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**Development and validation of numerical
modeling approaches for triaxially
braided composites**

TESIS DOCTORAL

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*A mi compañera de viaje Laura,
a mi hermana Julia,
y a mis padres,
Cecilio y María Inmaculada.*

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Master	09/2016	Raúl Herrero	Low-velocity impact characterization of triaxially braided composites
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Abstract

In recent years, the growing use of composite materials has acquired a great role, not only industrial, but also in fields such as construction, transportation or military industry, raising the necessity of an automated, cost-efficient and quality manufacturing process. The process of braiding composites allows automated production of variety of preforms due to its inherent drapability to adapt to complex shapes, in combination with a low material waste and good set of mechanical properties.

Two-dimensional triaxial braided stand out as very promising reinforcement architectures for composite materials with improved quasi-isotropic stiffness and strength, as well as resistance to crack propagation under fatigue and dynamic loads. Nevertheless, the textile yarn architecture controls the material behaviour, resulting in a variation of the mechanical properties on a complex composite structure.

Nowadays, no established numerical methodology is available for the study of the mechanical behaviour of triaxial braided composites. Thus, the main goal of this thesis is to create a multi-scale modelling framework for the prediction and analysis of triaxial braided composites. This requires an efficient unit cell modelling approach and macroscopic methods to predict and analyse the mechanical properties of large components. Furthermore, additional work has been carried out on the modelling of yarn architecture, and on mechanical characterisation and tests required for model input and validation, respectively.

The yarn architecture parameters for $\pm 60^\circ$ triaxial braided composites were robustly measured using optical microscopy and image analysis of surface scans. In-plane mechanical experiments showed that the mechanical behaviour was essentially nonlinear. Inelastic deformation and damage were identified as the underlying mechanisms, attributed to microscopic cracking at the fibre/matrix interface. The failure modes observed were dominated by yarn failure and matrix cracking, with the stress

on the bias yarns controlling the latter mechanism.

An innovative, generic, computationally efficient and parametric numerical finite element methodology based on unit cell modelling approach, using shell elements, was developed and applied to predict the mechanical behaviour of triaxially-braided composites. The stress fields obtained correlated well with classical continuum unit cell results.

A macroscopic modelling approach was developed to simulate in-plane strength, fracture behaviour and to study size effects on the tensile strength triaxially-braided composites with notches. The input parameters of the macroscopic modelling were calculated from the unit cell results. The macroscopic analyses showed that it is crucial for the failure prediction to use an equivalent laminate model providing the stresses in the yarn directions. The predictions, employing a continuum damage model based on Hashin's 2D failure criterion combined with a CDM based on the independent dissipation of fracture energies associated with the major damage mechanisms, were in acceptable correlation with the experiments.

An attempt to validate the model on the basis of simulation of low-velocity impact characterisation is described in an annex to the thesis. Results are presented as well as discussions as to why the used fracture energies associated with tensile fibre and matrix failure obtained in the virtual characterisation are incorrectly representing the energy dissipation mechanisms.

It is concluded that the proposed modelling approach can be considered validated through experimental tests, demonstrating a valid and fast approach for the prediction of the behaviour of triaxially-braided composites which could be extrapolated, with necessary adaptations, to other fibre architectures. Future works and possible optimisation techniques are discussed.

Resumen

En los últimos años, el uso creciente de materiales compuestos ha adquirido un gran papel, no solo industrial, sino también en campos como la construcción, el transporte o la industria militar, lo que plantea la necesidad de un proceso de fabricación automatizado, rentable y de calidad. El proceso de trenzado de materiales compuestos permite la producción automatizada de una variedad de preformas debido a habilidad inherente para adaptarse a formas complejas, junto con un bajo desperdicio de material y un buen conjunto de propiedades mecánicas.

Los tejidos trenzados triaxiales bidimensionales se destacan como arquitecturas de refuerzo muy prometedoras para materiales compuestos con una rigidez y resistencia cuasi-isotrópicas mejoradas, así como resistencia a la propagación de grietas bajo fatiga y cargas dinámicas. Sin embargo, la arquitectura del tejido controla el comportamiento del material, lo que resulta en una variación de las propiedades mecánicas en una estructura compuesta compleja.

Hoy en día, no se dispone de una metodología numérica establecida para el estudio del comportamiento mecánico de los compuestos trenzados triaxiales. Por lo tanto, el objetivo principal de esta tesis es crear un marco de modelado multiescala para la predicción y análisis de materiales compuestos triaxiales. Esto requiere un enfoque eficiente de modelado de celdas unitarias y métodos macroscópicos para predecir y analizar las propiedades mecánicas de grandes componentes. Además, se ha llevado a cabo un trabajo adicional en el modelado de la arquitectura del tejido y en su caracterización mecánica, necesarios para determinación de los datos de entrada del modelo y su validación, respectivamente.

Los parámetros de la arquitectura de haces para los tejidos compuestos trenzados triaxiales de $\pm 60^\circ$ grados se determinaron usando robustas técnicas de observación como microscopía óptica y análisis superficial de imágenes. Los experimentos

mecánicos en el plano demostraron que el comportamiento mecánico era esencialmente no lineal. La deformación inelástica y el daño se identificaron como los mecanismos subyacentes, atribuidos a la rotura microscópica en la interfaz fibra / matriz. Los principales modos de fallo observados estaban dominados por la rotura de los mazos de fibra y el agrietamiento de la matriz, siendo la rotura controlada por los mazos orientados a $\pm 60^\circ$ grados.

Una metodología innovadora, genérica, computacionalmente eficiente y paramétrica de elementos finitos basada en el modelado de celdas unitarias, utilizando elementos placa, fue desarrollada y aplicada para predecir el comportamiento mecánico de los compuestos trenzados triaxiales. Se obtuvo una buena correlación entre las distribuciones de tensiones experimentales y las obtenidas mediante los modelos continuos de celdas unitarias.

Una metodología de modelado macroscópico ha sido desarrollada para simular la resistencia en el plano, las propiedades de fractura y para estudiar los efectos del tamaño del agujero en la resistencia a la tracción de materiales compuestos trenzados triaxiales. Los parámetros de entrada del modelo macroscópico se calcularon a partir de los resultados de las celdas unitarias. Los análisis macroscópicos demostraron que es crucial para la predicción de fallos usar un modelo laminado equivalente que proporcione los esfuerzos en las direcciones del haz. Las predicciones, empleando un modelo de daño continuo basado en el criterio de fallo 2D de Hashin combinado con un modelo de daño continuo basado en la disipación independiente de las energías de fractura asociadas con los mecanismos principales de daño, estuvieron en buena correlación con los experimentos.

En los anexos se presenta un intento de validar el modelo usándolo para la simulación de la caracterización del impacto a baja velocidad. Los resultados y el por qué las energías de fractura usadas asociadas con los fallos de la matriz y la fibra a tracción obtenidas en la caracterización virtual representan incorrectamente los mecanismos de disipación de energía, son discutidos.

Se concluye con que la metodología de modelización propuesta puede considerarse validada mediante la realización de ensayos experimentales, lo que demuestra un método rápido y válido para la predicción de propiedades mecánicas que podría aplicarse a otras arquitecturas de fibra. Los trabajos futuros y posibles técnicas de optimización son discutidos.

Nomenclature

α, β, χ	Compliance curve fitting parameters
α_{sh}	Shear-stress coupling factor in longitudinal yarn failure
δ	Load point displacement
Δ_a	Total crack growth
γ_{xy}	Laminate in-plane shear strain
$\hat{\sigma}_1, \hat{\sigma}_2, \hat{\tau}_{12}$	Effective plane-stress tensor components
μ	Friction coefficient
ν_m	Pure matrix Poisson's ratio
ν_{12}	Laminate Poisson ratio
$\nu_{12}, \nu_{13}, \nu_{23}$	Yarn Poisson's ratios
ϕ	Filament diameter
σ_x, σ_y	Laminate longitudinal and transverse interrupted stress
σ_{ft}	Fibre tensile strength
σ_{mt}, σ_{mc}	Pure matrix tension/compression strength values
τ_{xy}	Laminate in-plane shear stress
θ	Braiding angle
$\varepsilon_L, \varepsilon_T$	Laminate longitudinal and transverse strain-to-failure
a, a_0, Δ_a	Crack length, initial crack length, total crack growth
C	Specimen compliance
d	Yarn spacing
d_f, d_m, d_s	Yarn damage variables
E_1, E_2	Laminate elastic constant
E_1, E_2, G_{12}	Yarn orthotropic elastic constants
E_c	Total fracture energy dissipation
E_m	Pure matrix elastic modulus
E_x, E_y, G_{xy}	Laminate in-plane elastic moduli

ECT	Extended compact tension
$F_{ft}, F_{fc}, F_{mt}, F_{mc}$	Yarn failure indices
F_{vf}	Fibre volume fraction
G_{Ic}^0	Interlaminar fracture toughness associated with longitudinal fibre breakage
G_{Ic}^I	Strain energy release rate for crack initiation
G_{Ic}^L, G_{Ic}^T	Laminate fracture energies for longitudinal and transverse loading directions
G_{Ic}^P	Strain energy release rate for steady-state crack growth
G_{Ic}^{lam}	Laminate critical strain release rate
G_m	Pure matrix fracture energy
$G_{ft}, G_{fc}, G_{mt}, G_{mc}$	Yarn fracture energies
$g_{ft}, g_{fc}, g_{mt} and g_{mc}$	Volumetric fracture energies for fibre and matrix
G_{Ic}	Critical energy release rate
G_{vf}	Geometrical volume fraction
K_{Ic}	Critical stress intensity factor
L_c	Characteristic element length
l_{fpz}	Length of the fracture process zone
P_c	Critical load to cause fracture
t, w	Specimen thickness and length
x	Axial direction
X_L, X_T	Laminate longitudinal and transverse tensile strength
$X_T, X_C, Y_T, Y_C, S_L, S_T$	Yarn unidirectional strength values
y	Perpendicular to axial direction
ν_{xy}	Laminate in-plane Poisson's ratio
CAD	Computer Aided Design
CC	Compliance Calibration method
CDM	Continuum Damage Mechanics
CFRP	Carbon Fiber Reinforced Polymers
CLPT	Classical Laminate Plate Theory
CMOD	Crack Mouth Opening Displacement
CNC	Computerized Numerical Control
CT	Compact Tension
CT	Computer Tomography
DCB	Double Cantilever Beam
DIC	Digital Image Correlation

FE	Finite Elements
FEA	Finite Element Analysis
FEM	Finite Element Model
FRP	Fiber Reinforced Polymers
GFRP	Glass Fiber Reinforced Polymers
GMC	Generalized Methods of Cells
LEFM	Linear Elastic Fracture Mechanics
MBT	Modified Beam Theory
MMS	Multi-Material Shell
MOC	Methods of Cells
MPC	Multi-point-constraint
OHT	Open Hole Tensile
PBC	Periodic Boundary Conditions
PMC	Polymer Matrix Composites
R-curves	Fracture toughness resistance curves
ROI	Region of interest
RTM	Resin Transfer Molding
RUC	Representative Unit Cell
RVE	Representative Volume Elements
SEM	Scanning Electron Microscopy
UD	Unidirectional
W/D	Width to hole diameter parameter
XCT	X-rays Computer Tomography

"Science can amuse and fascinate us all, but it is engineering that changes the world".

Isaac Asimov (1988 AD)

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Introduction and objectives

1

Throughout the years, the aeronautical industry has been to the vanguard of the technology in different aspects. One of them is the development of lightweight structures made of different combinations of materials. The demand for lighter and stronger materials, especially from the aeronautical industry, is constantly growing. There is an on-going change in using polymer matrix composite (PMC) materials rather than the traditionally-used metals. In the industry, PMC are the most used in small arm projectiles, for personal protection and as backing plates against larger projectiles and blast fragments, usually in combination with ceramics, [Ngo et al. \(2010\)](#).

PMC usually consist of two phases, one dispersed phase (reinforcing) and one matrix phase. Due to this combination, the resulting material conserves the benefits of the individual phases into a single material. Thus, the composite will remain the ability to absorb and mitigate kinetic energy, typical property of resin and, it will show high to ultra-high elastic modulus and strength, very common in high performance fibres. Thereby, the combination of both materials, presents higher specific strengths than their metal counterparts, and they are capable of providing equivalent ballistic impact protection at reduced weights.

The matrix phase of the PMC can either be a thermoset or a thermoplastic. During the curing, thermosets lose their viscosity when they are heated, but after some time at high temperature, several chemical reactions take place causing the polymer to cross link and thereby to form a solid material. Due to these irreversible chains, thermosets have high-temperature resistance, high dimensional stability and good resistance to solvents. On the other hand, thermoplastic resins reach low viscosity when they are heated and come solid when cooling down. This is the reason why thermoplastic matrix is not used at high temperature, [Naik \(1994\)](#).

However, they are more suitable to impact damage and cracking applications. The role of the reinforcing phase is to increase the mechanical properties. Commonly used reinforcing materials are glass, aramid, or carbon fibres. These fibres can be used in different architectures, either as single fibres or textile fabrics.

In general, the main advantages of polymer matrix composite materials are low density, high strength, high stiffness, high fracture toughness, good drapability, really high in-plane properties, and finally, poor impact properties. In general, properties, which are matrix dependent, are much lower than those which are governed by the fibre. Nevertheless, the role of the matrix is vital in composite behaviour; protecting the fibre, transferring stresses and, in some cases, alleviating brittle failure by providing alternative paths for crack growth.

A unidirectional (UD) fabric is one in which the majority of fibres run in one direction only. A small amount of fibre or other material may run in other directions with the main intention to hold the primary fibres in position. However, the other fibres may also offer some structural properties.

Carbon Fibre Reinforced Polymers (CFRP) and Glass Reinforced Polymers (GFRP) are polymer matrix composite materials reinforced by carbon and glass fine fibres, respectively. Their reinforced dispersed phase may be in the form of either continuous or discontinuous fibres commonly woven into a cloth. Both material characteristics as high modulus of elasticity, low density, high tensile strength, and high strength to weight ratio and corrosion resistance make them a highly rated and widely used PMC in the industry. The most popular matrix materials for manufacturing Fibre Reinforced Polymers (FRP) are thermosets such as epoxy, polyester and thermoplastics such as nylon (polyamide).

In contrast, the absence of fibres in z-direction, makes that the through-thickness characteristics, completely depend on the low mechanical properties of the resin and fibre-resin bond, resulting in one of the major disadvantage of the unidirectional PMC's, their brittleness, which can result in a disastrous way of failing. This situation causes the material to possess poor properties in that particular direction, exhibiting poor resistance to impact damage. Therefore, an impact can cause an important reduction of in-plane properties under tension, compression, bending and fatigue loads, [Tong et al. \(2002\)](#).

Laminates are prone to delamination damage when impacted by low speed projectiles; tools dropped during maintenance, bird strikes, hail impacts or stone im-

pacts, which they are very common situations, it makes the weak point of UD, [Tong et al. \(2002\)](#). To combat the problem of delamination damage, composite parts are often over-designed with extra thickness to provide in some cases moderate improvements to impact damage resistance. As a consequence, this increases the cost, weight and volume of the composite. Accordingly, it is the high price of these materials and their manufacturing process an important problem in contrast with metals.

1.1 Textile composites

There are several applications that required more than one fibre orientation; in these cases a fabric combining different fibre orientations is useful. Two and three dimensional composites are made using different textile processing techniques such as weaving, knitting, braiding and stitching. Among them, weaving and braiding are the most used techniques to manufacture composite materials.

Weaving is a method of textile production in which two distinct sets of yarns or threads are interlaced at right angles to form a fabric or cloth. The longitudinal threads are called the warp and the lateral threads are the weft or filling. The method in which these threads are woven affects the characteristics of the fabric, [Gill et al. \(2009\)](#); [Herrero García \(2016\)](#); [García-Carpintero \(2014\)](#). Woven technique has been used for years in textile industry, and the equipment can also be used to manufacture more intricate, net-shaped preforms that have three-dimensional fibre architecture, [Ayranci and Carey \(2008\)](#).

Braiding is a manufacturing technique where a braiding machine deposits continuous, intertwined, fibre tows to create desired reinforcing braid architecture before or during the impregnation of the fibres. When the fabric is compounded by two yarns, they are named bias, and if there is other yarn in the longitudinal fabric direction, it is called axial. The main advantages of braiding technology are high rate of strand deposition on the mandrel, ability to produce complex shapes, low capital investment cost and minimal labour cost compared to other manufacturing processes. The most important disadvantage is the difficulty to produce low braid angle preforms, [Ayranci and Carey \(2008\)](#).

In order to decrease manufacturing costs, increase through-thickness mechanical properties and improve impact damage tolerance, 2D and 3D composites with different architectures were developed. Each type has its own advantages and disadvantages and are illustrated in the following table (Table 1.1), [Department of](#)

1.1. Textile composites

Defense United States of America (2002).

Table 1.1: Advantages and disadvantages of different fibre architectures, Department of Defense United States of America (2002)

Fabric architecture	Advantages	Disadvantages
Non-crimp fabrics	High in-plane properties Good design flexibility Automated process	Low transverse and out-of-the plane properties Poor fabric stability
Multi-axial Warp Knitted fabrics	Good tailoring for balanced in-plane properties Highly automated process Suited for large area coverage Multi-layer high throughput	Low out-of the plane properties
2D Woven fabrics	High in-plane properties Excellent batch-to-batch consistency Often most economical choice in balance of properties Good drapability Highly automated process High design flexibility for complex shapes	Low transverse and out-of-the plane properties
2D Braided fabrics	Good balance in off axis properties High in-plane shear strength and stiffness Good Drapability High design flexibility for complex shapes Quasi-isotropic lamina Automated preform process	Low out-of-the plane properties Size limitations Tend to be limited to axis-symmetric shapes High set-up costs
3D Woven fabrics	Moderate in-plane and out-of-the plane properties Good drapability Automated process	Limited off-axis properties
3D Braided fabrics	Good in-plane and out-of-the plane properties Good drapability Highly automated process	Slow preform process Size limitations Tend to be limited to axis-symmetric shapes High set-up costs Difficult reproducibility (fibre content)

From Table 1.1, it can be seen that the trough thickness properties of the 3D preforms are better than the 2D because of the existence of z-direction fibres. However, 3D fabrics have size limitation which are conditioned by the machine that builds the fabric. Low/non-crimp fabrics present a highly unbalanced architecture, achieving the excellent mechanical properties of a unidirectional tape and some of the handling advantages of the fabrics. Therefore, high modulus fibres are only used, building a fabric with sterling in-plane properties. Multi-axial warp knitting technology is an advanced knitting method which can bind multiple sets of parallel straight yarns together by stitching yarns. Coated fabrics based on bi-axial or multi-axial warp knitted structures have been rapidly developing in recent years. They provide good mechanical properties in tension and tearing propagation resistance, [Hu and Xu \(2008\)](#). Two-dimensional woven fabrics are usually selected in the applications where good in-plane properties, drapability and large area coverage are required. The following figure (Figure 1.1) presents the most commonly woven architectures used in the industry.

1.2 Woven fabrics

2D woven fabrics are commonly used in fabrication of engine nozzle structures, thermal protection systems, and other relative non-complex hardware with relatively low out-of-plane loading requirements. Plain woven is manufactured passing each warp fibre alternately under and over each weft fibre. The fabric is symmetrical, with good stability and reasonable porosity. It is the most difficult of the weaves to drape, and the high level of fibre crimp imparts relatively low mechanical properties.

Twill weave has a looser interlacing and the weave is characterised by a diagonal line. In a twill weaving structure, each warp yarn floats over two consecutive weft yarns, and under the following one weft yarn. With reduced crimp, the fabric also has a smoother surface and slightly higher mechanical properties than the plain weave. Basket weave is a twill weave but with the difference that despite of 1 warp yarns floating over two consecutive weft yarns, are 2. Basket weave is flatter, and, through less crimp, stronger than a plain weave, but less stable.

Satin weaves are fundamentally twill weaves modified to produce fewer intersections of warp and weft. In a satin weave one warp yarn is woven over a number (value greater than 2) of successive weft yarns, and then under one weft yarn. This weave structure, with interlaced regions that are not connected, is called (the num-

1.2. Woven fabrics

---

ber used + 1)-harness satin weave, [Tan et al. \(1997\)](#). Its weaves are very flat, which have good wet out and a high degree of drape. The low crimp gives good mechanical properties. However, the style's low stability and asymmetry needs to be considered. The asymmetry causes one face of the fabric to have fibre running predominantly in the warp direction while the other face has fibres running predominantly in the weft direction.

Figure 1.1: Commonly woven fabric architectures. Images generated with TexGen software, [University of Nottingham \(2019\)](#).

Between the different weave architectures from Figure 1.1, in the field of manufacturing, satin fabrics are preferred over flat fabrics (plain, twill and basket) for their better adaptability to complex geometric shapes. This is called drapability and is an important factor in the manufacture of parts with complex geometry, such as a motor with dual blade curvature parameter.

Weaving is a process that has been used for over 50 years to produce single-layer, broad-cloth fabric for use as fibre reinforcement of composites. It is only relatively recently, however, that weaving techniques have been modified to produce 3D woven materials that contain through-thickness fibres binding together the in-plane fabrics. A variety of 3D woven composites have been manufactured using modified weaving looms with different amounts of x-, y- and z-direction fibres so that the properties can be tailored for a specific application, [Tong et al. \(2002\)](#).

The great flexibility of the 3D weaving process means that a wide variety of composite components have been developed for aerospace, marine, civil infrastructure and medical applications, . However, only a few 3D woven components are currently used; most of the components have been manufactured as demonstration items to showcase the potential applications of 3D woven composites. In the production of these demonstration items, it has been proven in many cases that it is faster and cheaper to manufacture 3D woven components than 2D laminates, particularly for complex shapes. Furthermore, 3D woven components have superior delamination resistance and impact damage tolerance, [Tong et al. \(2002\)](#).

There are basically three types of 3-D preforms currently used: polar, orthogonal and angle interlock. The three basic fibre directions can be customised forming a continuous structure tailored to specific design requirements. The 3-D structure has excellent damage tolerance, improved shear property performance, and designed specific strength and stiffness. The superior damage tolerance of 3D composites occurs because the through-thickness binder yarns are able to arrest or slow the growth of delamination cracks formed under an impact loading, [Mouritz et al. \(1999b\)](#).

Figure 1.2: Polar preform

Polar preforms include axial, radial, and circumferential fibre orientations. These weaves are used for applications having a body of revolution (cylinders, throat components, nose tip, exit cones, convergent and divergent shapes). Non axis-symmetric shapes (leading edges, conic/rectangular transitions) can be made by forming and then slitting the preform to the desired dimensions, [Mouritz et al. \(1999b\)](#). Figure 1.2 shows a 3D polar preform.

3-D orthogonal preforms are defined by the 90 degrees intersection of all the fibres to one another. Parts made from this process are typically formed in a billet and then subsequently machined to final dimension. These preforms are typically

1.3. Braided fabrics

Figure 1.3: a) Orthogonal and b) layer-interlock preforms woven fibre architectures commonly used in 3D woven composites, [Mouritz et al. \(1999b\)](#).

used in relative flat applications (brakes, rotors, and heat sinks). The 3-D angle interlock preforms are used where part complexity can be reduced by integrally woven parts such as seals, heat exchangers, stiffeners, truss structures, and related composite inserts (such as tee clip and pi clip). An application typically allows for some reduction in in-plane properties while improving the out-of-plane properties. These fibre preforms are commonly used in conjunction with the 2-D. Figure 1.3a and b show a 3D orthogonal and a layer interlock preforms respectively, [Mouritz et al. \(1999b\)](#).

1.3 Braided fabrics

Braiding has been used since the 1800s to produce textile fabrics. New demands for high production rate manufacturing of high quality composite materials have focused attention on braiding. A conventional braiding machine has fibre carriers moving in a circular pattern. Half of the carriers move clockwise, the others counter-clockwise, in an intertwining serpentine motion producing a desired braid pattern, such as 2-dimensional tubular and flat braids. Figure 1.4 presents a state-of-the-art 2D braiding machine.

Braiding is a manufacturing technique where a desired reinforcing braid architecture is created by depositing continuous intertwined fibre tows before or during its impregnation. Depending on the configuration of the braiding machine, different types of reinforcements can be produced on 2D braiding machines. The type of reinforcement is primarily distinguished by the number of yarn directions, the braid

Figure 1.4: 2D maypole braiding machine with 128 yarn carriers, [Cichosz \(2015\)](#).

interlacing pattern, and the type of preform, [LOMOV and VERPOEST \(2005\)](#). The addition of axial fibres along the mandrel axis, highlighted in blue in Figure 1.5b, it is called triaxial braid, and it increases bending, tensile strength and also, shear stiffness and strength of braided composite materials, [Kier et al. \(2015\)](#); [García-Carpintero et al. \(2016b\)](#). Conventional braiding machines produce preforms either vertically or horizontally. Most braiding machines are said to be Maypole-type machines due to the serpentine or maypole strand carrier path. There are also Rotary braiders which use two rotating tables. Although they have higher production rates than Maypole braiders, they cannot produce flat braids, [Ayranci and Carey \(2008\)](#). Because the fabric is manufactured in a cylindrical shape, the preform is cut parallel to the axial fibres and flattened afterwards. The following figure (Figure 1.5) illustrates the difference between biaxial and triaxial braiding.

The number of active yarn carriers defines the interlacing pattern of the braid. There are three commonly used braid architectures, [García-Carpintero \(2014\)](#); [Herrero García \(2016\)](#): Hercules braid, regular braid and diamond braid, Figure 1.6. 2x2 architectures, regular and diamond braid patterns (Figure 1.6a and b), are achieved by using all or half of the available yarn carriers on a regular braiding machine, Figure 1.4. However, to produce preform such as Hercules pattern, six to eight notches per horn-gear are needed, [Lomov et al. \(2007\)](#); [Cichosz \(2015\)](#). The waviness reduces with increasing length of the straight yarn region, yielding better in-plane properties. On the other hand, the stability of the braided decreases, which can be

1.3. Braided fabrics

Figure 1.5: Types of braid reinforcements. Images generated with TexGen software, University of Nottingham (2019)

an issue for complex parts with large cross sectional changes. Therefore, the choice of braid pattern is usually driven by the material requirement , such as component dimensions and shape, mechanical properties and preform stability.

Figure 1.6: Braided preform architectures. Images generated with TexGen software, University of Nottingham (2019)

The main advantages of braiding technology are high rate of strand deposition on the mandrel, ability to produce complex shapes, low capital investment cost and minimal labour cost compared to other manufacturing processes. The most important disadvantage is the difficulty to produce low braid angle preforms, Fedro and Willden (1992); Fouinneteau (2006).

There are several parameters that their variation can affect the in-plane and out-of-plane mechanical properties of the braided composites, such as braid angle, yarn size and axial yarn content. These changes in fibre architecture of textile composites were investigated for 2D triaxially braided, [Naik et al. \(1994b\)](#). *Naik* demonstrated that the mechanical properties are more sensitive to variability in braid angle than to variations in axial yarn content, and are not sensitive to change in yarn size. Increasing the braid angle decreases the longitudinal modulus but increases the transverse modulus, and the in-plane shear modulus values peak at the braid angle of $\pm 45^\circ$. An increase in the axial yarn content results in a higher longitudinal modulus, but a lower in-plane shear modulus and Poisson's ratio. The out-of-plane properties remain virtually unchanged with variations in the braid angle and the axial content.

The state of development of the two-dimensional triaxial braided composites is very poor, so their characteristics and benefits are currently in research. A summary of the benefits of this fabric architectures are the following, [Mouritz et al. \(1999b\)](#):

- Micro-cracks due to fatigue are stopped at the intersections of the reinforcing yarns, thus enabling this material to be easily adapted for use in the aerospace industry, [Mouritz et al. \(1999a\)](#).
- The efficient distribution of load in a braided structure enhances its impact resistance, [Staniszewski \(2007\)](#); [Roberts et al. \(2009\)](#); [Littell et al. \(2009\)](#).
- Braids improve interlaminar shear properties when combined together with other braids. Interlaminar adhesion in braided structures enables the layers to move together, which helps to prevent cracks, [Kelkar and Whitcomb \(2009\)](#).
- As braids are woven on a bias, they provide very efficient reinforcement for parts that are subjected to torsional loads. Braids can be easily and repeatedly expanded to fit over moulding tools or cores. They can accommodate straight, uniform cross-section forms as well as nonlinear, irregular cross-section components.

3D braided preforms have a number of important advantages over many types of 2D fabric preforms and prepreg tapes, including high levels of conformability, drapability, torsional stability and structural integrity. Furthermore, the 3D braiding process is capable of forming intricately-shaped preforms and the process can be

1.3. Braided fabrics

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varied during operation to produce changes in the cross sectional shape as well as to produce tapers, holes, bends, and bifurcations in the final preform. Despite these advantages, the applications for 3D braided composites have been limited for the following reasons, [Mouritz et al. \(1999b\)](#):

- Almost all 3D braiding machines are still under development and most of them are only capable of producing narrow preforms.
- 3D braiding machines have long set-up times and are slow and expensive.
- The spools in 3D braiding machines are small because they are continuously moving in the production of the preform; consequently production runs between machine set-ups are brief.
- Many of the mechanical properties of 3D braided composites have not been investigated yet.
- In-plane stiffness and strength of 3D braided composites are generally lower than 2D laminates.
- Scaling the results from mechanical test performed on small 3D braided composite specimens to large braided structures is difficult because of edge effects.
- The mechanical properties of braided composite structures are highly scattered.
- The predictive models for determining strength and fatigue performance have not been developed.
- The durability and long-term environmental ageing test on 3D braided composites have not been performed yet.

Three-dimensional braided preforms are classified based on various parameters depending on the yarn sets, yarn orientation and intertwining, micro and meso unit cells and macro geometry. These braid architectures have multiple layers and no delamination due to intertwine-type out-of-plane interlacement, high levels of conformability, drapability, torsional stability and structural integrity. However, the 3D braided fabrics have low transverse properties and they also have size and thickness limitations, [Tong et al. \(2002\)](#); [Bilisik \(2016\)](#). Almost all 3D braiding machines are still under development, these machines have long set-up time and,

they are slow and expensive. Many of the mechanical properties of 3D braided composites have not been investigated yet, so predictive models for determining strength and fatigue performance are underdeveloped. Three-dimensional braided composites can be grouped in three families: 3D braided fabrics, 3D axial-braided fabrics and multiaxis 3D braid. Besides, there are several techniques to manufacture 3D braided preforms: Four-step braiding method and Rotary braiding for 3D braided family, Maypole braiding method, four-step braiding and two-step braiding method for 3D axial-braided family and, finally, rotary braiding method and six-step braiding method for multiaxis 3D braided fabrics, [Bilisik \(2013\)](#).

1.4 Objectives and structure of the thesis

1.4.1 Objectives

The main objective of the research work presented in this thesis was to develop an efficient modelling framework for predicting the constitutive behaviour of triaxial braided composites. The modelling was intended to incorporate details of yarn architecture and its impact on the mechanical properties, while being applicable and practical for strength design of braided composite components. The core of the framework was multi-scale-modelling, including efficient unit cell models and macroscopic methods for structural simulation. Furthermore, the use of experimental techniques for input property determination and model validation were required. The major goals of the thesis can be summarised as:

- Characterisation of yarn architecture: development of an experimental method for robust internal geometry measurements of braided composites.
- Mechanical characterisation: experiments to determine the constitutive and failure behaviour of triaxial braided composites under combined stress states.
- Unit cell modelling: development of an efficient and parametric finite element unit cell modelling approach, typically required to predict the mechanical behaviour for a multitude of yarn architectures and load cases. Prediction of in-plane stiffness and strength and reproduction of fracture damage mechanisms.

1.4. Objectives and structure of the thesis

- Macroscopic modelling: simulation approach: formulation of an efficient and practical modelling approach, which is applicable for fracture simulation of large braided composite components.
- Validation and demonstration: numerical evaluation to ascertain and evaluate the applicability of the proposed modelling approach in structural and impact applications.

1.4.2 Structure of the thesis

The thesis follows the following structure with the purpose of accomplishing the mentioned objectives:

- **Chapter:2 State-of-the-art literature review.** Chapter 2 presents a literature review of technology and development of textile composites focused in the application to braided composites. The most relevant aspects of this fibre architecture in terms of geometry modelling methodologies and mechanical tests are presented.
- **Chapter 3: Characterisation and testing of braided composites.** In Chapter 3, the possibilities offered by braided composites in terms of loading capacity are evaluated. Mechanical tests at coupon level to determine the mechanical properties and to study the relevant damage mechanisms are carried out. Chapter 3 also covers the investigation of size effects presents at both meso- and macro-scale.
- **Chapter 4: Modelling the mechanical behaviour of braided composites.** In Chapter 4, an efficient and practical modelling approach for triaxially braided composites is developed and validated for elastic behaviour. The methodology includes the information of the fibre architecture and matrix regions.
- **Chapter 5: MMS modelling for the simulation of elasticity and strength of braided composites.** Chapter 5 covers the numerical analysis using the selected modelling methodology present in Chapter 4. Analysis to determine elasticity and strength of braided composites under tensile loadings are presented.

- **Chapter 6: MMS modelling for simulation of fracture of braided composites.** In Chapter 6, the MMS method is employed to predict fracture propagation in triaxially braided material using Compact Tensions (CT) specimens. The predictions are validated with the experimental results obtained in Chapter 3.
- **Chapter 7: MMS modelling for the prediction of notched strength and size effects in braided composites.** Chapter 7 covers the evaluation of the methodology presented in Chapter 4 and used in Chapter 5 and 6 under structural loading conditions using OHT specimens. Geometrical size effects are studied and their impact in the mechanical behaviour are explored numerically. The predictions are validated with the experimental results obtained in Chapter 3.
- **Chapter 8: Conclusions and future works.** This thesis concluded in Chapter 8 with an exposition of the lessons learned from the work reported in the previous chapters and a perspective on the research to be carried in the future based on the present contribution to the knowledge of the behaviour of braided composites.

