crack
Closure energy is significantly higher for case C compared to the other cases due to different constraining effects of material surrounding the bundles: Due to the free surface the opening of these cracks is significantly larger. The same argument applies also to the edge cracks (case D) but in this case the free surface is (due to the elongated cross-section of the bundle) much smaller and the effect is not significant. There are some bundles which belong to both group C and D. The energy release due to break of these bundles is even larger.

We can conclude that cracks in bundles located according to case C therefore will appear at lower tensile loads followed by case D, A and B respectively. In the first moment this conclusion seems to be confirmed by the larger stiffness reduction in type A composites which has outer layers with 0°-orientation of bundles. However, these conclusions are not consistent with the results obtained using optical microscopy: longitudinal bundle failure is statistically NOT more frequent in surface 0°-bundles than in other locations. At present we can suggest only a qualitative explanation of this apparent discrepancy. Most probably the longitudinal bundle breaks are related to imperfections, mainly in local waviness which is different in outer and inside layers.

Due to manufacturing features the curvature in the out-of-plane direction as can be seen in Fig. 6-7, is smaller for the bundles situated at the surface of the composite. The waviness leads local bending stresses in the bundle which are especially high if the bundle is partially debonded. The effect of this imperfection can be represented by different “effective” critical strain energy release rate $G_{IC}$ for bundles with different location in the composite. For surface bundles (Case C) which are comparably straight the “effective” $G_{IC}$ is higher which “compensates” for the higher released energy and reduces (or cancels) the predicted trend that most of the cracks should be in the surface layers.

Assuming that the above discussion is correct and that, indeed, the number of breaks in surface bundles is not larger than in inside bundles, we are back to the initial problem: why the stiffness degradation is much larger for type A composite with outside 0°-bundles if we just decided that the number of breaks in these bundles is not larger than in other locations? It seems that the difference is due to different behavior of inside bundles: in the type B composite two inside 0°-bundles are together and their resistance to local bending in the location of imperfection is much higher. There is much less bundle breaks in bundles which are together than in cases when they are separated by 90°-bundles.
The mechanics of the bending in imperfect bundles leading to bundle break requires a thorough investigation and the effect of the bending is strongly dependent on the degree of delamination (or perfect bonding) between bundles. In other words it depends on the actual sequence of failure events in the NCF composite.

### 4.2 Stiffness reduction due to bundle breaks and delamination

In this subsection, effect of $0^\circ$-bundle failure and delamination along the $0^\circ$-bundle interface on NCF laminate stiffness is analyzed. Since NCF composites consist of discrete bundles (not homogeneous layers as often assumed calculating stiffness of an undamaged laminate), it is more appropriate to consider failure of a separate $0^\circ$-bundle rather than failure of the whole $0^\circ$-layer when the stiffness degradation is analyzed. The reduction of modulus due to analyzed failure events in the $0^\circ$-layers should be estimated with respect to the number of breaks within $0^\circ$-bundles. This type of analysis for broken bundles with a location according to Fig. 10a is performed numerically as described in Section 3.2. In Fig. 12a, the result for a single bundle break is presented in the form of normalized Young’s modulus versus the normalized length of the $0^\circ$-bundle interface delamination. The modulus normalization was performed with respect to the Young’s modulus of the undamaged laminate (no break and no delamination of the bundle). The modulus of the undamaged NCF laminate was also calculated using FEM and due to free edge conditions on surfaces $S_1$, $S_2$ the result is slightly different than obtained from laminate theory. The normalized delamination length is defined as the delaminated length divided by the specimen length. Thus value 1 corresponds to totally delaminated bundle.

According to Fig 12a the modulus reduction of the NCF laminate due to a single $0^\circ$-bundle break without delamination is very small. With increasing delamination length the stiffness reduction is slightly nonlinear, reaching approximately 2% at the end when the bundle is totally delaminated. The values in Fig 12 and in the next section are obtained dividing the stiffness reduction calculated according to the FE model in Section 3.2 by 2. This procedure was necessary because the model in section 3.2, Fig. 10, is symmetric with respect to the laminate midplane and the calculation gives the stiffness reduction due to two bundle breaks in the specimen located symmetrically with respect to the midplane. The assumption that the stiffness reduction in the asymmetric case of one single bundle break in the specimen is equal to one half of the reduction due to two symmetric breaks.
is approximate: the bending forces which appear if only one bundle fail in
the laminate are neglected. To evaluate the error introduced by this
assumption a FE calculation was performed using a FE model where all 8
layers of the laminate are used and symmetry is not employed. In this way
the failure effect of one bundle could be modeled separately. The
normalized Young’s modulus from one bundle failure in the FE-model with
8 layers was 0.999565 whereas the stiffness reduction using the symmetric
model and division by 2 gave 0.999555. The agreement of results proves
that the symmetric case can be used to represent the effect of one single
bundle crack.

Fig. 12b shows the degradation of NCF composite modulus as a function
of number of broken 0°-bundles. For a cross-ply NCF specimen with four 0°-
layers shown in Fig. 10 the total number of 0°-bundles are 48 (N=48). In
Fig. 12b the number of broken bundles is denoted by n.

Figure 12. Modulus reduction in cross-ply NCF composite specimen due to
failure and delamination of 0°-bundles with location according to case A; a)
effect of one single crack with varying delamination length; b) effect of
number of failed bundles with two values of delamination.

The results presented in Fig. 12b, show a 46% reduction of the elastic
modulus when half of the 0°-bundles are broken and they are totally
delaminated from surrounding material. Using for this case “bundle
discount model” similar to the one given by equations (2), (3) we obtain

$$\frac{E_x}{E_{x0}} = \frac{N E_0^0 + E^0 \frac{N}{2}}{N E_0^0 + N E^0} = \frac{9.52 + 143.44}{9.52 + 143.44} = 0.53$$

(5)
Thus according to the bundle discount model the stiffness reduction is 47% which is in a very good agreement with the FEM results and proves that this simple rule can be used to evaluate the effect of totally delaminated bundles on stiffness.

