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Marie Sklodowska-Curie European Training Network

9 - 12 Dec.

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The FiBreMoD project has received funding from the European Union's Horizon 2020 research and innovation programme under the Marie Skłodowska-Curie grant agreement No 722626.

Abstracts for oral presentation

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Abstracts

For the oral programme

FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

STRENGTH AND TOUGHNESS TUNING OF FIBRE-REINFORCED COMPOSITES THROUGH INTERFACIAL TOPOGRAPHICAL OBSTACLES

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Keywords: Composite, Fibre reinforcement, Interface, Intermittent beading, Beaded fibre,
Session topics: Constituent properties, Micromechanics, Bio-inspired design

ABSTRACT

Natural composites display a superior balance of strength, toughness and resources through hierarchical design strategies that rely on the shape and arrangement of constituents in the composite rather than chemical means only. The presence of internal interfaces plays a particularly important role in this balance by governing toughness and deformation through non-linear deformations that channel cracks through configurations that ultimately arrest crack propagation. Implementation of such a concept, namely the incorporation of structural obstacles at the interface of synthetic fibre-reinforced composites, could be a promising way forward to simultaneously stronger and tougher engineering materials with tuneable properties dependant on the topography of the interface.

Bearing this in mind, a novel structural design for interfaces is proposed – one comprising of an array of cured polymer beads around and along the length of a fibre. The beads are proposed to act as intermittent topographical obstacles to simultaneously generate toughness and strength in fibre-reinforced composites through two main mechanisms - (1) anchorage of fibres to increase stress transfer and (2) improved resistance to crack propagation through increased pull-out resistance. Single fibre pull-out tests performed on model glass fibre–epoxy bead system support this hypothesis. The maximum fibre strength was achieved for beaded fibres under stress as compared to regular fibres without such obstacles. Pull-out of beaded fibres appeared to dissipate more plastic deformation energy compared to that of regular fibres.

The system has the potential to be highly tuneable. The polymer beads are formed by the Plateau-Rayleigh instability, through which successful control of the bead parameters (size, angle, spacing) has been achieved. An investigation into optimisation and fine-tuning the system is currently being conducted and results will be discussed.

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FRACTURE TOUGHNESS DETERMINATION OF POLYMERIC MATRIX MATERIALS WITH TDCB SPECIMENS AND ITS VALIDATION BY SIMULATION AND VCCT

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Keywords: Polymer, Fracture toughness, Mechanical properties, Finite elements
Session topics: Constituent properties, Fracture toughness, Novel experimental techniques

ABSTRACT

Polymer matrix composites (PMCs) are currently in research focus as lightweight materials for aircraft and automobiles constructions, because of their extraordinary mechanical properties. The polymer as the matrix material is responsible for holding the orientation and position of the reinforcing fibers and is responsible for the transfer of stresses between the fibers. Thus, the properties of the polymer and its interaction with the fiber surface are crucial for the optimization of the mechanical properties of PMCs.

Fracture toughness is a key property o resins used for PMCs. In order to study this property, compact tension (CT) test is commonly employed. This test involves a specimen with a special geometry. During this test, a crack is grown in the CT specimen. In order to ensure a most brittle crack, the test speed must not be high, which could make it difficult to monitor the crack extension length. Furthermore, at high speed large-scale deformations could occur in the specimen, which changes the local stress field at the crack tip. To ensure an almost constant stress intensity at the crack tip, Tapered Double-Cantilever-Beam (T-DCB) specimen is used as a CT specimen (Fig. 1)]. [1] This study presents fracture toughness with different initial crack lengths in order to find the range of stable conditions. In addition, the experimental results are validated by finite element analysis (FEA) simulation using the virtual crack-closure technique (VCCT).

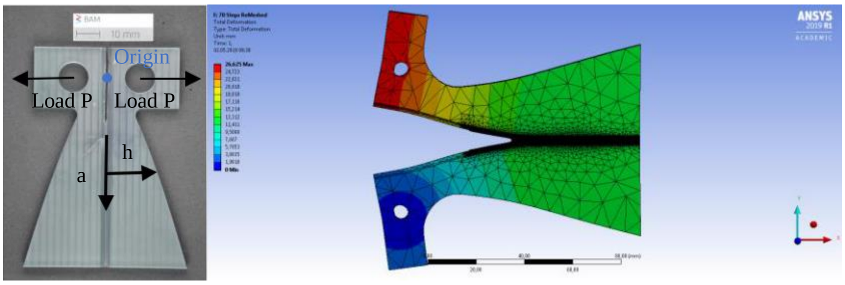

Figure 1: Tapered-double-cantilever beam (T-DCB) specimen and the related FEA model: a is the crack length, h is the distance between the crack growth path and the tapered curve.

The T-DCB specimen is basically a CT specimen with a special geometry factor m (1):

$$m = \frac{3a^2}{h^3} + \frac{3}{4}\left(1 + \frac{\gamma}{2}\right)\frac{1}{h} \quad (1)$$

The contour ensures a constant m even with the variation of the crack length a. Since the T-DCB specimens in this study are grooved, the fracture toughness, which is the critical stress intensity factor K_{Ic} can be written as (2):

$$K_{Ic} = 2P_c\sqrt{\frac{m}{BB_n}} \tag{2}$$

where P_c is the critical load for crack growth and B_n is the thickness of the specimen between the grooves.

In this study, the epoxy RIMR 135 / RIMH 137 from Hexion was investigated. The fracture experiments were performed at the displacement velocity of 1 mm/min.

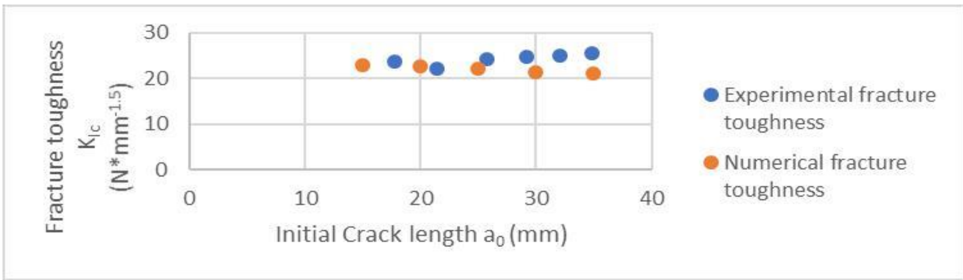

Figure 2: Fracture toughness of T-DCB specimens with different initial crack lengths.

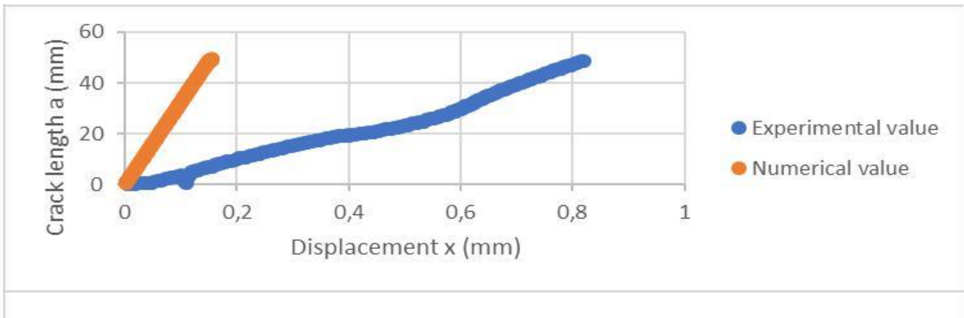

Figure 3: Comparison of experimental and numerical load–displacement diagrams with the initial crack length of 25mm.

The T-DCB shaped specimens were tested with different initial crack lengths. These were cut along the groove and extended for about 1mm with a shape blade, to generate a sharp tip. Then these specimens were loaded with force P_c, where the crack begins to grow. With P_c, the fracture toughness K_{Ic} was calculated and plotted as a function of the initial crack length a_0 (Fig. 2). From these tests, a constant range of K_{Ic} was obtained between 15 and 35 mm, and the average fracture toughness was 24 $N*mm^{-1.5}$. With the simulation, a value of 22 $N*mm^{-1.5}$ was found, which is about 10% lower than that of the experimental results. As shown in Fig. 3, a much larger displacement was required for driving the crack forward in the tests compared to the simulation. The reason lay in the fact that polymer was simulated as an ideal elastic and brittle material, yet it showed time-dependent deformation due to its viscoelasticity. Nevertheless, the crack begins to grow under the same crack intensity factor K_{Ic}.

The T-DCB specimen provides a constant range of K_{Ic}. Therefore, the determination of the fracture toughness K_{Ic} is easier than that with a standard CT specimen, where the stress intensity strongly depends on the crack extension length. Within the constant fracture toughness range, the experimental fracture toughness value agrees to the numerical value well.

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SURFACE MODIFICATION OF UHMWPE FIBRES WITH Ar-O_2 PLASMA TREATMENT

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Keywords: UHMWPE, plasma treatment, contact angle, surface modification, surface roughness

Session Topics: Constituent properties, Fibre-hybrid composites

ABSTRACT

An initial study on feasibility of surface modification was carried out on UHMWPE fibres which were supplied in the form of a yarn, by subjecting these to a high power Ar-O_2 plasma treatment for a short duration of time. The effect of plasma treatment was studied using various characterisation techniques that involved both qualitative and quantitative analysis of the chemical species as well as changes in surface roughness as a result of plasma treatment.

Morphological studies of UHMWPE fibres before plasma treatment confirmed their fibrillar nature [1], [2] whereas there were formation of cracks as a result of this treatment, the fibres in pristine condition showed a smooth and defect-free surface. The plasma treatment was carried out at a relatively high wattage and a longer time, surface deterioration was expected and observed using SEM. In some cases, the removal of flakes of material was observed. Although the surface seemed more roughened, it was useful for improved adhesion and wettability because it increases the mechanical interlocking phenomenon of fibres with the resin.

The wettability of the system was quantified by measuring the contact angle between tangents passing through the liquid and the solid phase, as explained by Chawla *et al.* [3]. However, in our study, both the phases were solid *i.e.* a droplet of resin was cured on a single fibre and was then observed under an optical microscope to determine the contact angle. A large variation in embedment length of the droplet on the fibre was observed with a slight decrease in contact angle as a result of plasma treatment. The embedment length of droplet increased and contact angle decreased suggesting that the resin was spreading more on the fibre as a result of plasma treatment.

The surface roughness of the fibres before and after the plasma treatment was measured using a High Speed - Atomic Force Microscope (HS-AFM). The data obtained were analysed in open source software (Gwyddion) and were treated with plane levelling, alignment of rows, removal of scars, and finally determination of statistical parameters. An increased surface area was observed as a result of plasma treatment and a large scatter in the measured statistical parameters, due to a rough nature of the samples. The peaks and valleys on the surface of treated sample seemed to be more pronounced, as observed by Gao [1]. The addition of oxygen bearing functional groups on the fibre surface were identified using FTIR and quantified by XPS analysis.

It was concluded that plasma treatment was responsible for surface roughness and addition of oxygen bearing functional groups. However, since the treatment was carried out on a higher wattage plasma setting, the environment was sufficiently harsh enough to degrade the fibre. An increase in oxygen functional group intensity was anticipated to lead towards higher affinity towards polar resins. Based on the current methodology used, the determination of contact angle requires more sophisticated measurement methods to reduce variability in the experimental observations.

This work is supported by the Engineering and Physical Sciences Research Council through EPSRC Centre for Doctoral Training at the Advanced Composites Centre for Innovation and Science (ACCIS, Grant number EP/L016028/1) and by DSM Netherlands in the form of a studentship. The author would also like to extend gratitude to the Interface Analysis Centre (University of Bristol) & Surface Analysis Laboratory (University of Surrey) for their comments & support in conducting the experiments at the respective facilities.

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NUMERICAL ANALYSIS OF THE EFFECT OF INTERFACIAL PARAMETERS ON LONGITUDINAL FIBRE-MATRIX DEBONDING

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Keywords: Finite elements, Cohesive zone, Interfacial strength, Interfacial fracture toughness
Session topics: Constituent properties, Micromechanics, Delamination, Fracture toughness

ABSTRACT

When the strength of the fibre-matrix interface is exceeded, the fibre detaches from the matrix. In longitudinal tension, fibre/matrix interfacial debonding initiates from fibre breaks or initial flaws and will propagate along the fibre. Debonding propagation leads to degradation of composite stiffness and can connect with the neighboring debonds and fiber breaks to form the critical fracture plane, which will lead to the final failure of the laminate. For interfacial analysis, a number of experimental tests involving specimens with a single fiber have been developed, such as single fiber pull-out/push-out, micro-bond test and single fiber fragmentation test. In the fragmentation test, where a single fibre embedded in a matrix is loaded under tension, the mean fibre fragment length and the debonding between the fibre and the matrix are measured as a function of the applied strain and then interfacial properties are back-calculated. Table 1 summarises the ranges for different interfacial properties required for debonding simulations.

Cohesive zone modelling in combination with interfacial friction has been used to simulate debonding process. This study gradually builds up the complexity by starting off from a finite element (FE) model with a single fibre and then evolving towards multi-fibre models in both hexagonal and random fibre packings [1] as shown in Figure 1. The model includes the effect of thermal residual stresses and matrix plasticity. The unconstrained broken fibre is surrounded by neighboring intact fibres all in tension. To assign cohesive properties and friction, a “seam crack” is introduced in between fibre and matrix to separate the overlapping nodes in a single partitioned part. The contact status, at the final state of the applied strain, can be divided into sticking (cohesion and friction are both acting on the interface) and slipping (cohesion is lost, but friction is still active) cases. The slipping length is considered as the debond length and represented by the initial linear region on the stress recovery profile for the broken fibre.

The effect of friction coefficient, interfacial shear strength and fracture toughness on the stress recovery profile in different fibre packings can be then studied. This provides an insight for the parameters that control the interfacial debond length and the ineffective length (the distance from the break where 90% of the nominal fibre stress is recovered).

This study is considered as an initial step in modeling translaminar fracture, since the main damage dissipation mechanisms during a translaminar fracture in unidirectional composites are fibre pull-out, debonding, and, to a lesser degree, tensile fibre fracture and delaminations. The information obtained from the debonding model will be transferred into a longitudinal tensile strength model proposed by Swolfs et al. [2].