State-of-the-art literature review

2

Composites made with textile fabrics can be classified as "traditional composites" because of their use over many years, and have remained popular because of their good in-plane mechanical properties when compared to traditional unidirectional composite laminates.

In textile composites, the fibres are the main components controlling the composite stiffness and strength in the fibre direction. The primary function of the polymer matrix is to give the composite strength in the fibre transverse and shear direction by binding the fibres together when loaded off-axis. Polymers are rate-dependent, nonlinear, viscoelastic materials that behave differently when tested under tensile or compressive loading rates and temperatures, [Goldberg et al. \(2003\)](#). Since polymers are rate-dependent materials, they account for the main differences found when loading composite materials. Having both the properties of the fibre fabric and polymer matrix, micro-mechanics based equations and techniques such as classical laminated plate theory (CLPT) can be used to determine the effective properties of the composite material. However, it is very complicated and many times impossible to measure the necessary fibre properties directly, so the properties of the composite material are measured instead. In addition, by testing composite material coupons, insight about what is happening can be gathered by evaluating the mechanical response of the constituents within the composite material. The knowledge gained from testing can provide valuable information about the fibre-matrix interface and the ways in which the interactions between the fibre and matrix affect the material response, [Littell \(2008, 2013\)](#).

2.1 Mechanical properties of textile composites

Generally, when material properties of a full textile composite material need to be known, standardised tests can be used. ASTM D3039, [ASTM International \(2000\)](#), gives specific guidelines in testing unidirectional or balanced composite laminates in tension; ASTM D7078 provide guidelines in measuring shear properties on V-notched composite coupons, [ASTM International \(2005\)](#); ASTM D5766 give guidelines to determine the tensile strength of open hole composite coupons, [ASTM International \(2011\)](#). ASTM suggests using either bonded strain gauges or extensometers to compute and measure the strain field and load cell to record the force applied on the test machine. These standards call for rectangular-shaped specimens geometries and were designed to give basic material properties of unidirectional or balanced laminates, such as strengths and stiffness, Poisson's ratio, etc., and are widely used in the industry. However, when the material is not homogeneous at meso-scale level, the scale of the fibre bundles, the geometries considered in the standards may not represent the overall behaviour of the textile material and could trigger some local effects caused by the in-homogeneity of the material, [Falzon and Herszbergb \(1998\)](#). Consequently, several authors, [Masters and Portanova \(1996\)](#); [Cox \(1999\)](#), developed their own geometries or modify the methods proposed for the standards to measure the mechanical performance of the textile composites. Other approaches are not standardised, such as the guidelines proposed by [Pinho et al. \(2006b\)](#) to determine the tensile and compressive fracture toughness of unidirectional composite laminates.

2.1.1 In-plane properties

Textile composites are characterised by exhibiting good stability and shear deformation resistance, produced by the crimp due to the highly interlaced and tightest fibre architecture. Depending of the type of the fibre architecture used in the laminate lay-up, the crimp factor will lead to better or worse in-plane mechanical properties compared to plain fabrics. Furthermore, the type of interlacing affects also to the flexibility, shear deformation resistance and tensile and flexural strengths, which allows their effective utilisation in the aerospace and automotive industry. In addition, their preforms exhibit durability, strength, reliability and structural integrity. Multilayer textile preforms are designed to overcome low delamination resistance and in-plane shear properties of traditional laminated composites, [M. Aly \(2017\)](#)

2.1.2 Out-of-plane properties

Notwithstanding, composites made from laminated fabrics have low through-thickness mechanical properties, poor impact damage tolerance, and low interlaminar fracture toughness, [Mouritz et al. \(1999a\)](#). [Compston and Jar \(1999\)](#), reported that the interlaminar fracture toughness defined by Mode I critical strain release rate of vinylester-based composites reinforced with woven roving fibreglass are between 0.5 and 1 kJ/m^2 , and therefore susceptible to delamination cracking under impact loading.

Another problem with architectural composites is their poor drapability in comparison with traditional unidirectional laminates, which makes it difficult and expensive to manufacture complex composite structural parts to the near-net shape. This combination of low interlaminar fracture toughness (resulting in poor impact damage tolerance) and potentially high production costs has contributed to the difficulty in implementation on structures subjected to critical loads for the automotive and aircraft industry.

Measurement of interlaminar fracture toughness is necessary due to the significant impact of the delamination effect into the integrity of the composite. Understanding of the effect of fibre volume fraction on a laminate's fracture toughness is important because there can be variations in the resin distribution. A common technique to raise the Mode I interlaminar fracture toughness is by using toughened resins, [Compston and Jar \(1999\)](#), nonetheless these are expensive and do not improve the drapability of the fabric. Another approach is to use advanced textile architectures such as knitting, braiding, through-the-thickness stitching and weaving. Textile composites made by these techniques have higher impact damage tolerance and in some architectural cases, outstanding drapability as mentioned before. Their multiple potential applications are due the growing number of structural components being manufactured, including wings and propellers and car chassis.

Despite their outstanding potential use in structural applications, the Mode I interlaminar fracture toughness properties for braided composites have not extensively evaluated. Most of the studies were directed to demonstrate that carbon fibre reinforced epoxy stitched composites can substantially improve the critical energy release of composites made from unidirectional or woven roving fabrics. [Briscoe et al. \(1993\)](#), investigated the effect of weave type, areal weight and surface texture on the mode I interlaminar fracture toughness (G_{Ic}) for aramid-fibre/epoxy com-

2.1. Mechanical properties of textile composites

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posites. They found that although weave type has little effect, larger values of G_{Ic} were observed for fabrics that contained higher densities of fibre-ends created by abrasion of surface and for fabrics with coarser weave patterns. Additionally G_{Ic} was found to increase with increasing crack length, behaviour attributed to fibre bridging. [Alif et al. \(1998\)](#), investigated the effect of weave structure and interface tow orientation. They found that while R-curves for plain weave were almost horizontal, indicating no change in the fracture mechanisms during crack propagation, twill and satin weaves had R-curves which exhibited an initial increase. This was attributed to the increase of fibre bridging in comparison with plain weaves, although the extent of fibre bridging was limited by the interlacing of the weave pattern and in all the cases, the R-curves exhibited a steady-state plateau toughness. Higher toughness was observed for crack propagation between surfaces with the majority of tows adjacent to the interface oriented at 90° to the crack direction than for predominately 0° surfaces. This was attributed to transverse tow delamination in the 90° surface pinning the crack and causing it to arrest, thereby increasing toughness. This does not occur in the 0° surfaces due to the constraint imposed on the transverse tows by the interlaced longitudinal tows.

[Bradley and Cohen \(1985\)](#), argued that the necessary resin plastic deformation zones are constrained by the fibres and cannot fully develop, consequently limiting toughness. Thus, it was concluded that the optimal thickness of a resin interply region depends on the type of resin used. Ductile resins would lead to maximum toughness with a thicker zone, whereas, for brittle epoxy resins, a very thin zone would be optimal. [Russell \(1987\)](#), studied the influence of local interlaminar fibre distribution on the fibre bridging for graphite-fibre/epoxy laminates and showed that the extent of fibre bridging is increased as the plies are brought together. [Hunston et al. \(1987\)](#), noted that the mode I fracture toughness of a number of laminates made of identical materials but manufactured by two different organisations (NASA and Hexcel) were different. Samples manufactured by NASA were around a 40-100% tougher than those by Hexcel. This was caused because the process followed by NASA resulted in an increase of fibre nesting and intermingling between layers which did not occur with Hexcel fabricated laminates.

[Gill et al. \(2009\)](#), investigated the effect of variations in fibre volume fraction on the mode I and mode II delamination behaviour for 5 harness satin (5HS) woven carbon-fibre/epoxy resin composites manufactured by resin transfer moulding. It was found that the toughness increased with the increase of fibre volume frac-

tion and that a strong correlation between fracture surface damage and measured interlaminar fracture toughness was observed.

2.1.3 Notched strength

Two-dimensional braided composites have been widely used in thin-walled structures recently, [Hwan et al. \(2011\)](#), it is thus very important to be able to predict the notched strength of 2D braided composite plates before they can be used as structural members for mechanical connection.

In order for braided composites to be utilised as structural components, it must be possible to connect braided members to other structural components. The introduction of holes in braided structures results in stress concentrations, albeit lower than those experienced in parts with post-production machines holes, that can affect the loading behaviour and loading-bearing capabilities of braided composites. Standard test methods are required to investigate the effect of stress concentrations and bolted connection on braided composites.

The open hole tensile test is important in deriving allowable stress levels for use in component design since open hole strength can be a limiting factor. It is also one of the simplest tests that combines unavoidable geometrical interaction with material properties. Thus it is on the border between a true material test and a simple or small scale structural test. For this reason, it is also sometimes used for material ranking to give an early indication of material's structural performance. Although standards exist for open hole tensile strength, results are strongly dependent on both layup and testing configurations eg. the well known hole (notched) size effect. This creates difficulties for designers and modellers alike since small changes in parameters will affect data values produced. There is thus an increased requirement for virtual testing to avoid expensive tests but paradoxically many predictive techniques require empirically derived calibration factors.

Several authors, [Lagace \(1986\)](#); [Naik et al. \(1990\)](#); [Shembekar and Naik \(1992\)](#); [Naik and Shembekar \(1992b\)](#), investigated the notch sensitivity and the influence of the stacking sequence on the unnotched and notched strength of fabric laminates. They pointed out the importance of the size effects, notch diameter to fabric architecture and layer configuration, onto the mechanical behaviour of fabric composite laminates. Therefore, this section aims to investigate the size effects onto the notched strength of braided composite plates.

2.1. Mechanical properties of textile composites

The state of development of two-dimensional triaxial braided composites is very poor, implying that this textile architecture is still under research. This composite fabric is more in-homogeneous than traditional unidirectional angle-ply laminates and mechanical analyses are more complicated. It can be observed that open-hole is one of the most critical defect within the geometry of a composite, producing high stress concentrations around the edge of the hole evolving to damage in the form of cracks, which nucleate and propagate towards the edge of the coupon following the fibre direction.

2.1.4 Size effects on textile composites

The introduction of textile composite materials in new applications relies on the development of accurate analytical and computational tools that are able to predict the thermo-mechanical response of composites under general loading conditions and geometries, [Camanho et al. \(2007\)](#). In the absence of accurate analytical models, the process has to rely on costly matrices of mechanical tests based on large numbers of test specimens, [Department of Defense United States of America \(2002\)](#), and empirical knock-down factors.

The prediction of ultimate strength remains the main challenge in the simulation of the mechanical response of composite materials, [Hinton and Soden \(1998\)](#). Most of the work has been centred in unidirectional composite laminates, in which the simulation of size effects on the strength of composites is of particular interest and relevance, [Green et al. \(2007\)](#); [Bazant et al. \(1996\)](#); [Wisnom \(1999\)](#); [Dvorak and Suvorov \(1999\)](#); [Bazant et al. \(2004\)](#): reliable analytical and numerical models must represent the decrease of the ultimate strength when the structural dimensions increase, [Bazant \(1999\)](#).

Size effects in laminated composites occur at different material and structural levels. At the meso-scale level, it is observed that the transverse tensile and in-plane shear strengths of a ply constrained by sub-laminates depend on the ply thickness, [Parvizi et al. \(1978\)](#). This size effect is normally called the *in-situ* effect and can be accounted for in the prediction of matrix cracking onset using the *in-situ* strengths in appropriate failure criteria. The *in-situ* strengths can be calculated from analytical closed-form solutions using ply elastic properties and fracture energies, [Reddy and Pandey \(1987\)](#); [Camanho et al. \(2006\)](#).

Size effects also occur at the macro-mechanical level. For example, the strength of notched quasi-isotropic composite laminates decreases for increasing notch sizes

when thin plies are used, [Waddoups et al. \(1971\)](#). This effect, usually known as the "hole size effect", is caused by the development and propagation of non-critical ply-level damage mechanisms that occur in the vicinity of the hole before the final collapse of the laminate. This damage mechanism is caused by a fracture process zone that develops before final failure of the laminate. For very small specimens, the fracture process zone affects the entire width of the laminate, whereas for larger specimens, the size of the fracture process zone is negligible when compared with the characteristic dimensions of the specimen. The relative dimension of the fracture process zone with respect to the specimen size justifies the different strengths observed in small and large specimens. Thus, to predict the hole size effect in quasi-brittle materials, methods that account for the energy dissipated by the propagation of non-critical damage mechanisms are required, [Bazant \(1999\)](#).

The calculation of macro-mechanical size effects is often based on semi-empirical methods that require calibration such as the point stress and average stress models proposed by [White \(1982\)](#), or models based on continuum damage mechanics, which do not require calibration, and could potentially provide the means for a truly predictive methodology for the strength prediction of composite laminates.

2.2 Yarn architecture of textile composites

Textile composites are hierarchical materials: the material definition has to be described on different length scales which are based on each other: the fibres in the yarns on the micro-scale, the internal geometry of yarns and fabric unit cells on the meso-scale, and the textile reinforcement on the macro-scale. Each scale has its characteristic length, 1-10 μ m for the fibres, 1-30mm for the fabric the unit cell, and centimetres to several meters for the composite structures, [Lomov et al. \(2001\)](#); [LOMOV and VERPOEST \(2005\)](#); [Potluri et al. \(2003\)](#); [Cichosz \(2015\)](#).

It should be noted that the term meso-scale is not consistently used in literature. Some researches, [Lomov et al. \(2001\)](#); [Potter et al. \(2012\)](#); [Crookston et al. \(2007\)](#), define the field of composite unit cell mechanics as meso-mechanics, the same type of models are defined as micro-mechanical in other publications, [Goyal \(2007\)](#); [Qu et al. \(2012\)](#); [Melro \(2011\)](#). [Ladevèze et al. \(2000\)](#) define single (unidirectional) plies in composite laminates as meso-scale. Throughout this thesis, the definition of [Lomov et al. \(2001\)](#) given in Table 2.1 will be used.

The term "yarn architecture" refers to the internal geometry of a textile com-

2.2. Yarn architecture of textile composites

Table 2.1: Definition for micro-, meso-, and macro-scale used.

micro	scale of fibre and matrix
meso	scale of yarns and fabric unit cells
macro	scale of textile composites and composites structures

posite described on the meso-scale. The yarn architecture of a braided composite comprises three components, as shown in Figure 2.1:

- Yarns, consisting of several thousand filaments impregnated with matrix.
- Matrix pockets, regions of pure matrix in-between the yarns.
- Voids: intra- or inter-yarn voids.

Figure 2.1: Optical cross-section of the internal geometry of a ($\pm 60^\circ$) triaxial braided composite

The architecture of a textile composite is complex and it is influenced by several parameters such as geometrical volume fractions, local fibre volume fraction, nesting, etc. Geometrical volume fraction, or the space occupied by the yarns in the laminate can be calculated from optical inspection over a region of interest (ROI) of the composite. Local fibre volume fractions can be also calculated by measuring the space occupied by the filament within an impregnated yarn, while global fibre

volume fraction can be defined as the total volume of fibre in the composite. These parameters affect the shape and mechanical response of the composite.

In the following subsection, an overview about previous works regarding the yarn architecture of textile composites with a focus on woven and braided composites will be given.

2.2.1 Volume fractions

Volume fractions can be defined according to the components in a textile composite, [Qu et al. \(2012\)](#). The total fibre volume fraction (φ_f) is the volume of fibres in the composites, while yarn volume fraction (φ_Y) describes the volume of yarns in the composite and can be considered as a measure of the tightness. Additionally, the packing density (p_d) describes the fibre volume fraction inside the yarn. These can be described by the following:

$$\varphi_f = \frac{V_f}{V_{ges}} \quad (2.1)$$

$$\varphi_Y = \frac{V_Y}{V_{ges}} \quad (2.2)$$

$$p_d = \frac{V_{f,Yarn}}{V_Y} = \frac{\varphi_f}{\varphi_Y} \quad (2.3)$$

The total fibre volume fraction can be measured directly from braided composites by using a variety of methods, such as digestion of the matrix, [ASTM International \(1999\)](#), or resin ignition, [International \(2002\)](#). The packing density can be calculated from the yarn area (measured, e.g. from micrographs) if the diameter and the number of filaments is known, [Birkefeld et al. \(2012\)](#). Typical values of the packing density have been reported to vary from 60% to 80% [Birkefeld et al. \(2012\)](#); [Byun \(2000\)](#); [Ivanov et al. \(2009\)](#) for braided composite fibre volume fractions of 50-60%.

2.2.2 Yarn cross-section and yarn paths

The shape of the yarns in a textile composite is usually described by using the yarn cross section, cross sectional orientation, and yarn path, [Verpoest and Lomov \(2005\)](#). Commonly elliptical, [Qu et al. \(2012\)](#); [De Carvalho et al. \(2012\)](#), and lenticular, [Byun \(2000\)](#); [Huang and Ramakrishna \(2002\)](#); [Matveev \(2015\)](#), cross sections have been reported in literature. [Birkefeld et al. \(2012\)](#) noted that both shapes of cross sections can be seen in micrographs of braided composites. While the braiding

2.2. Yarn architecture of textile composites

yarns typically manifest a mixture of both cross sections, elliptical cross sections dominate the axial yarns of triaxial braids. [Byun \(2000\)](#) reports consistent and regular shapes for the axial yarns in a triaxial braid, while the shapes of the braiding yarns were irregular. The cross sections of the braiding yarns located on the surfaces of the specimens tended to flatten. This is in agreement with the observations from [Birkefeld et al. \(2012\)](#), where the flattening was considered to be due to the contact of the yarns with the solid mould or the vacuum bag. [Ruijter et al. \(2007\)](#); [Ruijter \(2009\)](#) found that the average shape is close to lenticular, by applying an averaging routine to several cross section images obtained from micrographs of a woven fabric. It was noted that the yarn geometry was variable (aspect ratio between 1:8 and 1:15) and that the automated image averaging of cross sections only quantifies the variation, but cannot provide information about the possible reasons (e.g. yarn crossover).

The yarn path can be divided into crossover regions (straight regions, floats: AB, CD, EF in Figure 2.2), where the yarns run over/under the crossing yarns, and undulation regions (BC, DE in Figure 2.2), where the yarn runs from the upper to the lower surface and vice versa, [Qu et al. \(2012\)](#).

Figure 2.2: Division of the yarn path, [Qu et al. \(2012\)](#).

Different parameters are used to characterise the magnitude of undulation: the crimp angle is the maximum out-of-plane angle of the yarn path inside the undulation region, [Cox and Flanagan \(1997\)](#); the yarn crimp ratio c given in Equation 2.4 is most commonly defined by comparing the crimped yarn length l_{yarn} , to the length of the fabric l_{fabric} , [Adanur \(2001\)](#). It should be noted, that different definitions for the yarn crimp ratio have been described, e.g. by [Ivanov et al. \(2009\)](#). [Owens et al. \(2010\)](#) use the waviness ratio WR , which is the thickness of the ply, h_{ply} , divided through the wavelength of the undulation, l_{fabric} .

$$c = \frac{l_{yarn}}{l_{fabric}} - 1 \quad (2.4)$$

$$WR = \frac{h_{ply}}{l_{fabric}} \quad (2.5)$$

Besides average descriptions of yarn shape and path, considerable effort has been undertaken to describe of variability of the yarn architecture. [Vanaerschot et al. \(2013\)](#) presented a stochastic framework based on the period collation method for the analysis of variations in the yarn path and yarn cross sectional parameters based on micro-CT measurements. He aimed to calculate systematically and stochastically the variations of the yarn path from the geometric properties measured. It was observed that out-of-plane waviness and cross sectional parameters vary systematically, dependent on the relative positions on the yarn path, i.e. these values do not depend on the lateral and laminate position. However, this is different for the in-plane variation of the yarn path, but a reliable analysis of this would require a larger specimen. [Matveev \(2015\)](#) presented a similar evaluation method for variability of the yarn paths in a 2x2 carbon-epoxy woven fabric. It was found that the variations in the out-of-plane coordinates of the yarn path were of the same magnitude as the resolution of the micro-CT images used for evaluation.

2.2.3 Nesting

Nesting is a phenomenon typically described for textile composite laminates, [Birkefeld et al. \(2012\)](#); [Lomov et al. \(2003\)](#); [Schultz and Garnich \(2013a\)](#); [Chen and Chou \(2000\)](#); [Chen et al. \(2001\)](#). When several plies of a textile are stacked on top of each other to build a laminate and compacted in a vacuum or RTM process, the cured ply thickness in the laminate decreases with increasing number of plies. This effect is described as nesting and leads to an increase of the laminate fibre volume fraction (compared to a single ply) and is mainly attributed to two mechanisms, [Chen and Chou \(2000\)](#):

1. If textile composite plies are stacked on top of each other, the discrete yarn architecture comprises regions with free volume within one ply, which is filled by the adjacent plies. [Lomov et al. \(2003\)](#) presented a purely geometrical approach to model this.
2. The dry yarns deform under the compressive force applied to the preform during compaction. This may lead to yarn flattening or yarn cross section deformation (Figure [2.2.3](#))

The degree of nesting, η , is given by the following equation:

2.2. Yarn architecture of textile composites

Main factors	Uncompressed	Compressed
Yarn cross-section deformation		
Yarn flattening		
Yarn bending deformation		
Void / gap condensation		
Nesting		

Figure 2.3: Factor affecting fibre volume fraction, [Chen and Chou \(2000\)](#).

$$\eta = \frac{N_{plies} t_{ply}}{h} \quad (2.6)$$

where N_{plies} is the number of plies in the laminate, and t_{ply} and h are the thickness of a single ply and the laminate, respectively.

[Chen et al. \(2001\)](#) conducted compaction experiments with different textile reinforcements and found that the nesting of the laminate reached a stable state for more than 10 plies. The difference in ply thickness in laminates between 10 and 25 plies is negligible. Numerical studies regarding geometric nesting, [Lomov et al. \(2003\)](#), show that closely packed fabrics are less prone to nesting. The tightness, T , can be defined as a measure for this:

$$T = \frac{d_1 + d_2}{2p} \quad (2.7)$$

with d_1 , d_2 being yarn height and width and p being the yarn spacing. It has also been shown that larger patterns (i.e. 2X2 compared to 1X1, [Figure 1.1](#) and [Figure 1.6](#)) and shear deformation patterns (braiding angles $\theta \neq 45^\circ$ for braids), reduce the nesting. This has also been measured experimentally for biaxial and triaxial braided composites with 30-60° braiding angles, [Birkefeld et al. \(2012\)](#).

2.3 Modelling and simulation of Textile Composites

Textile composites are characterised by a complex internal geometry which is difficult to model. A variety of analytical and numerical model exist to predict the constitutive behaviour of textile composites. Their predictions offer advantages in comparison with the traditional testing to study the mechanical properties, type of reinforcement and composition. The models can be grouped into three categories:

1. Analytical methods, eg., Iso-Strain/Stress, Mori-Tanaka, and methods of cells.
2. CLT-Methods: classical laminate theory is adapted to model the textile reinforcement.
3. Finite element unit cell models: the yarn architecture is modelled within an unit cell.

The models require different depth of input and numerical effort and offer different levels of detail regarding the predicted response. Analytical models can include detailed information about the yarn-architecture geometry and provide a very efficient solution. However, such models often suffer problems in correctly representing the stress and strain fields, which are important to reproduce the failure and damage mechanisms, [Prodromou et al. \(2011\)](#). In comparison, FE unit cell models provide detailed stress and strain fields and can include constitutive models for plasticity and damage in the matrix and yarns. However, these models require a considerable number of input properties, can be very complex to implement and their computational cost is high, [Stig and Hallström \(2012a\)](#). Computational efficiency is the biggest advantage of the CLT-models, which enable the implementation of material constitutive equations and can be used to solve structural problems using FE methods, [García-Carpintero et al. \(2017b\)](#). The following section gives an overview of models for textile composite constitutive behaviour prediction, with a focus on models for braided composites.

2.3.1 Analytical models

Extensive studies have been conducted to investigate mechanical properties of textile composites, [Miravete \(1999\)](#). Among textile composites woven composites are mostly used for structural applications. Hence most of the studies were carried out in this area are related to this type of composites. *Ishikawa and Chou* developed

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Figure 2.4: Mosaic, fibre undulation and bridging model, [Ishikawa and Chou \(1982b\)](#)

three analytical models for two-dimensional woven composites based on Classical Lamination Theory (CLT), namely the *Mosaic Model*, *Fibre Undulation Model* and *Bridging Model* (Figure 2.4), [Ishikawa and Chou \(1982a,b, 1983\)](#); [Chou and Ishikawa \(1983\)](#). The mosaic model neglected undulation, and used cross-ply laminates assembled under the assumption of iso-strain and iso-stress to model woven fabrics. The fibre undulation model calculates the stiffness knock-down due to yarn crimp in a woven fabric by modelling the out-of-plane waviness of the yarn. The bridging model can be described as a combination of the mosaic and fibre undulation models. When compared with experimental data, the mosaic model provides a good estimation of the elastic properties, the undulation model is useful to model plain weaves and the bridging model is desirable to low crimping woven fabrics.

Extension and enhancements to the the model presented by [Ishikawa and Chou](#) were developed. [Naik and Shembekar \(1992a\)](#), evolved the fibre undulation model to a 2D representation using classical laminate theory. Later, [Masters et al. \(1994\)](#), presented an analytical model for predicting the mechanical properties of triaxially braided composites using a rule of mixture as a micromechanical model. [Dadkhah et al. \(1995\)](#), developed a modified laminate model based on Christensen's self consistent model, [Christensen \(1990\)](#), to predict the macroscopic in-plane elastic properties of triaxially braided composites. [Ruan et al. \(1999\)](#), presented an analytical methodology to predict the elastic, piezoelectric and dielectric properties of three-dimensional piezoceramic braided preforms.

Various extensions of iso-strain /iso-stress methods have been applied to braided composites including a more detailed geometrical description and some of them were extended to failure prediction. [Byun \(2000\)](#), developed an analytical model based on

the unit cell for the prediction of the geometric characteristics and three-dimensional engineering constants of two-dimensional triaxially braided composites. A micromechanical model was introduced by [Huang \(2000\)](#), for two-dimensional determination of mechanical properties of braided and woven composites. [Quek et al. \(2003\)](#), presented an analytical method in which an unit cell is considered representative of the composite material properties. In the model, the average stiffness of the unit cell is calculated by computing the contribution of each component and applying a volume average formula. [Shokrieh and Mazloomi \(2010\)](#), modified the rule of mixtures used in the calculation of the shear modulus by *Quek et al.* to improve the predictions. [Potluri et al. \(2006\)](#), presented an investigation of flexural and torsional properties of two-dimensional braided composites, with one or more layers, at different braid angles. [Tsai et al. \(2008\)](#), developed a parallelogram spring model to investigate the effects of fabric architecture parameters, such as braid angle, fibre volume fraction and the unit cell aspect ratio on the elastic properties of braided composite plates. An extensive overview and comparison of these models was given by [Hallal et al. \(2013\)](#).

Recently [Zhang et al. \(2014a\)](#), developed a three-dimensional analytical model to study the behaviour of triaxially braided composites. The model considered fibre undulation and architecture parameters obtained from optical microscopy of cross-sectional imaging. Results were compared with experimental data. Furthermore, [Kier et al. \(2015\)](#), developed an analytical model to estimate the in-plane mechanical response of two-dimensional braided composites of different architectures. The estimations compared with experimental data from uni-axial tensile and V-notched rail shear test showed good agreement.

Another approach to homogenisation of woven, braided and knitted textile composites using a Mori-Tanaka method was proposed by [Gommers et al. \(1998\)](#). The yarns within a fabric unit cell, were subdivided in smaller segments considering their fibre volume fraction, local curvature, orientation and elliptical or circular cross-section. [Lomov et al. \(2009\)](#), implemented this method into TexComp software, which is part of WiseTex package [Lomov et al. \(2001\)](#). Good agreement was found in comparing with experimental and Finite Element results, [Desplentere et al. \(2005\)](#); [Birkefeld et al. \(2012\)](#).

Not long ago, the use of MOC (the methods of cells) has materialised into analytical modelling of textile composites. [Prodromou \(2004\)](#); [Prodromou et al. \(2011\)](#), used the MOC to predict the stiffness and strength of a variety of 2D and 3D woven

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composites. The RVE (representative volume elements) was divided in sub-cells using the geometric description of WiseTex. The model was found to be computationally efficient and a promising correlation was achieved when compared to conventional methods and experiments. [Hutchins et al. \(2010\)](#), compared numerically two distinct braid architectures using the Generalised method of Cells (GMC): a) the "traditional" triaxial braid in which the axial tows are merely laid in between the woven biased ($\pm\theta^\circ$) tows and (b) the "true" triaxial braid in which the axial tows are interleaved through the biased tows. Good agreement was found between both model except for the axial modulus at higher braid angles.

2.3.2 Classical laminate theory methods

Classical laminate theory (CLT) was primarily designed for unidirectional composite laminates but it has been applied to textile composites using equivalent laminates that represent the architecture of the material, [Naik \(1994\)](#); [Cox and Flanagan \(1997\)](#).

[Smith and Swanson \(1994, 1995b,a, 1996\)](#), modelled the axial and bias tow failure with a strain criteria to predict the behaviour of triaxial braided composites. The fibre strain failure was computed from the composite strain and a maximum strain failure criteria was used under different load steps. The tow strain was obtained by correlation of test results showing good results.

A constitutive model for the prediction of non-linearities and damage in woven composites was presented by [Johnson et al. \(2001\)](#). A continuum damage-mechanics (CDM) model for fabric-reinforced composites derived from experiments was developed to model the failure during impact loading conditions. The CDM model was implemented in conjunction with a one-dimensional plasticity law under shear, in a commercial explicit finite element (FE) code, [ESI Group \(2018\)](#)). Tow straightening and delamination were included by [Fouinneteau \(2006\)](#), in a similar model also implemented in PAM-Crash, [ESI Group \(2018\)](#). However, the coupling between inter- and intra-ply damage was not considered due to the lack of experimental data.

[Goyal \(2003, 2007\)](#), investigated the effect of various parameters like braid angle, waviness ratio, stacking sequence and material properties on the effective engineering properties of the 2x2 braids. Equivalent laminated materials with angle plies and a resin layer were analysed to compare the difference in predictions from the full three dimensional finite element analysis of the 2x2 braided composites (Fig 2.5a). Comparison showed only minor deviations, exposing the potential of the CLT mod-

Figure 2.5: Classical laminate theory methods: a) Equivalent tape laminate including matrix layer, [Goyal \(2007\)](#); b) Sub-cell approach for triaxial braided composites, [Xiao et al. \(2011\)](#)

elling approach. Nevertheless, A stiffer response was obtained in the equivalent tape laminate because the fibre undulation was not considered.

[Zebdi et al. \(2009\)](#), presented an inverse modelling approach based on CLT to calculate the mechanical properties of braided composites. The algorithm calculates properties of a $\pm\theta$ angle-ply laminate made out of two unidirectional plies which is equivalent to a woven or braided composite oriented at $\pm\theta$. The virtual angle-ply and cross-ply laminates obtained by inverse calculation include the effects of undulation and strand shear, and therefore differs significantly from those obtained for UD plies. Applicability to braided preforms with different braiding angles is demonstrated.

[Cichosz et al. \(2016\)](#), proposed a model for failure prediction of $\pm 30^\circ$ biaxial braided composites considering material properties knock-down factors produced by the yarn waviness. The elastic parameters were obtained by the application of the inverse modelling approach proposed by [Zebdi et al. \(2009\)](#). The strength parameters used in the model were obtained from a calibration process based on conducted experimental tests. Plane-stress Puck criterion was used to represent the yarn failure obtaining an accurate prediction of the failure.

Another approach consist on modelling the macroscopic architecture divided into subcells with different lay-ups in FE calculations (Figure 2.5b), [Xiao et al. \(2011\)](#). Usually, two types of subcells, comprising and not comprising axial yarns,

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Figure 2.6: Schematic of the triaxial braided subcell model proposed by Littell (2008): a) triaxial braid unit cell showing discretisation into four subcells and b) Subcell discretisation of six-layer triaxial braided composites.

are distinguished for triaxial braided composites. The fibre tows are represented by plies of unidirectional (UD) composite. A laminate consists of UD composite plies corresponding to axial tow and braider tow, and plies representing the matrix phase.

Several authors, Littell (2008); Littell et al. (2009); Roberts et al. (2009); Littell (2013), used the measured two dimensional triaxial braided composite properties obtained from testing to create a macro-mechanical composite computer model capable of accurately simulating composite material response under static loading conditions. The braid geometry was divided into four subcells and stacking the layers of braid using a through-thickness integration technique, Figure 2.6. The unit cell of the braided composite is modelled as a series of shell elements, where each element is modelled as a laminated composite. The properties needed for the material model were obtained from the coupon tests performed using the optical measurement data and classical composite micromechanical techniques. The created computer model predicted strength and stiffness with 10 percent error for the static tests and offered a prediction of the a ballistic penetration threshold. The advantage of the model is that any number of fibre layers, any fibre angle, and thicknesses can be included to study the performance of triaxial braided composites.