However, microscopy observations (Fig. 7, 8) show that the broken bundles were only partially delaminated from the surrounding. It means that there is still some stress transfer possible from the broken bundles to surrounding layers and complete disregard of the whole bundle leads to an overestimation of the modulus decrease. From Fig. 12b, it can be seen that when the delamination length corresponded to half of the bundles total length (50% delamination), the total reduction in modulus corresponded to 28%. Obviously, bundle discount model does not give good results in this case. Another aspect of the problem is that the results are sensitive to the delaminated area which is an uncertain parameter. In order to predict the effect of delamination with higher accuracy, more detailed quantitative experimental information regarding delaminated areas and damage growth has to be collected and a thorough modeling of these phenomena is required.

4.3 Effect of location of the broken 0°-bundle on NCF composite stiffness

In Fig. 12 it is assumed that all bundles effect the stiffness reduction in the NCF laminate to the same extent as the bundles with location according to Fig. 10a in Section 3.2 (Case A).

However, in laminates with lay-up [0/90/0/90]s the situation is quite different: half of the 0°-bundles are situated at the surfaces of the laminate and those bundles experience different constraining effects of surrounding material compared to bundles belonging to Case A. This leads to a somewhat different stiffness reduction for such laminates which is discussed in this Section. We consider broken 0°-bundles in the NCF composite assuming that the delamination between the bundle and surrounding material is zero and perform a parametric study of the effect of the location of the broken bundle in context of the caused stiffness reduction. In Fig. 13, the results are displayed in form of normalized Young’s modulus. The results in Fig. 13a, corresponding to one bundle failure, reveal that the reduction in elastic modulus is quite similar for case A, B and D whereas the reduction for case C is more severe. The difference in stiffness reduction between these cases can be explained by different constraining effects from surrounding layers experienced by the broken bundle. Analyzing Fig. 10c in Section 3.2, it can be seen that the broken bundle belonging to case C is situated at the surface of the laminate and has a significantly smaller amount
of material surrounding the bundle compared to the case A, B and D which results in larger stress relaxation when the bundle breaks. The results displayed in Fig. 13a are in line with the experimental observations in Fig. 3 where the laminate with the 0°-bundles situated on the surface of the laminate experiences the most severe stiffness reduction when loaded in tension. The same amount of bundle breaks in surface layer gives two times larger stiffness reduction.

In Fig. 13b, the total stiffness reduction due to break of all 0°-bundles belonging to the same group (Case A, C and D) is presented for a laminate with the same lay-up as the Type A laminate discussed in section 2.2. The total stiffness reduction is presented as the last point on the corresponding curve. It can be seen from the Figure that the total stiffness reduction due to failure of all bundles belonging to case D are rather low (0.2%) compared to the other cases due to the fact that only 4 bundles belong to this case. Case A results in a total stiffness reduction of 0.9% whereas Case C shows a total reduction in modulus of 2.4%. In Fig. 13b, it is assumed that all bundles in the outer layer including the edge bundles belong to case C. In Fig. 13b, the effect of one bundle break (Fig. 13a) is multiplied with the number of broken bundles in the laminate and presented as a linear relationship.

Figure 13. a) Reduction of Young’s modulus due to one broken 0°-bundle for case A-D; b) Normalized Young’s modulus when all 0°-bundles with a given location are broken (N=48).

This methodology uses the assumption that the cracks in the bundles are not interacting with each other. If the opposite takes place, meaning that the cracks in several bundles are situated at the same or close cross-sections (same x and z-coordinate in Fig 14), the effect on stiffness reduction would certainly increase since the crack opening displacement would increase.
5 Conclusions

The presented experimental data show that the response to tensile loading of NCF cross-ply composites is very much dependent on the fabric layer stacking sequence. Comparing lay-up A: [0/90/0/90]_S with lay-up B: [90/0/90/0]_S one can see that much larger elastic modulus reduction with strain was found in lay-up A. Only transverse cracks were found in the 90°-bundles of type B laminates, where modulus decrease was approximately 5%, which is well in the range of reduction of modulus (typically 3-7%) caused by transverse cracking in carbon fiber cross-ply laminates. The around 40% stiffness reduction in composites of type A can not be explained by damage in 90°-bundles only and is attributed to failure and delamination of bundles oriented in the direction of the applied load. This assumption was verified by analysis of micrographs which showed extensive delaminations and 0°-bundle breaks in type A composites.

FE analysis showed that bundle breaks without any delamination at the bundle/matrix interface lead to modest stiffness reduction of the same magnitude as transverse cracking. The effect becomes much larger if the break is accompanied by delamination at the bundle/matrix interface. When the bundle is totally delaminated from the surrounding layers, the stiffness reduction may be predicted by “bundle discount” model.

FE calculations, performed to evaluate the possible causes of the observed stiffness reduction, showed that failure of bundles at the surface of the laminate is energetically favorable. Since microscopy observations do not show a larger density of broken bundles in surface layer, we conclude that
the fracture resistance of surface bundles is larger due to smaller bundle waviness on the surface as a result of the used manufacturing routine.

Further analysis disclosed two possible important reasons for the higher stiffness reduction in [0/90/0/90]s NCF composites compared with [90/0/90/0]s. The main reason is the larger number of inside 0°-bundle breaks in laminates where imperfect bundles of this orientation are supported by 90°-bundles compared to cases where two 0°-bundles from different layers are together. The mechanics is not clear at present but it may be related to larger bending resistance of two imperfect bundles as compared with one. The difference is larger if these bundles would be delaminated which may be initiated by transverse cracks in 90°-bundles. Another possible reason for larger stiffness reduction in [0/90/0/90]s composite is, according to FE calculations, directly related to the larger effect of each surface 0°-bundle break on the composite stiffness: due to less constraint from the surrounding material the opening of surface bundle breaks is much larger.

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6 References


Paper E
Internal structure parameters governing performance in NCF composites and methodology for their characterization

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Abstract

Mechanical performance of non-crimp fabric composites is very dependent on their internal meso- and micro- structure which is defined by the manufacturing process of the fabric and composite processing conditions. This paper identifies the most important parameters which control mechanical properties of these materials. The identification is based on experimental observations and available theoretical findings. Characteristics of the internal structure of non-crimp fabric composites are analyzed in context of their significance for in-plane elastic and failure properties. Methodology for determination of most typical geometrical parameters of composites using optical observations of cross-sections of manufactured laminates is described. The methodology is applied to characterize cross-ply and quasi-isotropic composite laminates. These results are analyzed and a comparison between the laminates is performed.

Discussion concerning advantages and disadvantages of the proposed methodology in terms of accuracy and usefulness along with practical recommendations of its application are presented.