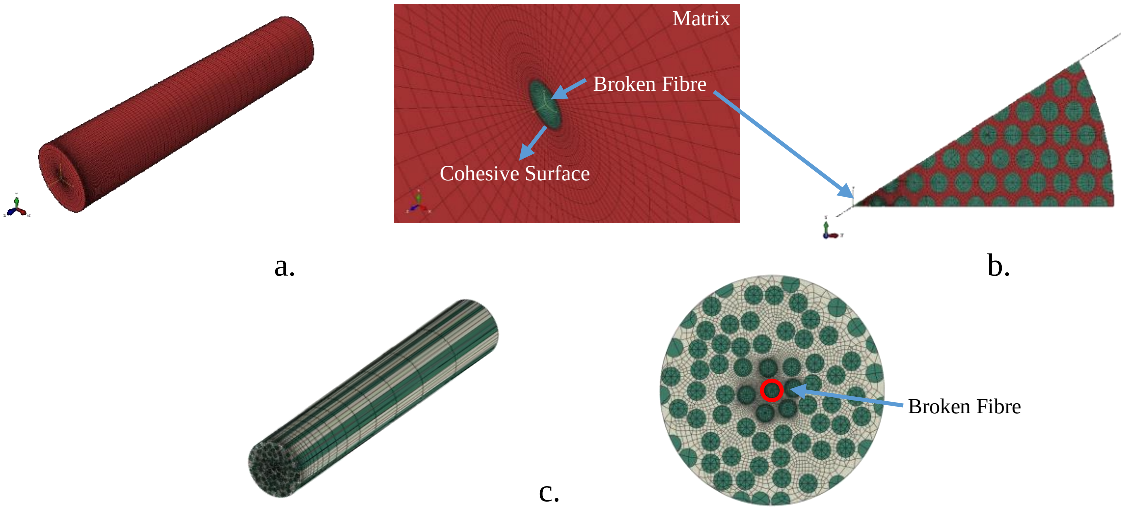

Figure 1: Debonding FE models with a) single broken fibre, b) broken fibre in a hexagonal packing, c) broken fibre in a random packing

Table 1: Range of the interfacial properties in the literature

Required properties	Values [3-11]
Interfacial shear toughness [kJ/m^2]	0.2 – 1.0
Interfacial shear strength [MPa]	6 – 53
Friction coefficient	0.1 –0.4

This project has received funding from the European Union’s Horizon 2020 research and innovation programme under the Marie Sklodowska-Curie grant agreement No 765881.

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FiBreMoD Conference
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Leuven, 11-12 December 2019

A SHEAR TRANSFORMATION ZONE MODEL TO PREDICT THE DEFORMATION AND FAILURE OF GLASSY POLYMERS IN FIBRE REINFORCED COMPOSITES

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Keywords: Shear transformation zone (STZ), constitutive modelling, deformation mechanisms, glassy polymers, confinement effects
Session topics: Constituent properties, Multiscale modelling, Micromechanics

ABSTRACT

Fibre reinforced polymers (FRPs) are attractive material candidates for load bearing structures requiring materials exhibiting high specific stiffness and high specific strength [1]. There is therefore a significant need to predict the deformation (and failure) behaviour of FRPs when subjected to complex load cases and/or a wide range of environmental conditions. Homogenisation schemes may be used to simulate the constitutive behaviour of a single ply or laminate by considering a representative volume element (RVE) of the microstructure. The accuracy of these calculations depends upon the choice of the constitutive laws for the FRP’s constituents and their interfaces. We focus in this talk on the prediction of the deformation and failure of the glassy polymer matrix.

The deformation and failure behaviour of a glassy polymer is complex. When subjected to uniaxial deformation below the glass transition temperature, the measured stress-strain curve[1] of a polymeric glass typically has an initial linear, (visco-)elastic part, followed by yield, strain softening, plastic flow, and strain hardening. Glassy polymers are also known to exhibit strong non-linear behaviour upon unloading. A vast amount of sophisticated (visco)elastic-(visco)plastic models are available to simulate this large deformation response [2-5]. They generally give good fits to measured uniaxial stress-strain curves. However, they are predominantly phenomenological, give limited insight into the micromechanical nature of deformation (and failure), and require the calibration of a large number of fitting parameters. Molecular dynamics (MD) simulations may serve to shed light on the role of molecular deformation mechanisms driving the deformation and failure of glassy polymers. These computations confirm that plasticity occurs through thermally activated molecular rearrangements and conformational changes of a collection of polymer chains as suggested in standard molecular yield models. However, MD simulations are restricted to short time and length scales, limiting their practical use when attempting to predict the response of glassy polymers in bulk or when confined between fibres.

The use of a mesoscale numerical model based on the activation of shear transformation zones (STZs) offers a practical avenue to bridge the state-of-the-art continuum and atomistic simulation approaches and thereby may help to improve the understanding of the interactions between discrete and elementary distortion mechanisms (and their collective organisation) during plastic deformation of a polymeric glass. The STZ framework was originally developed by Argon [6] to simulate the deformation of metallic glasses through thermally activated shear transformations around free volume regions present in the glass. We have used the implementation of Homer and Schuh [7] to develop a mesoscale finite element model for polymeric glasses based on random, heterogeneous activations of

[1] In the absence of failure due to, for example, fast fracture from pre-existing defects or nucleated crazes.

STZs. The model assumes that plastic deformation is governed by both conformational changes of molecular clusters (STZs) and the elastic interaction of these local events with the surrounding matrix. The mesoscale STZ framework only requires the calibration of 7 parameters (of which 5 can be directly measured) and successfully predicts the complex large deformation response of glassy polymers, including post-yield softening and non-linear unloading behaviour, while accounting for the thermally activated nature and the inherently heterogeneous microstructural state of the glass [8,9].

The scope of this talk is as follows. The theoretical framework of the STZ model is presented, and the implementation within a standard finite element solver package (Abaqus) is detailed. In addition, we demonstrate how the STZ model is able to predict the deformation response of a glassy polymer of interest (when calibrated via a limited set of measurements). Emphasis is placed on the prediction of strain localisation (i.e. shear bands), which is typically observed within the glassy matrix of an FRP when deformed in uniaxial compression, see Fig. 1. Future avenues involving mechanical rejuvenation and ageing effects are briefly discussed too.

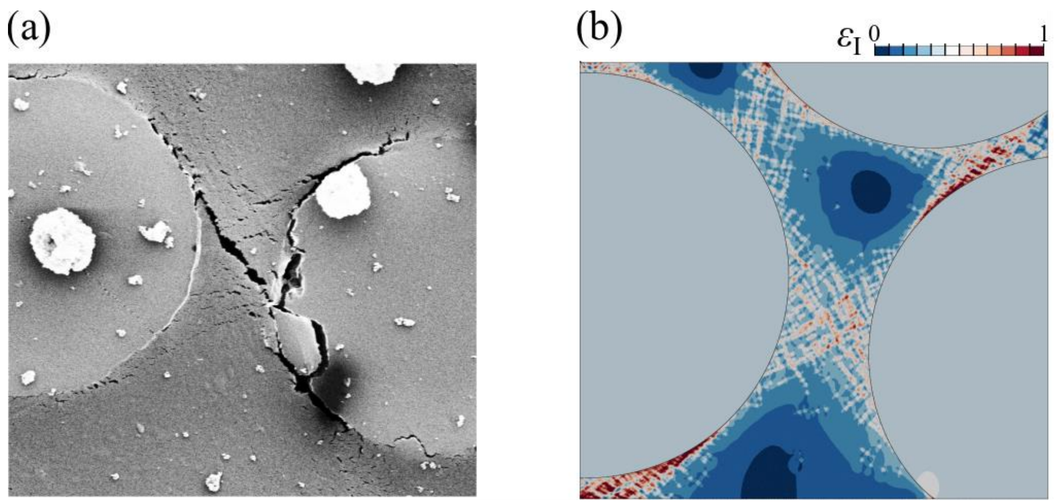

Figure 1: (a) Scanning electron micrographs of a unidirectionial FRP (epoxy-carbon fibre) in uniaxial compression. (b) Computation of the maximum principle strain ε_{I} of a representative volume element of the FRP when loaded in compression via the STZ model; the STZ model is able to predict the observed strain localisation, facilitated by the nucleation and growth of shear bands, in the glassy matrix between the fibres. Figures are adapted from the thesis of Chevalier [9].

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COMPARATIVE STUDY OF VOIDING AND FAILURE MODES IN THERMOPLASTICS AND THERMOSETTING MATRICES IN FIBRE REINFORCED COMPOSITES

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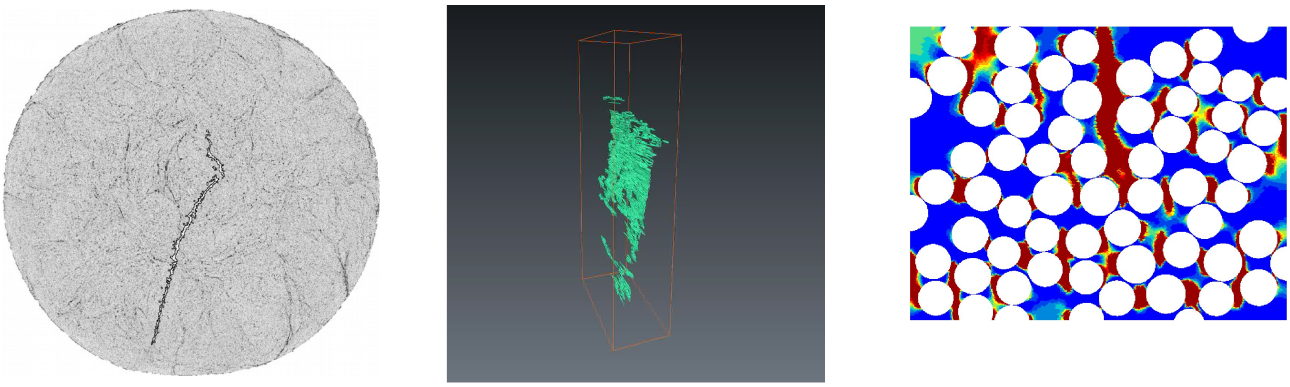
Keywords: Polymer composites, Thermoset, Mechanical properties, Finite elements
Session topics: Constituent properties, Micromechanics, Computed tomography, Multiscale modelling, Notch sensitivity

ABSTRACT

Continuous fibre reinforced composite parts are most often based on thermosetting matrices. Since thermoplastic fibre reinforced composites (TPFRC) have gained growing interest, at least for their recyclability, this study addresses, first, the micro-mechanisms of damage and cracking of these two types of composites.

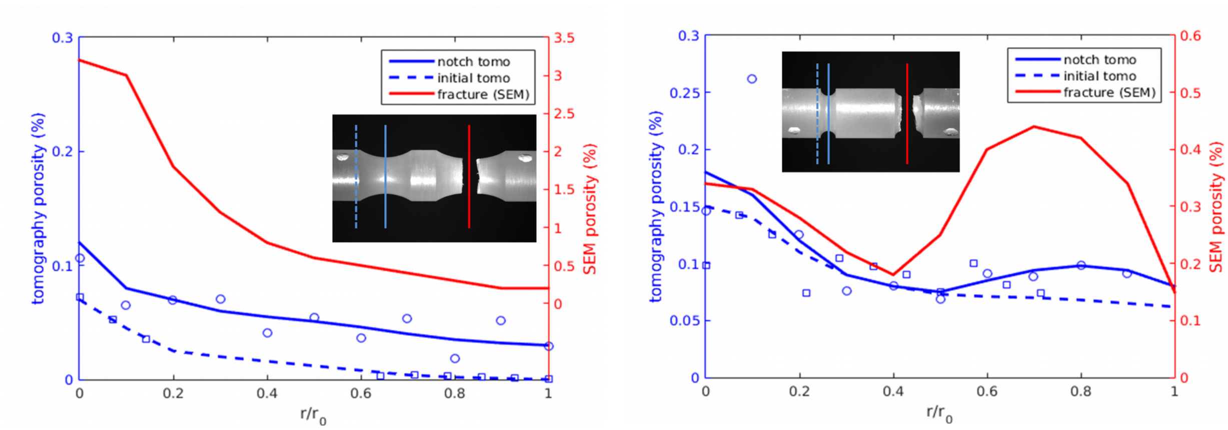


a) b) c)

Figure 1: Multiscale approach to study voiding and failure of a TPFRC: a) SRCT image showing matrix cracking at mid-thickness and mid-height of a rod subjected to a vertical transverse compression ; b) 3D shape of the matrix cracking; c) Contour map of the void volume fraction on a representative volume at the microscopic scale: in red broken elements mimicking matrix cracking.

Figure 1 shows results from a previous study on unidirectionnal rods (8 mm diameter) of Polyamide 6 reinforced by glass fibres [1]. Fig.1a was obtained after synchrotron radiation computed tomography (SRCT) at mid-thickness and at mid-height of a deformed rod. Fig.1b shows a rendering of the crack surfaces in 3D. The contour map of void volume fraction in fig.1c was produced by using FE modelling on a real microstructure where the fibres were meshed. The characterization of the PA6 matrix was performed by highlighting the effects of the cavitation as well as the time dependent deformation. To this end, tensile tests on axi-symmetrically notched round bars with two notch root radii : 4 mm and 0.45 mm respectively for specimens noted as NT4 and NT045. Void nucleation, growth and coalescence were observed using the SRCT technique [2]. Failure of the notched round bars was ductile for NT4 and brittle for NT045 specimens. In addition, the maximum void volume fraction was located at the centre of the net cross section for NT4 and in between the centre and the external notch for NT045. FE simulations, using a poro-elasto-visco-plastic model were able to take this difference in location into account.
Here, an attempt has been made to carry out the same extended characterization of an epoxy resin matrix. It should be noted that epoxy resin is amorphous (no microstructural signature), stiffer (less time dependent deformation) and brittle.

Recent work in the framework of FibreMoD project allowed the following analyses for a dedicated epoxy resin to be carried out [3]. NT6 and NT1 round bars were machined so as to study the effects of the notch root radius. Each round bar was double notched so that at the end of the tensile test, just one notch was broken. The non broken notch experienced the same deformation and damage mechanisms seen by the broken one. Fractography for the broken notch and ex-situ SRCT examinations on the unbroken one allowed the characterization of the void volume fraction as well as their evolution during the overall deformation up to the failure of the specimen. The material coefficients of the poro-elasto-visco-plastic model were then calibrated by combining the load versus notch opening displacement together with the profiles of the void volume fraction distribution through the net cross section.



a) b)

Figure 2: Void volume fraction distribution measured on the epoxy broken specimens: a) for NT6 ; b) for NT1. *Figures taken from [3]*.

Figure 2 illustrates the main results obtained for voiding and failure characteristics of the epoxy under study. Brittle fracture surfaces were systematically obtained where the maximum void volume fraction was located in the centre for NT6 (fig.2a red line) and near the notch root surface for NT1 (fig.2b red line). Additionally, the localization of the defect that triggered the crack initiation was clearly established. Sometimes, the initial porosity was not homogeneous. Indeed, as plotted by the dashed blue lines in fig.2a-b, some specimens exhibited a gradient of the initial void volume fraction, the maximum being located in the central axis of the round bar. The FE simulations with the poro-elasto-visco-plastic model led to a conclusion that the failure criterion is rather based on the maximum principal stress instead of the critical porosity as in thermo-plastics.
Future studies will consist of incorporating the poro-elasto-visco-plastic model in a representative volume element as in fig.1c. This volume will be submitted to the multiaxial stress state previously computed from a thick wall reservoir. The effects of the void volume fraction and the viscosity will then be discussed.