A more detailed approach to meso-modelling the structure of textile composites consists in discretise the entire fabric unit cell geometry by one and two dimensional elements. These two-dimensional structural elements lead to a substantial increase

Figure 2.7: Shell elements based unit cell (top) disassembled into idealised tows (left) and matrix regions (right) for a 2/2 twill weave composite, [Gager and Pettermann \(2012\)](#)

in the fidelity of the model, enabling computationally low-cost simulations. This strategy considers geometric idealisations to represent weaving patterns and uniform cross-sectional bundle shapes. However, only two-dimensional stress states are taken into account, not being the the interply (delamination) damage introduced in the simulations. Moreover, the failure propagation cannot be traced independently at the interface between the plies, so the results are typically mesh-dependent. [Sidhu et al. \(2001\)](#), developed a numerical approach for stamping analysis of a plain-weave textile composite. The mesh was generated with truss elements in combination with shell elements. The truss elements represented the fibre bundles and the shell elements were used to represent the inter-tow friction and fibre angle jamming. Numerical results showed a good correlation with results from the experiments. [Gager and Pettermann \(2012\)](#), represented the idealised geometry of a 2/2 twill weave using shell elements. The impregnated tows and the pure resin pockets within the unit cell volume are treated as UD laminates represented by shell elements, Figure 2.7. The shell thickness was varied appropriately to represent the simplified tow cross-section and matrix regions. Results from this modelisation strategy compared well with a three-dimensional discretisation of the same weave.

2.3.3 3D finite element unit cell modelling

Finite element (FE) unit cell modelling is another common approach used to examine the elastic and nonlinear behaviour of textile and braided composites. Finite element analysis (FEA) models can be used to study the stiffness, 3-D stress-strain fields, damage development and failure of textile composites. FEA models consists on a suitable geometric description of the yarn architecture of a representative unit cell (RUC). In a FE unit cell model, the depicted geometry of yarn and matrix pockets is imported into a finite element commercial-code, meshed and solved using periodic boundary conditions (PBC's). All FEA models focus on representing the textile geometry as a representative volume elements (RVE), which is the smallest subvolume that represents the bundle architecture. Due to the complexity of the textile architecture, high efforts for model generation and increased computational costs are required, which are challenging for macroscopic modelling, [Li et al. \(2010\)](#). The ability to reproduce the textile geometry using computer aided design (CAD) tools makes them advantageous over traditional analytical models. Furthermore, FEA models are able to detail the stress-strain distribution through the textile geometry, making them suitable for failure and damage prediction, [Ji et al. \(2014\)](#). In addition, results can be compared with experiments using techniques such as the "*virtual field method*", [Pierron and Grédiac \(2012\)](#). [Lomov et al. \(2008c\)](#), advertised their potential application by comparing the strain field calculated by FE unit cell models to experimental results registered by digital image correlation and digital phase shifting grating shearography.

[Lomov et al. \(2007\)](#), outlined the steps necessary to model a three-dimensional textile unit cell:

1. Generate a bundle architecture geometrical model
2. Transfer the geometry to a volume representation
3. Correction of geometrical interpenetrations
4. Meshing of yarns and matrix regions
5. Assignment of local material properties and yarn orientation
6. Definition of the RVE and application of PBC's.
7. Homogenisation procedure

8. Damage initiation and propagation modelling

The stages described above refer to a full three-dimensional discretisation of the textile unit cell using continuum elements. Nevertheless, some of the stages may not be necessary attending to the unit cell modelling technique used. Some models do not require a correction for tow interpenetrations, [Cox et al. \(1994\)](#); [Jiang et al. \(2008\)](#), and others do not include damage propagation, [De Carvalho et al. \(2012\)](#); [Potluri and Manan \(2007\)](#). This section gives a summary of the most relevant approaches for finite element modelling of textile composites unit cell.

Commonly unit cells are modelled using three-dimensional continuum finite elements. Yarns and matrix pockets are usually meshed with tetrahedral or hexahedral volume elements and need to be connected by opposing nodes. Therefore, a consistent geometry description is required. However, the yarn deformation occurring during compaction and preforming of the fabric and the nesting between fabric is, in most cases, neglected. As a consequence, the resin rich areas between the yarns are bigger than in the real composite, due to the idealised geometrical model, [Olave et al. \(2012\)](#); [Grail et al. \(2013\)](#). These models usually have a limitation for the global fibre volume fraction that can be modelled, typically $\approx 40\%$, which requires small and often distorted elements of coarse quality, [Potter et al. \(2012\)](#). Therefore, in order to preserve the overall fibre volume fraction in the composite (typically 50-60 %), the fibre volume fraction in the yarns must be set to higher values than observed experimentally (sometimes close to 90% or higher, [Ivanov et al. \(2009\)](#); [Zhang et al. \(2014b\)](#)).

Yarn shapes closer to those in the real composite can be obtained from finite element modelling of the dry preforming, [Badel et al. \(2007, 2008\)](#); [Lin et al. \(2008\)](#); [Stig and Hallström \(2012b\)](#); [Nguyen et al. \(2013\)](#), in which the contact iterations between yarns are explicitly taken into account. In addition, it is also possible to determine the yarn shapes by means of X-rays computer tomography (XCT) inspection, [Schell et al. \(2006\)](#). In spite of the reinforcement geometries created by these methods can be used in meso-scale modelling of textile composites, the periodicity is not guarantee, which makes them (without post-treatment) not suitable for periodic boundary conditions often used in multi-scale modelling, [Woo and Whitcomb \(1997\)](#); [Le Page et al. \(2004\)](#); [Verpoest and Lomov \(2005\)](#); [Lomov et al. \(2007\)](#); [Badel et al. \(2007, 2008\)](#); [Lin et al. \(2008\)](#); [Ivanov et al. \(2009\)](#); [Nguyen et al. \(2013\)](#).

The generation of a finite element model is therefore, assumed to be complex.

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The geometry needs to be properly meshed. However, due to the complex yarn shapes in a deformed fabric and the multiple contact zones between the yarns, the generation of a finite element mesh of the reinforcement in conjunction with the matrix complement and a consistent treatment of the multiple yarn contact zones is far from being obvious. A solution often used is to insert a thin matrix layer between all yarns to create modified yarn surfaces, Long (2006); Lomov et al. (2008a); De Carvalho et al. (2012); Melro et al. (2012); Stig and Hallström (2012a). In this way, an accurate mesh without problems of yarn interpenetrations can be generated using meshing tools available in commercial finite element codes or meshing softwares. Nevertheless, the introduction of thin layers causes, in general, either bad quality elements or a very fine mesh size within the layers, and a reduction of the fibre volume fraction that does not correspond to the reality. A few authors have generated FE meshes of composites with preformed fabrics after manual mesh modifications or using meshing software packages, Loix et al. (2008); Xu et al. (2014, 2015). To overcome these problems, several methods have been proposed to correct the yarn volume interpenetrations characteristic of the idealised geometrical models, Lomov et al. (2007); Sherburn (2007); Tabatabaei and Lomov (2015).

For higher volume fractions, the matrix pockets in the RUC's become very small, which requires much smaller and distorted elements. Thus, to achieve realistic total volume fractions, alternative techniques including compaction simulation have been proposed, Grail et al. (2013); Hsu and Cheng (2013), but they require an additional simulation step, which increases the computational cost. Other approaches, Stig and Hallström (2012a), are based in a quasi-physical approach, in which the RUC geometry is inflated to increase the yarns volume. The yarn cross-sections are allowed to vary, while the overlapping is avoided by introducing a contact clearance between the bundles.

Despite the complex mesh generation, continuum element unit cells are used to obtain with high accuracy and level of detail of the strain and stress fields in the RUC. In addition, they have the advantage that common constitutive models can be used to describe damage initiation and damage progression in the bundles. Simple damage models in which the stiffness is instantly reduced to value close to zero after damage is initiated were used to predict the mechanical behaviour of woven fabrics, Johnson et al. (1993); Zako et al. (2003). Zako et al. (2003), compared the unit cell simulations numerical results with experiments and obtained prediction of the transverse cracking within the yarns with fairly accuracy.

Ivanov et al. (2009), used a finite element model of a triaxially braided unit cell to predict damage initiation and crack orientation using Puck criterion, Puck and Schürmann (2002). Progressive damage and stiffness deterioration is modelled using the degradation scheme of Murakami-Ohno and the damage evolution law of Ladeveze, applied to the average stress state of the yarns. Good agreement was found with experiments.

A study of the non-linear behaviour of woven composites was presented by De Carvalho et al. (2012). A yarn failure criterion, debonding between tows and matrix, and a plasticity criterion in the matrix were used to predict the compressive strength. A dependency of the out-of-the plane boundary conditions was found in the compressive response. Goyal (2007), studied the non-linear behaviour of biaxial braids prior to final failure. Anisotropic Hill plasticity criterion in conjunction with hardening were used to model the non-linear behaviour of the yarns. The parameters included in the constitutive equations were obtained from numerical correlations of different load cases. The non-linear behaviour of glass/epoxy braids was predicted fairly well with 3D unit cell simulations, while for the carbon/epoxy braids reasonable agreement was obtained.

An alternative and very flexible method to mesh complex geometries is to generate a voxel mesh, Potter et al. (2012). Similar to meshes from micro-CT reconstructions, the volume of the unit cell is discretised into cuboids of equal size, which are labeled volumetric pixels. The disadvantage of this method is that the yarn surfaces are more smooth but rather have a stair-like shape (Figure 2.8, Green et al. (2014)). Kim and Swan (2003b,a), proposed a methodology based on the virtual strain energy to improve the accuracy of the voxel meshing along the yarn boundaries. Voxel meshing was found to be a fast and accurate strategy to generate unit cell models maintaining a reliable shape and aspect ratio of the elements, which is not the same for tetrahedral meshes. Crookston et al. (2007), studied the different tetrahedral and voxel meshing techniques used in woven composites. Tetrahedral meshing was found to be faster; nevertheless, when the overall fibre volume fraction increases, the reduced free spaces located in between the yarns makes challenging the meshing process. Thus, voxel meshing may be a good alternative for tight preformed textile composites. Not long ago, Potter et al. (2012), included surface smoothing in a voxel meshing algorithm which enables the tow/matrix debonding in the unit cell simulations. Even so, voxel meshes are reported to cause spurious stress concentrations, De Carvalho et al. (2011), therefore unsmoothed voxel meshes do not seem

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to be suitable for damage modelling. [Zhang et al. \(2014a\)](#), used the open source software TexGen, [University of Nottingham \(2019\)](#), to built the meso-scale geometry of a triaxial braided composite. The FE voxel mesh is generated automatically using a package included in the software. Edge damage effects were studied in the model, and was found to occur periodically along the free edge and resulted in reduced transverse strength and stiffness. In addition, the elastic modulus was found dependent of the number of unit cells simulated.

Figure 2.8: Voxel meshes for: (a) idealised model, (b) simulated model, [Green et al. \(2014\)](#)

[Cox et al. \(1994\)](#), presented the Binary Model, which is a reduced approach for unit cell modelling of textile composites. In the Binary Model, the yarns are modelled using truss elements embedded into continuum elements. The truss elements are used to model the longitudinal response of the yarn, while the transverse and shear properties of the yarns as well as the matrix constitutive behaviour are included in the continuum elements. In a following publication the constitutive model was improved, [Xu et al. \(1995\)](#), allowing to be calibrated based on UD experiments. Finally, the elastic response was found to correlate pretty well. In addition, models for tensile and compressive fibre failure and the effect of tow misalignment were proposed. [Yang and Cox \(2003, 2010\)](#), extended the Binary Model by introducing multi-point-constraint (MPC) coupling between the truss and continuum elements and an averaging technique for the strains in the effective medium. Matrix non-linearity was accounted with a plasticity law in the Binary model, [Flores et al. \(2010\)](#). A Drucker-Prager plasticity criterion with fully associated flow rule and perfectly plastic deformation was used in the effective medium. Uniaxial tests were carried out to calibrate the model material parameters. Simulations results demonstrated that the non-linearity was properly accounted using Drucker-Prager plasticity criterion when compared with open-hole tension experiments, as shown in

Figure 2.9. Blacklock et al. (2012), used the Binary Model and statistical data from XCT inspections to generate virtual textile composites.

Figure 2.9: Open-hole tension simulations using the Binary Model, Flores et al. (2010).

A extension to the Binary model was proposed by Jiang et al. (2008). This approach named domain-superposition technique (DST) embeds the yarns discretised by continuum elements into the matrix continuum mesh. The displacements of the yarns nodes are coupled with MPC's with the matrix nodes. This approach is similar to the *embedded elements* method, Yang and Cox (2003). Due to the doubling volume, the yarn stiffness was adjusted by subtracting the matrix stiffness. The error was reported to be less than 5% when compared to traditional FE analysis, Tabatabaei and Lomov (2015). Even so, the stress field were found to be in good accordance.

2.4 Beyond the state-of-the-art - simulation of braided composites

As described in Chapter 1, braiding offers significant possibilities for process automation, high-volume production, and cost reduction during the manufacturing process. High flexibility is available, as the type of reinforcement can be adapted to the mechanical needs and the over-braiding process allows for curved mandrels with changing cross sections. This makes the usage of braiding favourable for a multitude of lengthy structures. From the designers viewpoint, the flexibility of the

2.4. Beyond the state-of-the-art - simulation of braided composites

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process creates the challenge of a highly variable yarn architecture and thus changing material properties of the braid.

The published literature regarding the textile composite process, yarn architecture, characterisation methods, and mechanical modelling approaches including FE unit cell modelling and macroscopic approaches has been presented. In addition, different possibilities to characterise the yarn architecture and approaches to geometric modelling of the yarns have been reviewed. Different methods have been used to provide input for geometric models. However, little has been done for braided composites, and even less for triaxially-braided composites.

The mechanical behaviour of textile composites was reviewed, with many publications describing the elastic, nonlinear, and failure behaviour of textile composites under uniaxial loads. The mechanisms driving the failure and damage are similar to those of unidirectional composites, but influenced by the textile architecture. However, to date there is no evidence, which of the characterisation methods and models applicable to braided composites provides a suitable balance between effort, level of detail, and robustness.

Different approaches for the prediction of textile composite constitutive behaviour have been evaluated with a focus on: simple and analytical models for stiffness prediction; FE unit cell models for the detailed modelling of the nonlinear effects and failure within a braided composite; and macroscopic CLT based methods favourable for structural simulation. Besides the analytical models, which are only suitable for stiffness prediction, FE unit cell methods can be used to provide detailed predictions of different damage and failure events inside the textile. Structural simulations typically use macroscopic approaches with high computational efficiency and good possibilities for implementation into conventional FE analysis. It should be noted that simple failure and damage models are mostly used for braided composites on a macroscopic scale and to date no validation of these models exists regarding the applicability on multi-axial stress states. Furthermore, failure and damage models must to be validated under various combined load cases.

The literature review has emphasised the need for an integrated approach to the characterisation and simulation of braided composites focusing on the triaxially fibre architecture on different length scales. Therefore a practical methodology based on a experimental campaign to determine the relevant mechanical properties and to study the damage mechanisms present in braided composites is proposed:

- First, a mechanical characterisation by means of tensile tests to determine strength and stiffness and onset of damage.
- Second, a experimental campaign to evaluate the loading capacity under shear loadings and to determine the rest of the in-plane mechanical stiffness and strength.
- Third, translaminar and interlaminar tests to determine the fibre fracture toughness and Mode I critical energy release.
- Fourth, the material is evaluated under structural conditions by means of an additional Open Hole Tensile (OHT) test.

Once the mechanical properties are determined, a numerical methodology based on multiscale FEM approach is developed, starting with unit cell modelling and incorporating the fibre architecture and matrix regions information into a macro-scale representation of the coupon geometry. The objective is to set-up a FEM methodology which can be used to predict the mechanical behaviour of braided composites and to evaluate whereas the relevant damage mechanisms observed in the experimental campaign are captured.

Characterisation and testing of braided composites

3

The main goal of this chapter is to develop a property characterisation methodology for triaxially braided composites and analysis of the mechanical response and failure micromechanisms of two-dimensional triaxial braided composites. Nesting, which is present in laminates with different number of plies (such as 12 and 18) and drilled hole size effect in six-layered laminates, are also studied.

Tensile and shear stiffness and strength were determined by means of plain and shear V-notch coupons until failure in the longitudinal and transverse direction. Translaminar and interlaminar fracture toughness properties were determined using Compact Tension (CT) and Mode I DCB specimens respectively. Moreover, Open Hole (OH) test were performed on different geometries varied from the recommended by the standard to evaluate the structural response of the material under structural conditions and to serve as an independent test for model validation.

3.1 Triaxially-braided material

Composite panels of 300x600mm² were manufactured using resin transfer moulding (RTM). A 271g/m² triaxial braided fabric QISO-L-52 supplied by A&P Technology (Ohio, USA, [A&P Technology \(2016\)](#)) was used to produce laminated preforms containing six, twelve and eighteen layers. The fibre architecture corresponds to a balanced 2x2 diamond patterns formed by 6K AS4-C carbon fibre axial yarns interlaced with 3K bias yarns forming $\pm 60^\circ$ as it is showed in Figure 3.1. A representative unit cell (RUC) of 10x2.9mm² formed by axial and bias yarns is also plotted in the

3.1. Triaxially-braided material

same figure for comparison. This configuration results in a balanced and in-plane quasi-isotropic behaviour of the fabric.

Figure 3.1: QISO-L-52 $\pm 60^\circ$ triaxial braided fabric and $10 \times 2.9 \text{ mm}^2$ representative unit cell representation (RUC).

The preforms were placed inside a steel mould heated by electrical resistances cartridges. The mould has a central injection port connected to a pressure pot and two venting gates connected to a vacuum pump. RTM6 resin (Hexcel, UK) was used for manufacturing the panels. The resin and the mould were first preheated at 80°C and 100°C respectively. Then, the resin was injected using a pressure of 3bars until it fills completely the mould cavity. At this point, the outlet gates were closed and the temperature of the mould increased at a rate of $3^\circ\text{C}/\text{min}$ up to 180°C . This latter temperature was maintained for two hours until the composite panels were completely cured. Panels were extracted from the mould cavity and submitted to ultrasound inspection to ensure that they were free of initial delaminations or other defects.

The nominal thickness of the dry fabric was 0.25 mm and its areal density was $0.2 \text{ kg}/\text{m}^2$. The manufactured panel thicknesses were $\approx 1.8 \text{ mm}$ (0.3mm cured ply thickness), $\approx 3.2 \text{ mm}$ (0.27mm cured ply thickness) and $\approx 4.4 \text{ mm}$ (0.24mm cured ply thickness). The resulting global volume fraction of reinforcement in the composite was $\approx 51.5\%$, $\approx 55.1\%$ and $\approx 64.1\%$ respectively. The areal density of the individual plies was provided by the manufacturer and was used to determine the matrix volume fraction, calculated from the matrix density and the weight of the composite panel before and after infiltration.

3. Characterisation and testing of braided composites

Sections perpendicular to the longitudinal and transversal directions were prepared to determine important geometrical features as yarn thickness and width, local fibre volume fraction, etc. To this end, metallographic sections were first cut, embedded in a mounting resin, polished with 4000grit SiC paper and finished using a diamond slurry up to $1\mu\text{m}$. Micrographs were acquired using optical microscopy and images were treated using the ImageJ software. The fundamental results obtained for the three panel thicknesses were summarised in Table 3.1. The geometrical volume fraction, or the space occupied by the yarns in the laminate, was estimated from the optical images and corresponded to $\approx 25.3 \pm 0.5\%$ and $\approx 48.6 \pm 0.8\%$, for the axial and bias yarns, respectively for the six layered specimen. The local volume fraction inside the yarn was similar, $\approx 70\%$, in both types of tows, axial and bias.

Figure 3.2: Schematic of the process followed to determine the unit cell and fibre bundle architecture parameters of a six layers specimen using braided composite cross-sections examined with an optical microscope [García-Carpintero et al. \(2017b\)](#)

3.2. Experimental techniques

Table 3.1: Geometrical bundle relationships of six-, twelve- and eighteen layers panel obtained from optical microscopy, [García-Carpintero et al. \(2017b,a\)](#)

		G_{vf} (%)	F_{vf} (%)
6 Layers panel	Axial yarn	25.3±0.5	70%
	Bias yarn	48.6±0.8	70%
12 Layers panel	Axial yarn	27.7±0.4	70%
	Bias yarn	51±0.6	70%
18 Layers panel	Axial yarn	34.1±0.6	70%
	Bias yarn	57.4±0.9	70%

3.2 Experimental techniques

The microstructure of the material was inspected by using several techniques, namely optical microscopy, scanning electron microscopy (SEM), C-scan ultrasounds and X-Ray computed tomography (XCT). Optical microscopy is easy-to-use, cost-effective and can be used for any material, whereas SEM provides an excellent resolution. However, such techniques only provide information of a specific section and, moreover, damage can be introduced during sample preparation. C-scan is widely used in aerospace industry to measure delamination in composites, but is unable to determine the shape of the flaws nor distinguish between failure modes. XCT is particularly powerful, because it enables the visualisation of a 3D image so that the reconstructed volume can be cut by any plane; it is also a non-destructive technique which ensures that the sample is free of damage. The technique is further described below.

3.2.1 Digital Image Correlation

To measure strains during the mechanical tests, both extensometry and digital image correlation (DIC) were used. The latter is a non-contacting optical technique to measure the displacement field on the surface of a specimen at different stages during deformation. The full strain field on the surface can be obtained afterwards as the derivative of the displacement field. The foundation of the technique is very simple: the displacement field is computed by tracking the distribution of grey intensity on the specimen surface in images acquired at different stages upon loading, [Canal et al. \(2012\)](#).

3.2.2 X-ray computed tomography/radiography

XCT is a non-destructive imaging technique in which the 3D view of an object can be reconstructed from several X-ray images collected at different angles. X rays are generated by the acceleration of electrons towards a target material, normally made of a heavy element, such as tungsten or molybdenum. The electrons are extracted from a tungsten filament (cathode) with a V-shape when a high voltage is applied (Figure 3.3). The accelerated electrons travel inside a vacuum tube towards the anode and then to the end of the tube where they crash against the target, which is positioned in between the X-ray source and the detector. The interactions of the electron beam with the atoms of the target material produce the X-ray spectrum that is used for the tomographic scan. X-rays travelling through the sample are attenuated depending on the absorption coefficient of the material and the energy of the incident X-ray beam reaching the detector which records a radiography.

Figure 3.3: Scheme of working principle of an open X-ray tube with a transmission target, [Muñoz Sánchez \(2014\)](#).

The sample can be modelled as a two or three-dimensional distribution of the X-ray attenuation coefficient, $(x; y)$, which is a property that characterises the ability of the material to absorb X-ray from the beam source. The radiation intensity I transmitted through a layer of material, Figure 3.4, is related to the incident

3.2. Experimental techniques

intensity I_0 according to Lambert-Beer's law, equation 3.1. This equation relates the total attenuation $p(t)$ (ratio of transmitted to incident intensity of radiation) through the X-ray absorption coefficient of the material, $\mu(x, y)$:

$$p(t) = \ln \frac{I}{I_0} = \exp \left[\int_{\Gamma} \mu(x, y) ds \right] \quad (3.1)$$

where the line integral represents the total attenuation suffered by the X-ray beam travelling along a straight path $s(x, y)$ through the cross-section of the object and t the distance from each ray of parallel beam to the centre of rotation, Figure 3.4.

Figure 3.4: Principle of tomography and illustration of the Fourier slice theorem, Muñoz Sánchez (2014).

The procedure for the reconstruction of a sample volume from the radiographies collected at different angles θ of rotation is summarised briefly, and explained on a parallel beam configuration for simplicity. During radiography collection, sample rotates around the z-axis (perpendicular to the paper). The cross-section of the sample is described by the function $f(x, y)$. The X-ray beam is assumed to be formed by parallel rays. When each ray passes through the sample, part of the radiation is absorbed and the attenuated intensity, $p(t, \theta)$, is collected in the detector. The attenuation will depend on the absorption coefficient of the material crossed and on the length of the path s through the sample.

Once the different projections are recorded for a set of rotation angles, the next step is to obtain the tomographic reconstruction of the original object. The object is reconstructed by means of the projections-slice theorem, .

In spite of the advantages in terms of study the damage mechanisms of the samples, the reconstruction based on the X-ray projections is a time consuming

methodology, and the ROI that can be studied is limited by the resolution of the tomograph and the absorption coefficient of the material. Therefore, instead of reconstructing the volume by means of the projection-slice theorem, radiographs at the whole specimen were acquired to study the crack propagation along the fibre architecture. This methodology was reported very useful to study the damage progression of this type of fibre architectures with the possibility to not only study the damage mechanisms associated to fibre breakage and tow splitting/debonding but to study other geometrical effects of the coupon that triggers the appearance of damage zones due to the non-homogeneity of the material.

3.2.3 Interrupted testing

The macroscopic evolution of damage in the composite coupons was assessed by means of periodic unloading-reloading (to obtain the elastic modulus and the residual strain), whereas the microscopic mechanisms were established by means of X-ray computed radiography. To this end, specimens were periodically removed from the mechanical testing machine and infiltrated with ZnI-containing liquid to assess the main damage modes as a function of the applied strain. Damage inspection was completed with the examination of post-mortem specimens by means of optical microscopy, scanning electron microscopy and XCT. The experimental observations and the predictions of an isostrain model were used to understand the key factors controlling the elastic modulus, strength and notch sensitivity of two-dimensional braided composites in tension and shear. All this information provides a comprehensive picture of the effect of the fibre architecture, crimping and damage on the mechanical behaviour of two-dimensional braided composites under different loading conditions.

3.3 Property Characterisation

3.3.1 Longitudinal / transverse tensile stiffness and strength properties

3.3.1.1 Experimental procedure

Rectangular $[0^\circ]_6$ specimens ($250 \times 25 \text{ mm}^2$) were cut from the laminates to prepare specimens for tensile tests in both directions, axial x and perpendicular to axial y (x and y directions in Figure 3.1), according to the specifications of the ASTM standard D3039. Special consideration has to be taken into account when testing textiles composites to ensure that a representative volume of material is tested.

3.3. Property Characterisation

Falzon and Herszbergb (1998), suggested that the specimen width should contain, at least, two unit cells of the fabric. Therefore, 25mm (2.5unit cells) width can be considered representative to capture the fundamental failure modes in the material, Figure 3.1.

Special care was taken during specimen milling operations to avoid damage and delaminations. E-glass tabs were added to the specimens to avoid premature failure due to excessive grip pressure during tensile tests. The specimens were tested under uniaxial tension in a electromechanical universal testing machine (Instron 3340) under stroke control at a constant cross-head speed of 1 mm/min. The load was recorded with a 150kN load cell. The strain field on one side of the specimen was measured using digital image correlation (VicSNAP, Correlated Solutions, Inc.). A speckle pattern was created on the surface of the specimen by first, painting in white and second, by spraying black dots over the white background. The DIC system acquired high-resolution images of the specimen surface at a rate of 0.5s using a single camera. The images were analysed after each test to determine the displacement field on the specimen surface and the corresponding strains.

A minimum of five tensile tests until failure in each direction, longitudinal and transversal, were carried out (x and y directions according to Figure 3.1). In addition, some tests were stopped at different load levels, 30, 40, 50, 60, 70 and 80% of the material strength, to ascertain the evolution of damage during deformation. To this end, the frame displacement of the testing machine was maintained constant and the specimen middle part was immersed in a dye penetrant liquid (60g of ZnI in 10ml of distilled water, 10ml of ethanol and 10ml of Kodak Photo-Flo 200) for 30 minutes, as illustrated in Figure 3.5. After each step, the specimen is unloaded and inspected by X-ray radiography to determine the evolution of the crack density in the material upon the application of load. This method enhances the contrast between the cracks and the composite material and was proven efficient to scan damage evolution in composite laminates, Sket et al. (2014). The periodic unloading and the liquid immersion did not modify the laminate properties. Radiographs were acquired with an exposure time of 1000ms using a Nanotom 160NF (General Electric-Phoenix) XCT system using a tungsten target under 90kV and 100 μ A operational conditions. The radiography pixel size was set approximately to 10 μ m/pixel.

Finally, post-mortem specimens were inspected by using optical microscopy. a An optical microscope (Olympus BX-51) was used to inspect sections located parallel

Figure 3.5: Specimen immersion in dye penetrant liquid prior to testing, Muñoz Sánchez (2014).

and perpendicular to the load in the longitudinal and transverse directions. Samples were previously grounded and polished according to the standard procedure.

3.3.1.2 Results and discussion

The strains were determined by averaging the field quantities over a region of interest (ROI) of $25 \times 20 \text{ mm}^2$. The stress-strain curves obtained experimentally for both directions x and y are plotted in Figure 3.6. The response of the material was essentially linear and elastic up to failure when testing in the direction x that corresponds with the axial tows. However, the material exhibited a slight non-linear behaviour in the direction y in the last part of the test. In any case, the experimental scatter was minimum and all the curves superimpose together. The elastic modulus, obtained from the slope of the stress-strain curve in the deformation range of 0.2-0.5% is presented in Table 3.2 in combination with the strain-to-failure and the strength for both directions analysed. The Poisson ratio ν_{xy} was obtained using the averaged strains in the longitudinal and transversal directions from digital image correlation measurements.

Figures 3.7 a), b) and c) show the evolution of a set of cracks, measured from radiographs, corresponding to one of the specimens tested in the direction of the axial yarns. For the sake of clarity, cracks were marked in white to enhance the contrast. Cracks grew stably running parallel to the bias yarns direction and this damage mechanism did not affect significantly the shape of the stress-strain curve which was essentially controlled by the stiff axial yarns. Analysis of the crack at higher magnifications by means of optical microscopy is presented in Figure 3.7 d)

3.3. Property Characterisation

Table 3.2: Tensile mechanical properties.

Property	Experimental
Longitudinal modulus, E_1 (GPa)	42.9 ± 1.8
Transverse modulus, E_2 (GPa)	38.6 ± 0.9
Longitudinal Poisson ratio, ν_{12} ($\hat{A}\cdot$)	0.30 ± 0.01
Longitudinal tensile strength, X_L (MPa)	673 ± 55
Transverse tensile strength, X_T (MPa)	630 ± 23
Longitudinal strain-to-failure, ϵ_L (%)	1.6 ± 0.1
Transverse strain-to-failure, ϵ_T (%)	1.8 ± 0.1

Figure 3.6: Stress-strain curves of the braided composite tested in the longitudinal direction x (axial yarns) and transversal direction y (perpendicular to axial yarns).

and shows that this set of cracks running in the bias yarn direction were generated preferentially either at the yarn/matrix interface or grew from the inner region of the tows deflecting at the yarn/matrix interface [Lomov et al. \(2008b\)](#); [Ivanov et al. \(2009\)](#). These tow/interface cracks grew stably in the fibre direction mainly driven by the resolved shear forces acting in the tow direction due to the applied

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macroscopic stresses. At some point, bias yarns were severely cracked losing the stability provided by the presence of the resin and triggering the final failure by fibre breakage followed by yarn pull-out, Figure 3.7 e).

Figure 3.7: Radiographs of the triaxial braided composite tested in the longitudinal direction: a) $\sigma_x = 403 \text{ MPa}$, b) $\sigma_x = 471 \text{ MPa}$, c) $\sigma_x = 538 \text{ MPa}$, d) Cross-section perpendicular to the axial direction, e) Bias yarn extracted from tested coupon revealing shear cracking and fibre breakage.

The damage mechanisms observed in the specimens tested in the direction perpendicular to the axial yarns were totally different from the previously discussed for loading along the axial direction. The radiographs acquired also revealed the dominant crack patterns when the material is loaded in the transversal direction, Figure 3.8. Incipient cracks running parallel to the axial tows and emanating from the specimen edges were observed at the first stages of the loading process (Figure 3.8a).

The crack population in this region increased monotonically until a certain level of saturation. Those edge cracks tended to be distributed in triangular regions constrained by the bias yarn architecture following the $\pm 60^\circ$ directions of the material, Figure 3.8 b) and c). At some load level, some of these cracks progressed through the whole width of the laminate, but still grouped in packets following the regions where the axial yarns concentrated in the laminate, coinciding with the darker regions of the X-ray images (parallel axial yarns of $\approx 3 \text{ mm}$ of width separated $\approx 5 \text{ mm}$). Optical

3.3. Property Characterisation

Figure 3.8: Radiographs of the triaxial braided composite tested in the transverse direction: a) $\sigma_y=378\text{MPa}$, b) $\sigma_y=441\text{MPa}$, c) $\sigma_y=504\text{MPa}$, d) Cross-section parallel to the axial direction in the edge region of the specimen, e) Cross-section perpendicular to the axial direction in the inner region of the specimen.

micrographs of the damaged area confirmed the cracking mechanisms observed in the projections of the X-ray images. Figure 3.8 d) shows the edge area of the specimen parallel to the axial yarns and perpendicular to the loading direction. Cracks were formed at the edges by either tow debonding and matrix splitting. Sections perpendicular to the axial tows obtained in the middle region of the specimen loaded transversely are presented in Figure 3.8 e) and shows that through-the-width cracks in X-ray images, Figure 3.8 c), were produced by axial tow debonding.

X-ray radiographs allowed to determine the evolution of the material cracks during the applied load. Individual radiographs were taken along the entire free length of the specimen and stitched together using the ImageJ software, (USA). Cracks were automatically recognised by the image analysis software and the statistics referred to the number of cracks, their position, crack length, etc. determined. It should be mentioned that this methodology led to the determination of a segmented population of cracks corresponding to those filled by the dye penetrant liquid, and therefore, bulk cracks not connected were not included in the statistics.