Keywords: (A) Non-Crimp Fabric, (B) Mechanical properties, microstructures (D) Optical microscopy

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1. Introduction

Traditionally, structural composite parts with prescribed fiber lay-up are produced from pre-impregnated (prepreg) unidirectional (UD) tapes using the required stacking sequence. The curing cycle of such composites is usually vacuum assisted and is carried out under a certain pressure and temperature; therefore autoclaves are most suited for this purpose. Although this manufacturing technique allows producing very high performance composites with very few defects, it is also associated with high costs. This cost comes from time and energy required to assemble and cure laminate with desired stacking sequence as well as from price paid for prepregs and their storage. Moreover, there are other disadvantages of using prepreg materials, such as limited shelf life and limitation of size of the component that can be produced as well as rather high amount of wasted material. Majority of these problems can be avoided by using multi-axial fabrics which consist of a number of unidirectional plies arranged in various orientations. The individual plies in these fabrics are kept together by stitching yarns. In these materials there is ideally no crimp as in case of woven fabrics and therefore they are called Non-Crimp Fabrics (NCF). Production cost of NCF composites is relatively low because they can be produced in one step by Resin Transfer Moulding or Resin Film Infusion type of manufacturing techniques and shelf life of these fabrics is unlimited (environmental conditions for storage of these materials are not very demanding either). Also improvements in damage tolerance and out-of-plane fracture toughness of these materials have been reported [1–3]. NCF composites due to their excellent performance and relatively low cost have become an attractive alternative for aerospace, marine and even automotive applications. For example, glass fiber NCFs are widely used in USA to produce recreational ships (motor boats and yachts) and large, 72 m long, stealth ships used by the navy (corvette Visby) are manufactured from carbon fiber (NCF) fabrics in Sweden. Two large European Commission funded projects have been completed recently: FALCOM for aerospace and TECABS for automotive applications of these materials.

One of the features of the NCF composite is that it is an inherently multiscale material. Each layer due to stitching is divided in fiber bundles. On the microscale each bundle with orientation \( \phi \) is a UD composite with a certain fiber content \( V_f^\phi \) and the homogenized bundle properties may be calculated using micromechanics expressions for long fiber composites. On the mesoscale the bundle is considered as homogeneous transversely isotropic material surrounded by matrix and other bundles of the same or
different orientation. A mesoscale characteristic of the composite is bundle content $V_b^g$ in the composite. Micrographs presented in Fig. 1 demonstrate the hierarchical structure of the NCF composites. The geometrical shape of bundles (cross-section and axial alignment) is complex and depends on bundle orientation in the blanket, surface compression during production, resin pockets etc.

Figure 1. Hierarchical structure of the NCF composites.

The described mesoscale configuration determines the NCF composite properties on macroscale and is typically used in simulations of macro behaviour [4-8]. These studies demonstrated the importance of geometrical parameters of the micro/meso structure of the material on different mechanical properties of NCF composites. For instance, bundle waviness is a crucial parameter that defines compressive strength of composite laminates and it has also a strong influence on tensile strength and Young’s modulus of the composite. While cross-section characteristics and fiber content of bundles are required to analyze intrabundle cracks and delaminations. Although the internal structure of the NCF composite directly affects their mechanical performance, there is no comprehensive study which describes and identifies all microstructural parameters in a structured way and gives methodology and practical recommendation on quantification of these
parameters. There are a number of studies concerning the geometrical characteristics of unimpregnated fabric [9-12] which are mostly applicable to modelling and optimization of processing variables. However, mesoscale structure of NCF composites is affected as much by processing of composites (impregnation of NCF with resin) as by geometry of the fabric itself. Therefore, in order to deal successfully with prediction of mechanical properties and failure events in NCF composites, final structure of the material has to be well known. This investigation is intended to identify and characterize critical parameters of internal structure of NCF composites which are important for certain in-plane mechanical properties. Then the methodology is described for systematic characterization of the NCF composite using CF/EP cross-ply and quasi-isotropic composites as examples. This paper is focused on structural parameters and therefore it is not analysing the role of fiber and matrix properties which is, certainly, of great importance.

Also the stitching effect on the material performance is not discussed. It is mainly because the used optical inspection tools for analysis of different cross-sections in the composite are not the most convenient tool for such investigation. More convenient is to characterize stitches on the NCF fabric before impregnation. However, then the information regarding the disturbances introduced during manufacturing is not available. More about the stitch effect on NCF composite performance may be found in [9-10, 12].

2. Effect of internal structure on NCF composite properties

The mechanical behaviour of NCF composites differs from the behaviour of prepreg based composites. The difference between these two materials is in the mesoscale structure which is in the NCF case basically a bundle arrangement of fibers. In this section we will briefly summarize the available experimental observations and theoretical findings regarding the in-plane properties dependence on the composite internal structure on mesoscale.

2.1 Stiffness of the NCF composite

The stiffness of a multidirectional NCF composite is strongly dependent on the contribution of fibers. The main parameter characterizing fiber contribution is the average fiber content of a certain orientation \( \phi \) in the composite, \( V_{fa}^\phi \). Knowing the constituent density, the fiber content may be obtained from the area weight data for the fabric and the measured weight of
the composite. A more detailed mesoscale information would include the volume content of bundles with a certain orientation, \( V_{\phi} \), the fiber volume fraction inside a bundle of each orientation, \( V_f^{\phi} \), the shape of the cross-section of bundles etc. However, FE analysis performed in [8] showed that the effect of mesoscale details on the NCF laminate stiffness may be in any cases neglected and a “smearing out” routine leads to high accuracy of the determined stiffness matrix. The most appropriate homogenization routine is based on the average fiber content in a layer consisting of fibers with a given orientation. The bundle structure in the layer is replaced by a homogenized layer and the elastic properties of the layer are calculated using Hashin’s composite cylinder assemblage model [13, 14]. As the next step the classical laminate theory (CLT) is used to calculate the in-plane stiffness matrix. This approach was compared with 3-D FE modelling. The results even for NCF laminates with large matrix regions between bundles are within 1% accuracy which exceeds the typical precision in tests. The laminate analogy gives slightly lower accuracy for shear modulus, which may be an indication that the bundle mesostructure is of more importance for this property. However, the error is still negligible (2% for cross-ply type of NCF composite).