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NATURAL FIBRES AS RAW MATERIALS FOR SUSTAINABLE ALIGNED DISCONTINUOUS FIBRE COMPOSITES

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Keywords: Sustainable composites, Mechanical properties, Aligned discontinuous fibre composites
Session topics: Constituent properties, Micromechanics, Discontinuous fibre composites

ABSTRACT

Owing to public and industrial environmental consciousness, sustainably derived composites are an attractive alternative option for engineering applications [1], compared to composites derived from petroleum distillates. The natural fibres, abundantly available around the world in various types, are the ideal reinforcement constituent for renewable and recyclable composite materials; on one hand they can be sustainably produced, and on the other, they simplify the waste management processes [2].

The High Performance Discontinuous Fibre (HiPerDiF) method, shown in Figure 1, invented at the University of Bristol, is a new and, potentially, high throughput water-based manufacturing process to produce high-performance highly aligned discontinuous fibre composites [3]. Even though natural fibres are typically hydrophilic in nature, it has been shown that flax fibres can be processed *via* the HiPerDiF method, also in combination with reclaimed carbon fibres, to manufacture more sustainable composites with good stiffness and high vibration damping properties [4].

In this work, three different natural fibres: jute (Indian origin), kenaf (Indonesian origin) and curaua (Brazilian origin), are investigated as raw materials for the HiPerDiF method in terms of physical and mechanical properties to evaluate their processability, and the final composite properties.

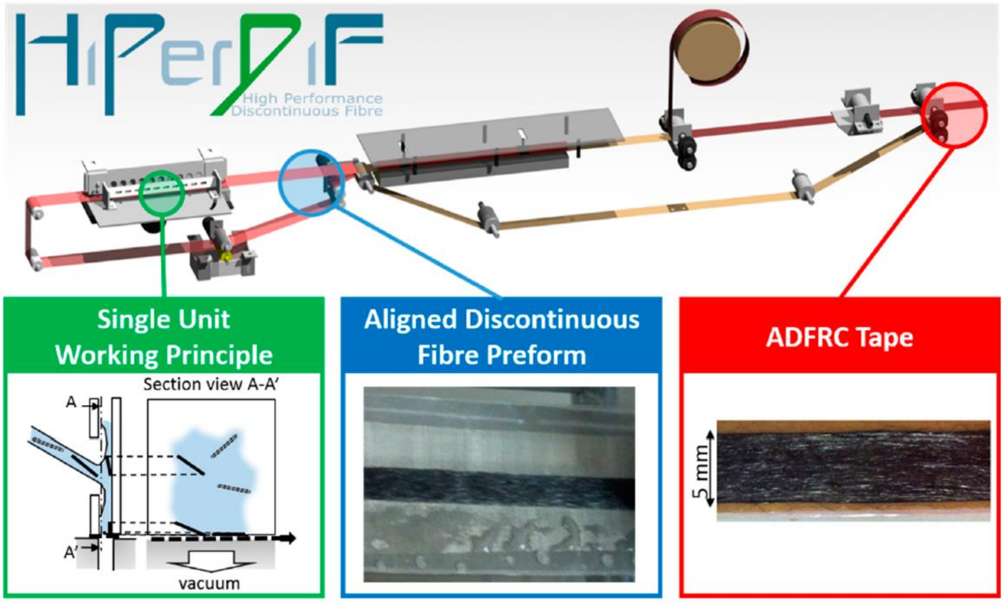

Figure 1: The HiPerDiF fibre alignment machine [3].

The mechanical properties of the fibres, *i.e.* stiffness, strength, and failure strain, were determined by single fibre tensile tests [5]. The interfacial shear strength between the fibres and a commercial diamine-cured difunctional epoxy matrix (PRIME™ 20LV, *ex* Gurit), was determined with an especially adapted micro-bond droplet method [6]. Both tests were performed using a Dia-stron LEX820 Extensometer machine with built-in 20 N load cell and strain measurement system. To ensure that the fibres would not be damaged during the final drying stage of the HiPerDiF process, *i.e.* exposure to a temperature of 90ºC, simultaneous thermal analysis (Netzsch STA 449 F1 Jupiter) was carried out in nitrogen atmosphere to determine thermal stability; it was concluded that the fibres would not be degraded up to 200°C.

Figure 2 shows Young's moduli and interfacial shear strengths of the fibres. It was found that curaua has the highest stiffness (~29.94 GPa) and interfacial shear strength (~13.17 MPa) among the three types. By using the mechanical, physical, and interfacial properties, the critical fibre lengths were also calculated to determine their reinforcement capabilities when used in aligned discontinuous format. In addition, jute has the lowest critical fibre length value (~0.94 mm) among those tested.

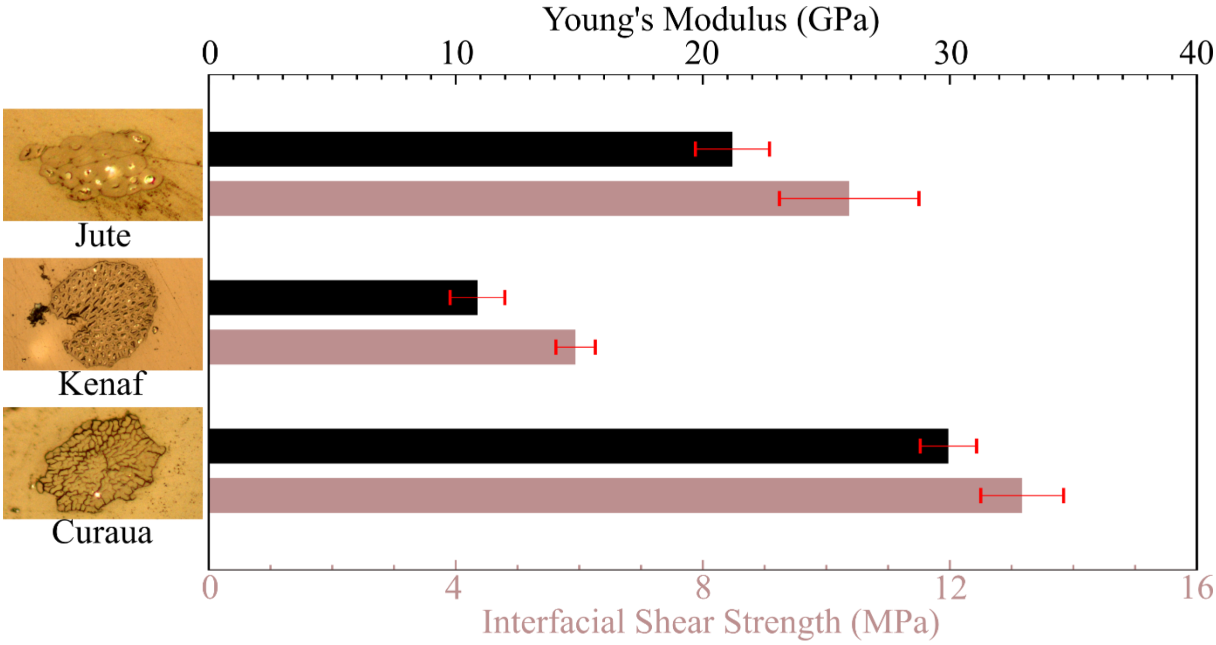

Figure 2: Young's moduli and interfacial shear strengths of the natural fibres (with PRIME™ 20LV).

In conclusion, considering the obtained physical, mechanical, and interfacial properties, jute, kenaf, and curaua fibres are deemed suitable to be processed with the HiPerDiF technology as they will not be degraded during processing. Moreover, it is foreseen that it is possible to obtain more sustainable composites with mechanical properties comparable with those of glass fibre reinforced plastic. The work will be continued by investigating interfacial properties between the natural fibres and matrices such as thermoplastics and covalently adaptable networks. Furthermore, the composites manufactured with the HiPerDiF technology will be characterised.

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FUNCTIONAL CELLULOSE FIBERS FOR SMART COMPOSITES

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Keywords: Functional fiber composites, Continuous cellulose fibers, Mechanical properties, sensors
Session topics: Constituent properties, Novel experimental techniques, Structural applications

ABSTRACT

Functional composites are the new trend in the composite science and technology especially for bio-based composites. Providing electrical or/and thermal conductivities besides the improved mechanical properties attained by incorporating fibers into polymer matrix makes composites available for wider range of applications with enhanced sustainability. Engineering solutions are made to achieve this functionality such as conductive coating or incorporation of conductive particles in the polymer matrix. In both cases, the prediction of the mechanical properties of the resulting composite requires modifications of the theories and models applied for the conventional composites.

In a previous work by authors, regenerated cellulose fibers (RCFs) were coated with copper in an industrial-friendly way using electroless copper plating [1]. The plated copper resulted in a significant improvement in electrical conductivity while the mechanical properties of the fibers were substantially degraded by the chemicals used during the coating process. Despite the fact that strength of the fibers is drastically decreased, they still provide stiffness reinforcement (partially due to the layer of stiff metal deposited over the fibers). These copper-coated fibers in the form of bundles can be incorporated into composite structures and act as sensors to monitor strains and damage accumulation during the service life of composite.

In this study, a bundle of the coated fibers was embedded in a slab of epoxy matrix with wires attached at the ends connected to a 4-probe setup for electrical resistance measurements. The sample was loaded in tension in a loading rate of 1%/min to allow in-situ measurements of the change of electrical resistance which is shown in Figure 1a. It can be seen that electrical resistance is increased gradually as the sample is being strained. The first part of the curve is linear where the bundle is being stretched in its elastic region and the average gauge factor for the resistance change (the slope of the line from 0 to 1.8% strain) is found to be 0.056 ± 0.021 Ω/Ω/%. At around 1.8 % of the strain, the linearity is lost, and this can be attributed to the damage in the copper layer as breakage occurs randomly within the bundle. From the Figure 1b, it can be seen that the copper presence is limited to the outer fibers in the bundle. Analysis of micrographs showed that the average thickness of copper measured at different points around the bundle is 3.72 ± 1.11 µm after 90 minutes of copper plating. Variation in coating thickness is inevitable along the fiber or between the different fibers within the bundle, thus, it is

necessary to find a reliable approach that describes the constituents' fractions within the composite. The coated bundle has been described by the model shown in Figure 1c. The known diameter and number of filaments in the bundle is used to calculate the volume fraction of fibers while the copper is assumed to be a continuous layer forming a hollow cylinder surrounding the bundle, having the measured thickness. Applying the rule of mixtures (ROM) approach on the single bundle composite, approximation gives a stiffness of the bundle within 10% error margin compared to the actual stiffness measured for non-impregnated bundle.
Effect of bundles incorporation on the mechanical properties of the conventional unidirectional composites of glass fiber (GF) or carbon fiber (CF) in epoxy matrix (EP) is estimated by the use of ROM. The ratio of coated bundles to conventional fibers was varied at a fixed total fiber volume fraction of 60%. Up to a ratio of 10%, the overall stiffness does not degrade by more than 8% for both composites. Such effect has been reported in literature before by the use of optics fibers with CF/EP systems [2] but quite the opposite with the GF /EP composite [3].

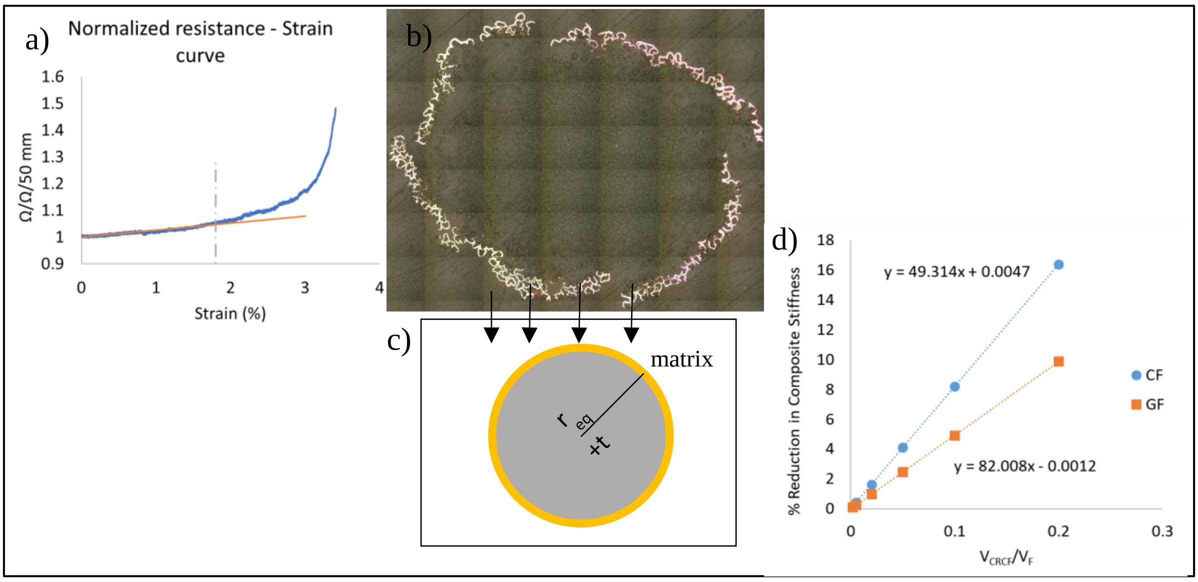

Figure 1: a) change of electrical resistance with respect to strain for a single copper coated bundle composite; b) micrograph of the cross-section of coated bundle; c) representative model of the composite in b; d) effect of the bundle volume fraction on the total longitudinal stiffness of conventional composites of GF/EP and CF/EP estimated by RoM.

From the results presented above, it can be concluded that these bundles can work properly as sensors for both glass fiber/epoxy composites as well as carbon fiber/epoxy composites since their effect on the composite stiffness is limited and their linearity limit for resistance change fall below the strain of failure for these composites. Compared to the use of optics fiber, the coated bundles are smaller in diameter which reduces the resin rich pockets around the bundles when positioned perpendicular to the fiber orientation. In addition, these are cheaper materials and ductile in nature and can achieve purpose with lower volume fractions than the optics fibers.

The Authors would like to acknowledge the financial support from the Swedish Foundation for Higher Education and Research and the excellence and innovation area of Smart Machine and Materials at Luleå University of Technology.

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RELATIONSHIPS BETWEEN PROCESSING PARAMETERS, MECHANICAL AND CHEMICAL PROPERTIES OF THICK GLASS FIBRE REINFORCED THERMOPLASTIC METHACRYLIC COMPOSITES

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Keywords: Polymer composites, Mechanical properties, Physico-chemical properties, Porosity
Session topics: Constituent properties, Thin ply composites, Micromechanics, Computed tomography, Structural applications

ABSTRACT

As the energy transition unfolds, the recycling targets applied to composite materials are on the rise, thus getting more and more challenging to meet. In this context, continuous fibre reinforced thermoplastic composites and their recyclable matrix have gained increasing interest over the past twenty years – especially for the design of lightweight and high-performance structural parts. In order to produce such materials, specific monomers are usually vacuum-infused through glass or carbon fabric before undergoing in-situ polymerization. While most parts obtained this way are only a few millimetres thick, some industrially important applications require the manufacturing of much thicker components – up to several centimetres.