The evolution of the number of cracks, irrespective of their individual length, as a function of the applied deformation is presented in Figure 3.9 a). The total number of cracks in the free length specimen area ($\approx 150 \times 25 \text{mm}^2$) was plotted for the specimens loaded in the longitudinal and transverse directions. The evolution

3. Characterisation and testing of braided composites

Figure 3.9: a) Evolution of total number of cracks irrespective to their individual length as a function of the applied deformation. Damage onset deformations of 0.64% and 0.49% for longitudinal and transversal loading, b) Distribution of crack lengths in specimens subjected to longitudinal tensile loading at 50% and 80% of the maximum strength, c) Distribution of crack lengths in specimens subjected to transversal tensile loading at 50% and 80% of the maximum strength.

was monotonic in both cases although with different growth rates. The deformation corresponded to the onset of matrix cracking was 0.64% and 0.49% for the longitudinal and transversal loading. The number of cracks was much higher in the specimen loaded transversely than in the longitudinal one. As it was mentioned previously, and shown in Figure 3.8 a), b) and c), cracks in transverse specimens appeared at the early stages of the loading process in the form of tow debonding/splitting cracks

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at the edge area of the specimen. At some deformation point, $\approx 1\%$, this initial rate decreased and this fact can be probably endorsed to the crack saturation phenomenon at the edge area of the specimen. Conversely, the evolution of the number of cracks in the specimen loaded in the longitudinal direction showed a slight rate increase at $\approx 1\%$, probably due to the bias yarns rotation translated to inter-yarn debonding, due to excessive shearing, Figure 3.9 a). The crack length distribution for the specimens loaded transversely is also shown in Figure 3.9 c) where most of the cracks detected by the X-ray system were relatively short lying within the range of $\approx 0 - 5\text{mm}$ which is consistent with the previously mentioned assumption. Moreover, a number of cracks that spanned the total width of the specimen (25mm) were also described in the same statistics. In any case, it was not possible to separate in this crack length distribution, those new cracks generated during the loading step from the former cracks that also grew stably. Similar results were found in the specimen loaded in the longitudinal direction, Figure 3.9 b). In this case, as pointed previously, the number of cracks was much lower than in the transverse loaded specimen. The maximum length measured was also much lower than the total diagonal length ($\pm 30^\circ$) of the specimen ($25\text{mm}/\cos(30^\circ) \approx 28\text{mm}$, and, therefore, no cracks spanned the diagonal in this configuration.

3.3.2 Longitudinal / transverse Shear stiffness and strength properties

3.3.2.1 Experimental procedure

There is not a single method for determining the shear mechanical response of braided composites, Carey (2016). The recommended standards, methods for determining the in-plane shear properties of textile composites, are Test Methods D4255/D4255M or D5379/D5379M, ASTM International (2015a, 2012). These methods were not designed explicitly to braided composites as have designed for general testing of polymer matrix composites materials. Nevertheless, several authors have been using them to characterise the shear properties of braided composites, Naik et al. (1994a); Minguet et al. (1994); Kier et al. (2015).

A comparison of different shear methods for composite materials was performed by Adams et al. (2007). Iosipescu beam V-notch test (ASTM D5379) and two- or three-rail shear tests (ASTM D4255) were evaluated in a selection process described by Lee and Munro (1986). In this study, focused on the evaluation of shear methods for composite laminates, Iosipescu, 45° and 10° of axis shear-loading tests are recommended over three-rail shear and slotted tensile tests methods.

3. Characterisation and testing of braided composites

In the three-rail shear test fixture, [ASTM International \(2015a\)](#), the strain gauges are oriented at 45° angle relatively to the axis of the rails. A load is applied to the middle rail, which results in an opposing load from the external two rails. The test configuration for the Iosipescu beam V-notch, [ASTM International \(2012\)](#), allows for the average shear strain to be measured within the notched section of the test specimen. The Iosipescu shear test allows for both in-plane shear as well as interlaminar shear to be evaluated.

Another method that has been employed to evaluate the shear properties of braided composites is the V-notched Rail Shear Test Method, [ASTM International \(2005\)](#). The geometry of the ASTM D7078 standard has been optimally designed to produce uniform stress field in the gage section for a variety of composite configurations, [Adams et al. \(2007\)](#). This method was employed by [Kier et al. \(2010\)](#) since a triple rail shear fixture resulted in artificially high results for shear modulus, caused by the direct loading of the fibres that bridged between the two gauged-sections. A schematic of the shear test methods described is shown in Figure 3.10.

For anisotropic materials, Test Methods [ASTM International \(2012\)](#) and [ASTM International \(2005\)](#) can be used to evaluate shear properties in any of the six possible shear planes, [Carey \(2016\)](#). Since braided composites typically exhibit orthotropic mechanical properties, these test methods may be advantageous, as mechanical properties in multiple shear planes can be evaluated. Therefore, ASTM D7078 was chosen as the basis for the shear testing performed during this work. All testing was performed with the V-Notched Rail Shear fixture developed by Wyoming Testing Fixtures Inc., Figure 3.11.

Six-layered rectangular specimens (76 x 56 mm²) were cut according to the dimensions specified of the cited standard. 90°V-notches with a notch root radius of 1.3 mm were machined using special tools to avoid damage and delamination during milling. Figure 3.11 shows a shear specimen clamped in the V-Notched Rail Shear fixture. This fixture was mounted in the same servo-mechanical machine used for tensile testing. A constant displacement rate of 0.5 mm/min was used. The applied load was measured continuously with a 30 kN load cell.

In the shear tests, the longitudinal direction was defined as the orientation in which the axial tows extend along the loading axis. The transverse orientation is rotated 90 degrees so bias tows are placed are more aligned with the loading direction.

3.3. Property Characterisation

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a) Three-rail shear test

b) Iosipescu shear test fixture.

c) V-notched shear rail test

Figure 3.10: Schematic of the fixture used to determine the shear properties of composite materials, [Carey \(2016\)](#).

Full field DIC was used to examine the strain field during loading and to identify the damage initiation and evolution in the specimens, [Totry et al. \(2010\)](#). The in-plane shear stress τ_{xy} was calculated from the test data and the cross sectional area of the ligament neglecting the effect of the stress concentration around the notches. The shear strain, γ_{xy} , was determined using digital image correlation by averaging the shear strains over a region of interest (ROI) of approximately $25 \times 5 \text{ mm}^2$ in between the notches. As shown in Figure 3.12, the region of interest was centred along the ligament of the specimen and provided a virtual strain gauge during the

Figure 3.11: Fixture and specimen for the V-notched rail shear tests, [García-Carpintero et al. \(2016b\)](#).

whole test. It should be noticed that the in-plane shear strain was nearly constant along the ligament despite the notches.

Figure 3.12: Transverse specimen shear strain γ_{12}

It was observed that in most specimens, the crack initiation started at the 45° notched edges. This effect causes the standard geometry to develop tensile and compressive strains along the free edges of the notch which have the potential for

3.3. Property Characterisation

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driving delamination resulting in failure, [Kohlman \(2012\)](#). In Figure 3.13 the failure produced by this edge effect in the specimens loaded in the longitudinal and transverse directions respectively is illustrated. This behaviour, mainly driven by the specimen geometry, is characteristic of this type of fibre architecture in which the crack fails to initiate in the designated notch. The more complex fibre architecture compared to unidirectional or woven composites, creates multiple crack fronts which follow the fibre direction, and therefore fail to propagate through the designed ligament between notches.

This behaviour was spotted by [Kohlman \(2012\)](#), in which different specimen geometries modified from the proposed by the standard were used to evaluate the geometrical effects on the response and measured strength on biaxial braided composites. Three types of geometries were evaluated ranging from ASTM D7078 standard, [ASTM International \(2005\)](#), to notched shear specimen through "H" specimen. It was found that the "7078" geometry developed tensile and compressive strains along the free edges of the notch which have potential for driving a delamination resulting in failure. The "Notch" shear geometry produced highly localised tensile and compressive strain near the notch root. The "H" geometry is a geometric intermediate and developed strain distributions as such. In general, the shear strain was concentrated in the notch region for all the specimen geometries, as expected. However, the orientation of the fibre architecture distorted the shear strain distribution in the narrower gage geometries producing strains parallel to the loading direction triggering the crack initiation and subsequent propagation.

To overcome this edge effect and to decouple the shear failure from the bias tensile failure, an artificial pre-crack of $\approx 5\text{mm}$ was introduced along the ligament between the notches to replicate the notched specimen behaviour by enforcing the propagation in both longitudinal and transverse specimens. Figure 3.13 shows the pre-crack introduced in the longitudinal and transverse specimens and their crack propagation after failure. It can be seen that the pre-crack induces a stress concentration in the ligament between the notches enforcing the crack propagation in comparison with the standard geometry.

To compute the strains in the pre-cracked geometry, DIC systems was also used. However, DIC systems only provides useful results for non-cracked in the samples. Therefore the new ROI was chosen to be placed centred in the ligament in between the non-cracked region of the specimen.

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Table 3.3: Longitudinal and transverse shear loading results.

Test configuration	Loading direction	Shear modulus [GPa]	Ultimate strength [MPa]	Strain to failure [%]
No-Notch	Longitudinal	17.6±0.3	309± 13	2±0.2
	Transverse	16.4±0.4	298±3.1	2±0.1
Notch	Longitudinal	20±0.1	310±9	1.9±0.3
	Transverse	20± 1	249± 15	1.6± 0.2

3.3.2.2 Results and discussion

Results from shear V-notch tests performed on braided specimens are given in Table 3.3. The stress/strain response of the specimens tested revealed a non-linear behaviour as shown in Figure 3.14. The linear region of the shear response in which the shear stiffness was measured is within the 0.2-0.5 % strain range.

From Table 3.3 it can be observed that despite the weakening effect to enforce the crack propagation, the shear stiffness is greater in the pre-cracked specimens than in the specimens without pre-crack. This stress concentration around the notches triggered an earlier failure in comparison with the specimens without pre-crack. Furthermore it can be observed that the ultimate strength, strain to failure, and shear stiffness modulus are alike in both longitudinal and transverse direction for the un-notched specimens due to the quasi-isotropic behaviour of the material.

Figure 3.15a and b illustrates the shear damage accumulation stages through the force/displacement response of the pre-cracked specimens tested in the longitudinal and transverse directions.

As shown in Figure 3.15b for the transverse loading, after the linear regimen, cracks start to form at the pre-crack tips and propagate along the fibre direction (bias fibres) in the zones liable to compression stresses resulting from the shear loading, indicated as point A. The architecture. The presence of axial yarns perpendicular to the loading direction acts as a fibre containment structure, disrupting or stopping the crack propagation and forcing to propagate through the axial yarn. As, the loading continues, the crack front bridges the axial yarn and continues its propagation along a neighbouring bias yarn. From this point, the crack front continues to propagate through fibre bridging and fibre branching. After the first significant failure, represented as point B, the emanating cracks generated at both tips joins, and the damaged area around the crack path is increased. At point B, the test

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can be considered to be finished, as the loading capacity of the sample is minimum, mainly produced by the unbroken axial yarns in tension and fibre self-penetration. The sample carries a residual load through the yarns which have not completely failed yet and the residual load decreases slowly with increasing damage along the cracked zones.

The longitudinal specimen exhibit a different crack path, as shown in Figure 3.15a. The crack starts to form at the pre-crack tips and propagates, localised in the ligament between the notches, affected by the presence of axial yarns, parallel to the loading direction, as indicated in point A. The loading capacity of the sample is $\approx 50\%$ more than the transverse specimen, mainly caused by the tensile contribution of the axial yarns. The crack emanated from the tips continues its propagation through the axial yarns till the entanglement of bias fibres stops the front, which barely disrupt the path. At this point, fibre bridging occurs till the sample failure, represented as point B, when both fronts joints. In contrast with the transverse specimen, the longitudinal specimen continues to bear a significant loading capacity after the axial yarns breakage, caused by the reorientation and tensioning of the of the bridged bias yarns up to their catastrophic failure, represented as point C.

As mentioned earlier, the complex architecture of the triaxially braided material dominates the crack propagation through different damage mechanisms and necessitates geometry alterations from the ASTM D7078 standard. Despite that the standardised geometry recommended by the standard fails to create a region of pure shear within the coupon, the stress prediction for both longitudinal and transverse specimen remains invariable, a typical characteristic of the quasi-isotropic behaviour exhibited by this fibre architecture. The proposed one, pre-cracking (5mm) to enforce crack propagation during test, might be valid to predict the shear stiffness and strength for longitudinal specimens in which the trend is less affected and the behaviour is similar to a pure shear response. However in the transverse direction, creating a notch produce a instability region which accumulates damage triggering the early failure. Therefore, it is recommended to use the standardised geometry for stress prediction and the 5mm pre-cracked version to stiffness determination. Moreover, studying and understanding the occurring damage mechanisms might provide insight to improve the sample geometry.

3.3.3 Translaminar fracture toughness

3.3.3.1 Experimental procedure

The use of fracture mechanics to analyse failure of composites is grounded on the hypothesis of the existence of a material property called *fracture toughness*, defined by either the critical stress intensity factor K_{Ic} or the critical strain energy release rate G_{Ic} , [Slepetz and Carlson \(1976\)](#). The test standard ASTM E1944, based on Extended Compact Tension (ECT) specimens, [ASTM International \(2004\)](#), is recommended to characterise the translaminar fracture toughness of laminated polymeric composites, however it covers only fracture initiation rather than the determination of the full R -curve, as intended in this work. Towards that end, most of the work on fracture characterisation on composites has been developed using CT experiments based on the test standard procedure ASTM E399, [ASTM International \(2006\)](#), recommended for fragile isotropic materials such as metals, e.g. [Pinho et al. \(2006b\)](#); [Laffan et al. \(2010, 2011, 2012\)](#). It is assumed the crack propagates in a stable and self-similar way, and that the fracture process zone is small compared with the specimen dimensions, hence Linear Elastic Fracture Mechanics (LEFM) considerations can be applied. Under these conditions, the critical stress intensity factor is given as, [ASTM International \(2006\)](#):

$$K_{Ic} = \frac{P_c}{t\sqrt{w}} f(a/w) \quad (3.2)$$

with

$$f(a/w) = \frac{2 + a/w}{(1 - a/w)^{1.5}} (0.886 + 4.64(a/w) - 13.32(a/w)^2 + 14.72(a/w)^3 - 5.6(a/w)^4)$$

wherein P_c is the measured critical load to cause fracture, t is the specimen thickness, w is the length measured, in the crack direction, from the loading line to the end of the specimen, and a is the crack length. The critical strain energy release rate of the laminate could then be calculated from the laminate critical stress intensity factor as [Tada et al. \(2000\)](#):

$$G_{Ic}^{lam} = \frac{K_{Ic}^2}{\sqrt{2E_x E_y}} \sqrt{\sqrt{\frac{E_x}{E_y}} + \frac{E_x}{2G_{xy}} - \nu_{xy}} \quad (3.3)$$

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wherein E_x and E_y are the elastic moduli in two orthogonal planar directions, G_{xy} is the laminate shear modulus and ν_{xy} its Poisson's ratio. However, J-integral studies conducted by Laffan et al. (2010), reached the conclusion that the above expression leads to important inaccuracies when applied to highly orthotropic laminates such as the cross-ply types proposed to extract the intralaminar fracture toughness associated with longitudinal fibre breakage, G_{Ic}^0 . For better results, the "Area Method" and the "Compliance Calibration Method" (CC) were proposed as practical alternatives to determine R -curves from CT experimental data Laffan et al. (2010). With the first, the property is calculated by dividing the area underneath the force-displacement curve by the fractured area associated with the crack growth. The CC method uses the rate of specimen compliance (C) due to the crack length growth. In this section, the application of these experimental procedures and data reduction methods to triaxial braided composites is presented.

The CT specimen geometry proposed by Pinho et al. (2006b), was adopted in this work, i.e. 65x60 mm^2 rectangular-based with two load-bearing holes (8 mm in diameter), a 4 mm wide notch (30 mm in length) prolonged by a 10 mm slit, as shown in Figure 3.16. The specimens were carefully milled from braided composite panels of two different layups ($[0^\circ]_{12}$ and $[0^\circ]_{18}$) manufactured by means of resin transfer moulding (RTM), resulting in two different volume fractions of fibre reinforcement. The resin used was RTM6 epoxy, Hexcel (2016a). The fabric was the QISO-L-52 (271g/ m^2) supplied by A&P Technology (Ohio, USA, A&P Technology (2016)), and corresponds to a balanced 2x2 diamond patterns formed by 6K AS4-C carbon fibre axial yarns interlaced with 3K bias yarns forming $\pm 60^\circ$, as shown in Figure 3.1a. This configuration results in a balanced and in-plane quasi-isotropic behaviour of the fabric. The direction of the axial yarns is identified as the 'longitudinal direction'. A Representative Unit Cell (RUC) of 10x2.9 mm^2 for this architecture is also represented in the Figure 3.1a.

The resulting thickness of the twelve layered specimens was $\approx 3.2mm$ (0.27mm cured ply thickness) while for the 18-layered specimens it was $\approx 4.4mm$ (0.24mm cured ply thickness). Results of the analyses of the yarn microstructures are given in Figure 3.1b. The resulting global fibre Volume Fraction (F_{vf}) was $\approx 55.1\%$ and $\approx 64.1\%$ respectively for 12 and 18-layered specimens. A 10 mm long slit notch was created on each of the specimens with a standard wire-cutting machine equipped with a 0.2 mm thick filament.

Specimens were cut in two different orientations to allow loading along longitu-

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dinal and transverse material directions (see Figure 3.16), in this way driving cracks to propagate perpendicularly and tangentially to the axial yarns, respectively.

The CT specimens were pin-loaded using an electromechanical universal testing machine (Instron 3340) under stroke control at a constant cross-head speed of 0.5 mm/min to produce the propagation of the crack from the root of the slit notch. The applied force and displacement were recorded from the readings of the 10 kN load-cell of the testing frame. A speckle pattern was created on the surface of the specimen to easily monitor the crack length during the tests. The speckle pattern also helped identifying damage not easily visible otherwise. A high-resolution CCD camera was used to record the tests and posterior detailed crack growth analyses by pixel count. A minimum of five translaminar fracture tests for each layup configuration and specimen orientation was performed. The typical test was run up to the observation of compression failure on the back side of the specimen. The data obtained at this stage and beyond was invalidated for the fracture process characterisation. Valid data was obtained for crack propagation lengths of about 15-20 *mm*.

Although the CT specimens seem to be a practical approach to obtain the translaminar critical energy release rate of laminated composites, the adopted geometry only allows limited crack growth before the results start to be invalidated by the occurrence of secondary damage mechanisms such as compression failure on the back side of the specimen. In the present case, this valid crack growth distance is equivalent to 5-7 RUC side dimensions for loading in the transverse direction, but only 1-2 RUC's for longitudinal loading. This might lead to a loss of representativeness of fracture propagation for this loading mode. A redesign of the CT specimen to allow longer crack growing in triaxial braided composites is recommended. The extended CT specimen, as proposed by Blanco et al. (2014a,b), can extend the validity of crack propagation results and lead to larger steady-state crack propagation regions reducing the scatter in the results. However, irrespective of possible modifications, the use of CT specimens always leads to a limitation in valid length of crack propagation. Catalanotti et al. (2014b,a), proposed the use of the size-effect law, Bazant and Planas (1998), to determine the crack resistance curves of composites. By using sufficient double edge notched specimens of different sizes, the full R-curve of a FRP material or laminate can be inferred, including initiation and propagation energy release rate values, based on values of the critical energy release rate for different crack lengths (corresponding to each specimen size). This approach does not have the limitations of the CT tests because it does not rely on the progressive

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propagation of a crack in a single coupon. The major drawback of the approach is that it is more expensive in terms of specimen preparation, test procedure and data reduction.

3.3.3.2 Data reduction, results and discussion

The load-displacement curves obtained for 12- and 18-layered CT specimens, and for both material directions are plotted in Figure 3.17, respectively. Essentially, the response of the material was linear and elastic up to the peak loads, corresponding to the propagation of the slit notch. The specimens stiffness and resistance were higher for longitudinal loading case due to the presence of the stiff and strong longitudinal yarns aligned with the loading direction. Variations in the peak loads are visible which can be attributed to different relative locations of termination of the slit notch within the RUC's, as the features of the braided architecture were likely to be shifted from specimen to specimen.

The propagation of fracture was not entirely stable. Instead, a stick-slip crack growth is observed, i.e. in the form of large load drops with corresponding increases in crack openings. Fracture under longitudinal loading seems to have been slightly more stable, with less and smaller load drops, than in the case of transverse loading. The digital image recording system allowed the correlation of the step-wise crack advance with the corresponding load step. The crack front was identified with the longest extension of visible damage in the direction of crack propagation.

In order to determine the translaminar critical energy release rate, (G_{Ic}), methods that require the relation between load and optically measured crack growth may lead to inaccuracies, specially when applied to triaxial braided composites. The main reasons for this are: i) it is very difficult to identify the crack front, as perceived from Figure 3.20 below, because of the complexity of the cracking patterns, i.e. there is actually no single crack but a branch of cracks advancing through the material; ii) the unstable nature of crack propagation that makes the measurements difficult at any given instant. To avoid the second issue and estimate an average value for G_{Ic} , the Area Method may be applied to the total crack growth, Δ_a , by which

$$G_{Ic} = \frac{E_c}{\Delta_a t} \quad (3.4)$$

wherein E_c is the corresponding total energy dissipated through the cracking process. This parameter can be computed as the area under the force-displacement curve

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swept by the decreasing stiffness line, as described in [Laffan et al. \(2010\)](#).

For the 12-layered configuration studied in this work, Equation 3.4 yields critical strain energy release rates of $G_{Ic}^L = 124.9 \pm 9.7 \text{ kJ/m}^2$ for longitudinal loading, and $G_{Ic}^T = 93.4 \pm 2.3 \text{ kJ/m}^2$ for loading in the transverse direction. For the 18-layered case, $G_{Ic}^L = 163.6 \pm 3.5 \text{ kJ/m}^2$ and $G_{Ic}^T = 97.0 \pm 5.6 \text{ kJ/m}^2$. The most remarkable trend in these results is the 35-65% higher material toughness for fracture under longitudinal loading than under transverse loading. This visible effect can be attributed to the anisotropic braided architecture and will be analysed below.

Another trend in the results is associated with specimen thickness, specially for longitudinal loading for which direction a 30% increase in fracture toughness is observed for the 18-layered specimens with respect to the 12-layered. This trend could be the fingerprint of the effect of ply nesting and of the increase in fibre volume fraction for higher ply compactness. Nesting is a misalignment of the textile architecture produced during the manufacturing process due to the compaction resulting in a lower cured ply thickness. It also produces an increase in the fibre volume fraction by means of a decrease in the relative content of matrix within the laminate, this being the reason for different F_{fv} 's in 12 and 18-layered specimens ($\approx 55.1\%$ and $\approx 64\%$, respectively). Then, it could be deduced that an increase in the fibre volume fraction would result in an increase of the critical strain energy release rate, but since the fracture toughness of single fibres was demonstrated to be relatively low, [Herráez et al. \(2016a\)](#), other factors may have a more pronounced effect. For example, the of the fracture process, due the the higher fibre compaction, leading to and increase in microcracking, fibre/matrix interface debonding and pull-out, crack branching and fibre bridging, i.e. effectively increasing the fracture surface. These considerations should, however be taken with care at this point since the affect, if it exists, is only inexplicably visible for longitudinal loading. Moreover, the reliability of the Area Method used in this way for data reduction and its representativeness with regards the fracture process are questionable, [Laffan et al. \(2010\)](#).

The Area Method used applied as above gives average values of the fracture toughness for the whole crack growth produced in the experiments. However, the fracture energy release rate only reaches a steady value when the fracture process zone has fully developed, growing from the initiation of fracture up to that point. This effect is captured in the material fracture resistance curve (R -curve). The Area Method could be used towards that end, by considering intermediate stages of crack propagation and the associated fractions of energy dissipation, but this method has

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been proved to be less accurate than others because it relies heavily on optical measurements of the unstable crack growth, [Laffan et al. \(2010\)](#). Alternatively, the Compliance Calibration (CC) approach leads to more reliable results without significant increase in the complexity of testing and data reduction procedures. From Figure 3.17 it can be deduced that steady-state crack propagation occurs in the zones in which the load is increasing through the opening of the specimen. Therefore, R -curves can be determined using the average crack growth in-between the points before a load drop (symbols in Figure 3.17). Using these points, the critical energy release rate corresponding the each steady-state cracking step can be determined by the Irwin-Kies relation, [Truss et al. \(1997\)](#):

$$G_{Ic} = \frac{P_c^2}{2t} \frac{dC}{da} \quad (3.5)$$

wherein dC/da is the change in specimen compliance over the corresponding change in crack length. Assuming that the specimen unloading at any point of the crack propagation would return it to the initial position, this relation can be determined with a fit to the experimental compliance, i.e. the displacement over the load measured during the experiment. The following fitting function is recommended by [Dávila et al. \(2009\)](#).

$$C = (\alpha a + \beta)^\chi \quad (3.6)$$

wherein α , β and χ need to be determined for the best fit to the experimental results. Following this approach, the critical strain energy release rate was then obtained from

$$G_{Ic} = \frac{P_c^2}{2t} \cdot \alpha \chi (\alpha a + \beta)^{\chi-1} \quad (3.7)$$

The fitted compliance curves for the test cases studied are represented in Figure 3.18. The curves become steeper as values of crack length increase, revealing the sensitivity of the CC method to crack length measurement. As a result, the error associated with this data reduction method also increases with the optically measured crack length, as shown in [Laffan et al. \(2010\)](#). To mitigate this effect and others related to the proximity of the crack tip to the back edge of the specimen, values of Δa greater than 15 mm were discarded from the analyses.

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The R -curves, resulting from the application of Equation 3.5 are given in Figure 3.19. A notable feature in all curves is the significant scatter, thought to be a result of the unstable crack growth, the use of optically-measured crack length, and the complex crack propagation mechanisms discussed below. The occurrence of these mechanisms, and associated damage dissipation, depends of the location of the crack tip within the architecture of the material, in other words, the crack propagation is not self-similar at local level. This is to say that the crack resistance curve is expected to show some oscillation by nature. Notwithstanding, extreme oscillations were filtered out by removing points with deviations from average data above a reasonable threshold. Due to the scatter and the lack of data points for small values of crack growth, strong R -curve effects are not noticeable in Figure 3.19.

The critical energy release rates for crack initiation and propagation along both material directions, determined according to the Compliance Calibration method, are given in Table 3.4. The values for crack initiation are calculated as the average of experimentally-obtained G_{Ic} values corresponding to the first mm of crack. The values for crack propagation are determined as the average of the remaining G_{Ic} values. The results hence obtained are taken to be more representative of the fracture process than the average ones obtained with the Area Method, but are associated with larger experimental deviations as a result of the scatter in the R -curves.

Table 3.4: Values of critical energy release rate for longitudinal (G_{Ic}^L) and transverse (G_{Ic}^T) loading directions obtained by the Area Method and Compliance Calibration (CC) Method.

Spec. Config.	Loading direction	CC initiation	CC propagation	Area Method
		$[kJ/m^2]$	$[kJ/m^2]$	$[kJ/m^2]$
12 Layers	Longitudinal (G_{Ic}^L)	88 ± 20.2	149.4 ± 53	124.9 ± 9.7
	Transverse (G_{Ic}^T)	93 ± 10.1	99 ± 24.1	93.4 ± 2.3
18 Layers	Longitudinal (G_{Ic}^L)	119.1 ± 7.2	154.2 ± 36	163.6 ± 3.5
	Transverse (G_{Ic}^T)	$61.2 \pm --$	92.2 ± 27.2	97.0 ± 5.6

The results of the CC method reveal 50-70% higher critical strain energy release rates associated with fracture propagation under longitudinal loading as compared with propagation under transverse loading. This goes in line with similar findings with the Area Method discussed above (see Table 3.4). However, contrary to what the results of the Area Method data reduction suggest, the CC method does not identify effects on G_{Ic}^L or G_{Ic}^T due to changes in laminate thickness. As the CC

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method can be considered more reliable, besides allowing the separation of crack propagation behaviour from that of crack initiation, it can be judged with reasonable confidence that there is negligible laminate thickness effect on critical strain energy release rate of triaxial braided composite materials. However, this means that either the effects of nesting and change in fibre volume fraction are negligible or cancel each other, a subject that needs to be further explored.

3.3.3.3 Failure mechanisms

The fracture process zones developed in AS4C/RTM6 triaxial braided composite specimens are represented in Figure 3.20, which includes visible-light and radiography photographs of 12-layered specimens. The fracture mechanisms in 18-layered specimens are similar at this level of observation.

As schematised in Figure 3.20b), in transversely loaded specimens the translamellar crack tends to propagate in a saw blade pattern aligned with the axial yarns. Each dent of the saw blade pattern corresponds to the longitudinal fracture of both bias yarns, and can be inferred to have an approximate angle of 60 degrees since each bias yarn seems to have fractured in a direction parallel to the other bias yarn. The longitudinal yarns were generally not fractured through the fibres and it can be inferred that they suffered at most intra-yarn fibre splitting. Non-coalesced intra-yarn microcracks can also be identified in the bias yarns around the fracture process zone.

In longitudinally loaded specimens (see Figure 3.20), the translamellar crack also zig-zags through the laminate following yarn architecture features. The dents of the saw blade fracture pattern can be inferred to have an approximate angle of 120 degrees and are wider than for transverse loading. In this case, each side of the dent corresponds to a crack through the fibres of one of the bias yarns and a longitudinal yarn, in a direction parallel to the other bias yarn. This means that all yarns were fractured through the fibres. Moreover, compared to the transverse load case, the fracture angle was larger with respect to the longitudinal yarn direction, hence resulting in a larger intra-yarn fibre failure path. A fracture pattern involving a larger amount of fibre failure is likely to be the main cause for the higher fracture energy for longitudinal loading. Minor contributions can be attributed to the occurrence of secondary cracking parallel to the main translamellar fracture and a wider region around the fracture process zone that develops intra-yarn microcracking.

Cross-sectional cuts of 12-layered specimens for both longitudinal and transverse

loading cases are shown in Figure 3.21. In both cases, cracks were also generated at the yarn/matrix interface or grew from the inner region of the tows producing yarn to matrix interface debonding as identified in [García-Carpintero et al. \(2017b\)](#) for tensile strength specimens. These tow/interface cracks grew stably in the fibre direction mainly driven by the resolved tensile and shear forces acting in this direction due to the applied normal stresses. At some point, axial and bias yarns were severely cracked losing the stability provided by the presence of the resin and triggering the advance of the crack front until the final failure by fibre breakage and yarn pull-out. As shown in Figure 3.21, these mechanisms develop within a 2-3 mm wide band. This reveals the spread of crack branching and also the level of objectivity with respect to the definition of a crack tip in these materials.

3.3.4 Interlaminar fracture toughness

3.3.4.1 Experimental procedure

All the previously documented experimental evidences suggest that interlaminar fracture toughness is greatly influenced by fibre volume fraction and fibre architecture. The aim of this section is to evaluate the effect of the fibre volume fraction on twelve- and eighteen-layered $0^\circ \pm 60^\circ$ triaxial braided carbon fibre architecture impregnated with epoxy resin.

RTM panel dimensions were 290 mm by 290 mm and mid-plane delamination films (each 290 mm by 60 mm) of ≈ 0.07 mm thick were inserted at the laminate edges to provide starter-cracks. After infusing and curing, all the panels were ultrasound C-scanned to verify quality. Resin transfer moulded (RTM) test panels were manufactured each with twelve- and eighteen-ply of $0^\circ \pm 60^\circ$ triaxial braided carbon fibre. The resulting thickness of the twelve layered specimens was ≈ 3.2 mm (0.27 mm cured ply thickness) while for the 18-layered specimens it was ≈ 4.4 mm (0.24 mm cured ply thickness). Results of the analyses of the yarn microstructures are given in Figure 3.1b. The resulting global Fibre Volume Fraction (F_{vf}) was $\approx 55.1\%$ and $\approx 64.1\%$ respectively for 12 and 18-layered specimens.

Mode I specimens were 135 mm in length and the width was 20 mm as recommended by the ASTM Standard D5528, [ASTM International \(2013\)](#). Mode I tests were performed using the double cantilever beam method (DCB) with data reduction using the modified beam theory method. This method include the compliance of the beam in postprocessing and thus it is independent of beam stiffness.

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Additionally, where appropriate, correction factors were used to account for crack rotation and end block rigidity in accordance with the standards.

For the DCB tests, 20 mm width piano hinges were bonded to both sides of the specimens on the side of the pre-delaminated region, in order for the testing machine (servo-mechanical Instron 3340 with a 1 kN load cell) to be able to hold the specimen. The initial crack length was 50 mm to the end of the aluminium hinges. The displacement rate for the DCB test was 1 mm/min and crack propagation was observed on surfaced drawn rulers with 1mm accuracy using high magnification cameras and lenses on both sides of the specimen. Force and displacement were digitally recorded by the machine, whereas, crack growth was visually observed every millimetre and recorded in a time table as recommended by the standard. Test time was also recorded to be able to correlate the machine force and displacement data with the crack growth.

In order to ensure and control the crack propagation, specimens were loaded until a pre-crack of about 5 mm was visually observed. At that point, the crack was marked in the specimen sides and the load was removed. After unloading, the specimens were loaded again to measure the fracture toughness over a crack growth of 50 mm, starting from the earlier created crack tip.

3.3.4.2 Results and discussion

In the mode I DCB tests, all the eighteen- and twelve-layered specimens exhibited similar saw-toothed load-displacement behaviour, as shown in Figure 3.22. It can be noticed a rapid and unstable crack growth after a load drop (named as *zone A* for clarification) followed by a loading with minimal crack growth (named as *zone B*) till forming the saw-toothed profile. This made it difficult to accurately determine the critical energy release rate values from the curves. Fracture toughness can be calculated from the points associated with the onset of unstable crack growth and for other points on the rising portion of the load-displacement curve associates with crack growth.