The above conclusions are valid if the thickness, \( h_\phi \) of a layer with bundle orientation \( \phi \) is constant and the layer (bundles) does not experience waviness along the fiber direction. However, experimental studies [4, 6, 15, 16, 18] have shown that bundles have rather large out-of-plane waviness, see Fig. 1 (upper figure). This feature is not affecting the bundle stiffness much in transverse direction, but the effective stiffness in the longitudinal direction is affected by the local waviness.

The effect of the local waviness on the NCF composite longitudinal modulus was studied by FE in [16] and was described by an analytical model in [7]. In [7] a knock-down factor was introduced to calculate the effective longitudinal stiffness of an imperfect bundle with waviness. The reduction of the modulus in a NCF cross-ply laminate due to 0°-bundle waviness can be as high as 20% [16]. This effect explains the significantly lower measured elastic modulus of the NCF composite comparing with CLT or FEM predictions (with straight 0°-bundles). The main mesoscale parameters of the local bundle waviness used in the stiffness model are the wavelength of the imperfection and its amplitude. This implies a simplified representation of the imperfection as a periodic phenomenon. The wavelength and amplitude may be considered as independent parameters. However they are not dependent only on the manufacturing parameters like
stitching technique, draping etc. They depend also on the mesostructure of the surrounding bundles. For example, the wavelength may depend on length of the repeating element in the neighbouring layer with different bundle orientation ratio, especially if the length of the resin region between bundles is large, see schematic picture of the resin channel in Fig. 2.

Figure 2. Schematic showing of the out-of-plane misalignment of the 0°-bundle, caused by the periodicity in the 90°-bundle structure.

The same discussion also concerns the amplitude of the wave. An accurate stiffness prediction of NCF composite with geometrical defects by numerical methods requires quantified information regarding the imperfect mesostructure. The elastic properties of bundles also must be calculated and this requires information about local fiber volume fraction in the bundles. It may be concluded that if the imperfect bundle alignment, which we describe as waviness, and its effect on NCF composite stiffness is analyzed, a more detailed mesoscale information is required: fiber content in the layer and in the bundle, the relative content of bundles and resin in the layer, the shape of the bundle and the resin region between bundles etc.

2.2 Damage in tensile loading

The first mode of damage observed during tensile loading of NCF composites is formation of multiple intrabundle cracks in layers with off-axis orientation with respect to the load [16], as shown in Fig. 3. Often these cracks are going through one fiber bundle but sometimes (if two 90°-bundle layers are together) they go through the bundles in multiple neighbouring layers. The stiffness reduction due to this damage mode has been analyzed in [16, 17] replacing the bundle structure in the layer with a homogeneous composite with the same number of cracks.
In these very successful predictions the information from experiment about the density of bundle cracks was used. Simulation of intrabundle crack evolution is a very complex task and requires more detailed strain distribution than just the average strain applied to the composite. It was shown in [8] that the strain in the bundle in the direction transverse to the bundle axes may be much lower than the applied strain. This strain ratio is crucial for damage evolution analysis. It depends on the bundle cross-section (height to width ratio), the ratio of the bundle width and the matrix channel size, elastic modulus of the matrix and the longitudinal and transverse modulus of the bundle. These dependences were described in [8] by a power law type of expression. The final failure of the NCF composite in tension is usually related to the failure of longitudinal bundles. In contrast to prepreg type of composites several longitudinally oriented bundles may break before the specimens final failure [18] leading to significant stiffness reduction with increasing strain. The bundles which break usually are misaligned: out-of plane waviness, see Fig. 4.
Figure 4. Cross-ply laminate with lay-up [0/90/0/90]s at a strain level of ~0.87% showing extensive delaminations and failure of the 0°-bundle.

FE analyses showed that the breaks of 0°-bundles on the specimen surface are energetically favourable and were expected to be observed more often. However tests do not confirm this trend because for inside bundles the out-of-plane waviness, which reduces the effective bundle strength, is larger than for surface bundles. The latter phenomenon is, certainly, manufacturing tool dependent. Another observation was that the misalignment of a longitudinal bundle was less dangerous if another bundle with the same orientation was adjacent to it. The damage pattern after longitudinal bundle break is rather complex and includes the broken bundle which is partially delaminated along the interface from the rest of the material. Many intrabundle cracks in adjacent bundles are also present and the real sequence of failure events is not clear. The two most probable sequences are: a) break of longitudinal bundles which is followed by delamination and multiple cracks in 90°-bundles (dynamic effect due to very large energy release); b) intrabundle cracks initiate delamination with the neighbouring 0°-bundle. The delaminated misaligned 0°-bundle when subjected to tension is not supported by the rest of the material and it breaks because of bending stresses which are significant due to its curved shape and the lack of side support. FE analysis performed in [16] shows that out-of-plane stresses may be significant at the bundle interface if the 0°-bundle has certain waviness. Combination of $\sigma_{zz}$ and $\sigma_{xz}$ may lead to interface delamination even without presence of intrabundle cracks. At the tip of an intrabundle crack (for example, at the interface between the 90°-bundle and 0°-bundle) the out-of-plane shear stresses are very large which often cause local delamination. This local delamination may initiate a large interface crack growth if the 0°-bundle is curved.
2.3 Compressive failure

Two failure mechanisms have been recognized as possible reasons of compressive failure in NCF composites. One of them is related to large out-of-plane stresses (normal and shear stresses) in the matrix region between imperfect bundles. They could lead to interbundle splitting which may be analyzed using Von Mises type of criteria for matrix failure. The second mechanism is related to kink-band formation in the longitudinal bundle. Fig. 5 shows a sample with an extremely misaligned fiber bundle in the composite subjected to compression before and after the failure. It is obvious that the largest misalignment is the locus of failure initiation.

![Figure 5](image)

Figure 5. Premature failure of NCF composite in compression due to bundle waviness.

To understand the kink-band formation specifics in NCF composites it has to be realized that the NCF composite has imperfections in form of waviness on two scales. On the microscale the bundle has an inherent waviness, \( \Phi \) exactly of the same nature as an UD prepreg composite has. Hence, the same compressive failure criterion can be applied for NCF composites as for kink band formation in UD prepreg composites. Due to the in-plane and out-of-plane waviness of the bundle on mesoscale the bundle is in its local axes subjected to compressive as well as shear stress. The combined action of these two stress components (\( \sigma_L \) and \( \tau_{LT} \)) facilitates the kink-band formation in the bundle (UD composite) according to [19]
\[ \sigma_L = \frac{k^* - r_{LT}}{\Phi + \gamma} \]  
(1)

In Eq. (1) \( k^* \) is the shear strength of the bundle material and \( \gamma \) is the yield strain in shear. On the mesoscale the combined stress failure criterion (1) has the form

\[ \frac{\sigma_L + r_{LT}}{\sigma_{cu}} = k^* \]  
(2)

Here \( \sigma_{cu} \) is the compressive strength in L-direction of the bundle material. The shear stress component in Eq. (2) is caused by the bundle mesoscale misalignment only.