The present work focuses on the links between the infusion parameters, physico-chemical state of the matrix and mechanical properties of 6-cm-thick glass fibre reinforced thermoplastic methacrylic composite plates, at both the micro- and macroscopic scales. More precisely, different composite plates were infused with Elium® resin and left to polymerize at different temperatures, while recording the temperature profiles at several locations. The evolution of the microstructure was then studied along the thickness of each sample, and emphasis was placed on the characterization of the porosity distribution and morphology by optical microscopy and computed microtomography.

Heating up the bottom part of the plates after infusion - in order to trigger the polymerization reaction - can in turn lead to monomer boiling and thus favour the formation of porosity inside the matrix, with more and larger pores observed near the surface. While it seems obvious that such cavities affect the overall mechanical response of composite parts, the decrease of modulus and strength as a function of the volume fraction of pores proved much more significant than first expected, at least under uniaxial compression.

In order to get a better understanding of the phenomena, the local mechanical response of the matrix was then measured further to various infusion conditions by carrying out nano-indentation tests inside the matrix pockets, while chromatographic analyses gave access to the molecular weight distributions and monomer conversions achieved in the Elium® matrix. The results revealed small but mutually consistent differences in matrix properties across the thickness of the plates. Though the latter were slight enough to have a limited influence on the chemical and micromechanical properties of the matrix, they give important insights into the complexity of the thermal and chemical phenomena occurring in the system during polymerization and cooling.

In a nutshell, this work suggests that changing the infusion temperature strongly impacts the amount and distribution of porosity in Elium® composites. This governs in turn the macroscopic mechanical properties of the final part, and implies significant variations in the course of the polymerization reaction along the thickness of the plate.

The authors would like to thank Arkema for funding this research, the Elium® team at Groupement de Recherche de Lacq for their guidance and support, and Dr Jeroen Soete at KULeuven for his help with CT measurements.

VARIABILITY OF STRAND-BASED SHEET MOULDING COMPOUNDS AS FAILURE DETERMINANT

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Keywords: Sheet Moulding Compounds, Discontinuous Reinforcements, Digital Image Correlation, Mechanical properties

Session topics: Discontinuous fibre composites

ABSTRACT

Carbon Fibre Sheet Moulding Compounds (CF-SMCs) are strand based composites that have recently spread in the automotive industry. CF-SMCs prepregs are sheets made of 25 mm long carbon fibre strands randomly dispersed in a thermoset resin. The sheets are compression moulded in the final desired shape. They allow fast manufacturing and good specific mechanical properties.

The inherently stochastic microstructure of SMC makes it a highly variable material. Using Digital Image Correlation (DIC) during tensile tests of SMC reveals a very high heterogeneity[1–3]: strain peaks can deviate significantly from the global one, which is often quite low at failure. Potential site for strain peaks are tows transversely oriented with the load direction and tow boundaries. The majority of the studies showing such characterisation are mainly focused on low in-mould flow, with mainly uniform random orientation.

Only recently the authors showed the effects of a high in-mould flow on the mechanical properties of CF-SMC [4]. During compression moulding, according to the amount of flow inside the mould, tows tend to re-orient and distort significantly. A prevailing tow orientation was achieved, and specimens with different prevailing orientation were obtained. Those showed an anisotropic material response both in tension and compression: the specimens with tows mainly aligned along the load direction (0° case) being stronger and stiffer than those with a perpendicular orientation (90° case). Failure mechanisms were also highly dependent on the orientation state, including fibre-dominated failure for the 0° case and matrix-dominated failure for the 90° case.

This work uses the same oriented specimens to focus on the effects of variability on the mechanical response of CF-SMC. DIC showed how local high strain regions lead to failure at low values of global strain (see Fig. 1). The effects are less evident in compression, due to the shorter gauge length. This explains the brittleness of CF-SMCs: for the tensile case, failure occurred at a global failure strain smaller than 1%, sometimes not exceeding 0.5%. Randomly oriented CF-SMC (from the literature) are also considered, showing that this heterogeneity is also failure determinant for the low in-mould flow case.

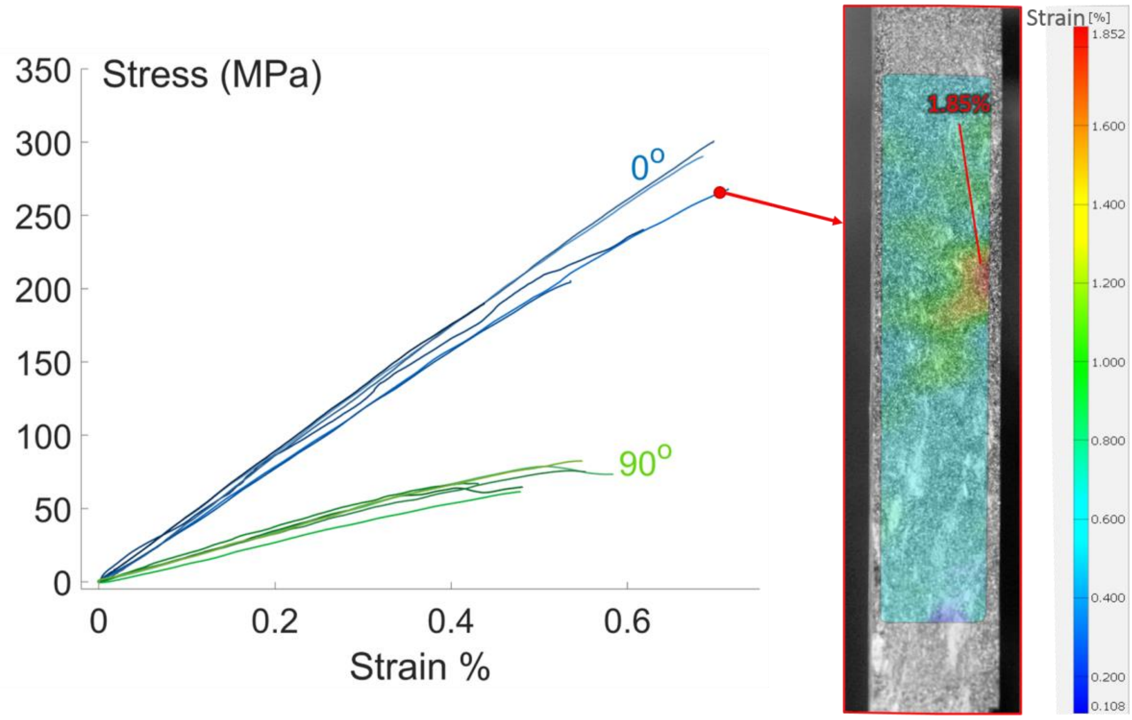

Fig. 1: Tensile stress-strain curves, with an example of a pre-failure state of a 0° specimen.

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ACKNOWLEDGMENTS

The research leading to these results has been performed within the framework of the FiBreMoD project and has received funding from the European Union’s Horizon 2020 research and innovation programme under the Marie Sklodowska-Curie grant agreement No 722626.
YS acknowledges FWO Flanders for his postdoctoral fellowship. SVL holds the Toray Chair at KU Leuven, the support of which is acknowledged.
The authors also acknowledge Mitsubishi Chemical Carbon Fiber and Composites GmbH for providing material and related information.

RECYCLED CARBON FIBRE REINFORCED PET AS FEED FOR ADDITIVE MANUFACTURING FOR OPTIMAL COMPOSITE PERFORMANCE

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Keywords: Fused deposition modelling, Polymer composites, Mechanical properties

Session topics: Fibre-hybrid composites, Discontinuous fibre composites, Computed tomography

ABSTRACT

Introduction

In modern life, polymers are indispensable because of their range of applications in the fields of packaging, construction and safety. The latter has prominent consequences for nature and therefore, recycling of plastics has gained increasing attention[1]. The current project therefore focusses on the development of more reliable composite structures with enhanced mechanical properties, starting from recycled PET (rPET) and recycled carbon fibres as fillers. The composites were extruded to obtain carbon fibre-reinforced (CFR) rPET filaments, which were subsequently used as feed for fused deposition modelling (FDM). Herein, carbon fibres with different lengths were applied (i.e. 0.1 mm, 6 and 12 mm).

Results and discussion

After optimization of the extrusion parameters with a DSM vertical mini-extruder, rather flexible filaments containing 1 wt. % CFR rPET were obtained. With the Ultimaker 3, the printing parameters were optimized and two sets of parameters were obtained (see Table 1) which resulted in structures being more amorphous and flexible versus more semi-crystalline and brittle. The latter observation was further investigated by comparing the crystallinity degree of the printed structures, which was calculated via the heat capacity of the cold crystallization peak in dynamic scanning calorimetry (DSC).

Table 1: Printing parameters for rPET with Ultimaker 3.

Printing Parameter	Brittle	Flexible
Nozzle temperature	275°C	265°C
Build plate temperature	80°C	30°C
Printing speed	40 mm/s	40 mm/s
Cooling	Off	On

Extensive characterization of the filaments and printed structures was performed using thermal gravimetric analysis (TGA), gel permeation chromatography (GPC) and DSC. The mechanical properties were evaluated via compression tests, which indicated that the Young's modulus increased when carbon fibres were incorporated and therefore a stiffer material was obtained. Via optical microscopy, a first indication about the fibre alignment could be obtained. Despite the carbon fibres being cut into smaller pieces during the blending and extrusion, the fibre alignment during printing still became apparent. Further in-depth evaluation of the correlation between fibre alignment and the associated printing parameters is currently ongoing using micro-computed tomography (µ-CT) and will be presented at the meeting.

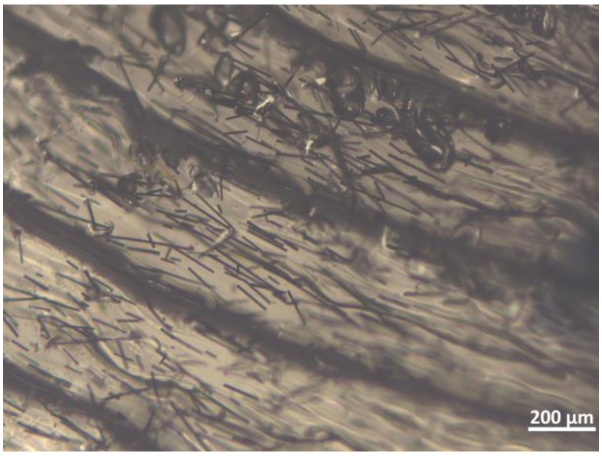

Figure 1: Optical microscopy image of the first layer of printed rPET with 1 wt. % 6 mm CF.

As a result, the parameters for CFR rPET filaments were optimized at lab scale. Therefore, the next aim was to switch to filaments processed on a more industrial scale. This was done at Centexbel by using a compounder for optimal mixing and the monofilament line to obtain filaments with a controlled diameter. To obtain CFR rPET filaments with various compositions, pellets containing 1 to 10 wt% CF in rPET were processed. Filaments with two diameter size were extruded, namely 1.75 mm and 2.85 mm, respectively for the Prusa i3 MK3S and the Ultimaker 3. Hereafter, a full characterization will be performed and during printing, the printing parameters will be varied (printing direction, build plate temperature, nozzle temperature, etc.). A full mechanical testing will be performed including interlayer and intralayer cohesion tests (DCB mode I and tensile testing) and furthermore, the fracture mechanics will be investigated with compact tension and single-end-notch bending tests (SENB).

Conclusion

In the current project, 3D printing with rPET reinforced with recycled carbon fibres was performed. For this purpose, an extensive optimization study was done for the extrusion of the filaments. Two sets of parameters were obtained for the 3D printing. Incorporation of carbon fibres yielded stiffer materials with a clear fibre alignment. Moreover, the mechanical properties were significantly improved.

The authors acknowledge the support of the Agency for Innovation and Entrepreneurship Flanders via the Strategic Initiative Materials in Flanders in the framework of the SIM-SBO RELFICOM project. Tridea is acknowledged for providing the recycled PET pellets.

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TWO DIFFERENT INTERLEAVING APPROACHES TO MITIGATE IMPACT DAMAGE PROPAGATION ON CARBON/EPOXY LAMINATES

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Keywords: Carbon Fibre Reinforced Polymers, Advanced Composites, Thin Veils, Low Velocity Impact, Damage Tolerance

Session topics: Impact resistance

ABSTRACT

The possibility to optimize laminate characteristics according the specific application, turn polymeric advanced composite materials very attractive to preform in high demanding environments in alternative to the traditional ones. In the last decades, advanced industries applications, such as aeronautic, aerospace, sports and defence, have been taking advantage of the high mechanical properties and low-density combination that these materials present to enhance better performances [1]–[3]. However, some issues, associated to their intrinsic nature, as high brittleness and layer-by-layer architecture, turn them quite vulnerable to solicitations dominated by shear and dynamic stresses [4]. One of the most common and dangerous examples of these solicitations are the low velocity impact events, under these conditions, some barely visible internal damages may be caused that can propagate into the interlaminar region of the composite part, compromising its mechanical performance along its service lifetime [5].

In an attempt to mitigate these problems some strategies have been already explored, among them, the reinforcement of the interlaminar resin rich region seems to be a promising approach to reduce damage propagation without seriously compromise in-plane properties. Some works may be found in literature reporting an improvement on interlaminar fracture toughness in mode I and II [6, 7] through the placement of thin veils between layers. However, only a few works may be found exploring the potential of these interlaminar structures to improve composites damage tolerance and resistance when subjected to low velocity impact solicitations [8]–[10].

Despite the recognized potential of these interleaving approach, the inclusion of interlaminar toughening thin veils necessarily means an increment on composite thickness. In order to minimize this problem, in this work, we propose a strategical interleaving of a carbon fibre/epoxy laminate typically used on aircraft components to reduce thickness overage. Based on the idea that an impact event is meanly dominated by shear and bending stresses, using the Classical Lamination Theory (CLT), a preliminary theoretical study on through-the-thickness stresses distribution of the laminate under a bending moment was first carried out to define where should be better interleave thin veils. A schematic representation of this selection process may be observed in Figure 1.

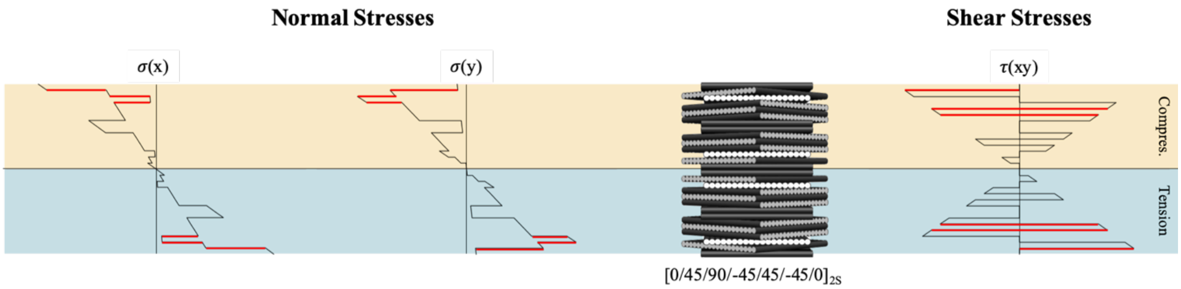

Figure 1 – Schematic representation of the interleaved interfaces selection process.