However, the toughness after the load drop and before the re-loading point (the point where the crack arrests), must be lower than at the peak load located at the tooth vertex (hence instability), indicating that at that point the toughness is a local maximum. Furthermore, the dynamics of the unstable crack growth mean that G_{IC} calculated at *zone B* is not necessary a material property. The term "local maxima" thus refers to areas of locally high toughness (such as at the bias $\pm 60^\circ$ tows) that

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cause the crack to temporary arrest and enforces the fibre bridging prior to fibre branching and subsequent fibre breakage. Thus, although G_{Ic} calculated at the peak load of *zone A* is representative of the material fracture toughness and of use of this investigation, it must be noted that it is a local maxima.

Delamination crack growth resistance curves (R-curves), as shown in Figure 3.23, were calculated using modified beam theory method MBT, [ASTM International \(2013\)](#), as shown in the following equation (Eq. 3.8).

$$G_I = \frac{3P\delta}{2b(a + |\Delta|)} \quad (3.8)$$

wherein P is the load, δ is the load point displacement, b is the specimen width, and a is the delamination length. The expression $(a + |\Delta|)$ is introduced in the equation to correct the overestimation of G_{Ic} due to the appearance of rotations in the delamination front considering that the beam is not perfectly built-in. Δ may be determined experimentally by generating a least squares plot of the cube root of compliance, $C^{1/3}$, as a function of delamination length. The compliance, C , is the ratio of the load point displacement to the applied load, δ/P .

Using the formula above, fracture toughness was calculated every 1 mm crack growth. However, the standard recommendation is to record the first and the last 5 mm crack growth and every 5 mm in between the 5-45 mm crack growth. A high amount of unstable crack growth was produced by the tow fibre failure during the loading drops, showing high scatter in between the calculated values. Therefore, the moving average technique was used to reduce the scatter with a window size of 5 points.

The resulting R-curves are presented in Figure 3.23. Two regions of fracture toughness can be distinguished within the curves. The first one is the initiation value, caused by the intertow debonding due to the pre-crack generated during the initial loading step. The second one, is the propagation value, which is higher than the initiation due to the increase of fibre bridging and tow branching.

In both cases, the R-curves increased over about the first 20 mm of crack growth, and this was caused by the gradual formation of a fibre bridging zone behind the crack front, as shown in Figure 3.25, which is the main toughening process in these type of fibre architecture, [Huang and Hull \(1989\)](#). Neglecting the scatter between coupons, the R-curves show that the interlaminar fracture toughness values remained reasonably constant beyond 25 mm, and during DCB testing it was observed that

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the delaminations grew at a reasonably constant speed and in a stable manner. In the case of the lower fibre volume fraction specimens, the twelve layered ones, the steady-state propagation or Length of the fracture process zone ($L_{f_{pz}}$) can be located at a crack length in between 25-35 mm, whereas in the thicker specimens with higher volume fraction, the steady-state crack propagation is found to be in between 30-40 mm.

The strain energy release rate for crack initiation (G^I_{Ic}) and the steady-state crack growth (G^P_{Ic}) of both thickness are listed in Table 3.5. The G^I_{Ic} value was taken to be the average over the first 5 mm of the R-curves, resulting in $890 \pm 99 \text{ J/m}^2$ and $1033 \pm 113 \text{ J/m}^2$ for the twelve- and eighteen-layered ones respectively. The G^P_{Ic} value was taken for an average of the G_{Ic} values over the region where reasonably steady-state crack growth occurs, deriving in $1131 \pm 151 \text{ J/m}^2$ and $1300 \pm 203 \text{ J/m}^2$ for the twelve- and eighteen-layered specimens respectively.

Table 3.5: Values of initiation and propagation critical energy release rate for 12/18 layered specimens obtained by the MBT Method

Spec. Config.	F_{vf} [%]	$L_{f_{pz}}$ [mm]	G^I_{Ic} [J/m^2]	G^P_{Ic} [J/m^2]
12 Layers	55.1	25-35	890 ± 99	1131 ± 151
18 Layers	64.1	30-40	1033 ± 113	1300 ± 203

Despite the scatter found when increasing the fibre volume fraction, it is visible the effect of the fibre architecture into increasing the interlaminar fracture toughness when comparing with other architectures made of similar fibre composition and fibre volume fraction. *Gill et al.*, [Gill et al. \(2009\)](#), reported that the the G_{Ic} propagation values of a carbon fibre/epoxy resin five harness satin (5HS) woven composite for 51.1%, 54.4% and 61.3% were 430 J/m^2 , 525 J/m^2 and 741 J/m^2 respectively.

The R curves show considerable variation in the G_{Ic} values as the single/multiple delamination fronts grew through the braided composites, and this is indicative of a reasonably unstable crack growth process. Crack branching is almost the main toughening mechanism responsible for the higher interlaminar fracture toughness of the braided specimens, [Minguet et al. \(1994\)](#); [Mouritz et al. \(1999a\)](#). This process was studied in greater detail by examining the braided specimens in cross-section using a optical microscope of post-mortem specimens, as illustrated in Figure 3.24.

Clearly two mechanisms can be distinguished, fibre branching and fibre bridging

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around the tow boundaries. This two mechanisms occur because the crack did not propagate exactly through the resin in the mid-plane of the coupon, caused by the nesting or architecture compaction during the manufacturing process. It was observed that the crack at the edge of the specimen did not follow a straight path but a wavy pattern over and under the fibre, as detailed in Figure 3.25, caused by the locking effect of the braided architecture. Fibre bundles laying in the surface which pass over and under other fibre bundles in the ply were pulled out, which results in delamination in the tow to tow interface. At a certain time, the bond between the filaments in the yarn fails which causes fibre branching, after which the crack can follow either in the delaminated interface between the yarns or in (still bonded) interface between other yarns creating more delaminations. Figure 3.26 shows a schematic top-view of the crack front propagation in the DCB tests of a triaxial braided composite summarised in the stages(A, B and C).

As the delamination front progresses between two layers of the triaxially braided composite, the crack is confined between two sets of bias yarns (Stage A). At a further crack propagation (Stage B), the crack jumps under (I on Stage B) or over (II Stage B) a bias yarn. Jumping underneath a bias yarn in the layer below the delamination front, results into that specific bias yarn bridging across the delamination. This bridged yarn starts loading with the increasing specimen opening and preventing the spreading or progression of the delamination. As result, significant build up can be seen in the calculated fracture toughness as the force required for propagating the crack increases. Failure of the bridged yarns releases the delamination front (Stage C) and results in a rapid crack propagation as shown in the saw-tooth behaviour of the DCB test.

By examining the fracture surface of a tested specimen, fibre bundles were observed both on the top and on the bottom fractured surfaces. This can be seen in Figure 3.27, where the two red boxes indicate breakage of bundle fibres. Moreover during testing, fibre bridging was observed as seen in in the red box in Figure 3.25. These two mechanisms caused a significant increase in the interlaminar fracture energy.

3.3. Property Characterisation

Figure 3.13: Shear test specimens loaded in both longitudinal and transverse directions.

Figure 3.14: Longitudinal (a) and transverse (b) specimen stress-strain shear response

3.3. Property Characterisation

(a)

(b)

Figure 3.15: Longitudinal (a) and transverse (b) crack propagation due to ASTM geometry alterations (5 mm notch)

Figure 3.16: Two loading configurations for Compact Tension (CT) coupons of triaxial braided composite materials. Dimensions are in mm.

3.3. Property Characterisation

(a) 12-Layer, Longitudinal loading

(b) 12-Layer, Transverse loading

(c) 18-Layer, Longitudinal loading

(d) 18-Layer, Transverse loading

Figure 3.17: Experimental load-displacement curves obtained for 12- and 18-layered CT specimens oriented in two material directions.

3. Characterisation and testing of braided composites

Figure 3.18: Compliance vs. crack length curves obtained by fitting Equation 3.6 to experimental results.

3.3. Property Characterisation

Figure 3.19: *R*-curves obtained from CT testing and Compliance Calibration data reduction method. The dashed lines are the mean values.

Figure 3.20: Crack propagation in the 12-layered triaxial braided AS4C/RTM6 composite for longitudinal and transverse load cases. Top: CT specimen visible-light photographs. Middle: CT specimen radiography photographs. Bottom: Schematics of crack propagation.

3.3. Property Characterisation

Figure 3.21: Cross-sections of 12-layered specimens perpendicular to the crack propagation: a) longitudinal loading and b) transverse loading. See Figure 3.16 for reference.

Figure 3.22: Mode I saw-toothed load-displacement behaviour for twelve- and eighteen-layered specimens.

3.3. Property Characterisation

Figure 3.23: R curves showing initiation and propagation fracture toughness values for the DCB test specimens

Figure 3.24: Optical micrographs showing delamination cracking in the $0^\circ \pm 60^\circ$ DCB 12 layered braided composite. The delamination has experienced extensive branching around the fibre tows (indicated in the region B) as well as through the tows (region A). Arrow indicates crack growth direction.

Figure 3.25: DCB specimen during testing. Green box showing the wavy pattern of the crack front and fibre bridging in the red box

Figure 3.26: Schematic of the crack front propagation stages through the mode I DCB test: A) crack confinement; B) fibre bridging; and C) fibre branching and bias tow tensile failure.

Figure 3.27: DCB specimen during testing. Green box showing the wavy pattern of the crack front and fibre bridging in the red box

3.4 Structural testing

Coupon level testing typically benefits all design areas that are fabricated using the material of interest. The required testing to support an analysis method should take advantage of established property characterisation tests and then select additional tests required for detailed analysis of the mechanical behaviour of the material under structural conditions. Structural testing is used to represent complicated details or to evaluate not visible effects which cannot be analysed using coupons, such as mechanisms (buckling, crushing, and compression after impact), effects of stress concentrations and geometrical non-linearities, size effects, etc.

3.4.1 Notched laminates

In this section, six-layered open-hole coupons were used to investigate the effect of hole scaling onto the tensile strength of notched composites and the relation with the RUC dimensions ($\approx 10 \times 3$ mm). Hole diameter was investigated as the independent variable, whilst keeping constant cure ply thickness, stacking sequence, hole diameter to width ratio and width to length ratio.

3.4.1.1 Experimental procedure

For this reason six-layered coupons with three different notched sizes (2,4 and 6 mm) and the driving W/D parameter kept constant, Width to diameter ratio, were manufactured in both longitudinal and transverse direction in accordance with ASTM D3039 [ASTM International \(2000\)](#). From publications, [Minguet et al. \(1994\)](#), and standard, [ASTM International \(2000\)](#), it is recommended to use $W/D=6$.

Specimens were cut by a Computerised Numerical Control (CNC) milling machine. This kind of machines are automated and follows a predetermined sequence of machining operations at the required speed to produce a workpiece of the right shape and size. The notched shapes were created by drilling using different drill sizes in the geometrical centre of the coupons. It has to be noted that no damage was observed after the cutting operations of the coupons.

Three homothetic coupon configurations were cut, where the the biggest sized rectangular specimen, named as OHT6, had 200×36 mm², as recommended by the ASTM standard D3039. The geometric characteristics are presented in Table 3.6.

The smaller sized homothetic coupons listed in Table 3.6, named as OHT4 and OHT2 for simplicity, had 2/3 and 1/3 of the initial width recommended from the

3.4. Structural testing

Table 3.6: OHT specimen dimensions

Type	Hole diameter (mm)	Width (mm)	Length (mm)	Effective length (mm)
OHT6	6	36	200	90
OHT4	4	24	135	63
OHT2	2	12	70	32

standard, resulting in $135 \times 24 \text{ mm}^2$ and $70 \times 12 \text{ mm}^2$ respectively. For each coupon manufactured, as previously denoted, the width to hole diameter (W/D) parameter was kept constant and equal to 6, resulting in 3 different hole sizes of 6, 4 and 2 mm as illustrated in Figure 3.28 and listed in Table 3.6.

Figure 3.28: OHT 6-layered triaxial braided AS4C/RTM6 scaled coupons. Speckle painted in a ROI to properly represent the strain field.

Falzon and Herszbergb (1998), suggested that the specimen width should con-

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tain, at least, two unit cells of the fabric to be representative of the material mechanical behaviour. However, in the longitudinal direction, from the three different scaled sizes proposed, only two fulfil this requirement, while the other one, 12 mm wide, it only contains one unit cell, and therefore its size effects are studied. In the transversal direction, as the unit cell is rotated (3x10 mm), at least 4 unit cells are contained in the 12 mm width coupons. The main goal of this study, is to determine the tensile strength of the textile architecture considered and to study the hole size effect with respect to the unit cell dimensions in both longitudinal and transverse directions.

The specimens were tested under uniaxial tension in an electromechanical universal testing machine, Instron 3340, under stroke control at a speed of 1mm/min. Different load cells have been used depending on the size of the specimens. For 36 mm width specimens, a 150 kN load cell was used to record the load; meanwhile for the other two sizes, a load cell of 30 kN was sufficient to record the loads with accuracy. In order to transfer properly the stresses from the clamps to the samples, double sided sandpaper was placed between them in both sides as recommended by the standard, being enough to avoid damage and delamination. Tabs were not necessary, which implies lower manufacturing costs.

Samples were painted in white and speckled with a black dots pattern over a ROI (see Figure 3.28) representative to capture the strain fields. VicSNAP (Correlated Solutions) was also used to digitally measure the strain field during the deformation process at a rate of 1 picture/second.

Uniaxial tensile test on $[0^\circ]_6$ specimens were conducted along the axial yarn and its perpendicular direction following the standard. The applied stress was computed by the readings of the load cell at a constant cross-head speed 1 mm/min and with the cross-sectional area of the samples. The strains were determined using DIC technique previously described, over a ROI sufficient to guarantee an uniform strain field. The ROI for the different sizes of samples were 40x30 mm² for 36 mm width specimens, 27x20 mm² for 24 mm width specimens and 18x10 mm² for 12 mm width specimens. A minimum of 5 samples of each coupon geometry (6,4 and 2 mm hole drilled) and on each loading direction (longitudinal and transverse) with respect to the axial direction were tested to represent accurately the mechanical behaviour.

3.4. Structural testing

3.4.1.2 Results and discussion

The response of the material was not linear and elastic as observed from the tensile test. In fact, the geometrical size effect of the notched hole produced a non-linear response from the beginning of the test till failure by fibre breakage produced by damage accumulation. The effect of the geometrical non-linearity onto the mechanical response of the material was more evident in the longitudinal coupons with higher hole diameter, varying not only the strength but the stiffness. This effect can be related with the local inhomogeneity of the material, being the unit cell dimensions (10x3 mm) the minimum required to represent the mechanical behaviour of the material. As consequence, for the longitudinal specimen, when the hole diameter is the same order of magnitude as the RUC dimensions the non-linear trend is not likely affected, but if the drilled hole is bigger than the RUC dimension (3mm height) the mechanical response is significantly modified. On the contrary for the transverse specimens, the RUC is therefore rotated 0° (3x10mm) and not being the trend affected in the same amount as in the longitudinal specimens. In any case, the experimental scatter was minimum and all the curves superimpose together, as illustrated in Figure 3.29. Ultimate strength and strain from the tested specimens are presented in Table 3.7.

Figure 3.29: Strength variation of braided composite laminates with the increase of drilled hole diameter.

According to the values presented in Table 3.7, the impact on the tensile strength

3. Characterisation and testing of braided composites

Table 3.7: OHT mechanical properties.

Type	Hole diameter (mm)	Tensile strength (MPa)	Strain-to-failure (%)
Longitudinal specimens			
OHT2	2	438±52	0.94±0.1
OHT4	4	429±41	0.92±0.1
OHT6	6	375±22	0.89±0.04
Transverse specimens			
OHT2	2	424±5	1.16±0.03
OHT4	4	433±7	1.05±0.02
OHT6	6	368±14	0.87±0.04

from the 2 and 4 mm diameter hole is negligible. However, the strength in the specimens with a drilled hole of 6 mm has a significant variation caused by the increase of initial damage. Nevertheless, it is clear that even a small defect is affecting the mechanical response when compared to the specimens with no defect or negligible defects, as previously reported (673±55 MPa and 630±23 MPa for the longitudinal and transverse specimens).

Figure 3.29a and b represent the strength variation with the increase of drilled hole diameter for longitudinal and transverse direction. It can be noticed in the first part of the graph, the strength of the composite decreases abruptly with the presence of an small defect, caused by the increase of the natural inhomogeneity of the material. Between 2 and 4 mm hole diameter, the strength is almost flat for both directions. It can attributed to the fact that the defect is inferior or approximate o the unit cell dimensions. Above 4 mm, It can be noticed that the strength starts again to decrease due to the growth in damage accumulation causing an inhomogeneous response. The same response was observed by Hwan et al. (2011), in which the normalised notched strength decreases with the increase in the size of hole. In other words, the notch sensitivity of braided composite plates increases with the increase in the size of the hole. Furthermore, both curves superimpose, proving again the quasi-isotropic response of the material even in the presence of mechanical defects, as shown in Figure 3.30.

In general, there is no obvious damage and delamination around the edge of a drilled hole in a specimen macroscopically just after drilling if reasonable working conditions are provided. To investigate the possible damage caused by drilling,

3.4. Structural testing

Figure 3.30: Variation of OHT strength with the increase of drilled hole diameter in comparison with the tensile strength of unnotched specimens (673 ± 55 MPa and 630 ± 23 MPa for the longitudinal and transverse)

the drilled specimen was sliced near the rim of the circular hole along the axial yarn and bias yarn direction. These slices were then observed through using optical microscopy. It was noted that delamination occurs near the rim of the hole and especially obviously in the lower layer of the specimen for both cases. This initialisation of delamination and crack in matrix near the rim of the hole caused by drilling is usually referred to as the damage zone, [Awerbuch and Madhukar \(1985\)](#). No propagation of the initial damage caused by drilling was observed along the direction of the yarns. This initial damage is previously described in the unnotched tensile specimen and defined as edge effect caused by the coupon machining onto the fibre architecture, generating local material inhomogeneous regions prone to initiate damage.

[Nakai et al. \(2001\)](#), investigated the initialisation of damage on the tensile test of braided composite plates with machined holes. It was observed that matrix crack occurred along the discontinuous fibre bundle from the rim of a hole with the debonding between fibre bundles and propagated with changing the direction from the direction of fibre orientation to the perpendicular direction to the loading direction. It was found from observing the post-mortem specimens (see [Figure 3.31](#))

3. Characterisation and testing of braided composites

that the matrix crack initiated at the stress concentration point, which is located in the region with matrix concentration in between two overlapping bundles at the edge of the rim hole. The propagation process was similar to the one occurred in the unnotched specimens, creating debonding and delamination between fibre bundle and matrix till the final failure by fibre breakage.

Figure 3.31: Fracture specimens with a centre hole of radius 2, 4 and 6 mm on longitudinal direction x (axial yarns) and transversal direction y (perpendicular to axial yarns).

3.5. Conclusions on characterisation and testing

When the applied load reaches its maximum value, the notched specimen breaks suddenly and the load drops instantaneously to zero, as in general composite material under tensile loads. The failure is therefore the same as the unnotched coupons, failing catastrophically with a zig-zag fracture surface, typically of triaxial braided architecture with the axial yarn oriented in the loading direction, that propagates perpendicular to the direction of the applied load and almost no visual indication of damage development in the outer surfaces prior to failure. Macroscopically, breakage of fibre bundles can be clearly found in Figure 3.31a). For the transverse specimens, the presence of fibre bundles perpendicular to the loading direction, produced a dominant crack propagation parallel to the fibre bundles through the width of the coupon. This propagation pattern is similar to the one obtained from an unidirectional composite laminate oriented perpendicular to the loading direction.

As observed from Figure 3.31, the fracture process zone is similar to the tensile test on unnotched coupons and the mechanical behaviour is only varied because the inclusion of a drilled hole acting as an additional free edge, which creates damage accumulation and triggers a more non-linear response. Therefore, damage mechanism are suspected to be the same as in unnotched specimens.

3.5 Conclusions on characterisation and testing

3.5.1 Yarn architecture characterisation

From the different methods for the characterisation of the yarn architecture of textile composites reported in the literature, optical microscopy was used as baseline technique and additional investigations using X-ray CT and image analysis of surface scans were conducted.

Optical microscopy was found to be an appropriate technique to acquire most of the yarn architecture parameters needed to build up a geometric model of the braided composites. Nesting and bundle geometrical relationships for six-, twelve- and eighteen-layered panel were determined using image treatment with Image J software. Nevertheless, micrographs were found to be not suitable to reliably measure the braiding angle.

Additional information on the yarn architecture regarding three-dimensional effects and nesting could be acquired by using micro-CT scanning. The shape and dimensions of the yarns were different near the surfaces, differing with the yarn cross-section dimensions (e.g. yarn width) obtained from the surface scanning. Qualitative

information of the yarn architecture could be extracted from the X-ray CT scans, but a quantitative information was found to be more difficult to obtain, as an automated and robust segmentation of the CT-scans is required. This proved to be difficult due to the low contrast between matrix and carbon fibre bundles.

All investigations presented here have been conducted on plane braided composite panels, but the strategy presented for yarn architecture characterisation can also be applied to complete braided structures with variable yarn architecture on different regions of the component.

3.5.2 Property characterisation

A property characterisation and analysis methodology for two-dimensional triaxial braided composites with different thicknesses was developed. In order to do so, tensile and shear tests were performed to determine the relevant in-plane mechanical properties, fracture toughness measurements were obtained through translaminar compact tension and Mode I DCB tests, and in addition, the structural capability was evaluated through independent testing using Open Hole coupon geometries. Nesting and size effects present on structural laminates were also studied.

The in-plane tensile behaviour of six-layered RTM6 injected triaxial braided carbon fabric QISO-L-52 under tensile loads was determined experimentally. Interrupted tensile tests were carried out in composite coupons cut in the longitudinal and transverse directions to ascertain the damage mechanisms occurring up to failure using the radiographs acquired sequentially. The stress-strain curves of the triaxial braided loaded in the longitudinal direction were almost linear and elastic up-to failure controlled by the strain-to-failure of the axial yarns. In the case of specimens loaded in the transverse direction, the material exhibited a more non-linear response and this effect was attributed to the induced damage due to edge effect. The response of the material in terms of elastic moduli and strength in both directions did not differ significantly, indicating a good mechanical balance between the axial and $\pm 60^\circ$ tows of the material.

The shear strength and stiffness of six-layered triaxially braided composites was determined experimentally following one of the proposed ASTM Standard Methods. Among them, [ASTM International \(2005\)](#) was chosen due to the orthotropic nature of the material. It was observed damage accumulation triggering early failure at the 45° notched edges instead at the V-notch. This effect caused a development of tensile and compressive strains along the free edges of the notch resulting in delamination

3.5. Conclusions on characterisation and testing

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till final fibre failure. This behaviour was reported in the literature and it is characteristic of complex fibre architectures, which develop multiple crack fronts following the fibre direction, failing to propagate through the ligament between notches. To overcome this effect, an artificial pre-crack of $\approx 5\text{mm}$ was introduced in between the notch tips, which induced further stress concentration in the ligament between notches allowing a enforced the propagation between the notch tips. It was observed that the ultimate strength, strain to failure, and shear stiffness remained invariant for both longitudinal and transverse loading directions of the un-notched specimen. This can be endorsed to the quasi-isotropic nature of the material. Nevertheless, despite the weakening effect introduced by the artificial pre-cracking, the shear stiffness was found greater in comparison with the standard geometry specimens. The complex architecture of the triaxially-braided material dominates the crack propagation through different damage mechanisms, and it was concluded on the necessity of geometry alterations from the ASTM D7078 standard. The proposed geometrical modification, 5mm pre-cracking, to enforce the crack propagation and a more pure shear response during the test, might be valid to predict the shear response and strength for longitudinal specimens. However, in the transverse direction, early failure was triggered by the instability introduced by the artificial notch creating damage accumulation. Therefore, it is recommended the use of standardised geometry for stress prediction and the pre-crack geometry alteration for stiffness determination.

The translaminar fracture response of twelve- and eighteen-layered triaxial braided composites was analysed by means of Compact Tension (CT) experiments and two reliable data reduction procedures. The fracture mechanisms and values of critical strain energy release rate were found to be dependent on the direction of crack propagation with respect to the material architecture. The toughness for fracture transverse to the longitudinal yarns was found to be 50-70 % higher than that for fracture along those yarns. This effect was correlated with the mechanisms of fracture. It was found that, in longitudinal loading, the crack cuts through all yarns by means of fibre breakage - the yarn failure mode associated with the highest level of energy dissipation. The fracture mode parallel to the longitudinal yarns only cuts through the bias yarns leading to a lower value of the translaminar critical strain energy release rate. Furthermore, it was concluded that the laminate thickness has negligible effect on the translaminar fracture toughness.

Mode I DCB tests were performed to evaluate the influence of the fibre volume

3. Characterisation and testing of braided composites

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fraction and fibre architecture onto the determination of the interlaminar fracture toughness of twelve- and eighteen-layered specimens. The specimen exhibited similar saw-toothed load-displacement behaviour with rapid and unstable crack growth after load drop, which influenced the accurate determination of the critical energy release rate values from the force-displacement curves. Delamination crack growth resistance curves, (R-curves) were derived using the modified beam theory method (MBT) at every 1 mm crack growth. Unstable crack growth was observed due to the tow fibre failure during the loading drops influencing calculated values and therefore showing high scatter. Thus, the moving average technique with a windows size of 5 mm was used to reduce the scatter. Furthermore, two regions of fracture toughness could be distinguished within the curves, being the first one the initiation value, caused by the intertow debonding during the pre-crack generation at the initial loading step, and the second one, reflects the propagation value, which is higher than the initiation value due to the increase if fibre bridging and tow branching. Despite the scatter found when increasing the fibre volume fraction, it is remarkable the effect of the fibre architecture into increasing the interlaminar fracture toughness, when comparing with other architectures with similar fibre composition and fibre volume fraction. This is caused by the multiple delamination front growing through the braided composites, creating an unstable crack growth which yielded into an increase of crack branching, which is the main toughening mechanism responsible for high interlaminar fracture toughness.

Additionally, to the property characterisation, an independent OHT testing campaign was carried out on six-layered triaxial braided composites to evaluate its structural capacity and to demonstrate whereas the material possess adequate performance while investigating the effects of hole scaling with respect to the RUC dimension. Three homothetic coupon configurations were evaluated, in which the width to hole diameter ratio was kept constant. The response of the material was not linear and elastic as in the un-notched tensile tests, caused by the geometrical size effect of the notched hole creating damage accumulation and producing non-linear response from the beginning of the test till failure by fibre breakage. This effect did not produce the same results independently of the fibre alignment. For longitudinal specimens, when the hole diameter is in the same order of magnitude as the RUC dimensions, the response remained invariant, but when increasing the hole diameter (larger than 3mm), the strength and strain-to-failure are significantly affected, dropping up to a 55% of the tensile strength of un-notched tensile tests for

3.5. Conclusions on characterisation and testing

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the 6mm hole diameter. On the contrary, for the transverse specimens, the trend is not affected in the same amount, showing similar non-linear response when the hole diameter is smaller than half of the longer transverse RUC dimension (10x3mm), but the strength unquestionably decreased, up to a 58% when comparing with unnotched transverse tensile tests, with 6mm hole diameter. Nevertheless, the observed fracture process zone and crack propagation pattern is similar to the tensile tests on unnotched coupons and the mechanical behaviour is only affected because the inclusion of a drilled hole, which creates additional damage accumulation around the free edges of the specimen triggering a more non-linear response and early failure.

Computational modelling of the mechanical behaviour of braided composites

4

The objective of this chapter is to develop a numerical methodology to predict the mechanical response of a triaxial braided laminate given the mechanical behaviour of its constituents as well as the textile architecture. The model will be validated using experimental results presented in previous chapters and its ability to address the different deformation and failure mechanisms addressed.

The prediction of the mechanical response of textile composites is not new in the scientific and technical literature. Characterising the mechanical behaviour of any material, and in particular composites, always requires costly and extensive testing campaigns so models to predict mechanical and physical properties will be necessary to accelerate design processes.

Among all the numerical techniques, the Unit Cell Approach, is a well established method for the prediction of mechanical properties of textiles, specially when non-linear behaviour is expected. Such geometrical models assumes a basic repeating unit cell of the textile to be representative of the overall behaviour of the material. The fibre architecture, yarn-to-yarn interactions, are included explicitly in the model.

In this chapter, the geometrical model of the ($\pm 60^\circ$) triaxial braided composites is based on the textile and yarn architecture parameters that were measured by means of optical microscopy in the previous chapters. WiseTex software package, [Lomov et al. \(2001\)](#), was used to represent independently the bias and axial yarns

4.1. Geometric modelling of triaxial braided textiles

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in a single unit cell of the textile. WiseTex calculates automatically for a set of global parameters, the geometry of yarns, cross-section and its variation along the direction of the yarn, taking into account yarn-to-yarn non-penetration constraints. A comparison between the resulting geometry and the X-ray computed tomography measurements is carried out.

4.1 Geometric modelling of triaxial braided textiles

Most geometric models for braided and woven composites make use of idealised descriptions of yarn paths and cross-sections [Goyal \(2007\)](#); [Huang and Ramakrishna \(2002\)](#); [Aggarwal et al. \(2002\)](#); [Robitaille et al. \(2003\)](#). Several software packages such as WiseTex [Ivanov et al. \(2009\)](#) and TexGen [Robitaille et al. \(2003\)](#) were developed for geometry modelling of textile composites offering a general framework to define the corresponding unit cells. WiseTex computes the internal geometry of the textile from the user input, such as yarn properties (e.g. shape of cross-section) and weave patterns, based on an energy minimisation of algorithm, [Lomov et al. \(2001\)](#). The yarn path is described in terms of elliptical arcs and 5th order polynomial functions. The cross-section can be chosen from either circular, elliptical or lenticular. WiseTex also serves inputs for various interfaces, micromechanics (Texcomp, [Huysmans et al. \(1998\)](#)), permeability modelling (FlowTex, [Verleye et al. \(2010\)](#)), and finite element (FETex) software. Recently, a command-line interface and an open XML-data format have been implemented in WiseTex, [Lomov et al. \(2014\)](#). This allows for straightforward integration of WiseTex geometry models into simulations performed using 3rd-party software.

TexGen, [University of Nottingham \(2019\)](#), is an open-source software package for textile modelling, providing a Python scripting interface and various export options to Abaqus FE software, [Dassault Systèmes \(2013\)](#). TexGen uses spline interpolation for the yarn paths with master nodes defined by the user as well as a library with yarn cross-sections including elliptical, power-elliptical and lenticular, [Sherburn \(2007\)](#). The models from TexGen can also be exported to 3rd-party software using common geometric formats as IGES or STEP, or as a tetrahedral or hexahedral continuum mesh to the FE-software Abaqus, [Crookston et al. \(2007\)](#).

Generally speaking, geometrical models are fast and simple to obtain using the aforementioned software. However, the complexity increases considerably when high-performance textile architectures, as triaxial braided, including large volu-

metric fraction of fibre reinforcement are required, because the large problems related with yarn-to-yarn interpenetrations, Figure 4.1, Lomov et al. (2007); García-Carpintero et al. (2015). In this case, finite element discretisation using continuum elements requires consistent yarn geometry avoiding interpenetration and volume overlaps.

Figure 4.1: Volume interpenetration for triaxial braided textile: left, generated with a CAD software, right, generated with a textile fabric generator (TexGen), García-Carpintero et al. (2015)

Researchers proposed several improvements to overcome the problem of volume interpenetration by using analytical considerations during the finite element contact analysis. Sherburn (2007) used an idealised geometry of a 4-harness satin weave modelled with TexGen as a basis, refining the yarn path definition, and changing the cross-section shape and its orientation at the cross-over points of the unit cell. The improved geometry was validated against X-ray tomography measurements and provided good correlation of yarn path, cross-section shape, and yarn orientation angle. In a second publication of the same authors, an automated algorithm for correction of yarn volume penetration based on a FE contact analysis using plate elements was proposed, Sherburn et al. (2012). The approach has been implemented in TexGen and led to substantial improvements in 2D textiles. On the other hand, the results for a 3D orthogonal weave were less accurate, and still showing some interpenetration problems.

Lomov et al. (2007), presented a different approach to solve interpenetration problems into the MeshTex software, Zako et al. (2003). With this method, the unit cell is divided into several sub-volumes, each one representing one contact pair. The corresponding yarn volumes are moved in such a way that interpenetration is relaxed. The yarns are defined to be volume-constant and pressed together to create tight, non-interpenetrating volumes. In the last step, the yarns are re-assembled into the complete unit cell.

4.2. 3D representative unit cell (RUC) modelling

Most of the methods in the literature made use of compaction algorithms. In a first step, the textile architecture is generated based on a low volume fraction of reinforcement avoiding hard contact problems between adjacent yarns. In a second step, the RVE to the textile is compressed in the thickness direction increasing the volume fraction until the target value. Different methods modelling of compaction phase have been presented in the literature, all of them using FE as contact solver. Discretisations used in a previous simulation can be used as inputs in a subsequent compaction simulation leading to realistic volume fraction of fibre reinforcement. For instance, [Hsu and Cheng \(2013\)](#) presented an general approach to generate the textile geometry using a multiple-step FE analysis. The method included FE simulation for the compaction phase, the reconstruction of the geometry, and removal of geometric inconsistencies as interpenetration and voids. [Stig and Hallström \(2012b\)](#) recently presented a new FE-based approach for realistic geometric modelling of 3D woven textiles. Based on TexGen output geometry, yarns were modelled as inflatable tubes in an explicit FE analysis. The validation of the calculated yarn geometry with X-ray tomography scans showed good agreement.

4.2 3D representative unit cell (RUC) modelling

Three-dimensional unit cell models are used to predict the mechanical properties of textile composites. This requires a realistic representation of the yarn architecture in the RUC and to reproduce the correct fibre volume fraction. Cross-sections of the composite material were observed by optical microscopy (see Figure 3.2) and the relevant parameters for the generation of the unit cell, yarn shape and cross-section dimensions, local volume fraction, etc, were measured. All these parameters were gathered in Table 4.1.