FE analysis performed in [6] demonstrated that the compressive failure strain due to kink band formation in the imperfect bundle depends on many mesoscale parameters, without a doubt the misalignment angle being the most crucial one. For instance, bundle misalignment angle of 14 degrees, depending on the extent of support from adjacent bundles, may reduce the failure strain by 50%. It was shown in [6] that the average fiber content alone does not govern the kink-band formation strain. At fixed average fiber content and mesoscale waviness, the composite with more evenly distributed fibers (lower fiber content in the bundle and larger bundle content in the composite) has larger resistance to compressive failure. The examples analyzed in this section show that dependent on the properties investigated different level of details regarding the mesostructure of the NCF composite is required. A summary of the requirements is presented in Table 1. Significance of the parameter for certain property is indicated by a number on scale from 1 to 5 (number 5 refers to the highest significance).

<table>
<thead>
<tr>
<th>Property</th>
<th>( V^\phi_{fas} )</th>
<th>( V^\phi_b )</th>
<th>( V^\phi_f )</th>
<th>Shape of the cross-section</th>
<th>Waviness</th>
</tr>
</thead>
<tbody>
<tr>
<td>Stiffness</td>
<td>5</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>5</td>
</tr>
<tr>
<td>Failure in tension</td>
<td>1</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>5</td>
</tr>
<tr>
<td>Failure in compression</td>
<td>2</td>
<td>3</td>
<td>3</td>
<td>1</td>
<td>5</td>
</tr>
</tbody>
</table>
3. Methodology

The laminates studied in this investigation were both quadriaxials with lay-up [(45, 0, -45, 90)₂, (90,45,0,-45)₂] and cross-ply with lay-up [0, 90, 0, 90]ₜ. The laminates were produced using either fabric of base configuration (standard production settings) or fabric with high stitch tension. All laminates were manufactured using resin film infusion (RFI) technique except for the cross-ply laminate with high stitch tension which was produced by resin infusion under flexible tooling (RIFT). The carbon fiber Tenax HTS 5632 (12k fabric) was used together with epoxy resin to produce the laminates. Layers with different orientation are in this study analysed separately (ply by ply) as well as by pooling measurements and averaging them for all layers with the same orientation together. Data for layers with +45° and -45°-bundle orientation were added together and averaged.

3.1 Classification of bundles by shape

Characterizing bundle geometry, the bundle is assumed to belong to one of the three idealized groups depending on its shape: half of an ellipse, an ellipse or a rectangle (see Fig. 6).

Figure 6. Classification of bundles into three different types depending on their shape.
In order to obtain the actual width of the bundles with 45° orientation, the bundle width measured on the specimen edge has to be projected to the direction transverse to the bundle axis using the following expression:

\[ W = \frac{W_{\text{meas}}}{\sqrt{2}} \]  

(3)

It is noticed, see Fig. 7, that the resin areas between the bundles are significantly smaller if two layers containing bundles with the same orientation are situated on top of each other. This phenomenon occurs for the two 90°-layers situated in the middle part of the laminate, especially when they are shifted horizontally against each other. Since there is almost no gap between the bundles, the area of the resin region between these bundles is almost zero. This means that bundles of one 90°-layer have penetrated into the space between two bundles in the other 90°-layer, creating a distorted shape of the bundle shown in Fig. 7. Thus, the total thickness of such double 90°-layer is smaller than the sum of thicknesses of two “separate” 90°-layers. The distorted shape of the bundles can be sufficiently well approximated by a “half ellipse” according to the definition in Fig. 6 for the analysis performed in the following sections.

![Figure 7. Distorted shape of the bundles in the two 90°-layers adjacent to the symmetry line.](image)

Ideally the shape of any bundle should be satisfactory described by these three shapes. However, in some cases there were also “unusual” or “degenerated” bundles, see Fig. 8, which had a form that did not belong to any of shapes defined in Fig 6. The determination of bundle dimensions in these cases in a satisfactory way to include them in the statistics was problematic. To account for this situation, these bundles are also classified into two additional categories: “continuous” and “tapered” which are described below and presented in Fig. 8.
“Continuous bundles”: The bundle structure in this region is not visible and several bundles are joined together. These bundles often reach from one side of the “window of observation” to the other without any interruption (in other words, the width is approximately 5 times larger than for a standard bundle). In the statistic analysis, these bundles are divided into smaller rectangles based on a slight variation of thickness and/or indication of stitches from neighbouring layers. In this way they are included in the bundle statistics as a certain number of rectangular bundles. However, the actual width of the continuous bundle is unknown because the edges are not visible in the “window”.

“Tapered bundles”: The borders between bundles in this region are indistinguishable, as in the case of “continuous bundles”. However, these regions can be divided into smaller parts (bundles) due to a large variation in thickness. Substantial change of thickness is being used as criterion to detect borders which leads to construction of a number of smaller bundles with somewhat constant thickness. Actually the only difference between “tapered” and “continuous” bundles is that thickness is almost constant across the whole length of the region for the latter ones.

The number of bundles in each layer can then be determined using the categories “half ellipse”, “ellipse” and “rectangle” defined in this Section. For the frequency of particular bundle shape, bundle geometry and distance between bundles, however, only “normal” bundles could be used. It means that bundles in the regions defined as “tapered bundles” or “continuous bundles” in Fig. 8 are not used in the statistics except for the case of volume fraction of bundle.