Thereafter, the six interfaces which presented larger discrepancies on normal and shear stress values were identified and both were interleaved by four different material thin veils (glass, carbon, aramid and polyester). Then, all the new interleaved laminates produced by vacuum bag infusion, were characterized and compared to a non-interleaved one, produced in the same conditions, according their thickness increment and the void presence was evaluated under the scanning electron microscopes (SEM). Moreover, three-point-bending, interlaminar shear strength (ILSS) and low velocity impact (LVI) experimental tests were also preformed and their mechanical properties were compared. LVI tests were performed at three different impact energy levels (13.5, 25 and 40 (J)) and the damage areas were evaluated by the meaner of ultrasonic visualizations.

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FULLY-UNCOUPLED MULTI-DIRECTIONAL DELAMINATION SPECIMENS: A PRELIMINARY VALIDATION

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Keywords: Delamination, Multidirectional laminates, Double Cantilever Beam test
Session topics: Delamination, Fracture toughness, Novel experimental techniques

ABSTRACT

Thanks to their attractive specific mechanical properties, structural composite materials are particularly suited for the design of high-efficiency structures. Hence, many industrial sectors (e.g.: automotive and aerospace) have largely exploited such materials, even for safety-critical applications. On the other hand, composites show highly complex mechanical behaviour and damaging processes, due to both anisotropy and heterogeneity intervening at different scales.

In long-fibre composite laminates, in particular, interlaminar fracture (delamination) is one of the most critical damaging modes: it is often difficult to detect and it strongly affects mechanical performances, especially in compression, which is already a weak point of these structures. A vast literature has been devoted to the study of delamination. Today, standard practices exist to perform delamination tests and thus to characterize the interlaminar behaviour of composite laminates in terms of critical Energy Release Rate (ERR) [1]. Unfortunately, standards are restricted to unidirectional (UD) specimens, where all the layers are oriented with fibres aligned to the longitudinal direction of the specimen (0°).

ERR of an interface, however, might depend on the orientations of layers embedding it [2]. Most importantly, real structures are built using multidirectional (MD) laminates, where delamination may appear and grow at any interface. Several researchers dealt with delamination in MD specimens, but the scientific community has not reached consensus on best practices to test delamination in MD laminates [2]. In particular, depending on their stacking sequences, MD laminates may show a complex thermo-mechanical behaviour, that can influence test conditions, thus invalidating results. Indeed, a generic MD laminate may induce mixed-mode conditions at delamination front even when performing a (globally) pure mode test.

To address this problem, Fully-Uncoupled Multi-Directional (FUMD) delamination specimens have been proposed [3], exploiting superposition criteria for quasi-trivial stacking sequences [4, 5]. Such innovative specimens allow to test any type of delamination interface, while having a thermo-mechanical response which is identical to that of a UD specimen. This prevents undesired mechanical behaviours and cure-induced residual stresses. Therefore, the use of FUMD specimens lets standard tests, currently available for UD interfaces, be extended to interfaces between differently oriented plies. FUMD specimens may therefore represent ideal candidates for a standard devoted at evaluating ERR of such interfaces.

While analytical and numerical proofs of the effectiveness of FUMD specimens were given in [3], experimental investigation is still ongoing. The aim of this communication is therefore twofold:

1. Briefly review and explain the principles underlining FUMD delamination specimens and their design;
2. Present the results of a first proof-of-concept experimental activity that has been carried out.

As it will be shown, FUMD specimens allow to investigate many aspects of delamination in MD laminates. Therefore it appears of paramount importance to get the scientific community involved in the development and exploitation of this concept.

The first author is grateful to region Nouvelle-Aquitaine for supporting this research activity through the SMARTCOMPOSITE project.

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

MANUFACTURING AND CHARACTERIZATION OF BIOINSPIRED CFRP

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Keywords: Polymer composites, Digital Image Correlation, Mechanical properties, Finite elements
Session topics: Delamination, Discontinuous fibre composites, Fracture toughness, Bio-inspired design, (Pseudo-)ductile composites

ABSTRACT

Bio-inspiration has already demonstrated its value as a viable approach for designing new materials. Specifically, the inherent nacre's remarkably high toughness despite being mostly (95%) composed by brittle calcite nanoplatelets has been focus of study. The outstanding mechanical properties of this biomaterial are associated to the layered arrangement of the crystals and proteins into a "bricks and mortar" structure [1].

This strategy has already been applied in the carbon fibre reinforced polymer materials (CFRP) field showing promising results [2,3]. In addition, great advances have been done in the recent years in the understanding and analysis of discontinuous CFRPs [3-5] what have opened a wide range of possibilities for developing of new materials and structures different to the conventional CFRPs.

In this study, this hierarchical structure has been implemented to unidirectional CFRPs made of AS4/8552 prepregs through an automated process, with an automated tape lay-up machine (ATL). Different materials with "brick lengths" between 5 and 50 mm have been manufactured successfully and cured in autoclave.

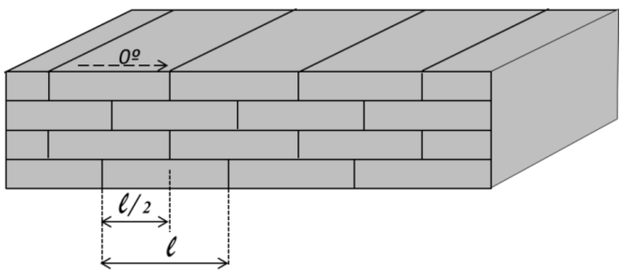

Figure 1: A) Brick and mortar structure.

The fracture toughness of these materials has been characterized by tensile, Mode I, Mode II and Compact Tension tests.

The results show an enhancement in the fracture toughness and significant changes in the fracture modes. Delamination has been avoided for Mode I and Mode II tests, and Compact Tension Test fracture mode present signals of pull out.

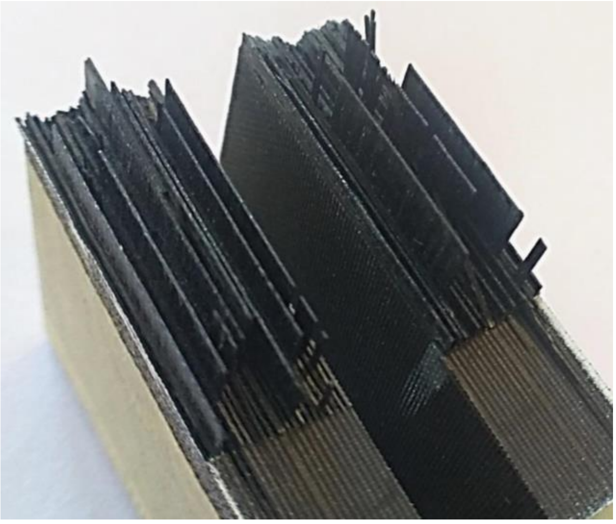

Figure 2: Failure mode obatined in Compact Tension test.

Successful manufacturing of “brick and mortar” laminates with various have been achieved. This study scales up the manufacturing of these structures with an automated process. In addition, fracture toughness tests have proven that these hierarchical structures prompt a tortuous path effect. Considering this, we aim to bring new possibilities of energy dissipation and crack deflection mechanisms of bioinspired structures within the composite industry.

We would like to thank J. L. Jiménez, M. de la Cruz and J. Martin Muñoz for their assistance with the sample preparation. We would also like to thanks C. Cenamor for his help with the execution of the tensile tests.

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HYBRID EFFECT IN CRITICAL ENERGY RELEASE RATES IN ALL-CARBON INTERLAYER UNIDIRECTIONAL FIBRE-HYBRIDS

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Keywords: hybrid laminates; all-carbon hybrids; fracture toughness; hybrid effect
Session topics: Fibre-hybrid composites

ABSTRACT

Critical energy release rates (CERR) in mode I and mode II for carbon-carbon unidirectional hybrid composites are reported in Table 1. The composites are reinforced with ultrahigh-modulus and high-strength carbon fibres. The CERR is measured as well for the reference non-hybrid composites to enable measuring hybrid effects. Standardised testing techniques are employed. The apparent fracture toughness is considered a material characteristic, and hence should be applicable in practice for laminate thicknesses of the same range as studied here. Important applications are the definition of cohesive laws or the creation of damage mode maps with the purpose of designing pseudo-ductile interlayer hybrids.

The “hybrid effect”, which is the difference between the CERR value of the hybrid interface and the average of the values for the reference composites, has been estimated. For mode I CERR, the hybrid effect is negative, corresponding to the lowest of the values for the two components (namely laminate with ultrahigh-modulus fibres). For mode II CERR, the hybrid effect is strongly positive. The CERR value for the hybrid is close to the sum of the values for the components. The mechanism of the high hybrid effect in mode II is only partially explained by SEM fractography and will therefore be the subject of future work.

Table 1 CERR values

Sample type	*G_{Ic}, J/m^2, initiation average and std. dev.*	*G_{Ic}, J/m^2, propagation average and std. dev.*	*G_{IIc}, J/m^2 average and std. dev.*
T/T	235 ± 23	718 ± 83	934 ± 77
D/D	105 ± 10	121 ± 15	469 ± 34
T/D	106 ± 6	127 ± 18	1376 ± 234

A MICROMECHANICAL PROGRESSIVE FAILURE MODEL FOR PREDICTING THE TENSILE FAILURE AND DAMAGE DEVELOPMENT IN HYBRID UNIDIRECTIONAL COMPOSITE MATERIALS

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Keywords: Polymer composites, Fibre-hybrid composites, Micro-mechanics, Modelling, Fragmentation
Session topics: Fibre-hybrid composites, Micromechanics, (Pseudo-)ductile composites

ABSTRACT

Fibre Reinforced Polymers (FRP) are widely employed to design lightweight structures mainly thanks to their excellent specific strength and stiffness. However, their inherent quasi-brittle behaviour and low toughness leads to fibre tensile failure with nearly any damage symptom [1]. Consequently, composite materials are often overdesigned, limiting the costs and weight savings. Fibre hybridization is a potential solution to overcome the low toughness of FRP. In a hybrid composite, a Low Elongation (LE) fibre is mixed with a High Elongation (HE) fibre in a single matrix. Thanks to this mixture, the failure of the LE fibre in the hybrid can be delayed compared to that of the baseline non-hybrid composite, resulting in hybrid effects, and an increase of ductility [2, 3]. Currently, this phenomenon is principally attributed to changes in the failure development compared with non-hybrid composites.

Nowadays, the longitudinal tensile failure process of hybrid composites is not yet entirely understood [2]. Derived from this insufficient knowledge, there is a lack of tools to predict their failure. At present, there are some models in the literature that attempt to simulate the failure and damage development in hybrid composites. Nevertheless, the majority of the models assume simple load redistributions around fibre breaks and consider that the fibres are distributed in a regular square or hexagonal packing, instead of being randomly allocated [4]. Therefore, the development of more advanced models is necessary. In this work, an efficient model for predicting the tensile failure and damage development in hybrid unidirectional FRP under fibre tensile loading is presented.

The micromechanical model developed, here called Progressive Failure Model (PFM) [5, 6, 7] consists in a Representative Volume Element (RVE) of width a, height b and length L, which contains parallel fibres randomly distributed. The fibres may be all of the same type, leading to a non-hybrid material, or may be of different types (such as glass and carbon) leading to a hybrid composite. The fibres are split into elements of length l along their longitudinal direction, thus leading to a domain of parallel springs divided into planes in series. The fibres are denoted with subindices $q \in [1, \dots, N_q]$, whereas the planes are referred with $p \in [1, \dots, N_p]$, where N_q and N_p are the number of fibres and planes respectively. A schematic representation of the model is shown in Figure 1.

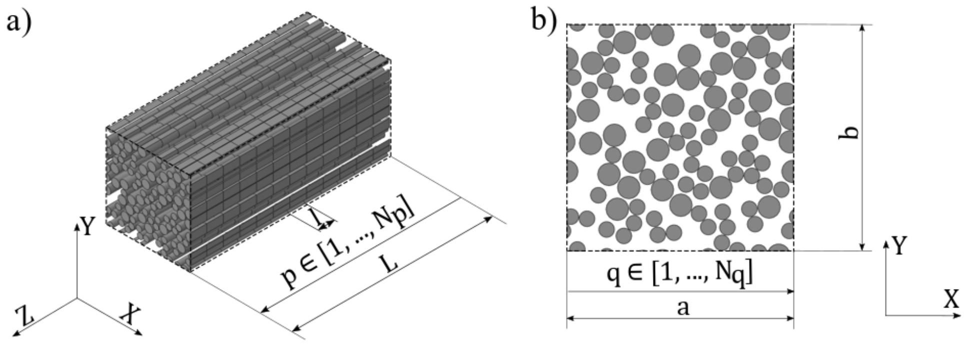

Figure 1: Schema of the RVE considered by the PFM. a) 3D view, b) cross-section view.

At the beginning of each simulation, a strength is generated and assigned for each fibre element in the RVE following a statistical distribution describing the strength scatter of the fibres. After that, an uniaxial strain controlled simulation is started until the failure of the composite. At each loading step, the stress of each element p, q is calculated with

$$\sigma_{p,q} = \frac{SCF_{p,q}}{\Omega_p} E_q \left(1 - D_{p,q}\right)\left(\varepsilon_p + \varepsilon_q^r\right) \tag{1}$$

where $SCF_{p,q}$ is the Stress Concentration Factor (SCF) of element p, q, E_q is the Young's modulus of fibre q, $D_{p,q}$ is a damage factor which is equal to 0 for intact elements, equal to 1 for broken elements and in between for elements in any stress recovery, ε_p is the mechanical strain of the plane, ε_q^r is the thermal residual strain of fibre q and Ω_p is a stress ratio to enforce load equilibrium at each plane [5, 7].

When an element fails, a damage factor is distributed along the ineffective length of the broken fibre, whereas stress concentration is applied onto the intact fibre elements surrounding the broken fibre. This load redistribution is computed with analytical equations that are entered into the PFM. These equations take into account matrix yielding, the cluster size, RVE size, volume fractions, fibre radius and elastic properties of each fibre population in the hybrid material [7]. Dynamic effects may also be considered by temporarily increasing the SCF with a magnification factor when new elements fail [6]. Fibre-matrix debonding is omitted. To calculate Ω_p and ε_p, a load equilibrium condition is satisfied, taking into account the damage factors and failed elements along the RVE. This approach allows to capture a different deformation along the model, since ε_p is larger for planes with a higher damage [5].

The PFM captures individual fibre breaks and the formation of clusters of multiple broken fibres. Due to the inclusion of the damage factor, the stiffness loss of composite materials is modelled, allowing to predict pseudo-ductility in fibre hybrid composites [5, 6, 7]. Moreover, since the model is semi-analytical, the PFM is computationally more efficient than other similar models [7].