The yarn architecture parameters given in Table 4.1 are used as an input for the geometric models. With these inputs, the undulated yarn paths are calculated using WiseTex by means of energy minimisation algorithms: the fabric architecture given by the braid pattern is divided into smaller parts, with the yarns being represented by structural elements. The yarn path is approximated by a parametric function $z(x)$ and its parameters are determined by minimisation of the elastic energy of the structural elements. A detailed description of the method can be found in [Verpoest and Lomov \(2005\)](#). As a instance, this section is focused on the generation of a three-dimensional RUC to capture the mechanical response, stiffness and strength, of a six-layered braided composite material with a fiber volume fraction of $\approx 51.5\%$.

4. Computational modelling of the mechanical behaviour of braided composites

Table 4.1: Input parameters for WiseTex unit cell modelling using a 6 layers specimen as reference.

name	variable	value (axial/bias)	source
geometric data			
yarn shape	–	lenticular	micrographs
yarn height	t	0.126/0.087	micrographs (Figure 3.2)
yarn width	w	3.16/2.03	micrographs (Figure 3.2)
spacing	d	5.01/2.89	micrographs, datasheet A&P Technology (2016)
braiding angle	θ	60°	micrographs, datasheet A&P Technology (2016)
fiber data			
filament diameter	ϕ	7 μ m	micrographs, datasheet Hexcel (2016b)
number of filaments	–	6k/3k	micrographs, datasheet A&P Technology (2016)

4.2.1 Geometric and FE model

The RUC geometry for the $\pm 60^\circ$ triaxial braid generated in WiseTex is shown in Figure 4.3. In order to validate the geometric model calculated from WiseTex, the yarn geometry was compared to the experimental results from X-ray images and optical microscopy, as shown in Figure 3.2 and in Figure 4.2.

The maximum fibre volume fraction attained using WiseTex is $\approx 40\%$ which deviates considerably from the experimental one. Among the main reasons, discrepancies can be endorsed to the deformation of the yarn cross-section, yarn cross-section rotation or differences in the yarn architecture in the outer plies (as shown on Figure 3.2). All these problems are not considered using WiseTex, [Lomov et al. \(2001\)](#); [Verpoest and Lomov \(2005\)](#); [Cichosz \(2015\)](#). As a consequence, volume interpenetrations are present in the RUC in the idealised geometry model giving unrealistic values of overall fibre volume fraction. To overcome these problems balancing the low geometrical volume fraction (volume fraction of tows in the RUC), a fictitious local volume fraction of fibres from 60-80% can normally be used, [Ivanov et al. \(2009\)](#); [Schultz and Garnich \(2013b\)](#); [Cichosz \(2015\)](#). This approach is reasonable to ensure good predictions of the elastic constants of the textile composite. However, such approximations might result in unrealistic stress concentrations preventing its use when damage need to be simulated accurately.

The WiseTex geometric model can be saved as either binary or XML file types. If solely the geometry of the model shall be exported, WiseTex allows exporting the yarn geometry to an ASCII (*.cfl) file. The *cfl* file saves the information of the unit cell and the yarns in a structured format, including unit cell dimensions,

Figure 4.2: Schematic view of sections along the axial fibres (Section B-B') and along the bias fibres (Section A-A')

braiding angle and information about the geometric and material properties of the yarns defined in the model. The information about the yarn architecture is saved in segment-wise form for all the yarns of the textile. Every segment of the yarn includes information about the yarn location, cross-section and orientation of the cross-section. Once the model is generated, the geometry is exported to Abaqus FE software package, [Dassault Systèmes \(2013\)](#), via a Python scripting. For instance, the resultant model for the triaxial braided textile is shown in Figure 4.4.

As it can be observed, the geometry contains some volume/yarn interpenetrations due to the input parameters introduced in WiseTex and due to idealisation of the architecture generation. Several approximation were reported in the literature and described in Chapter 2 to overcome such yarn interpenetrations. Among them, a quasi-physical thermo-mechanical approach was used, similar to the ones described in [Stig and Hallström \(2012a,b\)](#); [García-Carpintero et al. \(2016a\)](#).

In this thesis, a method based on explicit FE analysis is proposed for generat-

Figure 4.3: Schematic model generated with WiseTex software using the geometrical data from Table 4.1.

Figure 4.4: Geometric model for the triaxial braided composite obtained imported into Abaqus

ing the realistic textile geometry (yarn trajectories, cross-section size and shape) while avoiding spatial overlaps and obtaining the desired F_{vf} . The basic idea of the modelling strategy is to first generate a rough nominal geometry of the fabric, based on the initial one obtained from WiseTex software. In this case, the principal yarn paths are defined with no interpenetrations without substantial efforts. The initial overlaps present in this initial geometry generated with WiseTex are removed by reducing the elliptical strand-cross section areas until all the strands are totally separated and no overlaps remains. The reduced strands are then modelled as solid

4.2. 3D representative unit cell (RUC) modelling

bundles, which are expanded using an artificial temperature increase until the free spaces within the RUC are filled. The desired ply thickness is then achieved by compressing the expanded mesh using two parallel planes under general contact conditions. The proposed scheme involves the use of different commercial software (Abaqus, HyperMesh and WiseTex), managed by scripts for reading and processing the data necessary to generate the different input files, and transfer data between the software packages. The procedure is schematically illustrated in Figure 4.5 and performed through the following steps:

Figure 4.5: The sequence of the proposed geometry modelling framework

Step 1. In the first step, the nominal strand distribution with no overlaps is generated using WiseTex textile generator. The fiber bundles are all assumed to have an elliptical cross sectional shape. The nominal cross-section areas are set sufficiently small, or uniformly shrunk to $\approx 75\%$ of their initial dimensions listed in Table 4.1, to eliminate any strand overlap.

Step 2. The nominal geometry containing no overlapping is then exported

from WiseTex to Abaqus via Python script to visualise the geometry and then to an advanced/complex mesh generator software (HyperMesh 14, [Altair Engineering \(2016\)](#)). The mesh generator software is used to form the dry-fabric RUC and create periodic surface meshes of the strands in the RVE. A volume mesh is generated using second order ten-noded tetrahedral elements (C3D10 in Abaqus) ensuring the periodicity over the opposite faces of the resin-free RVE, as it is shown in Figure 4.6.

Figure 4.6: Artificially thinned resin-free RVE

Step 3. In the third step, all nodes and elements are imported from the mesh generator to Abaqus, and organised as objects. Master-slave contact pair surfaces are then generated from the faces of the bundles solid elements and kinematically coupled to ensure contact during subsequent expansion. Surface contact penalty algorithm and a relatively small friction coefficient of 0.15 was used to improve convergence during the analysis. To ensure that the final result does not contain any interpenetrating bundles, and to simplify volume meshing in subsequent modelling, small gaps are introduced between all bundles in contact by setting a contact clearance of 0.01mm governed by an exponential law. In a real fabric the strands are obviously brought to contact where they interlace, hence zero distance would be preferred. On the other hand, without some play there will be huge problems to mesh the entire RVE with volume elements. In practice, this trade-off is governed by what quality of finite elements is required and how fine mesh one can afford.

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Periodic boundary conditions were applied at the opposite faces of the RUC, as shown in Eq. 4.1, to ensure continuity between neighbouring RUC's which will deform as jig-saw puzzles, according to the following displacement relationships:

$$\vec{u}(x_1, x_2, 0) - \vec{U}_1 = \vec{u}(x_1, x_2, L) \quad (4.1)$$

$$\vec{u}(x_1, 0, x_3) - \vec{U}_2 = \vec{u}(x_1, L, x_3) \quad (4.2)$$

$$\vec{u}(0, x_2, x_3) - \vec{U}_3 = \vec{u}(L, x_2, x_3) \quad (4.3)$$

wherein x and y stand as the Cartesian coordinate axes parallel to the RUC edges and z as the out-of-the plane direction. The periodic boundary conditions can be expressed in terms of the displacement vectors $\vec{U}_1$, $\vec{U}_2$ and $\vec{U}_3$ of a set of master nodes which relate the displacements between opposite faces.

Step 4. A Explicit multi-step FE-analysis under PBC's is, therefore, defined:

- First, a pretension of the bundles to avoid wrinkling formation during the subsequent thermal expansion. A controlled small displacement of $0.01 \mu\text{m}$ and $0.03 \mu\text{m}$ over the x and y master nodes respectively, parallel and perpendicular to the axial yarns. The aim was to ensure homogeneous behaviour of the strands and to avoid any local stress concentrations caused by the non-uniformity of the mesh which could yield to in-homogeneous cross-sectional shape of the bundles.
- Second, an artificial expansion of the cross-section of the yarns is carried out (see properties listed in Table 4.2) to improve the bundle-to-bundle contact while filling the free spaces within the RUC. The longitudinal yarn thermal expansion coefficient was set to zero as the bundles only need to expand orthogonally to the fiber direction. The resultant geometry is shown in Figure 4.7a.
- Finally, a compression using rectangular rigid surfaces of $15 \times 15 \text{ mm}$ to the virtually expanded geometry to regain the 0.3 mm ply thickness, corresponding to a F_{vf} of 51.5% .

Step 5. The mesh on the deformed configuration is then exported back to Hypermesh, Altair Engineering (2016), in which geometrical cuts are applied to create a rectangular dry-fabric $10 \times 3 \text{ mm}$ RUC, as shown in Figure 4.7b.

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Table 4.2: Elastic constants used for the unidirectional yarns. The properties are based from the traditional Stainless Steel grade 301, [AZO Materials \(2001\)](#), and fitted for the analysis.

E (GPa)	G (GPa)	ν	α_1 ($\mu\text{m}/\text{m}/^\circ\text{C}$)	α_2, α_3 ($\mu\text{m}/\text{m}/^\circ\text{C}$)
193	72.5	0.33	0	17

Figure 4.7: Dry fabric expanded geometry: a) Displacement of the yarn during the thermal expansion and b) resultant geometry after the FE method.

Step 6. The matrix regions are then defined by a boolean operation between the RUC ($10 \times 3 \times 0.3 \text{ mm}^3$) and the volume occupied by the yarns. As result (see Figure 4.8), the discretisation mesh resulted in a 1.56 million C3D10 tetrahedron elements with a fibre volume fraction of $\approx 35\%$, which is far from the one archived

4.2. 3D representative unit cell (RUC) modelling

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experimentally (51.5%). The fibre volume fraction obtained is only representative of a single ply model while in reality, factors such as ply-nesting and fibre shifting increase the fibre volume fraction. All these ply factors are not taken into account in the model.

Figure 4.8: Periodic RVE obtained from the FE simulation using rectangular compression plates with a $F_{v_f} \approx 35\%$ and discretised with 1.5 million C3D10 elements.

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In order to improve the global fibre volume fraction obtained with the method ($\approx 35\%$), a modification was applied. The rigid surfaces used in the Step 4, were replaced with an adaptive membrane modelled with 4-noded with reduced integration M3D4R membrane elements. The aim is to homogenise the pressure distribution over the surface during the FE-compaction process and to increase accordingly the fibre volume fraction. The geometry in the deformed configuration with the membranes was imported back into HyperMesh, [Altair Engineering \(2016\)](#), to mesh the matrix region and to guarantee the periodicity of the RVE Mesh. As result, the desired fibre volume fraction (51.5%) was archived and typical limitations of the meshing process of boxed RVE were overcome, Figure 4.9.

Figure 4.9: Periodic RVE mesh with a 51.5% fibre volume fraction corresponding to a 6 layered laminate.

The model was properly discretised with 4.5 million C3D4 tetrahedral elements. The mesh was re-imported into Abaqus in which periodic boundary conditions were applied to perform uniaxial tensile tests in both longitudinal and transverse directions. This methodology considers the realistic geometry information, the influence of fibre undulation and fibre volume ratios in the axial and bias yarns as well as the

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total fibre volume fraction of the unit cell. As can be seen from Figure 3.2, nesting between the plies that may cause variations of the final fibre volume fraction throughout the panel were not considered in the proposed model.

4.2.2 Elastic Behaviour of constituent materials

4.2.2.1 RTM6 matrix

The RTM6 epoxy resin was assumed to behave as an isotropic elasto-plastic solid following the Lubliner damage/plasticity model as implemented in Abaqus/Standard, Dassault Systèmes (2013). The epoxy resin in this model behaves as a brittle material when it is loaded in tension undergoing a softening behaviour controlled by a damage variable. However, when the material is loaded under compressive stresses, plastic deformations may appear controlled by a Drucker-Prager plasticity flow rule. This material behaviour was successfully used to model quasi-brittle behaviour of polymers as reported by Herráez et al. (2015); Canal et al. (2012); Naya et al. (2017). For the sake of simplicity, the corresponding plasticity/damage parameters used in the simulations are presented in Table 4.3 including the elastic modulus and Poisson ratio, the tensile and compressive strength and the fracture toughness, according to values reported in the material data sheet, Hexcel (2016a), or literature, Gerlach et al. (2008); Herráez et al. (2015). More details regarding plasticity/damage constitutive equations can be found in Chapter 5.

Table 4.3: RTM6 damage plasticity model parameters used in the simulations.

E (GPa)	ν	σ_t (MPa)	σ_c (MPa)	G_t (J/m ²)
2.89	0.35	75	180	89

4.2.2.2 AS4C/RTM6 tows

The set of parameters describing the mechanical behaviour of the AS4C/RTM6 tows were estimated from micromechanical analysis or well-established values reported in the literature for similar materials. Carbon fibres were assumed to behave as transversely isotropic solids. Their corresponding elastic constants are reported in Table 4.4 and were obtained from either the material data sheet Hexcel (2016b) and literature values Herráez et al. (2015); Totry et al. (2010).

The fibre tows were also modelled as a transversely isotropic solids. The corresponding five elastic constants were determined using the simplified Chamis mi-

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Table 4.4: AS4C carbon fibres mechanical properties used in the simulations.

E_1 (GPa)	E_2 (GPa)	G_{12} (GPa)	ν_{12}	ν_{23}	σ_t (MPa)
231	12.97	11.28	0.3	0.46	4385

Table 4.5: Elastic constants of the AS4C/RTM6 unidirectional yarns obtained from micromechanical analysis

Material	E_1 (GPa)	E_2 (GPa)	G_{12} (GPa)	ν_{12}	ν_{13}
AS4C/RTM6	163.0	8.3	4.4	0.315	0.315

cromechanical model [Chamis \(1989\)](#); [Barbero \(2008\)](#) with the elastic properties of AS4C fibres and RTM6 matrix shown in Table 4.4 and Table 4.3 respectively. The local volume fraction of fibre reinforcement within the tow was $V_f = 70\%$ as measured from optical micrographs previously presented. The calculated values using this methodology lie within the expected range of elastic properties for such kind of unidirectional composite materials, Table 4.5.

4.2.3 Prediction of elastic behaviour using representative unit cell RUC

Uniaxial tensile tests were carried out in the longitudinal and transverse direction on the three-dimensional RUC using the set of properties described above. To this end, displacements along the longitudinal and transverse directions were imposed and the reaction forces at these pair of nodes calculated with the finite element model. The pair of reaction forces and conjugate displacements allows obtaining the elastic constants of the material as a function of the mechanical properties of the constituents and the textile architecture. The periodic boundary conditions applied to the model produce a uniform distribution of the displacement fields and no artificial stress concentration at the RUC edges were found (see Figure 4.10). The results obtained with the RUC model were compared with the experimental tests showing a good correlation, as it is shown in Table 4.6.

One of the main drawbacks of geometrical models based on RUC representation is related with the tremendous computational efforts needed to perform the analyses due to the high level of FEM discretisation within the RVE (≈ 4.5 million elements). A simple elastic computation needs several hours (≈ 20 hours with a 20 cores Intel Xeon with 32 Gb RAM memory machine). This level of accuracy can be obtained using simple analytical methods such as Classical Laminate Theory [Chamis \(1989\)](#);

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Figure 4.10: Displacement field on the three-dimensional RVE on both transverse (top) and longitudinal (bottom) directions

Barbero (2008). An implementation of such simple CLT model is presented in Appendix A.1 and the results gathered in Table 4.6 using the same set of properties of the constituents.

Table 4.6: Elastic properties prediction using 3D modelling approach and analytical methods based on CLT compared with results obtained from experimental tests

Property	Experimental	3D RVE	CLT	Error (%)
Longitudinal modulus, E_1 (GPa)	42.9 ± 1.8	44.2	48.2	3-12
Transverse modulus, E_2 (GPa)	38.6 ± 0.9	38.4	42.3	0.5-11
Longitudinal Poisson ratio, ν_{12} ($\hat{A}$)	0.30 ± 0.01	0.31	0.30	—

One of the major advantages of the use of geometrical RUC's is the possibility to include directly non-linear behaviour of the material constituents. Thus, damage in the form of matrix cracking as well as fibre fracture could be easy to implement in the model allowing the determination of the strength properties of the composite laminate. However, the method is extremely demanding in terms of computational resources, cores and FE licenses, so it was preferred to explore other alternatives that can provide similar accuracy level while maintaining robustness and efficiency. This will be illustrated in the following paragraphs.

4.3 Multi material shell model (MMS)

The objective of this section is to present a computationally efficient model based on a multi material shell approximation that overcame the difficulties encountered using a full 3D discretisation of the material. Three-dimensional representation of RUC's are very demanding in terms of computational resources and computing time are usually very large. In this model, the complex fiber architecture of the triaxial braided is simplified and represented using standard shell elements with material variation in the thickness included at Gauss point level. Rather than modelling the braided textile using the classical laminate subcells, the proposed model discretises the projected mesostructure including the different yarn orientations at the Gauss points of the finite elements and including also the pure resin regions.

A two-dimensional rectangular representative unit cell (RUC) of length $B=10\text{mm}$ and height $H=2.9\text{mm}$, as plotted in Figure 3.1, is modelled using a multi-material shell approximation (MMS). In this approach, the axial and bias tows are first projected over the rectangular surface of the RUC using the individual intersections between them to define the geometry of the multi-material shell. The geometry of the transverse sections of the tows was simplified in the modelisation. To this end, the axial and bias tow sections were assumed to be rectangular in shape being the width and thickness obtained from the optical micrographs presented in Figure 3.2. Each of the intersected regions can be formed by assuming a local lay-up combining different layers of the axial tows, bias tows and resin pockets depending on their position within the RUC. The geometry formed in this manner is then discretised using 4-noded shell elements with reduced integration (S4R in Abaqus, Dassault Systèmes (2013)) endorsing the material behaviour and lay-up configuration to the different through-the-thickness Gauss points of the individual elements. For instance, the element located at point A in Figure 3.2 contains the Gauss points endorsed to one of the bias tows and the interyarn resin layer, while the element located at point B the Gauss points are defined for the crossing axial and bias yarns.

The input file of the model for the finite element solver was generated automatically by means of Python script code. A finite element discretisation with element size of $\approx 0.15\text{mm}$ (≈ 1800 S4R elements) was enough to ensure a good resolution of the mechanical fields within the RUC. Figure 4.11 shows schematically the different lay-ups obtained in the MMS model of the triaxial braided composite. For instance, lay-up A corresponds with a region of the shell model where only a single

4.3. Multi material shell model (MMS)

bias tow (0.087mm thickness) is present being the remaining gap up to 0.3mm of the cured ply thickness filled completely with resin. The lay-up B is defined using the three reinforcement tows, axial and bias, being 0.126mm and 0.087 mm each corresponding thickness respectively. It should be mentioned that this material definition in the thickness led to unsymmetrical lay-up configurations that may alter the expected behaviour for a single ply subjected to in-plane loads and, in particular, non-linear geometric and out-of-plane displacements as noticed in [Cater et al. \(2015\)](#). In this work, the braided was simulated therefore by means of symmetric configuration including the six-layer material. However, ply nesting present in the composite laminate is not taken into account.

Figure 4.11: Schematic plot of the multi-material shell model (MMS) representing the unit cell of the triaxial braided composite material.

This methodology allowed a fast simulation of the in-plane behaviour of the

triaxial braided composite although with some important simplifications. Firstly, the three-dimensional structure of the braided textile is neglected and the yarn undulations were not taken into account. This effect could probably introduce some effects in the overall material behaviour specially in the bias tows in which the amplitude of the undulations is higher. Secondly, the kinematic assumption imposed by the multi-material shell configuration does not allow for tow debonding and all the damage is lumped to behaviour of yarn. Finally, other important effects such as ply nesting are not taken into account. Even though, the efficiency of the proposed MMS methodology makes this an interesting tool for fast simulation not only for braided textiles but also for any other textile architectures as will be demonstrated. Periodic boundary conditions, as shown in Equation 4.1, were applied to the edges of the RUC to ensure continuity between neighbouring RUC's which will deform as jig-saw puzzles.

The multi material shell model (MMS) was used to predict the elastic properties of triaxially braided composites. The elastic predictions were similar to the ones obtained by classical laminate theory CLT, as it is shown in Table 4.7, with the advantage that the computational cost was fairly reduced ($\approx 6-7$ min) when compared with the 3D RVE model. The efficiency in terms of degrees-of-freedom reduction allows the application of the model to higher length scales. However, the ability of the MMS model to address damage and fracture problems needs to be demonstrated. This will be done in the subsequent chapters of the thesis.

Table 4.7: Elastic properties and modelling predictions.

Property	Experimental	MMS	3D RVE	CLT	Error (%)
Longitudinal modulus, E_1 (GPa)	42.9 $\pm$ 1.8	46.8	44.2	48.2	3-12
Transverse modulus, E_2 (GPa)	38.6 $\pm$ 0.9	42.7	38.4	42.3	0.5-11
Longitudinal Poisson ratio, ν_{12} (·)	0.30 $\pm$ 0.01	0.31	0.31	0.31	–

MMS modelling for the simulation of elasticity and strength of braided composites

This chapter is aimed at applying the multi material shell (MMS) approach to characterise the orthotropic elastic and strength behaviour of triaxially braided composites. The results experimental campaign carried out at coupon level (Chapter 3) were used to validate the proposed approach.

To this end, tensile simulations of 6-layer coupons loaded in both longitudinal and transverse directions were performed to predict the relevant mechanical properties (elastic constants and strengths) and to evaluate the model ability to capture some of the characteristic damage mechanisms associated with this fibre architecture, such as damage localisation at the bundle overlapping, damage concentration due to edge effects and others.

In order to do so, a continuum damage model based on Hashin failure criteria to represent the damage progression using dissipation softening laws according to material direction was used to model the braided tows.

5.1 Constitutive behaviour of AS4C/RTM6 tows

AS4C/RTM6 tows were assumed to behave as linear and elastic solids undergoing damage and neglecting, therefore, the elastoplastic behaviour of the epoxy resin at the microlevel. A continuum damage mechanics model (CDM) is used to describe

5.1. Constitutive behaviour of AS4C/RTM6 tows

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the behaviour after failure initiation. The model used is based on the CDM implementation in Abaqus/Standard, [Dassault Systèmes \(2013\)](#), of the Hashin failure criteria allowing damage progression using energy dissipation softening laws according to material loading directions. The stress-strain relationship is given by the damaged stiffness matrix of the material material according to

$$\begin{pmatrix} \sigma_1 \\ \sigma_2 \\ \tau_{12} \end{pmatrix} = \frac{1}{D} \begin{bmatrix} (1-d_f)E_1 & (1-d_f)(1-d_m)\nu_{21}E_1 & 0 \\ (1-d_f)(1-d_m)\nu_{21}E_1 & (1-d_m)E_2 & 0 \\ 0 & 0 & (1-d_s)G_{12} \end{bmatrix} \begin{pmatrix} \epsilon_1 \\ \epsilon_2 \\ \gamma_{12} \end{pmatrix} \quad (5.1)$$

wherein E_1 , E_2 , ν_{12} and G_{12} stand, respectively, for the elastic constants of the unidirectional tows, being $D = 1 - (1-d_f)(1-d_m)\nu_{12}\nu_{21}$. The corresponding damage variables for the longitudinal (d_f), transverse (d_m) and shear (d_s) deformation modes follow linear evolution normalised damage laws ($0 \leq d_i \leq 1$). The damage variables control the evolution of damage once the onset condition or failure criterion is fulfilled and may depend on material internal variables as the stress, strain or the elastic energy, etc. The activation of the damage variables is triggered by the effective stress state corresponding to the reference configuration computed in the absence of damage. According to [Hashin \(1980\)](#), four main failure modes can be established for unidirectional composites, namely fibre tension (F_{ft} , $\hat{\sigma}_1 \geq 0$), fibre compression (F_{fc} , $\hat{\sigma}_1 < 0$), matrix tension (F_{mt} , $\hat{\sigma}_2 \geq 0$) and matrix compression (F_{mc} , $\hat{\sigma}_2 < 0$), which can be described by the following functions.

$$F_{ft} = \left(\frac{\hat{\sigma}_1}{X_T} \right)^2 + \alpha_{sh} \left(\frac{\hat{\tau}_{12}}{S_L} \right)^2 \quad (5.2)$$

$$F_{fc} = \left(\frac{\hat{\sigma}_1}{X_C} \right)^2 \quad (5.3)$$

$$F_{mt} = \left(\frac{\hat{\sigma}_2}{Y_T} \right)^2 + \left(\frac{\hat{\tau}_{12}}{S_L} \right)^2 \quad (5.4)$$

$$F_{mc} = \left(\frac{\hat{\sigma}_2}{2S_L} \right)^2 + \left[\left(\frac{Y_C}{2S_T} \right)^2 - 1 \right] \left(\frac{\hat{\sigma}_2}{Y_C} \right)^2 + \left(\frac{\hat{\tau}_{12}}{S_L} \right)^2 \quad (5.5)$$

wherein $\hat{\sigma}_1$, $\hat{\sigma}_2$ and $\hat{\tau}_{12}$ stand for the plane-stress effective tensor components computed in the absence of damage ($d_f = d_m = d_s = 0$). The failure functions depend

5. MMS modelling for the simulation of elasticity and strength of braided composites

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on the strength of the lamina in different loading directions: X_T - longitudinal tensile strength; X_C - longitudinal compression strength; Y_T - transverse tension strength; - Y_C transverse compression strength; S_L - in-plane shear strength; and S_T - out-of-plane shear strength. The parameter α_{sh} controls the effect of the in-plane shear stress on the final strength of the tow in the fibre direction. A typical value of $\alpha_{sh} = 0.6$ was used, according [Zhang et al. \(2014b\)](#), in the absence of more reliable data.

Once damage onset is predicted for one of the failure modes, the corresponding damage variable starts to evolve if deformation is further applied, following a softening behaviour controlled by the fracture energy of the material. The area under the stress-strain curve corresponds to the volumetric fracture energy, or energy dissipated during failure per unit of volume of material. Therefore, four different volumetric fracture energies can be defined corresponding to the energies dissipated in longitudinal loading damage in tension and compression (g_{ft} and g_{fc}), and transverse loading damage in tension and compression (g_{mt} and g_{mc}), according to the failure modes defined by the Hashin failure criteria, [Hashin \(1980\)](#). The softening parameters should ensure mesh objectivity of the model in order to dissipate the same amount of energy irrespectively of the size of the FE discretisation. To this end, Abaqus 6.13, [Dassault Systèmes \(2013\)](#), introduces a characteristic length L_c proportional to the element size to alleviate mesh dependency and provide the correct global dissipated energy parameters G_{ft} , G_{fc} , G_{mt} and G_{mc} .

The set of parameters describing the mechanical behaviour of the AS4C/RTM6 tows were either estimated by micromechanical analysis or obtained from well-established literature reporting similar materials. The five elastic constants of the orthotropic yarn were determined using the simplified Chamis micromechanical model [Chamis \(1989\)](#); [Barbero \(2010\)](#) with the elastic properties of AS4C fibres and RTM6 matrix shown in [Table 4.4](#) and [Table 4.3](#) respectively. The local volume fraction of fibre reinforcement within the tow was $V_f = 70\%$, as measured from optical micrographs previously presented. The calculated values using this methodology lie within the expected range of elastic properties for such kind of unidirectional composite materials, shown in [Table 5.1](#).

[Table 5.1](#) also reports the properties used in this work to model the strength of AS4C/RTM6 yarns. The fibre dependent strengths, X_T and X_C , were determined by linear extrapolation, to the actual fibre volume fraction of the tow, of the experimental values reported in the AS4C data sheet for unidirectional composites

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manufactured with 8552 epoxy resin, [Hexcel \(2016b\)](#), and $V_f = 60\%$. The matrix dependent strength, Y_T , Y_C and S_L were determined using micromechanics as described by [Barbero \(2008\)](#) and material manufacturer datasheets, [Hexcel \(2016b,a\)](#).

The same extrapolation logic does not apply to the critical energy release rates, G_{ft} , G_{fc} , G_{mt} and G_{mc} , because the influence of through-thickness effects, especially of crimping, on the damage micro/meso mechanisms cannot be predicted straightforwardly. A numerical meso-modelling approach, e.g. as developed by [Doitrand et al. \(2015\)](#), involving the three-dimensional discretisation of the fibre architecture to account properly for the different damage mechanisms associated with the textile composite as well as to represent other type of failure mechanisms such as subsurface delamination, would be the appropriate way to predict failure initiation in detail and estimate the energy dissipated by all mechanisms. Such homogenised information could then be used at to calibrate the fracture energy values used at the MMS level.

Table 5.1: Mechanical properties of the unidirectional AS4C/RTM6 yarns used in the analyses. Elastic constants were obtained from micromechanical analysis. Strengths of unidirectional fibre tows: X_T and X_C extrapolated from AS4C carbon fibre data sheet [Hexcel \(2016b\)](#) for 70% fibre volume fraction, Y_T , Y_C and S_L obtained from Chamis micromechanical models [Barbero \(2008\)](#). Intra-yarn fracture toughness values G_{mt} and G_{mc} were assumed to be similar to the ones obtained in-house for unidirectional AS4/8552 pre-impregnated plies, and G_{fc} was assumed to be equal to the value obtained by [Catalanotti and Xavier](#) for IM7/8552 [Catalanotti and Xavier \(2015\)](#). G_{ft} varies within a range of values obtained for UD plies.

E_1 [GPa]	E_2 [GPa]	G_{12} [GPa]	ν_{12}	ν_{13}
163.0	8.3	4.4	0.315	0.315
X_T [MPa]	X_C [MPa]	Y_T [MPa]	Y_C [MPa]	S_L [MPa]
2463	2079	61	290	100
G_{ft} [kJ/mm ²]	G_{fc} [kJ/mm ²]	G_{mt} [kJ/mm ²]	G_{mc} [kJ/mm ²]	
80-120	61	0.31	1.55	

Simplified micromechanical models such as the one mentioned above are practical but generally not accurate enough in the prediction of strength values of unidirectional plies, tows or yarns, [Barbero \(2008\)](#). An alternative reliable technique to overcome their limitations is based on computational micromechanics using RVE's containing dispersion of parallel fibres subjected to different loading conditions. This modelling technique is able to realistically represent the physical deformation and damage mechanisms occurring at the microlevel as fibre/matrix interface debonding,

matrix shear banding or matrix splitting, e.g. [Herráez et al. \(2015, 2016b\)](#); [Naya et al. \(2017\)](#).

Finally, in the absence of experimental data for the intraply critical energy release rates of the AS4C/RTM6 unidirectional, the values G_{mt} and G_{mc} were assumed to be similar to the ones obtained in-house for unidirectional AS4/8552 pre-impregnated plies, and G_{fc} was assumed to be equal to the value obtained by [Catalanotti et al. \(2017\)](#) for IM7/8552 (see Table 5.1). As demonstrated in the following section, the CT simulation results proved to be very sensitive to the values of G_{ft} . Reported values for this property for different unidirectional fibre composites range widely from as low as 65 kJ/mm^2 to as high as 237 kJ/mm^2 [Laffan et al. \(2012\)](#). The simulation results in the following section are presented in the form of a parametric study on G_{ft} by varying this property between the value measured in-house for AS4/8552 (80 kJ/mm^2) and 120 kJ/mm^2 , hence eliminating the extremes of the distribution reported in [Laffan et al. \(2012\)](#).

5.2 Constitutive behaviour of RTM6 resins

The polymer matrix material is simulated as an isotropic linear and elastic solid with its elastic properties listed in Table 4.3. In addition, the matrix is able to undergo plastic deformations with the possibility of damage by cracking under tensile loads. This approach has been adopted by other researchers in the literature [Melro et al. \(2013\)](#); [Yang et al. \(2012\)](#); [Ashouri Vajari \(2014\)](#); [Canal et al. \(2009\)](#) as it represents a realistic behaviour of a polymer [Fiedler et al. \(2001\)](#). The damage-plasticity model, available in ABAQUS/Standard [Dassault Systèmes \(2013\)](#) and schematically illustrated in Figure 5.1, is a modification of the Drucker-Prager plasticity yield surface [Drucker and Prager \(1952\)](#) by including a damage variable in order to capture the quasi-brittle behaviour of the polymer under dominant tensile loads. The constitutive equation is based on the yield function proposed by [Lubliner et al. \(1989\)](#), including modifications proposed by [Lee and Fenves \(1998\)](#) to account for strength evolution under tension and compressive loads. The yield function defined in terms of the I_1 and J_2 invariants of the stress tensor is:

$$\Phi(I_1, J_2, \sigma_I, \beta, \alpha) = \frac{1}{1 - \alpha} \left(\sqrt{3J_2} + \alpha I_1 + B \langle \sigma_I \rangle \right) - \sigma_{myc} = 0 \quad (5.6)$$

wherein I_1 stands for the first invariant of the stress tensor, J_2 is the second invariant of the deviatoric stress tensor, α is the pressure-sensitivity parameter of the Drucker-

5.2. Constitutive behaviour of RTM6 resins

Figure 5.1: Schematic of the uniaxial tension and compression response of the epoxy matrix according to the damage-plasticity model for quasi-brittle materials, [Naya et al. \(2017\)](#).