Finally, we have to mention “Partial bundles”. These bundles have one of the main shapes introduced in Fig.6. However, these bundles are partially outside the “window”, meaning that the width of these bundles can not be measured. However, they occupy a certain area fraction of the “window” and must be included in the bundle analysis. To include them in the general statistics we assume, after identifying their type, that the width of these bundles is the same as the average width of bundles with the same geometry and orientation.
3.2 Volume fractions

3.2.1 Volume fraction of bundle in the composite

The measurements of the volume fractions are based on measurements performed on images taken from the specimen edge. These measurements are based on the assumption that the average volume fraction is the same for any randomly cut-out piece in the laminate. In other words, we assume that the “window” is a representative element of the composite. This assumption however, is only valid if the “window” used in the analysis is large enough. Performed numerous measurements on different NCF composite systems have shown that a “window” size along the in-plane coordinate which corresponds to 5 bundle widths is sufficient to represent average values. The “window” in the out-of-plane direction covers the whole thickness of the laminate. This has been verified by measurements on several windows. Calculation of the area of the bundle is performed according to equations suitable for the particular bundle shape, approximated by an ellipse, half an ellipse or rectangle. The areas are:

\[
A_e = \frac{\pi H^e W^e}{4}, \quad A_{he} = \frac{\pi H^{he} W^{he}}{4}, \quad A_r = H^r W^r
\]

where \(A_e\) is the area of the ellipse, \(A_{he}\) is the area of a half ellipse, \(A_r\) is the area of the rectangle. The dimensions of the ellipse and the rectangle are shown in Fig. 9. The thickness (height) of the half ellipse is \(H^{he}\) and its width is \(W^{he}\).
The volume fraction of bundles is defined as the cross section area fraction of bundles in each layer according to the approximations described in Fig 6. Volume fraction of a bundle with 90°, ±45° or 0° orientation in the composite is calculated using the expression

$$V_b^\phi = \frac{v_{\text{tot}}^\phi}{v_{\text{tot}}} = \frac{\sum_k A_{b_k}^\phi + \sum_i A_{b_i}^\phi + \sum_l A_{r_l}^\phi}{A}$$  \hspace{1cm} (5)$$

Here $V_b^\phi$ is the volume fraction of bundles with a certain bundle orientation denoted by $\phi = \pm 45^\circ, 90^\circ$ or $0^\circ$ in the composite. $v_{\text{tot}}^\phi$ is the total volume of bundles with a certain orientation also denoted by $\phi$. $v_{\text{tot}}$ is the total volume of the composite. In Eq. (5) $A$ is the area of the “window” and areas of bundles are denoted in accordance to Eq. (4).

It should be noted that the volume fraction of bundles for both 45°-layers is considered to be the same. Thus the volume fraction of bundles of $\pm 45^\circ$ is calculated adding both volume fractions together. The total volume fraction of bundles in the composite is determined as

$$V_b^{\text{tot}} = V_b^0 + V_b^{\pm45} + V_b^{90}$$  \hspace{1cm} (6)$$

where $V_b^{\text{tot}}$ is the total volume fraction of bundles in the composite. $V_b^0$ is the volume fraction of 0°-bundles in the composite, $V_b^{\pm45}$ is the volume fraction of $\pm 45^\circ$-bundles in the composite and $V_b^{90}$ is the volume fraction of 90°-bundles in the composite.
3.2.2 Volume fraction of fibers in the bundle

The volume fraction of fibers in the bundle is measured using the software “Analysis” distributed by SIS (Soft imaging systems). Five high resolution images were analysed for each layer with the same orientation of the bundles for each specimen according to Fig. 10. Measuring areas, an average of the volume fraction of fibers in the bundles can then be calculated from these results. The volume fraction of fibers in the bundle is denoted by $V_f^\phi$.

![Figure 10. Example of picture used by Analysis for volume fraction of fiber inside the bundles.](image)

3.2.3 Average volume fraction of fibers in the composite

The average volume fraction of fibers with the same orientation $\phi$ in the composite is determined using the expression

$$V_{fa}^\phi = V_b^\phi V_f^\phi$$

were $V_{fa}^\phi$ is the average volume fraction of fibers with a certain orientation, $\phi = \pm 45^\circ, 90^\circ, 0^\circ$ in the composite (volume fractions for both $45^\circ$ orientations are added together). The total average volume fraction of fibers in the composite, $V_{f}^{lam}$ is determined as

$$V_{f}^{lam} = V_{fa}^0 + V_{fa}^{\pm 45} + V_{fa}^{90}$$

Here $V_{fa}^0$, $V_{fa}^{\pm 45}$ and $V_{fa}^{90}$ are average volume fractions of fibers (in the laminate) with corresponding orientations.
3.3 Characterisation of the matrix channel between bundles

Area and shape of the matrix channel between 0°, ±45° and 90°-bundles can be defined using the known shape of the bundles. The number of possible profiles of the matrix channel between the bundles is defined by the approximation of the bundle shapes introduced in section 3.1. The possible cases that can arise are presented in Fig. 11.

The average horizontal linear size of the resin channel between the bundles in a layer is calculated using the expression

\[ L_{CH} = \frac{W_{tot}^{bundle} - W_{tot}^{window}}{Nr of channels} \]  

(9)

where \( W_{tot}^{bundle} \) is the total width of the bundles in the layer in the window of observation and \( W_{tot}^{window} \) is the total width of the window.

It should be noted that the average area of the resin channel is underestimated by using the described technique. For example, although the linear size of a resin channel will be the same for the case of two rectangular bundles and two ellipses, the area is rather different and is larger in case of two ellipses. On the other hand, volume fractions in this study are entirely based on the measurements of bundles and therefore the significance of the error introduced by measurements of dimensions of resin channels is somewhat diminished. However, if more detailed information is required on
dimensions of channels then one can introduce correction factor for the bundle shape which can be easily obtained from geometrical analysis of the presented cases (see Fig. 11).