The authors acknowledge the financial support of the grant RTI2018-097880-B-I00 from the Spanish Ministerio de Ciencia, Innovación y Universidades.

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FIBRE HYBRID COMPOSITES CONSISTING OF DISCONTINUOUS WASTE CARBON FIBRE AND CONTINUOUS GLASS FILAMENTS WITH IMPROVED IMPACT STRENGTH

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Keywords: Waste carbon fibre, Fibre hybrid composites, Mechanical properties
Session topics: Fibre-hybrid composites, Discontinuous fibre composites

ABSTRACT

To achieve tailored composite properties precisely matching the needs of the structure under consideration, fibre hybrid composites (i.e. a composite in which at least two types of fibres are used to reinforce a common matrix) have recently received great attention by researchers. The methods employed for the combination of constituent materials can be generally categorized into three groups: (i) interlaminated hybrids by stacking two or more layers of constituents, (ii) intraply or yarn-by-yarn hybrid by combining two or more constituent types of continuous filament yarns within one layer through co-weaving, and (iii) intermingled hybrids by randomly and intensively mixing two fibre types (intrayarn or fibre-by-fibre) in the same ply.

The literature review shows that the majority of reported investigations address fibre hybrid composites based on interlaminated or intraplay hybrids consisting of continuous filament yarns [1]. In contrast, experimental studies on discontinuous fibre hybrid composites are sparse. Nevertheless the concept of fibre hybridization would be beneficial to the enhancement of mechanical as well as functional properties of composites made from waste carbon fibre (CF). With the increased use of CF in different areas such as in the aerospace, automotive, sports and wind energy sectors, waste CF is increasingly available from difference sources, e.g. production offcuts, selvedges, and bobbin ends. CF is also being reclaimed from end-of-life composite components using different techniques (e.g. pyrolysis and solvolysis). Due to the strict control for composite disposal and limitations regarding the landfilling of CF materials (EU Directive 1999/31/EC), the re-use of waste/reclaimed CF is becoming a key concern.

Within our previous studies, the development of core-sheath hybrid yarns and the tensile properties of composites produced from these hybrid yarns consisting of waste CF and polyamide 6 (PA 6) are described [2]. In [2], it is shown that composites manufactured based on such hybrid yarns consisting of waste CF and PA 6 of 60 mm length possess approximately 85% of the tensile strength of composites consisting of virgin filament tow. These composites based on waste CF offer an enormous potential for application in load-bearing structures compared to the achievable composite strengths of injection-moulded components and nonwovens. High fibre orientation and compaction are responsible for the high tensile strength of composites manufactured from hybrid yarns consisting of waste CF. However, the impact properties of composites manufactured from waste CF of defined lengths are still not satisfactory.

Therefore, to improve the impact properties of composites manufactured from waste CF without hampering their high tensile properties, a novel approach of producing fibre hybrid composites based on multi-material hybrid yarns is reported in this paper. The multi-material hybrid yarn is developed from waste CF (obtained from bobbin ends) mixed with PA 6 fibre (fibre length 60 mm) by combining with continuous glass filament (GF) yarn in a core-sheath structure on a DREF-3000 friction spinning machine (Figure 1). The tensile and Charpy impact properties of the unidirectional hybrid composites (Figure 2 (left)) produced from the developed multi-material hybrid yarns are compared with those of a non-hybrid composite reinforced exclusively with waste CF.

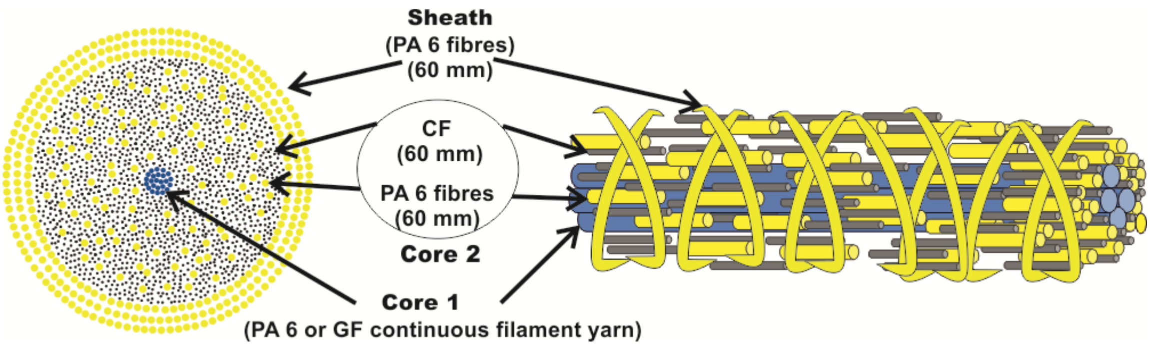

Figure 1: Sketch of cross-sectional (left) and longitudinal view (right) of the developed hybrid yarns

The results of Charpy impact testing suggest that the impact strength of waste CF reinforced composites can be increased by 18.4% by including 6 volume% of GF filament yarns in the waste CF composite structure, while keeping the total fibre volume content of the hybrid composite identical with that of the CF reinforced composite. Examples of microscopic images showing the fracture zone of the composites after Charpy impact testing are presented in Figure 2 (right). The elongated GF filament yarns in the fracture zone indicate that the hybrid composites absorb energy even after the waste CF is broken. The ultimate failure of the hybrid composite occurs only after the failure of GF filament yarns. In this case, the reduction in tensile strength and E-modulus of the hybrid composite is statistically insignificant. Moreover, an increase in approximately 50.5% in impact strength in hybrid composites can be achieved by including only 6 volume percentage of GF filament yarns additionally to waste CF. Therefore, the impact properties of UD composites are found to be significantly affected by hybridisation due to the incorporation of GF filament yarn in the hybrid yarn core.

A promising feature of this new type of hybrid composite is that its impact properties can be improved significantly based on a minimum compromise in terms of the tensile properties. Other advantages include the ease of manufacturing of these hybrid structures and their suitability for the mass production of thermoplastic composites based on waste/recycled CF.

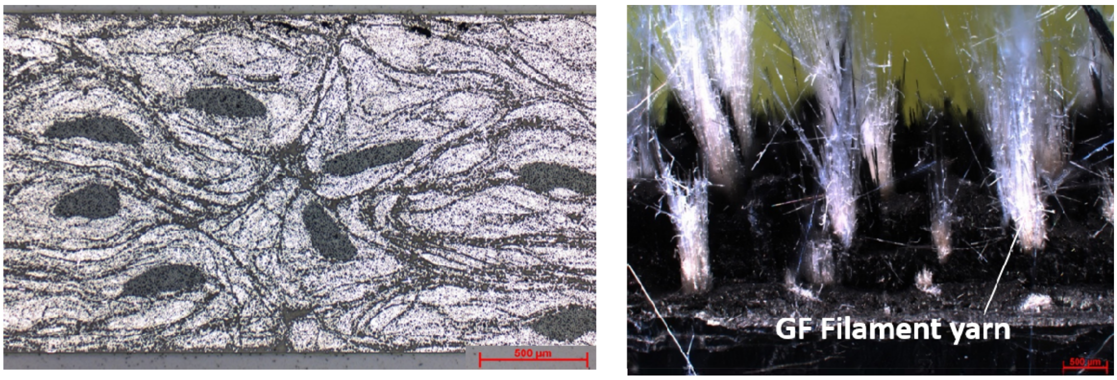

Figure 2: Cross-sectional image of hybrid composite (left) and the fracture zone of hybrid composite after Charpy impact testing (right) [1]

This article presents parts of the results from the research program ZIM ZF4008318BL7 of Forschungskuratorium Textil e.V. supported by the AiF by the Federal Ministry for Economic Affairs and Energy at the Technische Universität Dresden, Germany.

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MATERIAL ENGINEERING FOR COMPOSITES ACROSS SCALES AND PHYSICS

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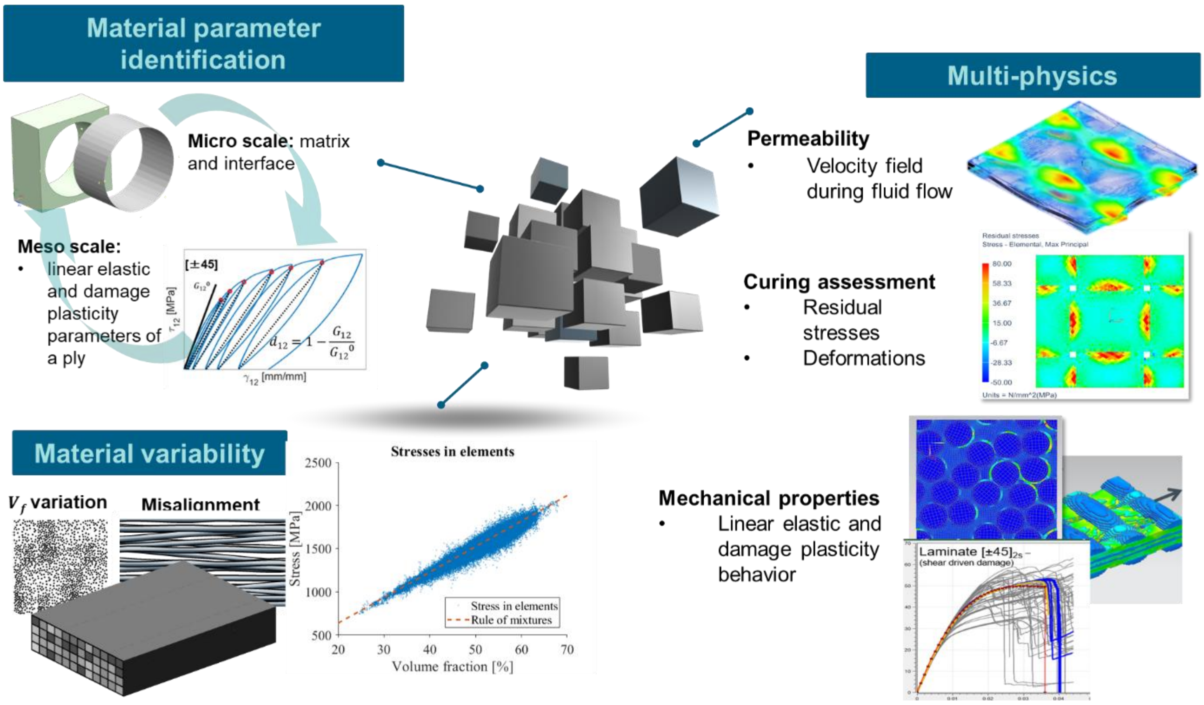
Keywords: Composites, Multiscale, Parameter identification, Variability, Permeability, Curing
Session topics: Multiscale modelling

ABSTRACT

Material engineering for composites is a complex process covering different stages from application-driven design and manufacturing method selection to composite production and performance assessment via iterative optimization loops. Large and expensive test campaign is needed before the final design decision is made. With the Simcenter 3D (Siemens PLM Software) Virtual Material Characterization (VMC), the material engineering process becomes more efficient and can be completed with fewer tests. This contribution demonstrates different stages of a VMC process across different scales and physics as depicted in the Figure below.



'The material parameter identification' case study focuses on the material parameter identification at the micro (fiber scale) and mesoscales (ply/yarn scale) of the carbon fiber unidirectional (UD) ply. First, fiber/matrix interface properties and damage-plasticity behavior of the matrix at the microscale are reverse-engineered from the tests on coupons. Static tensile test on a $[\pm 45]_{2s}$ laminate is used as a reference curve to fit matrix and interfacial behavior in shear, as well as mode II fracture toughness. To determine interfacial normal strength and mode I fracture toughness and to further refine properties of the matrix, specimens with [90] layup should be considered as well (not covered in this work). To identify damage-plasticity parameters at the mesoscale needed for a damage model developed in LMT-Cachan (Ladevèze et al., 1992), loading-unloading-reloading virtual tests are performed on $[\pm 45]_{2s}$ and $[\pm 67.5]_{2s}$ laminates.

‘The multi-physics’ cases are linked to the manufacturing simulations. The effect of curing on the development of residual stresses and deformations is studied in thermosetting composite materials on the microscale. In order to gain insight into the infusion process, saturated permeability is computed for woven composite at the mesoscale and first steps towards unsaturated permeability assessment on the microscale are discussed.

‘The material variability’ case focuses on the effect of the material variability by the example of a UD composite. The longitudinal tensile strength of a UD ply is predicted, considering the stochasticity of the fiber strength and microstructure (misalignment and fiber volume fraction variability).

The demonstrated Simcenter 3D VMC ToolKit, with a link to permeability calculations, curing simulations, assessment of the variability and virtual material parameter identification across different material scales, provides an industrial solution for the efficient linking of the manufacturing-induced material features to the materials’ performance.

Acknowledgement:
The authors gratefully acknowledge SIM (Strategic Initiative Materials in Flanders) and VLAIO (the Agency Flanders Innovation & Entrepreneurship) for their support of the ICON project M3Strength (Grant no.140158), which is part of the research program MacroModelMat (M3), coordinated by Siemens Industry Software (SISW). O. Shishkina thanks VLAIO for financing her work in the framework of the Innovation Mandate Project “MicroCT-based Model Generation Engine for Virtual Material Characterisation” (Grant no. HBC.2017.0189). F. Malgioglio thanks the EU Horizon 2020 research and innovation programme under the Marie Skłodowska-Curie grant agreement No 722626 for funding his work through the FiBreMoD project and his supervisors Prof. S.V. Lomov, Dr. Y. Swolfs, Prof. W. Desmet (KU Leuven, Belgium). The Mechanics of Materials and Structures research group (Ghent University, Belgium) coordinated by Prof. W. Van Paepegem is acknowledged for the support. We thank H. Egami (University of Tokyo, Japan) for his contribution to the research on unsaturated permeability during his internship at SISW through the program “Vulcanus in Europe” by EU-Japan Centre for Industrial Cooperation. We also thank G. Romano (University of Calabria, Italy) for performing study on curing in the scope of the Master thesis internship at SISW.

DISTINCT ELEMENT METHOD (DEM) FOR FIBROUS COMPOSITES: TOWARD COMPUTATIONAL GUIDED MANUFACTURING

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Keywords: Polymer composites, Mechanical properties, Distinct Element Method
Session topics: Multiscale modelling

ABSTRACT

We are exploring the new application of DEM to fibrous composites (FCs). In general, manufacturing of FCs (such as carbon fibre reinforced polymers, carbon nanotube polymer composites, textiles made from active fibers) is subject to substantial challenges, which are mainly caused by the discontinuities of the fibres. In this respect, it is inspiring to know that the most performant structures in Nature exhibit discontinuous architectures at various length of scale. These architectures are able to combine the benefits of strength and stiffness with outstanding toughness and damage tolerance. For FCs, the optimal discontinuous fibre architecture are not known, but can be proposed by advanced DEM computations.