Prager yield criterion, σ_I is the maximum principal stress, $\langle \rangle$ the Macaulay brackets (returning the argument if positive and zero otherwise) and B is a function of the tensile and compressive yield stresses, σ_{myt} and σ_{myc} , defined as:

$$B = \frac{\sigma_{myt}}{\sigma_{myc}}(1 - \alpha) - (1 + \alpha) \quad (5.7)$$

Under biaxial compression stress state, with $\sigma_I = 0$, equation 5.6 reduces to the initially proposed Drucker-Prager yield condition [Drucker and Prager \(1952\)](#), wherein α can be expressed in terms of the internal friction angle of the material (β) according to $\tan \beta = 3\alpha$. The internal friction angle controls the hydrostatic pressure dependence of the plastic behaviour of the material. Simultaneously, α can be related to the biaxial compression behaviour according to:

$$\alpha = \frac{\sigma_{b0} - \sigma_{c0}}{2\sigma_{b0} - \sigma_{c0}} \quad (5.8)$$

After the onset of damage in tension at σ_{tm} , the softening behaviour is controlled by an exponential cohesive law, characterised by a single normalised scalar damage variable, to ensure the correct energy dissipation of the matrix G_m . More details about the constitutive model and the numerical implementation can be found in [Canal et al. \(2012\)](#); [Dassault Systèmes \(2013\)](#).

Even though this damaged-plasticity model requires a complex calibration from detailed experiments, good results can be obtained by the assumption of default parameters while measuring the key properties. More details about the determination of the experimental properties using a micromechanics approach can be found in [Rodríguez et al. \(2012\)](#); [Naya et al. \(2017\)](#).

5.3 Simulation of plain tensile loading

In addition to the simulations performed in both longitudinal and transverse direction on a single RVE, as shown in Figure 4.11, periodic RVEs containing several RUCs were set up to analyse the effect of the width and free edges of the specimen on the stress-strain curve using the multi-material shell approximation. In this case, the RVE is formed by a number of RUCs necessary to approximate the finite width of the specimens tested (25mm) and depends on the direction selected. Since the triaxial braided RUC dimensions are $10 \times 2.9 \text{ mm}^2$ (see Figure 3.1 in the axial direction), the RVE used to analyse the behaviour in the transverse direction is composed by a total of 8.5 RUCs ($\approx 25 \text{ mm}$) while a total of 2.5 RUCs (25mm) is used in the longitudinal direction. For instance, when loading in the transverse direction y , the periodic boundary conditions are only applied to the nodes belonging to $y = 0$ and $y = H$ leaving the lateral constraints at $x = 0$ and $x = B$ under zero traction conditions to simulate the free edge condition. The master nodes are used to impose the tensile load in the direction y of the RVE similar to the standard single RVE case aforementioned.

5.3.1 Predictions with single RUC representative volume elements

The numerical stress-strain curves obtained with the single RUC representative volume element subjected to uniaxial tension in the longitudinal and transverse directions are presented in Figure 5.3. The results are compared with the experimental data previously reported in Chapter 3. Reasonable accuracy level is achieved despite the strong kinematic assumptions involved in the multi-material shell model. The stress strain curves are essentially linear and elastic up to the brittle failure of the composite although some slight non-linearity was observed in both types of loading due to the damage progression during the applied deformation.

The summary of the elastic moduli, strength and strain-to-failure predictions are also presented in Table 5.2 for comparison with the corresponding experimental

5.3. Simulation of plain tensile loading

Figure 5.2: Multiple RUC representative volume elements for finite width specimens loaded in: a) Transverse direction, b) Longitudinal direction.

results. In both cases, the model over predicts the elastic modulus of the material in both directions and this effect can be directly endorsed to the 2D nature of the model presented in this work. The yarns were assumed to be totally straight neglecting, thus, any effect of the real textile architecture as crimping and undulations. This results, therefore, in a slightly stiffer numerical response ($\approx 8.3\%$ and 9.6% , respectively for the longitudinal and transverse directions). However, the model was able to fairly reproduce the tensile strength, which is essentially controlled by the strength of the yarns, being in this case the relative errors of $\approx 1\%$ and $\approx 11\%$, respectively (Table 5.2).

Similar accuracy level in terms of strength and strain-to-failure were found by other authors, see [Cater et al. \(2015\)](#), when using subcell models for multiply braided laminates. However, these authors endorsed the non-linear behaviour observed in the transverse direction to non-linear geometrical effects caused by the local tension-twist as in a single ply non-symmetric braided ply rather to free edge delamination and damage [Littell \(2008\)](#); [Kohlman \(2012\)](#).

The damage variables for fibre d_f and matrix cracking d_m at ultimate load corresponding to the single RUC models loaded in the longitudinal and transverse directions are presented in Figure 5.4. In the case of longitudinal loading, the ma-

5. MMS modelling for the simulation of elasticity and strength of braided composites

Figure 5.3: Numerical and experimental stress-strain curves in tension and the fraction of matrix cracking damaged area in the RVE for the triaxial braided composites: a) Longitudinal loading, b) Transversal loading.

Table 5.2: Mechanical properties and modelling predictions for longitudinal and transverse directions.

Property	Experimental	Single RUC	Multiple RUC
Longitudinal modulus, E_1 (GPa)	42.9 ± 1.8	46.8	46.3
Transverse modulus, E_2 (GPa)	38.6 ± 0.9	42.7	40.5
Longitudinal Poisson ratio, ν_{12} ($\hat{\cdot}$)	0.30 ± 0.01	0.31	0.31
Longitudinal tensile strength, X_L (MPa)	673 ± 55	679	664
Transverse tensile strength, X_T (MPa)	630 ± 23	702	590
Longitudinal strain-to-failure, ϵ_L (%)	1.6 ± 0.1	1.49	1.49
Transverse strain-to-failure, ϵ_T (%)	1.8 ± 0.1	1.67	1.43

trix cracking variable, d_m , shows a uniform diffuse damage in both axial and bias tows although a strong concentration in the bias yarns in the free space between the axial tows ($\pm 60^\circ$ with respect to the axial loading) can be observed. These damage mechanisms in the form of matrix cracks running parallel to the bias tows were also observed experimentally in the radiographies shown in Figure 3.7a) to c). However, contrary to the experimental results previously described, the matrix cracking variable d_m is more concentrated in the space between the axial yarns which can be endorsed to the absence of ply nesting in the models. Finally, fibre failure was triggered at the highly damaged matrix areas generated in the space between adjacent

5.3. Simulation of plain tensile loading

axial tows producing the final catastrophic failure of the composite material. These stress concentrations generated in the damaged areas of the bias yarns promoted the fibre failure once the corresponding Hashin criterion was fulfilled according to equation 5.2.

Figure 5.4: Damage variables in the single RUC representative volume element, axial and $\pm 60^\circ$ bias yarns, at ultimate load for longitudinal and transverse loading: a) Fibre damage variable d_f , b) Matrix cracking variable d_m .

The damage patterns corresponding to the transversal loading of the triaxial braided composite are also represented in Figure 5.4 for comparison with the previous loading condition. In this case, matrix cracks are more concentrated (less diffused), than the previous longitudinal case in regions where the bias and axial yarns overlap. This matrix cracking mechanisms was also observed in the radio-graphs presented in Figure 3.8 a, b and c). Matrix cracks, white lines in Figure 3.8, were concentrated in dark grey bands separated by the distance corresponding

to the axial tow-to-tow pitch ($\approx 5\text{mm}$). The final catastrophic failure of the triaxial braided composite under transverse loading is produced when the fibre damage variable of the bias yarns fulfil the corresponding Hashin criterion (see Equation 5.2) and this event occurred by the combination of normal and shear stresses at the corner points of the bias and axial yarns crossing.

5.3.2 Predictions with multiple RUC representative volume elements

It should be noted that this strategy based on the simulation of a single periodic RUC represents an infinite material formed by the periodic translation of the RUC in the longitudinal and transverse directions preventing, thus, any damage emanating from the free stress edges. Multiple RUC representative volume elements, Figure 5.2, were submitted to longitudinal and transverse loadings to study the effect of damage emanating from the free edges of the coupons. The width of the representative volume elements corresponded approximately with the width of the specimens tested ($\approx 25\text{mm}$) and was obtained by stitching in the width direction 2.5 single RUC in the case of longitudinal loading and approximately 8.5 single RUCs for the transverse loading case.

The numerical stress-strain curves obtained for the multiple RUC representative volume element are also included in Figure 5.3 for comparison with the standard single RUC models. Not surprisingly, the elastic moduli obtained were almost equal in both models, as it is shown in Table 5.2. No significant differences between single and multiple RUC models were observed and the curves were practically superimposed when loading in the longitudinal direction as the strength of the composite is dictated by the strain-to-failure of the axial tows, Figure 5.3 a). However, some differences were found in the case of transverse loading perpendicular to the axial tows, Figure 5.3b). In this case, the stress-strain curves were coincident in the initial elastic regime and started to diverge for strains $\approx 0.5\%$ due to the appearance of damage by matrix cracking emanating from the coupon edges. The predicted strength of the material decreased significantly from 702MPa, for single RUC models, to 590MPa for multiple RUC ($\approx 19\%$ reduction) and this effect can be probably attributed to the presence of edge effect damage generated during the deformation.

The fibre and matrix damage variables of the multiple RUC representative volume element loaded in the longitudinal direction is shown in Figure 5.5. No significant differences are observed between the damage patterns obtained in this case and the single RUC representative volume element (Figure 5.4). As a result, the

5.3. Simulation of plain tensile loading

Figure 5.5: Damage variables in the multiple RUC representative volume element, axial and $\pm 60^\circ$ bias yarns, at ultimate load for longitudinal and transverse loading: a) Fibre damage variable d_f , b) Matrix cracking variable d_m .

obtained stress-strain curves for single and multiple RUCs are practically superimposed, Figure 5.3 a). However, the damage patterns obtained for the multiple RUC models loaded transversely (Figure 5.6) differed with respect to the single RUC ones (Figure 5.4). The damage patterns far from the edges of the coupon were similar to the single RUC models. However, the effect of edge damage was evident in the simulations of the multiple RUC models. The extension of the damaged zone obtained in the simulations was approximately the width of one individual tow (2-3 mm) similar to the length of the cracks visualised in the radiographs in Figure 3.8 and Figure 3.9c).

An estimation of the average damage progression in the triaxial braided composite was determined from the numerical simulations using the fraction of total area of the RVE affected by matrix cracking. To this end, those elements fulfilling the Hashin $F_{mt} = 1$ condition (equation 5.4) were monitored and the ratio between

Figure 5.6: Damage variable in the multiple RUC representative volume element at ultimate load in transversal direction: a) Fibre damage d_f , b) Matrix cracking d_m .

their total area and the corresponding area of the RVE presented in Figures 5.3 a) and 5.3 b), for longitudinal and transverse loading, respectively. The damage initiation strain obtained from the simulations, 0.69% and 0.42% for longitudinal and transversal direction respectively, compared quite well with those values measured from the radiographs carried out on loaded specimens in interrupted tests (Figures 3.9 a).

5.4 Conclusions to elasticity and strength simulations

A multi-material shell (MMS) model was developed to simulate the in-plane behaviour of the composite material. This model is based on a bi-dimensional approximation in which the axial and bias tows are projected over the rectangular planar surface of the RUC using the individual intersections between them. As a result, the regions corresponding to the different possible combinations, axial and bias yarns as well as matrix pockets, were defined. The RUC defined in such a way is discretised with 4-node shell elements with reduced integration containing the corresponding layer information at the Gauss point level. Yarns were assumed to be perfectly straight neglecting, therefore, the effect of crimping and undulations. A continuum damage model based on the Hashin failure criteria is used to address

5.4. Conclusions to elasticity and strength simulations

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the damage at the tow level. It should be remarked that one of the main drawbacks of the MMS model is the inability to simulate failure of the composite material by yarn-to-yarn delamination and only intra yarn damage by either fibre failure and/or matrix cracking is taking into account. The mechanical properties of the tows, stiffness, strength and toughness, were taken from values obtained from the literature for similar carbon composite materials.

The response of the single RUC representative volume element was very good for the longitudinal loading and differed slightly for the transverse loading case. To overcome this limitation, a multiple RUC model was generated to study the effect of the finite coupon width on the mechanical response of the material, specially during the transverse deformation. The response of the multiple RUC model to the transverse loading case was better and the strength reduction due to the finite coupon width was captured accurately. However, the experimental transverse tensile non-linearity of the material was not fully capture and could be generated by the appearance of internal asymmetric behaviour triggered by damage accumulation produced by matrix cracking. The model results in terms of the damage variable distribution were able to explain the origin of the cracking patterns appearing during the applied deformation that were detected in the radiographs. Specially, those cracks emanating from the edges of the coupon were well captured by means of the multiple RUC representative volume element using the free stress boundary conditions instead of the periodic ones. Despite the simplicity of the mesostructural representation of the current triaxial braided composite made with the multi-material shell, the results were in good agreement with the experimental behaviour obtained in the mechanical tests, and open possibilities to map the textile architecture into the structural response of the braided composite part.

MMS modelling for simulation of fracture in braided composites

This chapter is focused on the demonstration and validation of the MMS modelling and simulation methodology for simulation of fracture in triaxially-braided composites. To this end, virtual simulations of the Compact Tension (CT) experimentally characterised in Chapter 3 were carried out to assess if the proposed methodology was able to capture the mechanical stress-strain trend and the observed damage mechanisms. For this purpose, actual testing conditions such as cross-head speed and fixations were reproduced for simulations of the 12- and 18-layered CT specimens.

6.1 Simulations of Compact Tension (CT) specimens

The geometries of the CT specimens described in Chapter 3 were fully discretised using mosaic patterns of representative unit-cells (RUC) oriented in longitudinal and transverse directions, as schematically shown in Figure 3.16. The geometric patterns formed in this way are then further discretised using 4-noded shell elements with reduced integration (S4R) endorsing the material behaviour and lay-up configuration to the different through-the-thickness Gauss points of the individual elements. Figure 4.11 shows schematically the different lay-ups obtained in the MMS model of the unit cell of one layer of a triaxial braided composite for a global fibre volume fraction $\approx 51.5\%$, [García-Carpintero et al. \(2017b\)](#). This material definition leads to locally unsymmetrical lay-up configurations that may alter the expected behaviour for a single ply subjected to in-plane loads and generate non-linear geometric and

6.1. Simulations of Compact Tension (CT) specimens

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out-of-plane displacements as noticed in [Cater et al. \(2015\)](#). To overcome this issue, the 12- and 18-layered triaxial braided layups were obtained by replication and symmetrisation of the arrangement shown in Figure 4.11.

In order to reproduce the boundary conditions applied in the experimental tests and avoid unrealistically large deformations around the pin-loaded holes of the virtual specimens, rigid clamps were fit to the holes of the specimens avoiding any clearance. A penalty contact interaction between the clamps and the virtual specimens was used with a friction coefficient of $\mu = 0.15$. One of the clamps was prevented of all translations and rotations. These conditions allowed the specimen to rotate around the clamps during the virtual tests. Load was applied by means of a smooth displacement curve, to avoid peak accelerations, prescribed to a master node linked to the movable clamp.

The fracture simulations were performed by means of ABAQUS/Explicit [Dassault Systèmes \(2013\)](#). The FE input files were generated automatically by means of a Python script code. In spite of the analyses being dynamic, quasi-static loading conditions were guaranteed by ensuring that the kinetic energy applied to the model was several orders of magnitude lower than the model internal energy and that the obtained load-displacement behaviour was quasi-linear before damage initiation.

As with the simulations of tensile coupons in Chapter 5, the maximum element size was set based on the limits imposed by the crack-band model of [Bazant and Hwan Oh \(1983\)](#). Accordingly, the width of a crack band, here discretised by a single band of reduced-integration elements, needs to release the fracture energy at the rate characteristic of the damage mode being produced. Because the damage mechanism that is bound to occur at each element and integration point is not known a priori, the element is designed cover all possible scenarios. In the present case, the minimum element size required is $\approx 0.3\text{mm}$. Linear elements with a single integration point (S4R element in ABAQUS) were preferred to higher order elements with higher number of integration points because experience demonstrates that for similar computational cost in the modelling of crack propagation by means of continuum damage models, a higher number of degrees of freedom leads to better results than improved integration or field interpolation schemes. After the establishment of the minimum element size in the previous simulations, several mesh refinements were tested. A characteristic element size of $\approx 0.15\text{mm}$ was finally adopted to facilitate the generation of conforming meshes around the small pure matrix regions and to ensure a progressive and smooth softening at all model locations. This mesh re-

finement is also enough to ensure a good resolution of the mechanical fields in the presence of a crack within bias and axial yarns, which are with over 12 and 20 elements across the width, respectively. The complete CT specimen discretisation resulted in a total of ≈ 120000 S4R elements.

6.2 Prediction of translaminar fracture and correlation of results

6.2.1 Fracture behaviour

The simulated force vs. clamp displacement of the CT test specimens differ from the experimental results because of the flexibility introduced by the test fixtures, machine grips and load cell, and local damage around the load bearing holes, among other possible secondary effects that have not been taken into account in the . Hence, instead of using the displacement of the clamps to represent the mechanical response of the specimens, fair comparisons are obtained by using experimental vs. simulated Crack Mouth Opening Displacement (CMOD), as represented in Figure 6.1.

The initial elastic behaviour of the tested CT specimens are fairly captured by the models revealing that the geometries, boundary conditions and elastic properties of the virtual specimens are well represented. Furthermore, this correlation shows that the MMS modelling approach is appropriate to capture the elastic behaviour of triaxial braided composites, as demonstrated in [García-Carpintero et al. \(2017b\)](#). Variations in the peak loads, corresponding to the propagation of the slit notch, are obtained from the simulations as a result of varying G_{ft} . Variations within similar ranges are obtained from experiments because of the termination of the slit-notch at different relative locations within the RUC of the braided material architecture, what in effect causes a variation in the fracture toughness for crack initiation. The effect of shifting the material architecture from specimen to specimen was not simulated in this work.

Even more pronounced than in the experiments, the simulated crack propagation is unstable with large load drops and corresponding increases in crack growths. This is because features such as inter-tow deformation and debonding are not taken into account in the representation of the through-the-thickness behaviour of the material. In this way, all plies within the composite lay-up are in perfect solidarity until accumulated through-the-thickness damage causes instability of the coupon and the virtual all-through-thickness crack propagates. Due to the absence of these through-

6.2. Prediction of translaminar fracture and correlation of results

Figure 6.1: Correlation between experimentally- and numerically-obtained Load vs. Crack Mouth Opening Displacement (CMOD) curves for CT specimens.

thickness deformation and dissipative progressive failure mechanisms, the numerical analyses are even more unstable than the experimental results. This is reflected in a sudden crack propagation over several elements when damage has sufficiently accumulated over a single through-thickness element. When the crack is arrested and static equilibrium is re-established, then another crack-tip element needs to be sufficiently loaded and damaged for the crack to propagate further.

6. MMS modelling for simulation of fracture in braided composites

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For each of the configurations, the simulations with $G_{ft} = 80 \text{ kJ/mm}^2$ and $G_{ft} = 120 \text{ kJ/mm}^2$ constitute, respectively, lower and upper bounds of the scattered experimental crack propagation results. This indicates that the longitudinal fracture toughness of the braided yarns should in fact be within this range.

Deformation and damage results of simulations of 12-layered CT tests in the longitudinal and transverse loading directions are shown in Figure 6.2, wherein they are correlated with experimental results for similar CMOD's. For 18-layered specimens, the correlations are similar, hence not shown here. The instants represented in the figure correspond to experimental translaminal crack propagations of 16.95 mm and 17.3 mm for longitudinal and transverse loading directions respectively. These values compare with simulated crack growths of 16.25 mm and 14.15 mm, respectively, for similar values of CMOD in each case. The simulation of a translaminal crack is represented by a single band of fully damaged finite elements in fibre tension mode. Wider regions of matrix damaged elements around the fracture process zone reveal secondary damage modes such as intra-yarn transverse cracking.

By comparison of Figures 3.20 and 6.2, it can be judged that the experimentally-observed saw-blade patterns of crack propagation for each of the loading directions are remarkably well captured by the simulations. For transverse loading, this fracture mode involves the longitudinal fracture of both bias yarns but not of the longitudinal yarns which only show transverse damage. For longitudinal loading, each side of the dent in the simulated saw-blade damage pattern corresponds to a crack through the fibres of one of the bias yarns and a longitudinal yarn, in a direction parallel to the other bias yarn, hence involving the rupture of all yarns in the triaxial braided architecture, as obtained experimentally. The region of matrix damage accumulation, corresponding to non-coalesced matrix and intra-yarn microcracking, is predicted to be wider in the longitudinal loaded specimens. This result also correlates well with experimental findings.

6.2.2 Fracture toughness

The Area Method described above was applied to the simulation results for the calculation on average values of the critical energy release rates corresponding to the four configurations studied in this work. The summary of these results and their correlation with corresponding experimental data is given in Table 6.1. For a fair comparison between experiments and simulations, the specimen thickness values considered in both cases, t in Equation 3.4, are the ones actually measured

6.2. Prediction of translaminar fracture and correlation of results

Figure 6.2: Kinematic and damage results of simulations of 12-layered CT tests in the longitudinal and transverse loading directions, and correlation with corresponding experimental results (for similar CMOD's).

from real test specimens instead of the ones modelled. In this way, the effect of ply nesting is considered in the numerical results. The CC method was not used for the purpose of validation of the numerical approach because, as mentioned above, the simulated crack propagation is highly unstable, leading to a large reduction in number of load peaks in the numerically-obtained curve from which to extract data for the CC reduction method, in comparison with the experimental approach. This compromises the reliability of the simulated compliance curves.

The fracture toughness values obtained experimentally fall within the ranges simulated for $80 \text{ kJ/mm}^2 < G_{ft} < 120 \text{ kJ/mm}^2$. However, there is a notable issue in this correlation. Whilst, the G_{Ic}^L values obtained experimentally are in the upper range of values predicted by the simulations, the experimentally-characterised G_{Ic}^T values fall in the lower range of the predictions. The first case is more representative

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Table 6.1: Correlation between experimental and simulated values of critical energy release rates for longitudinal (G_{Ic}^L) and transverse (G_{Ic}^T) loading directions, obtained with the Area Method.

Spec. Config.	Loading direction	Experiments	Simulations
		$[kJ/m^2]$	$[kJ/m^2]$
12 Layers	Longitudinal (G_{Ic}^L)	124.9 ± 9.7	86 - 128.5
	Transverse (G_{Ic}^T)	93.4 ± 2.3	89.8 - 146.2
18 Layers	Longitudinal (G_{Ic}^L)	163.6 ± 3.5	82 - 168.3
	Transverse (G_{Ic}^T)	97 ± 5.6	99 - 155.3

of what could be expected of the simulations since they do not take into account some energy absorbing mechanisms such as intertow delamination and fibre/yarn pull-out. The second case denotes the limitations of the modelling approach in capturing with enough accuracy the elementary damage mechanisms associated with fracture under transverse loading, i.e. the fracture of the bias yarns. Due to their orientation, the fracture of the bias yarns under global transverse loading is likely to occur under a combination of local longitudinal tensile and shear loadings. Whilst the Hashin failure criteria used in this work (Hashin (1980); Lapczyk and Hurtado (2007); Dassault Systèmes (2013)) takes this stress combination into account in the prediction of failure initiation (see Equation 5.2), the Continuum Damage Model (Lapczyk and Hurtado (2007); Dassault Systèmes (2013)) does not include mixed-mode damage propagation laws, i.e. it does not taken into account that different values of energy dissipation might be associated with fractures occurring under different ratios between $\hat{\sigma}_1$ and $\hat{\tau}_{12}$. Actually, there is experimental indication that longitudinal (fibre) breakage of unidirectional plies or yarns under shear loads dissipates significantly less energy than under longitudinal loads, Catalanotti and Xavier (2015). Therefore, a damage model that captures this damage interaction behaviour, such as the one proposed by Böhm *et al.* for 2D and 3D-textile composites (Böhm *et al.* (2011, 2016)) would be needed to correctly simulate the fracture of the bias yarns and provide higher accuracy in the prediction of fracture under the transverse loads using the proposed methodology. To obtain this effect with the current modelling approach, the simulations of the CT specimens under transverse loading were repeated with the longitudinal tensile fracture energy, G_{ft} , of the bias yarns modified from 80 kJ/mm^2 to 34.5 kJ/mm^2 . As shown in Figures 6.1.b and 6.1.d, the simulated CT behaviour improved remarkably. With this approach, the predicted

6.3. Conclusions to translaminar fracture simulation

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values of the critical energy release rates are 94.4 kJ/mm^2 and 105.3 kJ/mm^2 for the 12- and 18-layered virtual specimen, respectively.

6.3 Conclusions to translaminar fracture simulation

The Multi-Material Shell (MMS) modelling strategy developed in Chapter 4 was used to simulate the translaminar fracture behaviour of the braided architecture. This approach is based on a bi-dimensional approximation in which the axial and bias tows are projected rectangular planar surface of a Representative Unit Cell (RUC) creating the characteristic triaxial braiding pattern. Multiple RUC's were used to represent the full CT specimens, or to represent MMS regions embedded in homogenised domains. A Continuum Damage Model (CDM) based on the Hashin failure criteria was used to address the damage at the tow level as described in Chapter 4. A continuum damage/plasticity model simulated the resin matrix.

The application of this modelling approach to fracture of triaxial braided composites was validated by correlation between simulations of the Compact Tension tests and the corresponding experimental results. The fracture mechanisms for two different loading modes were captured with remarkable representativeness, and the simulated average values of the critical strain energy release rates correlate fairly well with the ones obtained experimentally, taking into account the uncertainty with regards to the yarn fracture energy and the limitations of the corresponding constitutive model. Due to the level of discretisation offered, the MMS approach is not reliable in the prediction of the full crack resistance curves and energy dissipation values associated with crack initiation and propagation. Nevertheless, the remarkable results achieved reveal an efficient virtual testing approach that can be used for the analysis fracture of triaxial braided materials at structural level.

MMS modelling for prediction of notched strength and size effects in braided composites

This chapter is aimed to investigate the use of Multi Material Shell model for the prediction of size effects in carbon-epoxy braided laminates containing open holes loaded in tension. The results are then compared with the experimental program conducted in Chapter 3, which was focused to measure the relevant material properties and to identify size effects occurring in laminates with different hole diameters.

7.1 Prediction of strength in composites laminates containing notches and holes

The behaviour of composite laminates containing holes, notches or cut-outs is of great importance in design because of the resulting strength and life reduction due to the stress concentrations. A number of models has been developed in the literature for predicting the notched strength of composites, [Awerbuch and Madhukar \(1985\)](#). These models can generally be classified into three groups: (1) stress-based fracture models, [Whitney and Nuismer \(1974\)](#); [Pipes et al. \(1979\)](#); [Tan \(1987\)](#); [Hwan et al. \(2011\)](#); (2) fracture-mechanics based models, [Tan \(1988\)](#); [Gillespie and Carlsson \(1988\)](#); [Waddoups et al. \(1971\)](#); and (3) progressive damage models, [Mar and Lin \(1977\)](#); [Backlund and Aronsson \(1986\)](#); [Chang and Chang \(1987\)](#); [Tan \(1991\)](#); [Su et al. \(2015\)](#).

7.1. Prediction of strength in composites laminates containing notches and holes

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Among them, the stress-based methods are preferred due to their simplicity and easiness of implementation. According to these theories, failure is produced when a critical stress is attained at a given point close to the notch root or hole region. [Whitney and Nuismer \(1974\)](#) was the first author who proposed the point stress criterion (PSC) and the average stress criterion (ASC) based on the normal stress distribution adjacent to the notch root. These models require two parameters, namely the unnotched strength of the laminate and a characteristic distance, assumed as a material property and constant for all notch sizes. These criteria are capable of predicting the notch strength of composite laminates without resorting to classical concepts of linear elastic fracture mechanics, [Hwan et al. \(2011\)](#). Similar concepts were used in finite fracture mechanics models, [Camanho et al. \(2007\)](#). In this case, failure of the notched laminate is produced when two criteria, stress-based and energy-based, are fulfilled at the same time using the unnotched strength and fracture toughness as material parameters.

Cohesive zone models (CZM) were also employed to predict failure in quasi-brittle materials ([Dugdale \(1960\)](#); [Barenblatt \(1962\)](#)), and in particular, composite materials. Cohesive zone models represent the damage occurring in the area affected by the notch, crack or holes using a softening law. The softening law controls the way stress is transmitted by the cohesive crack and depends on the maximum strength of the material and its fracture toughness. More recently, continuum damage models (CDM) has emerged as an adequate method to predict failure load of composite materials due to its ability to be implemented in finite element code as constitutive material laws. [Chang and Lessard \(1991\)](#), developed a two-dimensional progressive damage model to predict damage accumulation and the strength of open-hole composite laminates using the failure criterion of Yamada-Sun and Hashin, [Hashin \(1980\)](#). Similar progressive damage models have been developed for three-dimensional open-hole laminates including different failure criteria, [Wang et al. \(2004\)](#); [Labeas et al. \(2008\)](#). Such progressive damage models could lead to mesh dependence problems and could result in inaccurate predictions. For instance, [Labeas et al. \(2008\)](#), employed a progressive damage model using Hashin criteria to study the interaction effects between the post-buckling behaviour and the various failure modes of notched composite laminates. Their models show strong mesh dependency and the results predicted are deemed acceptable only for certain cases. More realistic progressive damage models involving continuous material property degradation have been developed and employed to study OHC problems,

7. MMS modelling for prediction of notched strength and size effects in braided composites

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Lee and Kim (2007); McGregor et al. (2008); Qing and Mishnaevsky (2010). A micromechanical constitutive model based on the concept of the ensemble-average volume for laminated composites was employed to characterise the compressive response and damage evolution in laminated plates containing a cutout, Lee and Kim (2007). It was denoted that the model can predict accurate results for cross-ply laminates, while it overestimates the stiffness and final failure load for angle-ply and quasi-isotropic laminates. McGregor et al. (2008), developed a continuum damage mechanics model of nonlinear constitutive behaviour to model the compressive failure of notched composites. An analogue model composed of spring, gap, fuse and slider elements is constructed as to represent the complete force–displacement response of a representative volume element of the material. The analogue model is however given as a predefined nonlinear force–displacement relationship which is difficult to determine. Moreover, most of these progressive damage models deal with the in-plane failure only. Failure mechanisms such as micro-buckling/local buckling, delamination, and splitting can hardly be incorporated into these constitutive models or the corresponding finite element models. Pinho et al. (2006b,a), developed a comprehensive progressive damage model based on a set of physical failure models and criteria for fibre-reinforced composites, which can distinguish the matrix and fibre failure under tensile and compressive loading. This model was employed to predict the compressive failure stresses for open-hole composite laminates for various stacking sequences by Qing and Mishnaevsky (2010). The results were in a good agreement with the experimental values for some of the stacking sequences considered, whilst for others, the predictions shown between a 10-25% discrepancy when compared to the experimental results.

Unlike the lack of efficient progressive damage modelling for OHC composite laminates, several computational models have been successfully developed to predict the tensile strengths and progressive damage for notched composite laminates, Hallett et al. (2009); Chen et al. (2013); Ridha et al. (2014). Chen et al. (2013), employed the smeared crack model and cohesive elements to study the size effects on OHT composite laminates; their predictions correlate well with the experiments by Wisnom et al. (2010). Ridha et al. (2014), developed a progressive damage model with in-plane damage and delamination between plies to predict the OHT strengths and failure progression for notched composite laminates with varying sizes and stacking sequences. It is found that only if both the progressive delamination and in-plane damage are modelled, the OHT strength can be predicted accurately. It should be

7.2. Simulation of braided OHT specimens

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noted that the most important feature of these progressive damage models is the softening after the stress of an element reaches the peak value, commonly, linear softening governed by the fracture toughness. The tensile fracture toughness in the fibre direction is regarded as a lamina property which can be measured using compact tension experiments.

Although many references with regard to notched strength prediction of composite laminates and woven fabric composites have been found and briefly introduced above, there is still a lack of research related to notched strength prediction of braided composites. Up to now, most studies in this regard are limited to the investigation of the effect of machined versus braided holes on both the bearing strength and microscopic damage mechanism of braided composites, [Nakai et al. \(2001\)](#); [Brookstein and Tsiang \(1985\)](#); [Brookstein \(1986\)](#); [Wang \(1994\)](#); [Fujita et al. \(1994\)](#); [Ohki et al. \(2000a,b\)](#). The main reason for lack of papers related to prediction of notched strength of braided composites might be owing to the same difficulties encountered in predicting the effective elastic moduli for 2D braided composites as mentioned by [Tsai et al. \(2008\)](#).

Because 2D braided composites have been widely used in thin-walled structures recently, it is thus very important to be able to predict the notched strength of 2D braided composite plates before they can be used as structural members for mechanical connection. Therefore, this section aims to investigate if the proposed Multi Material Shell (MMS) methodology is able to predict the notched strength of braided composite plates and hole size effects.

7.2 Simulation of braided OHT specimens

The geometries of OHT specimens from Figure [3.28](#) (36x200mm, 24x135mm and 12x70mm) were fully discretised using mosaic patterns of representative unit-cells (RUC) oriented in longitudinal and transverse direction as shown depicted Figure [7.1](#). The multiple RVEs were fully discretised using 4-noded shell elements with reduced integration similar to the previous tensile and interlaminar fracture toughness simulations, yielding to a maximum of ≈ 190000 S4R elements for the larger geometry (36x90 mm). As the total dimensions of the coupons were considered for virtual testing, no periodic boundary conditions were used maintaining the free edges conditions. Instead, machine clamp boundary conditions were reproduced by simulation of the machine grip and a master node located in the top edge was used

7. MMS modelling for prediction of notched strength and size effects in braided composites

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to impose the tensile load in the y direction on the specimen while the bottom edge was constrained in the same direction. Constitutive models for yarns and resin as well as material parameters were the same than those presented in the previous chapters.