3.4 Waviness

Waviness is in Section 2 identified as one of the most important parameters controlling the mechanical performance of NCF composites. Therefore, characterization of the waviness is essential in order to analyze its effect on mechanical properties. Miller [15] showed that the waviness of the $0^\circ$-bundles was mainly determined by the adjacent layers of different orientation. For biaxial laminates this means that during manufacturing the $0^\circ$-bundles are nesting into the matrix channels between each two adjacent $90^\circ$-bundles as schematically shown in Fig. 2. Since the matrix regions between the $90^\circ$-bundles are periodically distributed, the waviness of the $0^\circ$-bundles was approximated by a periodic wave. Drapier and Wisnom [4, 5] used the results presented by Miller as input to numerical modeling of both the compressive strength and interlaminar shear strength of biaxial NCF composites. In this case, the amplitude of the wave in the $0^\circ$-bundles was determined from the measured standard deviation of orientation (SDO) with respect to the theoretical horizontal direction of the bundles. The maximum angle of misalignment of the $0^\circ$-bundles can then be calculated as

$$\phi_{\text{max}} = \pm \sqrt{2} \text{SDO}$$

From Eq. (10) and the assumption of sinus shaped waviness, the amplitude of the waviness can be expressed as

$$A = \sqrt{2} \ast \text{SDO}$$

This approach was also used by Edgren et al [7] where a knock down factor for stiffness predictions was developed as briefly described in Section 2.1. The wavelength is a good characteristic of the misalignment of the $0^\circ$-bundles only if some periodicity can be observed in the laminates. If the waviness is not periodic, another parameter of imperfection should be used instead of wavelength and amplitude, for example, the maximum inclination angle.
3.4.1 Measurements of waviness in cross-ply and quasi-isotropic laminates

Measurements of the waviness in 0º-bundles are performed by dividing the bundles along their length into 20 equidistant intervals where the length of each interval is approximately two times the average thickness of one bundle. Thus, 21 measurement points along the middle line of the bundle are obtained in this manner. Micrographs over quasi-isotropic and cross-ply laminates are presented in Fig. 12 and 13 respectively. It should be noted that the waviness in NCF composites have 2 different length scales, micro and meso-scale as described in Section 2.3 and that the analysis performed in this Section only treated the meso-scale waviness.

Figure 12. Example of the microstructure in the quasi-isotropic laminate produced with high stitch tension.
Figure 13. Example of the microstructure in the cross-ply laminate produced with high stitch tension: a) mesoscale view; b) detail showing stitch induced in-plane waviness.

3.4.2 Stitch induced waviness

As can be seen in Fig. 13 (dashed circle), the waviness in cross-ply composites is in some cases induced by the stitching tread in the sense that the tread separates the 0°-bundle from the adjacent bundles in the area closest to the tread (at a distance of approximately 2 bundle thicknesses). This phenomenon can introduce waviness both in the in-plane and out-of-plane direction. The case with the waviness introduced by the stitching tread in the in-plane direction can be seen in Fig. 13 b) as varying orientation of the fibers in the 0°-bundle.

However, in many cases there is no influence of stitching tread on the waviness of the bundles. For example, analysing the micrograph of the quasi-isotropic laminate in Fig. 12, it is clear that the effect of stitching treads on bundle waviness is insignificant for quasi-isotropic laminates with high stitch tension. Therefore, at present stage of understanding the role of stitches on the waviness is unclear and therefore it is difficult to establish a measure of the waviness induced by the stitching tread. Thus, a more thorough analysis is required in order to establish the dependence of wavelength of 0°-bundles on manufacturing parameters of non-crimp fabrics.
4. Application of the characterization methodology

4.1 Classification of bundle shape

4.1.1 Size and shape of bundles

These properties, measured for quadriaxial laminates, are shown in Fig. 14. Note; in this and following sections, standard deviation is not specified in the figures if the number of measurements is less or equal to 4. The numbers of bundles for both 45°-layers are considered together and the results are added and denoted ±45°. The conclusion regarding the bundle width is that it is larger for 0°-bundles but the effect of the stitch tension on this parameter is not noticeable. Bundles of 90° and 45° orientation are about two times smaller in width and the size is almost independent on the orientation and the stitch tension. It is interesting to note that, as can be seen in Fig. 14 a), there were no bundles shaped as “half ellipses” in the 0°-layer of the base configuration. Only the 0°-bundle thickness is affected by the stitch tension. The thickness of half ellipse bundles is the largest. The thickness of bundles in other layers is almost unaffected. In result the conclusions regarding the bundle cross-section area, see Fig. 14c) are similar: the area is about 50% smaller for off-axis bundles and there is no statistically proved difference between areas of bundles with different shapes.

The relative frequency of the three basic bundle shapes is analyzed in Fig. 14d). In case of high stitch tension material the bundle shapes are more irregular. This is indicated by a higher number of half-ellipse bundles in the high stitch tension fabric compared to base configuration (especially for 0°-bundles). The number of rectangular bundles is decreasing with higher stitch tension. It is reasonable to assume that an undeformed bundle most likely would have elliptical or rectangular cross-section. Therefore, it can be concluded that stitches applied with higher tension most likely cause deformation of bundles by pressing them against neighbouring layers, thus creating more half-ellipses.

The observed small differences between bundles in the inspected two materials may indicate that higher stitch tension is affecting geometry only in points of insertion but not so much in between the stitches.
Figure 14. Size and shape of the bundles with indicated standard deviation for NCF quadriaxial composites produced with base configuration and high stitch tension; a) Bundle width, b) Bundle thickness, c) Bundle area, d) Frequency of particular bundle shape.

4.2 Volume fraction measurements

4.2.1 Volume fraction of bundle, $V_{b}^{\phi}$

The volume fraction of bundles with certain orientation in the composite has been characterized for quadriaxial laminates. Contrary to the case with bundle dimensions and shapes of the different fabrics studied above, it was expected that the difference between volume fraction of fibers in these materials with base configuration and high stitch tension will not be significant. These fabrics are made out of the same types of fibers and total amount of reinforcement used in manufacturing of both composites was also the same.

Moreover, the same value for fiber volume fraction in the laminate (0.55) was planned and used to calculate the amount of the reinforcement material.
by weight employed to manufacture all composite plates. Therefore, the obtained volume fraction of fibers can be actually used to somewhat verify accuracy of measurements and the accuracy of manufacturing. Indeed, the total volume fraction of fibers in the laminate, see Fig. 15d), is very close to the projected value of 0.55.

Figure 15. Volume fractions in quadriaxial composites produced with base configuration and high stitch tension; a) Volume fraction of bundles with certain orientation in the laminate, b) Volume fraction of fibers within the bundles with certain orientation, c) Average volume fraction of fiber in layers with a certain orientation, d) Total volume fraction of bundle $V^{tot}_b$ and total average volume fraction of fibers in the composite $V^{lam}_f$.

Nevertheless, some differences due to stitch tension may be expected for fiber volume fractions within the bundle, due to different dimensions of bundles.