DEM for FCs, or mesoscopic DEM [1], has the potential to advance the manufacturing of FC structures. Since an exhaustive exploration of the large parameter space (fiber length, fiber diameter, filler material etc.) by direct manufacturing of FCs is impractical, the guidance offered by mesoscopic DEM is critical. Here we address simulations for guiding the manufacturing of ultra-strong carbon nanotube (CNT) composites [2]. During synthesis, the parameters (diameter, length, number of inner walls) of the individual CNTs can be adjusted, but the consequences on the materials' properties are not known a priori. mesoscopic DEM simulations are used to predict the variations in mechanical response caused by these parameters, and to guide the synthesis conditions toward the optimal parameters of the CNTs.

This talk will focus on the complex stretching process of CNT mats that involves subtle microstructural evolution, including de-bundling, waviness, zipping, etc. The factors that most decisively impacts the transition from the rather stochastic network structure to the yarn structure are not well understood. With the help of mesoscopic DEM, we simulate the stretching of various networks consisting of several hundred CNTs. Aiming at engineering highly aligned CNT yarns with reduced porosity, we explore by mesoscale simulations the impact of CNT length, friction, entanglement and bundling, and identify the significant contributing factors to the yarn formation.

We then stretched the network up to 150% strain through displacement-controlled tensile test. The stretched network undergoes significant restructuring, with the formation of a central yarn comprising aligned CNTs, Figure 1a. To understand this remarkable process, Figure 1b shows the calculated stress-strain curve. In correspondence to the structure evolution, we have identified four distinct regimes: (i) elastic, for $\varepsilon < 3\%$, (ii) softening, for $3\% < \varepsilon < 30\%$, (iii) stiffening, for $30\% < \varepsilon < 60\%$, and (iv) softening and failure, for $\varepsilon > 60\%$.

The elastic regime gives a Young's Modulus of 8.6 GPa. With the load transfer just begins to develop, zipping relaxation activation is limited. As stretching progresses (regime ii), the network softens significantly. The behavior can be seen in the slope changes presented by the stress-strain and

stretching energy curves in Fig. 1b. At the microstructure level, the local bundling and the pore sizes are increasing. Pores undergo squashing, especially in the central region of the network, under the transverse contraction caused by the Poisson's effect and the zipping relaxation directed by the strain direction. The hardening (regime iii) features energetic elasticity of wavy CNTs, visible in the evolution of the stretching energy, Figure 1b. Energetic elasticity is enabled by the increase in entanglement under pore squashing. The yarn formation also begins in this regime as alignment becomes significant. At regime iv, entangled CNTs can no longer sustain the load transfer, leading to its final softening and failure. As the stretched CNTs begin to straighten their severely bent portions located around larger pores, the bending energy drops. Above 75% strain, the dominance of elastic deformation is substituted by inter-tube CNT sliding corresponding to the thinning of the bundle and increase in packing density. The final stage shows organization into three main aligned bundles, each containing about 30 CNTs, separated by nm size pores.

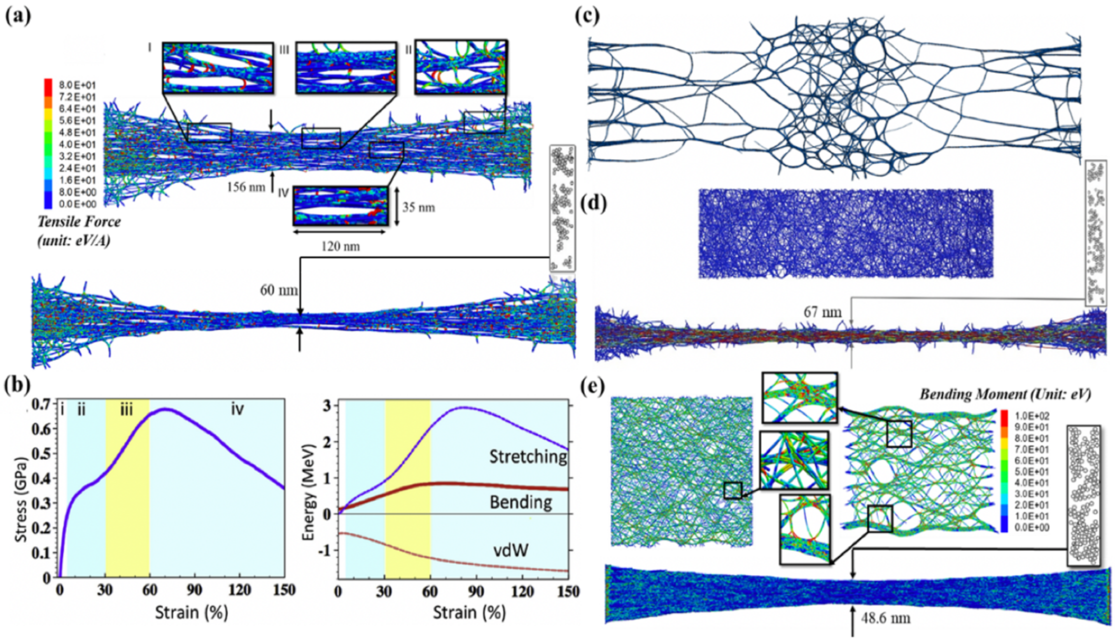

Figure 1. (a) CNT network stretched to 75% and 150% strain showing the closure of pores and packing of the yarn cross-section. (b) Strain-stress relation and energy evolution during the tensile test of (a). (c) Stretching of CNT network without nanofriction shows big pore opening. (d) A CNT ribbon stretched to 90% strain. (e) Anisotropic networks relaxed with/without friction (left/right). The network with friction is stretched to 75% strain showing a better packing at yarn's cross-section.

We conclude by noting that the application of DEM to CNTs is an important development because it fills a gap in the available simulation methods for FCs. Indeed, finite element modeling (FEM) methods can successfully analyze the behavior at continuum scale. However, extending FEM down to the "fibre" scale is challenging due to the complex topologies, inhomogeneities, and the dynamical inter-tube sliding. Nevertheless, as the current work shows, accounting for this mesoscale is critical. The current simulations are at the basis of a "Fiber Flow" toolbox of *PFC3D* of Itasca, Minneapolis.

This work was supported by NASA's Space Technology Research Grant NNX16AE03G, by the Institute for Ultra-Strong Composites by Computational Design, Grant NNX17AJ32G, and by University of Minnesota's MnDRIVE RSAM program.

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WHAT HOLDS CELLULOSE NANO-FIBRILS TOGETHER?

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Keywords: Molecular Dynamics simulations, Cellulose nano-fibrils, water, molecular interactions
Session topics: Fibre-hybrid composites, Multiscale modelling

ABSTRACT

Paper is produced by compressing cellulose pulps in a moist condition, followed by a drying cycle. Hirn and Schennach, in a recent work [1], have experimentally showed that in contrast to a general belief, it is the van der Waals interactions, and not the hydrogen bonding, which contribute more to the total energy dissipated in breaking fibre-fibre bonds. Nevertheless, the aforementioned interactions can be affected by the presence or absence of water. Water can cause an enormous swelling in cellulosic fibres. This implies that mechanical properties of fibres can be controlled by adjusting the moisture content. Yet, it is not still clear how this process operates at the nanoscale. In other words, a better understanding of the role of nano-confined water layer on the surface of the fibrils is required.

An atomistic model of cellulose nano-fibrils (CNF) is modelled in GROMACS, using GLYCAM06 parameter set [2]. Each fibril has a degree of polymerization of 30 (DP 30), containing a 36-chain Iβ cellulose stacking. The prediction of the elastic modulus of the fibrils is used as a parameter to validate the proposed model. A configuration is generated containing two fibrils placed at a distance lower than that of the van der Waals cut-off (see Fig.1). The system is then solvated in water, equilibrated at 300 K and atmospheric pressure.

Bonded fibrils are separated by applying a force on the centre of gravity (COG) of each. The change in the van der Waals (vdW) and Coulombic interactions, and consequently the contribution of the hydrogen bonding energies are calculated. The procedure is done both in vacuum and in presence of water.

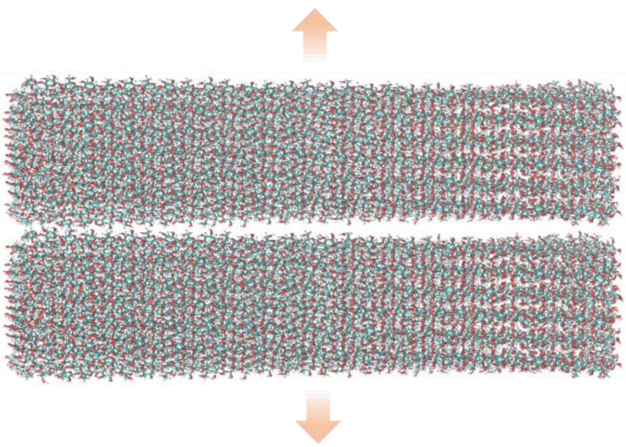

Figure 1: Two CNFs of DP 30 at a distance lower than the vdW cut-off radius. A force is applied on the COG of each fibril to detach them.

Results of the study show that the behaviour of the material is governed by presence of water. While in vacuum (no water in the simulation box) interactions are dominated by the hydrogen bonding between the fibrils, water alters the governing interactions by building a confined layer on the surface of the CNFs, leading to loss of a fraction of the hydrogen bonds. Presence of the water layer in between the fibrils however increases the inter-fibrillar distance, decreasing the magnitude of the vdW energies. Yet, the contribution of these forces are more than twice the hydrogen bond energies in the early stages of the separation (see Fig.2). The results of this study is particularly of a great import in defining the mechanical properties of papers and relevant studies.

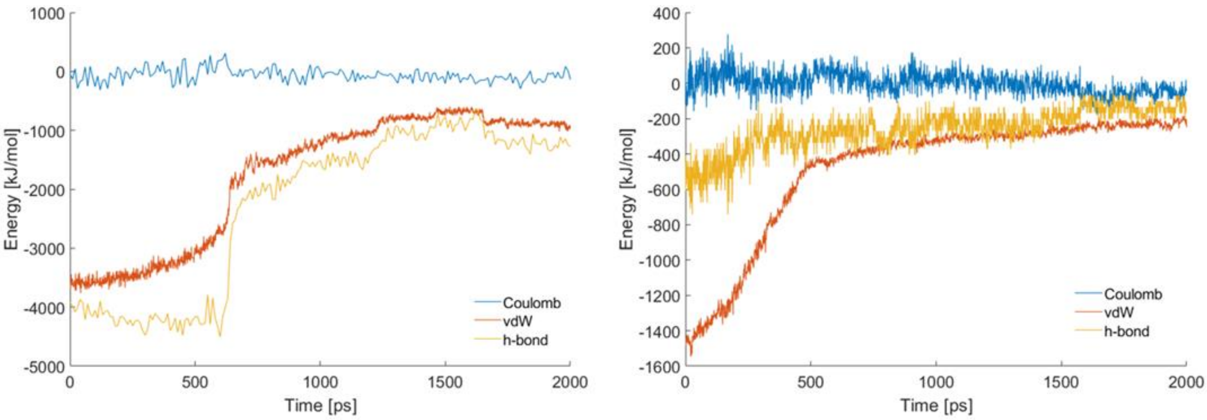

Figure 2: Interaction energies of the fibrils during separation in vacuum state (left) and in solvated state (right).

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

A MACHINE LEARNING INSPIRED APPROACH TO PREDICT THE LONGITUDINAL TENSILE FAILURE IN UNIDIRECTIONAL COMPOSITES

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Keywords: Polymer composites, Digital Image Correlation, Mechanical properties, Finite elements
Session topics: Multiscale modelling, Structural applications

ABSTRACT

The virtual characterization of composite materials properties is of great industrial interest, enabling considerable time and money savings. This is desirable in the design stage, where a large number of test repetitions would be required to avoid the use of large safety margins (and the related manufacturing costs).

To accurately predict the mechanical properties and the variability of the mechanical response, the material variability should be included. A strength model for longitudinal tension of unidirectional composites was developed in [1]. The model consisted in a finite element model of a ply. The model includes the stochasticity of misalignment and fibre volume fraction by assigning different homogenized properties and local material orientation to each finite element. Local strength in each element is sampled from a strength probability distribution calculated with a fibre break model [2]. A large number of realisations of the virtual coupons are generated and loaded in longitudinal tension. The outcome of the Monte Carlo simulation is the distribution of longitudinal properties.

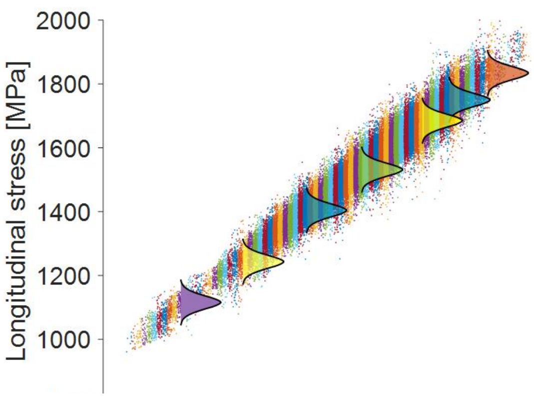

Figure 1 Longitudinal stress response of individual elements versus the volume fraction.

In this work, we focus on making this virtual testing framework computationally efficient, by means of a Machine Learning inspired strategy. The strength model is used to generate data related to the local material response of each finite element. Such material response is very complex due to the stochastic nature of the problem and can be described by probability distributions. The training process is performed on a few virtual coupons. The statistical parameters describing the material response are related to the local volume fraction with a linear regression strategy (Figure 1). When the parameters are identified, it is possible to generate a statistically equivalent stress response for new virtual coupons without solving new finite element models by sampling the statistical distributions identified in the training. The size of the training set needed is defined with an online convergence criterion for the regression parameters. When the variation of regression parameters is smaller than a user defined threshold, the training is stopped. For a threshold of less than 1%, the training set size is smaller than 5% of the total number of simulations required (500).

This strategy proves to be very effective in reducing the computational time (Figure 2), while returning the same results of the full FE based model (Figure 3).

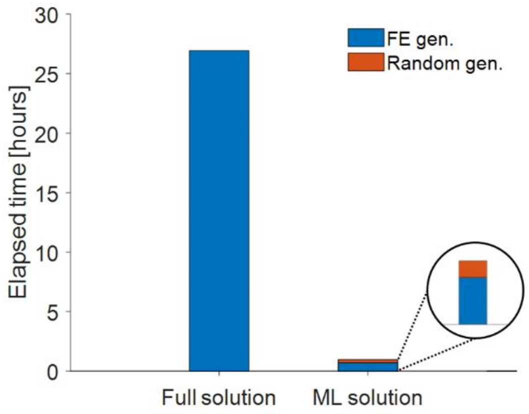

Figure 2 Comparison of computational time between the full FE based solution and the Machine Learning one.