Figure 7.1: MMS discretisation for notched coupons loaded in: a) Transverse direction, b) Longitudinal direction.

7.3 Correlation of OHT results

The numerical stress-strain curves obtained for the three different coupon and hole sizes and two braiding directions are presented in Figure 7.2. The results are compared with the experimental data previously reported in Chapter 3. The numerical results obtained are in good agreement with the experimental results and the non-linear elastic slope was properly captured despite the strong kinematic assumptions involved in the MMS model. The stress-strain curves are greatly affected by the

7.3. Correlation of OHT results

presence of the drilled hole in both loading directions due to the incipient stress concentrations and further damage progression during the applied deformation.

Table 7.1: OHT mechanical properties and modelling predictions.

Hole diameter (mm)	Exp. strength (MPa)	Exp. Strain-to failure (%)	Num. strength (MPa)	Num. Strain-to-failure(%)	Strength Error (%)	Strain Error (%)
Longitudinal specimens						
2	438±52	0.94±0.1	426	0.9	3	4.4
4	429±41	0.92±0.1	435	0.93	1.4	1.1
6	375±22	0.89±0.04	423	0.94	13	5.6
Transverse specimens						
2	424±5	1.16±0.03	497	1.28	17	10
4	433±7	1.05±0.02	482	1.19	11	13.3
6	368±14	0.87±0.04	467	1.16	27	33.3

The summary of the notched tensile strength and strain-to-failure predictions are also presented in Table 7.1 for comparison with the corresponding experimental results. In the transverse orientation, the model capture the non-linear elastic slope independently of the hole size but over-predicts the strength. This behaviour occurred also in the unnotched predictions, in which the relative error is about 11%. This divergence between the stress-strain curves, as reported for transverse loading on unnotched tensile simulations, may be caused by the appearance of damage by matrix cracking emanating from the notched and coupon edges.

In the longitudinal direction, the trend is fairly reproduced for certain hole dimensions (2 and 4 mm), while for larger hole diameters the non-linear response is not so well captured. Nevertheless, the strength prediction is in good agreement for the 2 mm and 4 mm hole diameters, as the non-linearity is not as pronounced as in the transverse direction. However, in the case of 6 mm hole diameter, the numerical curves exhibit a stiffer response than experimental ones but the strain-to-failure is predicted accurately. As in the unnotched tensile simulations, the presence of axial yarns aligned with the loading direction produces a stiffer response and the accumulated damage induced by the hole does not create sufficient matrix cracking to affect the stress-strain curve. It is clear that there is a size effect, in which the growth in the hole diameter induces more matrix cracking emanating from the notched edges creating a more non-linear response, but this is not captured properly by the model. This can be possibly attributed to the unit cell discretisation, in which the mesh directionality makes the model unsuitable to represent the relevant mechanisms when the hole diameter is larger than the RUC dimension ($\approx 3\text{mm}$) in

7. MMS modelling for prediction of notched strength and size effects in braided composites

Figure 7.2: Correlation between experimentally- and numerically-obtained stress strain curves for braided OHT specimens.

7.3. Correlation of OHT results

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the loading directions.

The damage variables for fibre d_f and matrix d_m in the longitudinal and transverse directions are shown in Figure 7.3 and 7.4 respectively. No significant differences are found with the damage patterns obtained from experimental tests. As in the unnotched tensile simulations, the evident presence of the free edge, triggers the damage initiation by strong accumulation in the free spaces between the overlapping of the axial and bias tows. This free-edge initiated damage mechanism in the form of matrix cracks is a behaviour characteristic of this fibre architecture when submitted to tensile loadings. Nevertheless, the concentration of matrix cracking variable d_m is more severe in the regions with high matrix material concentrations than in the experimental results previously described and can be settled to the absence of ply nesting. Fibre failure was promoted by the highly damaged matrix regions along the bias bundles producing catastrophic fibre failure once the corresponding Hashin criterion was fulfilled.

7. MMS modelling for prediction of notched strength and size effects in braided composites

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Figure 7.3: Correlation between experimental and numerical results of OHT coupons in the longitudinal braiding direction.

7.3. Correlation of OHT results

Figure 7.4: Correlation between experimental and numerical results of OHT coupons in the transverse braiding direction.

7.4 Conclusions to OHT simulation

The tensile notched behaviour of the QISO-L-52 triaxially braided architecture injected with RTM6 resin was determined numerically. The same damage mechanisms were observed to occur up to final failure as the ones ascertained in the radiographs of post-mortem tensile specimens (Chapter 3). The numerical strain-stress curves exhibited non-linear loading curve till final failure due to the geometrical non-linearities caused by the notched free edged effect.

The numerical curves obtained with the MMS superimpose with the ones obtained experimentally for hole sizes smaller than 6 mm. It was found a drilled hole size effect, in which its growth induces more matrix cracking emanating from the notched edges but not sufficient enough to capture the damage accumulation of the specimens with hole diameter larger than the smallest dimension of the RUC ($\approx 3\text{-}4$ mm). This can be endorsed to the unit cell discretisation in which the mesh is enforced to follow the bundle directions which makes it unsuitable to fully capture the non-linear trend with larger sized drilled holes.

In spite of the results, the error committed in the ultimate stress prediction using MMS method is less than 22% for transverse direction and 6.5% percent for the longitudinal direction, making it an interesting tool not only for mechanical property determination but for the study of architectural damage mechanisms.

Conclusions and future work

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During the last years of activity in the Composite Material Group at IMDEA Materials Institute, significant work was carried out in the field of development materials and protections subjected to dynamic and impact loads in aerospace applications. Triaxial braided composites emerged as good damage tolerant materials for such applications, specially under events produced by the impact of ice, composite and steel fragments. However, the design of laminates produced using these materials was very complex because the lack of reliable experimental results as well as robust and accurate models able to predict the mechanical performance in a limited and reasonable time.

This thesis was intended to solve both problems: first, by analysing experimentally the mechanical behaviour of a family of triaxial braided composites used to manufacture structural parts for lightweight applications under impact conditions. Second, to understand the main deformation and failure mechanisms necessary to deliver accurate and robust predictions. As a summary, the main developments and conclusions of the thesis are drawn in the following paragraphs:

- Triaxial braided laminates were manufactured using 271g/m² triaxial braided fabric QISO-L-52 supplied by A&P Technology (Ohio, USA, [A&P Technology \(2016\)](#)) and RTM6 (Hexcel) using resin transfer moulding (RTM). Laminates containing six, twelve and eighteen layers were produced with different overall volume fraction of fibre reinforcement. The fibre architecture corresponds to a balanced 2x2 diamond patterns formed by 6K carbon fibre axial yarns interlaced with 3K bias yarns. The geometrical parameters required for modelling purposes were determined by means of optical microscopy as well as radiography, [García-Carpintero et al. \(2017a\)](#).

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- The mechanical behaviour of the triaxial braided composites was analysed in detail by means of tension, shear and fracture tests that revealed the main deformation and fracture mechanisms, [García-Carpintero et al. \(2017b,a\)](#). Although the textile architecture was designed to deliver a quasi-isotropic behaviour irrespective to the loading direction, damage mechanisms were different. In tension, the mechanical behaviour in the longitudinal direction was totally controlled by the axial yarns that dictated the onset of damage and fracture. However, when testing in the perpendicular direction, specimen edge damage promote matrix cracking in the axial yarns being the failure produced by stress concentrations in the bias yarns. Interrupted test were proven to be useful to highlight the damage mechanisms occurring up to failure in the tensile composite coupons tested in the longitudinal and transverse direction. However this technique was difficult to be applied with other types of specimen which require complex fixtures, not allowing sufficient space to build a Teflon bag to contain the penetrant liquid.
  - The shear behaviour of this material was studied using the geometry proposed from the [ASTM International \(2005\)](#) standard, which failed to originate a crack at the 90° notch tip and was initiated at the 45° notched edges instead. A geometry modification was introduced to enforce the crack propagation and to show a more pure shear response in the longitudinal loading direction. However, it triggered an early failure in transverse loading, failing to predict the strength. A recommendation to use the standard geometry for strength determination and the proposed modification for stiffness prediction is given.
  - The effects of nesting, fibre volume fraction and fibre architecture were studied on the translaminar and interlaminar fracture toughness of triaxially braided composites with different conclusions, [García-Carpintero et al. \(2017a\)](#). While in the translaminar fracture toughness tests the fibre volume fraction and laminate thickness showed negligible effect, in the interlaminar fracture toughness tests, the increase of fibre volume fraction produced an increase in the critical energy release rate caused by the multiple delamination fronts yielding into fibre bridging and crack branching. Nevertheless, the increase in both interlaminar and translaminar fracture toughness of braided composites in comparison with other fibre architectures is remarkable. This makes it an outstanding material for energy dissipation applications.

- Structural evaluation and study of hole scaling size effects was performed using an independent test configuration (OHT) on triaxial braided composites. It was observed an additional non-linearity in the response of the material caused by further damage accumulation introduced by the free edge of the drilled hole. Depending on the fibre alignment and size of the hole, different response and level of damage accumulation were observed. The tensile strength decreased with the increase of the hole size when comparing with the unnotched specimens, being more severe in transversely aligned specimen, caused by the extra damage accumulation. As a conclusion, the material is suitable to be used in application in which the size of the hole is similar to the smallest dimension of the RUC (10x3mm) to avoid a decrease in its strength.
- Unit cell modelling is a technique to reproduce an idealisation continuous and repetitive portion of the material in which the mechanical properties remains invariant. The model offers more accurate property prediction attending to the level of detail applied to the composite material, e.g. yarn shape, fibre volume fraction, nesting, etc. However, the main drawback is the level of computational resources which increases directly with the level of detail considered into the model, making them in certain application prohibitive.
- Different computational unit cell models applicable for textile composites and triaxial braided composites in particular were compared in Chapter 4. Among them, MMS model, [García-Carpintero et al. \(2017b\)](#), present an outstanding advantage in terms of mechanical behaviour prediction, not only because it was proven to be a fast and reliable tool for property prediction keeping the computational resources to a minimum, but continuum two-dimensional damage models could be used to study and reproduce the associated damage mechanisms at coupon level, [García-Carpintero et al. \(2017a\)](#). Despite the simplicity of the approach, the inability to simulate yarn-to-yarn delamination and fibre/yarn pull-out, the numerical results were in good agreement with the ones obtained in the mechanical tests at coupon level. The model offered the possibility to capture macroscopically the damage mechanisms associated with different mechanical tests, Chapters 5-7. Nevertheless, the continuum damage model used did not incorporate a mixed-mode propagation law with different values of energy dissipation attending to the fracture level associated to the ratio between tow tensile and shear loadings, reducing the accuracy in the transverse direction, in which the model was unable to fully capture the non-

8.1. Proposed lines for future research

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linearity of the material. Furthermore the model seems only valid to predict accurately the tensile mechanical response of the material in the presence of defects with smaller or similar dimensions as the RUC.

8.1 Proposed lines for future research

Even though studies on textile composites materials have been done by many researchers, the field of triaxial braided composites development, characterisation and analysis are still open with the introduction of the different angle of bias braided fibers and analysis of different components. The present research work aimed to develop a numerical modelling methodology to represent the mechanical behavior of carbon fiber triaxial braided composites and it is validated through experimental tests at coupon level. Every new method or model opens the door for a number of possible improvements. Some points for potential future research, arising from the research work conducted in the framework of this thesis, are given below.

Concerning fibre architecture

- Study the tow deformation effects which occur during consolidation, such cross-sectional deformations and nesting by evaluating micro-CT scans quantitatively for yarn architecture characterisation. This requires contrast enhancement and 3D segmentation methods for yarns and matrix, especially when the tows are co-aligned. The most successful contrast enhancement procedures used up until now involve gold sputter coating and soaking of tows in an aqueous iodine medical tracer, [Djukic et al. \(2013\)](#). The procedure involves a tow coating with a material of relatively high density prior to braiding, to cause a difference in X-ray attenuation compared to the rest of the composite. Another approach to enhance the contrast could be by using glass tows as substitute for carbon, since both material have different density.

Concerning mechanical characterization

- Effects of sublaminate such as nesting and fibre volume fraction, and drilled hole size effect present in structural working conditions have been studied for certain mechanical tests. Therefore to have more insight about the behaviour of this fibre architectural composites and its effect on the mechanical prop-

erties, it is recommended to characterise the material at least in three fibre configurations used, 6- , 12- and 18-layered.

- Shear test using the [ASTM International \(2005\)](#) and the proposed modified version revealed the necessity of develop a methodology or standard procedure to characterise textile composites under shear stress states.
- Further low- and high-velocity impact characterisation are proposed to determine in more detail the energy absorption mechanisms and ballistic thickness.

Concerning unit cell modelling

- Investigate refined geometric modelling approaches with CT-scan results to obtain more realistic fibre volume fractions in the unit cells.
- Investigate experimental methods for measuring yarn strength and numerical methods to scale strength parameters to different fibre volume fractions.
- Develop a CDM to predict the strength mechanical properties and onset of damage using the three-dimensional unit cell.

Concerning macroscopic modelling

- Incorporate the effect of crimping and undulations in the behaviour of the bundles using the MMS methodology. In addition, improve the MMS model by incorporating cohesive behaviour between stacked plies to be able to simulate subsurface delamination.
- Promote and integrate into the ([Hashin, 1980](#)) CDM, mixed-mode propagation laws with different values of energy dissipation attending to the fracture level associated to the ratio between tensile and shear loadings. Study their effects on the transverse loading mechanical prediction.
- The applicability of the proposed macroscopic modelling methods for braided laminates with a small number of plies ($n < 6$) needs to be investigated. For these cases, the local out-of-plane deformation can become more prominent. Extend the investigations with comparisons to DIC measurements of the out-of-plane deformation. This requires a higher DIC- resolution on the specimen and experiments with laminates of different ply numbers.

8.1. Proposed lines for future research

- Investigate if the proposed MMS methodology is also valid to predict other mechanical properties such as Mode I interlaminar fracture toughness. Extend the investigation using continuum shell elements coupled with cohesive elements or surfaces to represent the crack opening.
- Evaluate the performance of the material under impact loading cases and determine if the MMS methodology is also valid to predict the mechanical behaviour and damage pattern.

Annexes

A.1 Analytical method for the mechanical properties prediction.

For simple estimation of the triaxially braided composite in-plane elastic properties an analytical model has been developed. The used constituent material properties are defined in Table 4.4 and Table 4.3 and the yarn volume fractions are obtained experimentally, as described in. Using the independent elastic constants, as obtained by means of Chamis micro-mechanical model, Chamis (1989), the transversely isotropic yarn compliance tensor for axial and bias yarns (S^a and S^b) can be written in Voigt notation as:

$$S^{a,b} = \begin{bmatrix} 1/E_{11} & -\nu_{12}/E_{11} & -\nu_{12}/E_{11} & 0 & 0 & 0 \\ -\nu_{12}/E_{11} & 1/E_{22} & -\nu_{23}/E_{22} & 0 & 0 & 0 \\ -\nu_{12}/E_{11} & -\nu_{23}/E_{11} & 1/E_{22} & 0 & 0 & 0 \\ 0 & 0 & 0 & 1/G_{23} & 0 & 0 \\ 0 & 0 & 0 & 0 & 1/G_{12} & 0 \\ 0 & 0 & 0 & 0 & 0 & 1/G_{12} \end{bmatrix} \quad (\text{A.1})$$

From the compliance matrices of the matrix material S^m , which is considered to be isotropic, the axial yarns S^a and bias yarns S^b , the corresponding stiffness matrices Q^m, Q^a and Q^b are obtained by matrix inversion.

The reduction of stiffness resulting from tow crimping has been taken into account by rotating the yarn stiffness matrices over the average undulation angle.

A.1. Analytical method for the mechanical properties prediction.

From micrograph cross sections the undulation properties for a six layer sample are acquired as $z(y') = A \sin(\omega y')$ with an amplitude A of $0.147 \pm 0.023 \text{ mm}$ and a wavelength $k = 2\pi/\omega$ of $10.181 \pm 1.383 \text{ mm}$ for the bias yarns and $A = 0.013 \pm 0.004 \text{ mm}$ and $k = 4.580 \pm 1.755 \text{ mm}$ for the axials, as illustrated in Fig. 4.2. The crimping angle ϕ can be expressed as:

$$\tan(\phi) = \frac{dz}{dy'} = A\omega \cos(\omega y') \quad (\text{A.2})$$

The reduced stiffness matrix of the undulated bias yarns $Q_{B_{und}}$, is obtained by integrating the transformed stiffness matrix Q_B over half a wavelength:

$$Q_{B_{und}} = \frac{2}{k} \int_0^{k/2} (T_1)^{-1} Q_B T_2 dy' \quad (\text{A.3})$$

with T_1 and T_2 as the Euler rotation matrices defined in Eq. A.4 and A.5, as described by Ruan et al. Ruan et al. (1999).

$$T_1 = \begin{bmatrix} a_{x1}^2 & a_{x2}^2 & a_{x3}^2 & 2a_{x2}a_{x3} & 2a_{x1}a_{x3} & 2a_{x1}a_{x2} \\ a_{y1}^2 & a_{y2}^2 & a_{y3}^2 & 2a_{y2}a_{y3} & 2a_{y1}a_{y3} & 2a_{y1}a_{y2} \\ a_{z1}^2 & a_{z2}^2 & a_{z3}^2 & 2a_{z2}a_{z3} & 2a_{z1}a_{z3} & 2a_{z1}a_{z2} \\ a_{y1}a_{z1} & a_{y2}a_{z2} & a_{y3}a_{z3} & a_{y2}a_{z3} + a_{y3}a_{z2} & a_{y1}a_{z3} + a_{y3}a_{z1} & a_{y1}a_{z2} + a_{y2}a_{z1} \\ a_{z1}a_{x1} & a_{z2}a_{x2} & a_{z3}a_{x3} & a_{z2}a_{x3} + a_{z3}a_{x2} & a_{z1}a_{x3} + a_{z3}a_{x1} & a_{z1}a_{x2} + a_{z2}a_{x1} \\ a_{x1}a_{y1} & a_{x2}a_{y2} & a_{x3}a_{y3} & a_{x2}a_{y3} + a_{x3}a_{y2} & a_{x1}a_{y3} + a_{x3}a_{y1} & a_{x1}a_{y2} + a_{x2}a_{y1} \end{bmatrix} \quad (\text{A.4})$$

$$T_2 = \begin{bmatrix} a_{x1}^2 & a_{x2}^2 & a_{x3}^2 & 2a_{x2}a_{x3} & 2a_{x1}a_{x3} & a_{x1}a_{x2} \\ a_{y1}^2 & a_{y2}^2 & a_{y3}^2 & 2a_{y2}a_{y3} & 2a_{y1}a_{y3} & a_{y1}a_{y2} \\ a_{z1}^2 & a_{z2}^2 & a_{z3}^2 & 2a_{z2}a_{z3} & 2a_{z1}a_{z3} & a_{z1}a_{z2} \\ 2a_{y1}a_{z1} & 2a_{y2}a_{z2} & 2a_{y3}a_{z3} & a_{y2}a_{z3} + a_{y3}a_{z2} & a_{y1}a_{z3} + a_{y3}a_{z1} & a_{y1}a_{z2} + a_{y2}a_{z1} \\ 2a_{z1}a_{x1} & 2a_{z2}a_{x2} & 2a_{z3}a_{x3} & a_{z2}a_{x3} + a_{z3}a_{x2} & a_{z1}a_{x3} + a_{z3}a_{x1} & a_{z1}a_{x2} + a_{z2}a_{x1} \\ 2a_{x1}a_{y1} & 2a_{x2}a_{y2} & 2a_{x3}a_{y3} & a_{x2}a_{y3} + a_{x3}a_{y2} & a_{x1}a_{y3} + a_{x3}a_{y1} & a_{x1}a_{y2} + a_{x2}a_{y1} \end{bmatrix} \quad (\text{A.5})$$

The terms a_{ij} ($i = 1, 2, 3, j = x, y, z$) are the cosines between the axis of the global en local coordinate systems, defined as $a_{ij} = \cos(i, j)$, which can be written as $\cos(\phi) = \frac{1}{\sqrt{1+\tan^2\phi}}$ for the undulation angle.

Figure A.1: Definition of local (1,2,3) and global (x,y,z) coordinate systems with corresponding braid angle θ in the x-y plane and the undulation angle ϕ in the y-z plane and z-x plane

Because of the braid angle $\theta = -60^\circ$ and $\theta = +60^\circ$ with respect to the axial yarns, the undulated bias stiffness matrices are transformed from local to global coordinates (see Fig. A.1) using again the Euler transformation matrices as shown in Eq. A.6:

$$\bar{Q}_B = (T_1)^{-1} Q_{B_{und}} T_2 \quad (\text{A.6})$$

Since Q^m is isotropic and Q^a is already aligned with the global coordinates, no transformation is required.

The total unit cell stiffness matrix $\bar{Q}_{RUC}$ is obtained by volume averaging the individual stiffness matrices, for which the Chamis properties are displayed in Table A.1, with the geometric volume fractions for the axial and bias yarns V_g^{fa} and V_g^{fb} and the matrix V_g^m as shown in Eq. A.7 while assuming isostrain conditions. Geometrical volume fractions for yarns and matrix are obtained from the area yarn cross-sections in a region of interest of micrographic samples (as shown in Figure 3.2)

In Equation A.7 $\bar{Q}_{b_{-60^\circ}}$ and $\bar{Q}_{b_{+60^\circ}}$ are the rotated undulated bias stiffness matrices. The geometric volume fractions are obtained by quantifying the area fractions of bias yarns, axial yarns and the matrix material using the optical microscope cross

A.1. Analytical method for the mechanical properties prediction.

Table A.1: Mechanical properties of composite components. Yarn properties are obtained via Chamis micro-mechanical model [Chamis \(1989\)](#)

Material	E_1 (GPa)	E_2 (GPa)	G_{12} (GPa)	ν_{12} (-)
Yarns	163.0	8.3	4.4	0.315
Resin	2.89	-	1.07	0.35

sections, as shown in Figure 3.2 and assuming constant bundle sections throughout the unitcell.

$$\bar{Q}_{RUC} = V_g^{fa} \bar{Q}_a + \frac{V_g^{fb}}{2} (\bar{Q}_{b_{+60^\circ}} + \bar{Q}_{b_{-60^\circ}}) + V_g^m \bar{Q}_m \quad (A.7)$$

By considering in-plane uniaxial tensile and shear stress states, the in-plane effective elastic properties can be obtained from $\bar{Q}_{RUC}$ [Quek et al. \(2003\)](#). Table A.2 shows the obtained elastic properties and the effect of undulation on the results, assuming the constituents properties from Table A.1 with a yarn local volume fraction of 67%.

Table A.2: In plane effective elastic properties acquired using the analytical model

Elastic property	No undulation	Undulation	Difference (%)
$E_x(GPa)$	48.22	48.16	-0.1
$E_y(GPa)$	43.87	42.29	-2.0
$G_{xy}(GPa)$	16.68	16.27	-1.4
$\nu_{xy}(-)$	0.31	0.30	-2.3

As expected from the micrographs inspection, the reduction in stiffness resulting from undulation is most evident in the y direction, as this direction is dominated by the bias fibers which show more significant undulation than the axials.

A.2 Towards simulation of low-velocity impact on braided composites

In the aeronautical and automotive industries, the structures are vulnerable to sustain damage due to tools dropped during fixing and low-velocity impacts (LVI). Susceptibility to damage from concentrated out-of-plane impact forces is one of the major design concerns of many structures made of advanced fiber architectural composite laminates. Knowledge of the damage resistance properties of braided composites is useful for product development and material selection.

The objective of this specific study is to evaluate if the proposed MMS method is able to predict the ballistic behavior of braided composites and if the model is able to reproduce the energy dissipation mechanism occurred in a real impact loading. Thus, low-velocity impact characterization on triaxially braided coupons to study the damage mechanisms following the applicable standard, [ASTM International \(2015b\)](#), were performed.

A.2.1 Low-velocity mechanical characterization

Drop weight impact testing is the most common technique for low-velocity impact characterization, usually selected for composite laminates. The test consists on raising a mass to a known height and release, impacting the specimen, as schematized in Figure [A.2.1](#).

Figure A.2: Schematic of a drop weight impact test, [Duell \(1995\)](#)

A.2. Towards simulation of low-velocity impact on braided composites

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wherein H is the initial drop height, W is the impact weight, and v_0 is the impact velocity. The tip is usually a hemispherical impactor that measures the strain during impact.

Following literature recommendations, [Falzon and Herszbergb \(1998\)](#), which were validated with the results of the tensile tests described in Chapter 3, a minimum of six plies should be considered in order to obtain a in-plane quasi-isotropic behavior. Therefore, and following the minimum thickness required by the standard (between 3 a and 4 mm), RTM panels of twelve layers with cured ply thickness of $\approx 0.3\text{mm}$ and a fiber volume fraction of 55.1% were manufactured. After manufacturing, ultrasonic C-scan inspection was applied to ensure the panel was suitable to testing consideration.

A.2.1.1 Test set-up and procedure

Drop-weight tests were carried out following ASTM D7136-15, [ASTM International \(2015b\)](#). In this standard, it is described the specimen preparation before impact, test procedure and data analysis method. The first step is to record the specimen dimensions, such as length, width and thickness. An adequate impactor should be selected, the mass of the drop-impactor and the maximum load that the machine has to support during the test have to be checked and adapted, the environmental conditions has to be controlled and corrected if it was necessary and, finally, set the data acquisition variables.

Once the specimen is properly placed inside the machine, it is clamped by four rubber tips to avoid any undesirable displacement during the impact testing and the mass is raised. In the machine, the mass and the impact energy are established and, automatically, the machine goes to the required height to achieve the required potential energy. In Equation A.8 , it can be seen the relationship between mass, gravity, height and the total energy.

$$E_{TOTAL} = mg(h_0 - h_f) \quad (\text{A.8})$$

where E_{TOTAL} is the potential energy, m is the mass, g is the gravitational acceleration, h_0 is the original height, and h_f is the final height, [Duell \(1995\)](#).

In order to select the initial impact energy, an approximate relation between specimen thickness and the specified ratio between absorbed energy to the thickness was followed, [ASTM International \(2015b\)](#). Based on the equation for non-embedded

testing from the standard, a modified version was proposed to account for the energy dissipation of braided composites. A factor ε with a value of 1.5 was added to the initial equation, yielding:

$$E_{PRIOR} = \varepsilon \beta t \quad (\text{A.9})$$

Where E_{PRIOR} is the impact energy of the impactor, ε is the incremental energy factor with a value of 1.5; β is the specified ratio of the impact energy to specimen thickness which the value of 6.7 J/mm and t is the specimen thickness, with a value of 3.2 mm. After introducing all the values in the equation, the resulting energy was 32J.

A.2.2 Data evaluation

Five tests with energies at 30J, 35J, 40J, 42J and 45J were carried out to determine the ballistic limit. It was achieved with an impact energy at 42J, as it can be observed in the Table A.3. The first test (SP01-30J) was performed at 30J, close to the initial energy calculated using equation A.9.

Table A.3: Drop tower test results on braided composites

Specimen	Thickness (mm)	Impact Energy (J)	Indentation (mm)	Result
SP01-30J	3.2	30	1.77	Rebound
SP02-35J	3.3	35	2.17	Rebound
SP03-45J	3.3	45	N/A	Embedded/Penetration
SP04-40J	3.3	40	4.05	Rebound
SP05-42J	3.3	42	6.96	Rebound

In Figure A.2.2 the ultrasonic C-Scan inspection of the damaged area of the five specimens tested is shown, scaled from black, the minimum, up to white, the maximum delamination level.

As it can be observed, blue is the background for these specimens and black areas are zones without material, such as the centre of the impact area in specimen SP03-45J, where the material was totally damaged. In the impact area, the centre is plotted by white, so the level of delamination is maximum. In addition, higher amount of impact energy made larger damaged area till reaching the ballistic limit,

A.2. Towards simulation of low-velocity impact on braided composites

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Figure A.3: Ultrasonic C-Scan inspection at the impact point on the tested specimens, [Herrero García \(2016\)](#).

in which the higher the impacting energy will cause the fibre breakage by shearing concentrating the damage around the impacted area.

Damage area can be likened to a rhomboid, with the maximum diagonal parallel to the large edge of the specimen. The edges of the specimen affected the crack propagation, being the specimen stiffer in the direction parallel to the small edge. The axial yarns were also absorbing large amount of energy, making damage goes easily in this direction. The specimen SP01-30J was trimmed with axial yarns in a different direction in comparison with the rest, and it can be observed that the damage did not follow the large edge direction.

The specimens SP01-30J and SP03-45J, present two bubbles of air on the surface, that they are not affecting the integrity of the specimens but they can be seen in the respective pictures.

A.2.3 MMS for low-velocity impact simulation

The Multi Material Shell (MMS) method was used to simulate the impact behaviour of 12-layered triaxial braided composites reproducing the test conditions recommended by the ASTM D71306, [ASTM International \(2015b\)](#), as illustrated in Figure A.2.3.

Figure A.4: Drop test schematic in Abaqus, [Herrero García \(2016\)](#)

The impactor was a spherical tip with 16 mm diameter, modelled as 3D analytic rigid shell. The initial velocity was set in the z axis, with rotations constrained to force the moving always vertically against the panel, which was placed in the plane XY. The clamps are rigid bodies rotationally constrained in every axis and they are applied a load of 1100N to the specimen. The support is placed below the

A.2. Towards simulation of low-velocity impact on braided composites

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150x100mm² target specimen, representing the machine, being totally constrained in displacements and rotations. The plate is made of steel and has a window in the middle, which allows the impactor to go through the specimen during the impact event.

The constitutive properties of the AS4C/RTM6 tows and RTM6 matrix pockets that were used in the numerical simulations are described in Chapters 4 and 5.

A.2.4 Numerical results

The numerical simulations showed that the energy to perforate the target panel is around 250J, higher than the one obtained in the experimental tests (42J, listed in Table A.3). This is caused because due to the two-dimensional modelization approach, the through-the-thickness damage behaviour is not represented, and therefore, the energy dissipated by delaminations and shear failure is ignored. Thus, the fibre tensile and compressive fracture energies, correlated in Chapter 5, were scaled down until the impact energy for perforation fitted with the experimental results, being these energies a tenth of their original value. In addition, these energies are verified to be higher than the minimum to avoid mesh dependency showed by the following equation:

$$G_{t,c_{1,2}} = \frac{X_{t,c_{1,2}}^2 L_c}{2E_{1,2}} \quad (\text{A.10})$$

where $G_{t,c_{1,2}}$ is the tensile and compressive fracture energy of the material in the 1,2-direction, $X_{t,c_{1,2}}$ is the tensile and compressive strength of the material, $E_{1,2}$ is the Young modulus and L_c is the characteristic element length.

The ballistic limit of the specimen was achieved with a penetration energy of 40 J, resulting from an initial impact velocity of 3,8 m/s. The following figure, Figure A.2.4, shows fiber and matrix tensile behaviour represented by the Hashin variables for damage initiation and evolution (*HSNFTCRT*, *HSNMTCRT*, *DAMAGEFT* and *DAMAGEMT*). Both indices should range from 0 to 1, but to avoid convergence problems caused by excessive element deformation, the damage evolution variable was limited to 0.99. Once the damage initiation value is equals to 1.0, the material starts to damage exhibiting a softening response until the damage evolution variable reach a value of 0.99, in which the element is considered failed and removed from the visualization.

Figure A.5: Damage initiation and evolution variables for fibre and matrix tension at the ballistic thickness, [Herrero García \(2016\)](#)

A.2. Towards simulation of low-velocity impact on braided composites

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From Figure A.2.4 it can be seen how due to the architecture geometry, the spherical impactor caused rhomboid-like shape, in a similar way to what was obtained experimentally. A high number of cracks started from the impact hole and move along the yarns preferential directions avoiding mesh dependency and element dominant crack propagation. The numerical simulation reproduced the typical shear behavior exhibited at the ballistic limit in conjunction with a ply-petalling at the bottom side caused by the fiber and matrix tensile failure. As a consequence, results are in good agreement with the observed damage from the tested specimen at the ballistic limit. Furthermore, tensile matrix failure can be related with delaminations, being the coloured red zone represented by the Hashin criteria damage initiation variable for matrix tension, the region prone to be delaminated. Similar response is observed from the ultrasonic C-scan inspection as shown in Figure A.2.2 for the impacting energy of 42J.

A.2.5 Conclusions to LVI experiments and simulations

An experimental characterization was carried out in order to understand the real behaviour of the triaxially braided composites. Five twelve-layered specimens were manufactured by RTM process. A low-velocity impact testing under ASTM D7136 standard was performed using the drop test tower machine. A penetration energy of 42J was the result achieved, demonstrating that this architecture exhibits an outstanding behaviour under impact loading.

It was found that the fracture toughness correlated through the numerical validation affected the results, over-predicting of the perforation energy. A sensitivity analysis on fracture energy properties was carried out in order to determine the energy dissipation levels inherent to these models avoiding snap-back and mesh dependency. It was observed that the numerical results exhibited a dependency from the mechanical properties and fracture toughness of yarns and matrix.

Nevertheless, the MMS methodology seems a valid and fast approach to accurately predict the perforation energy and to reproduce the energy dissipation mechanisms observed in the experimental tests.

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