However, as concluded in the previous section, differences in bundle dimensions are marginal and as direct result of this, fiber volume fraction
within the bundles is also very similar in both materials, see Fig. 15b). It
also should be noted that volume fraction within a bundle is rather high,
within 67-72%, which is higher than one normally would find in high
performance composites (60-65%).
The content of bundles with ±45° orientation is about 2 times larger than
for other orientations. Since we have + and – 45° bundles together in the
statistics the conclusion is that the bundle contents are rather equal and
independent on the stitch tension.

4.3 Resin channels

4.3.1 Distance between bundles with certain orientation

quadriaxial laminates are considered in this section. In Fig. 16, the channel
width for both 45º-layers is considered the same and denoted ±45º in Fig.
16. In this case it can be seen that the average distance between the bundles
in the 90º-layers are larger for the laminates produced with high stitch
tension.
However, there is no consistency in these trends because for 0º and 45º-
bundles the distance between the bundles is higher in case of base
configuration, which is opposite to 90º-bundle results. To clarify these
trends the results in Fig. 16 have to be analyzed together with the bundle
shape statistics presented in Fig. 14d) and with the conclusion from Fig.
15a) that the bundle content is the same for all bundles and stitching
techniques.
For example the 0º-bundles in the base configuration composite are mostly
rectangular or elliptical. In high stitch tension composites only a few are
rectangular which means that most of the bundles are half elliptical or
elliptical. Certainly, provided the bundle content is constant, the channel
area content is also the same and, hence, the distance between closest points
of two elliptical 0º-bundles in high stitch tension material has to be smaller.
In 90º-bundles the shape differences between base and high stitch tension
materials are not obvious (may be slightly more rectangular bundles in the
high stitch tension case) and the stitch effect on the distance becomes
different. In ±45º-bundles the amount of rectangular bundles is much
smaller in high stitch tension material (similarly as for 0º-bundles but the
difference is not so dramatic). As the result we see for ±45º-bundles in Fig.
16 the same trend as for 0º-bundles, just it is less explicit. In average for
fabric with base configuration the distance between the bundles is slightly
higher. Since the total bundle content is the same, see Fig. 15d), it is a direct
consequence of the larger amount of rectangular bundles in the base configuration.

![Average distance between bundles (mm)](image)

**Figure 16.** Average distance between bundles (mm).

### 4.4 Waviness

A comparative investigation of 0°-bundle waviness is performed between cross-ply and quadriaxial laminates with two different sets of manufacturing parameters for each lay-up (base and high stitch tension configuration). Results describing the out-of-plane waviness of the 0°-bundles are presented in Fig. 17 using x-z coordinate system as shown in Fig. 2. In this case each bundle is represented by its middle line and thus displays the waviness. Analysing the results presented in Fig. 17, it can be seen that the waviness of the 0°-bundles is rather disordered, which complicates the definition of a characteristic wavelength. This is certainly the case for the cross-ply laminates (Fig. 17a and 17c), where it is difficult to distinguish any periodic pattern of the waviness. It seems that waviness of bundles is on two scales: “global” wave with large wavelength and amplitude and “local” waviness (small wavelength and amplitude) which is superposed to the global wave and “hidden” in it. Schematically this is shown in Fig. 18 (scale of the drawing is purposely exaggerated for better demonstration).
However, “local” waviness is of such small amplitude that effect of it might be negligible. Unfortunately none of the waves (either local or global) has an obvious periodicity and therefore it is very difficult to obtain values of the wavelength. It might be useful to perform some kind of frequency analysis (Fourier transformation for example) in order to obtain more accurate quantitative characteristics of the waviness.
Since it is not evident from Fig. 17 a)-d) that there is a sine shaped waviness in the bundles; the approach suggested by Miller [15] would require a more thorough investigation of the nature of the waviness. Therefore, we use the standard deviation of the distance between the actual position of the bundle and a horizontal line (ideal orientation). The standard deviation of orientation as the measure of misalignment for cross-ply and quasi-isotropic composites is presented in Table 2. It can be noted that cross-ply laminates have approximately 3 times lower standard deviation (which represents wave amplitude, see Eq. 11) of orientation compared to the quasi-isotropic laminates.

Table 2. Measurements of standard deviation of orientation in the 0°-bundles.

<table>
<thead>
<tr>
<th>Laminate Type</th>
<th>Standard deviation of orientation (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cross-ply Laminate</td>
<td></td>
</tr>
<tr>
<td>Base configuration</td>
<td>27</td>
</tr>
<tr>
<td>High stitch tension</td>
<td>22</td>
</tr>
<tr>
<td>quadriaxial Laminate</td>
<td></td>
</tr>
<tr>
<td>Base configuration</td>
<td>71</td>
</tr>
<tr>
<td>High stitch tension</td>
<td>86</td>
</tr>
</tbody>
</table>

Although approximation of bundle waviness by a sinus function can be questionable in some cases, for modeling purpose it can still be sufficient to use standard deviation of orientation as a measure of waviness amplitude. However, in situations when waviness of the bundles is very large (as it is for the laminates studied here) probably it is more appropriate to use maximum misalignment angle as a characteristic for imperfect orientation.
5. Conclusions

Mechanical performance of NCF composites in contrast to prepreg tape composite laminates is not uniquely determined by fiber volume fraction in layers and by volume fraction of layers of different orientation. In addition to above parameters the mesoscale internal structure is as important. The main mesoscale parameter is related to imperfect bundle out-of-plane orientation which affects the compressive strength, the longitudinal bundle breaks in tension and composite stiffness. The structural parameters characterizing composite layer in the cross-section transverse to the bundle axis are of major importance analyzing matrix dominated damage mechanisms like intrabundle cracks. Intrabundle cracks and the imperfect mesoscale bundle structure are responsible for the most of secondary damage mechanisms serving as initiators for more severe damage modes. A clear and consistent methodology is suggested to determine the identified basic parameters by optical microscopy investigation. Since the number of studied micrographs is always limited by feasibility of practical application of methodology, degree of representation of actual material is briefly discussed. Errors and artifacts introduced by abnormalities in internal structure and practical limitation of equipment are also considered. Proposed methodology is applied to cross-ply and quadriaxial NCF composites to demonstrate details and range of results dependent on the used manufacturing parameters. Parameters of internal structure of these laminates are compared between each other. Approach and results presented in this paper demonstrate simple, yet useful method of reasonable accuracy to characterize structure of NCF composites. Due to its affordability in terms of equipment and time, this technique can be used not only by scientific community but also by manufacturers of NCF composite structures to monitor quality of their products.

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6. References

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