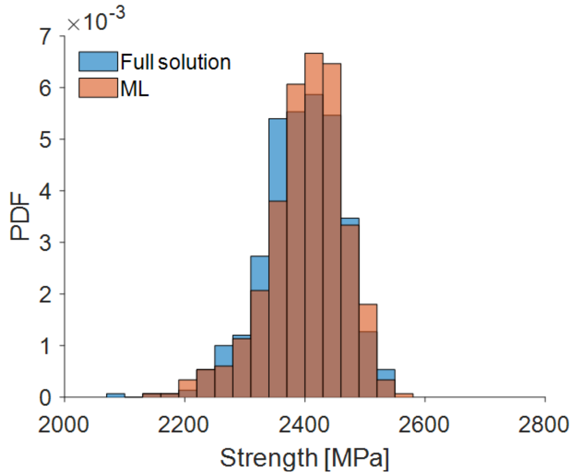

Figure 3 Comparison of strength predictions with the full FE based solution and the Machine Learning solution.

ACKNOWLEDGEMENTS

The research leading to these results has been done in the framework of the FiBreMoD project and has received funding from the European Union’s Horizon 2020 research and innovation programme under the Marie Skłodowska-Curie grant agreement No 722626. S. V. Lomov holds Toray Chair for Composite Materials at KU Leuven, support of which is acknowledged.

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EFFECTS OF DYNAMIC FAILURE AND LOCAL PEAK STRESS IN UD COMPOSITES

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Keywords: Fibre reinforced polymer, Stress concentration factor, Finite elements, Dynamic failure
Session topics: Micromechanics, Delamination, Fracture toughness, Structural applications

ABSTRACT

This study proposes a 3D finite element modelling approach to predict the phenomena associated with longitudinal tensile deformation and failure events in detail for unidirectional (UD) composites. The approach relies on a periodic micromechanical Representative Volume Element (RVE) with a random distribution of fibres, to capture the interaction between the constituents and the progression of damage. The carbon fibre reinforced plastic (CFRP) AS4/8552 was chosen for the purpose of demonstration of the methodology [1, 2].

Besides the detailed simulation of longitudinal fracture mechanisms, this approach allows the study of important effects on longitudinal failure, such as dynamic failure and matrix damaged-plasticity. Moreover, it allows the determination of important parameters such as the Stress Concentration Factor (SCF) caused by the failure of single fibres, which has significant effect on the failure probability of adjacent fibres, and the critical fibre cluster. In the end, the effects of local peak stress on the starts of damage and fracture is studied and compared with average stress.

In most studies, SCFs are computed by averaging the stress over the fibre cross-section in spite of the fact that the stress will decay from the crack tip of the broken fibre, likely according to $1/d^{\frac{1}{2}}$, d is radius of fiber [1]. In this case, the zone of the fiber close to the broken fibre would be more loaded than the opposite zone, Figure 1. The hypothesis behind averaging the stress over the fiber cross section is that the fibre behaves as a ductile material rather than as a brittle material. This may seem reasonable in view of the small diameter of the fibre, which points to the fact that any failure process zone (FPZ) would be larger than the fibre. The size of the FPZ ahead the notch tip l can be estimated according to the well-known Irwin relation as,

$$l = \frac{1}{\pi}\left(\frac{K_{IC}}{\sigma_0}\right)^2 \tag{1}$$

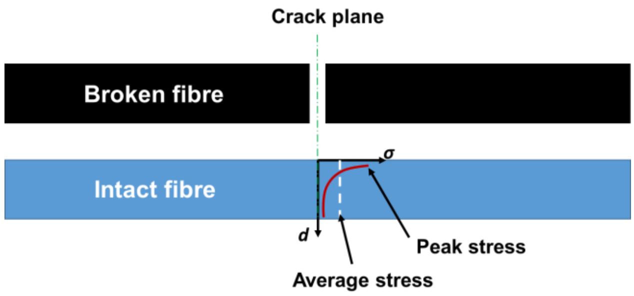

Figure 1: Definition of local peak stress and average stress.

where K_{IC} stands for the fracture toughness and σ_0 is the average tensile strength of the Weibull distribution. If Irwin length be taken as a rough estimation of the FPZ, it can be concluded that it might be developed inside the fiber diameter.

This study achieved the following results:

- The dynamic failure has a significant effect on the SCF, matrix damaged-plasticity, ineffective and debonding length, therefore can reproduce more realistic behavior of the failure process.
- The interface debonding has a negative impact on the SCF. That means, with increasing interface debonding the SCF reduces and vice versa.
- Local stress (peak stress) acts as the main role on the damage and crack onset.

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FIBRE-RELATED DAMAGE IN UNIDIRECTIONAL COMPOSITES UNDER CYCLIC TENSION LOADINGS

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Keywords: Fatigue, damage evolution, fibre failure
Session topics: Fatigue, Micromechanics

ABSTRACT

The fatigue behaviour of a multidirectional laminate made of unidirectional (UD) plies is characterised by a progressive damage evolution involving three main interacting mechanisms:

i) Multiple off-axis crack initiation and propagation;
ii) Delamination initiation and propagation;
iii) Fibre failure.

Off-axis cracking is the first observable phenomenon, occurring since the earlier stages of the fatigue life, triggering the initiation and propagation of delamination from their tips at the plies' interfaces. These two mechanisms do not cause directly the laminate failure. They are, however, responsible for the degradation of the apparent laminate stiffness. In addition, cracks and delamination act as stress concentrators for the 0° plies, promoting the failure of the fibres and, eventually, of the laminate. The prediction of the onset and evolution of the first two mechanisms was recently addressed by the authors [1, 2].

This work focused, instead, on the experimental observation of the damage evolution in unidirectional glass/epoxy plies under longitudinal cyclic loads. This is a fundamental step for understanding and modelling the underlying mechanisms at the micro-scale and developing a tool for predicting the fatigue life of a laminate.

To this aim, two testing campaigns were carried out, on unidirectional $[0]_8$ infused and $[0/50_2/0/-50_2]_s$ autoclave-moulded glass/epoxy laminates.

$[0]_8$ specimens were tested with a load ratio R=0.05 and a maximum cyclic stress of 200, 300, 320 and 340 MPa. The specimens, polished on the front surface, were periodically removed from the testing machine and scanned under an optical microscope. Isolated fibre breaks were observed after the first cycle. Then, two scenarios were observed:

- for load levels of 320 and 340 MPa, longitudinal debonds started to grow, promoting new fibre failures, fragmentation and clustering (see figure 1a);
- for load levels of 200 and 300 MPa, matrix yielding was sometimes observed at the fibre crack front, but no debond propagation and clustering were seen (figure 1b).

It was thus possible to conclude that:

- The fibre failure process starts at the first cycle with randomly distributed fibre breaks;
- For high enough load levels, damage evolves spatially through the propagation of debonds and clustering;
- For load levels below a certain threshold (here around 300 MPa), damage does not evolve, recalling the ideas already proposed by Talreja and Gamstedt [3] on the existence of a "region III" in the fatigue life diagram, associated to the fatigue limit in the longitudinal direction.

The detailed results of this campaign are reported in Ref. [4].

$[0/50_2/0/-50_2]_s$ laminates were tested with a load ratio R=0.05 under maximum stress values of 135, 195 and 245 MPa.

As expected, off axis cracks were the first visible damage mechanism. Increasing the number of cycles, the crack density evolution reached a saturation condition and some clusters of broken fibres became visible on the external 0° plies. These clusters propagated along the transverse direction and, when a certain width was reached, caused the initiation and stable propagation of splits (figure 2a).

As shown in figure 2b, the formation of a macro-scale cluster occurs through the connection of different micro-scale clusters of few fibres on different planes. This connection resulted from the propagation of debonds and matrix cracks (as on UD samples). In addition, the cluster originated in a region immediately above the underlying off-axis crack, as it causes a stress concentration on the 0° ply. However, the process of the macro-scale cluster formation did not occur above the off-axis crack only, but spanned a wider region covering one or two crack spacings.
These damage observations will be extremely useful in the definition of a damage-based framework to predict the final failure of a multidirectional laminae under cyclic loadings.

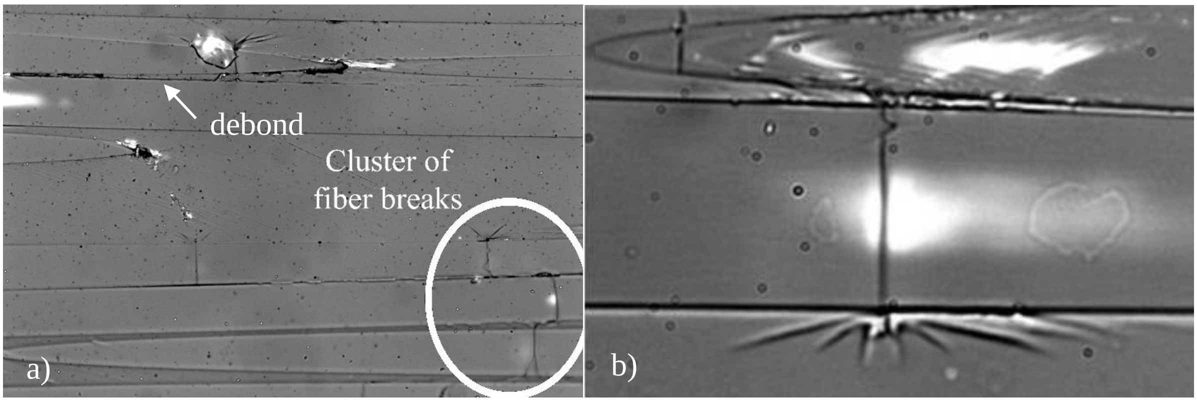

Figure 1: a) damage evolution and b) matrix yielding

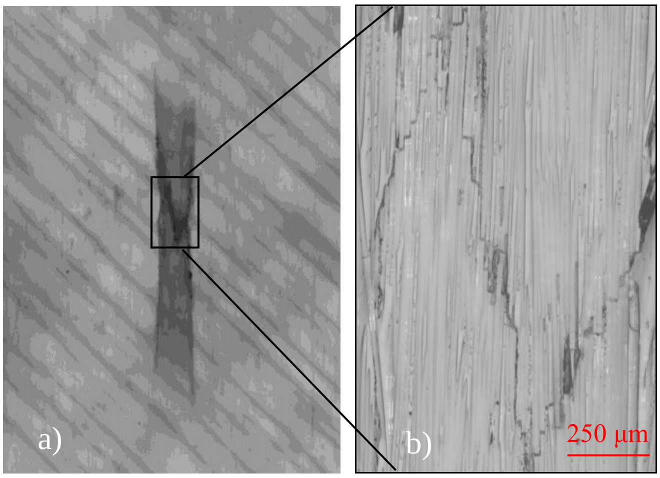

Figure 2: a) front image and b) micrograph of a macro-scale cluster

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

PREDICTION OF A FATIGUE LIMIT OF UNIDIRECTIONAL FIBRE COMPOSITES

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Keywords: Polymer composites, Mechanical properties, Finite elements
Session topics: Fatigue, Micromechanics, Computed tomography, Multiscale modelling

ABSTRACT

Composite materials are widely used in load-carrying parts of light-weight structures such as aeroplanes, transportation vehicles and wind turbine rotor blades. Most of these are designed against fatigue (failure under cyclic loading). For instance, a wind turbine rotor blades undergoes more than 100 million rotations it its design life of about 20 years. Typically, such structures are designed using so-called S-N data, i.e. empirical relations between maximum cyclic stress and the number of cycles to failure. The S-N data are usually established for failure up to a few million load cycles, the use of the S-N approach for high-cycle fatigue involves extrapolation outside the measured data. In the design of structures again high-cycle fatigue (e.g. 100 million cycles), it is relevant to consider if the material could possess a fatigue limit, i.e., a maximum stress (or strain) level, such that if the material is never loaded above this level, the material does not degrade under cyclic loading and thus can sustain an infinite number of load cycles.

Earlier studies of microscale damage evolution in unidirectional fibre composites subjected to cyclic loading have shown that isolated fibre failures are always present even in specimens that do not fail under cyclic loading (run-out for a few million load cycles) [1]. It was found that debond cracks propagated along the fibre/matrix interface, but the debond crack growth was retarded and the cracks arrrsted after about one million load cycles [1]. This raises the question if the situation of isolated broken and debonded fibes can be a stable situation under long-time cyclic loading. Recently, Sørensen and Goutianos [2] made a micromechanical model to provide an answer this question. The model considers breakage of a single fibre while its nearest neighbours remain intact. The broken fibre is assumed to debond a certain distance along the fibre/matrix interface. The length of the debond crack depends on several parameters including the fracture energy of the debond crack and the interfacial frictional shear stress acting along the debonded interface. Moreover, the debond crack tip is assumed to propagate under cyclic loading in accordance with a Paris-Erdogan law with a threshold. Also, the interfacial frictional shear stress is assumed to decrease from an initial value (first slip) to a lower stationary value after many load cycles.

It is found that the presence of frictional along the debonded interface has a significant influence of the crack tip stress intensity [2]. Just after load reversal (e.g., start of unloading), a reverse slip zone develops along the interface, starting from the location of the fibre break. There will be sticking fraction ahead of the crack tip. Then, it turns out, the crack tip stress intensity factor remains the same. With increasing unloading the length of the reverse slip zone gradually increases, obliterates parts of the original sticking zone. For short debond cracks, the reverse slip zone reaches the crack tip, and then the stress intensity decreases. Then, the stress intensity range will be non-zero and the debond crack tip will advance in accordance with the Paris-Erdogan law. Eventually, the debond crack tip has advance so much that the reverse slip length does not reach the crack tip. There will then be sticking friction ahead of the debond crack tip and the stress intensity range will be zero. The debond crack growth then stops advancing under cyclic loading. This explains the observation [1] that debond cracks stop growing under cyclic loading.

Next, it is considered if the situation of a broken, debonded fibre can be stable. The broken fibre can influence its six nearest neighbours by the additional stress induced by the debond crack tip stress fields that has moved from the location of the fibre breakage to the debond obtained after many cycles, when the interfacial frictional shear stress has decreased to its lowest value. Describing the fibre strength variation in terms of a two-parameter Weibull model, a fatigue limit was defined as the

stress (or strain) level at which the configuration of the nearest six fibres has a probability of failure of 1/6 [2]. An equation was derived in closed analytical form or the prediction of the fatigue limit in terms of strain. The equation was solved using Newton-Ralpson method.

Figure 1 shows results from the model applied to unidirectional glass fibre composites [2]. Microscale parameters (e.g. interfacial fracture energy, mismatch strain and frictional sliding shear stress) were calculated from an analysis of single fibre fragmentation tests experiments. The fatigue limit, expressed in terms of strain, is shown as a function of fibre volume fraction. The parameter $\mathscr{A}_1$ is a function of the so-called R-ratio ($\mathscr{A}_1$ increases with increasing R-ratio) and the ratio between the final and initial frictional shear stresses. It is seen that the fatigue limit is predicted to decrease with increasing fibre volume fraction and decreasing $\mathscr{A}_1$, corresponding to decreasing R-ratio. Experiment data for fatigue limit from various studies are superimposed. The microscale parameters from these materials could differ from those identified using the single fibre fragmentation tests. Still the experimental data confirms the major trend: The fatigue limit, expressed in terms of strain decreases with increasing fibre volume fraction. More results from a parameter study will be shown.

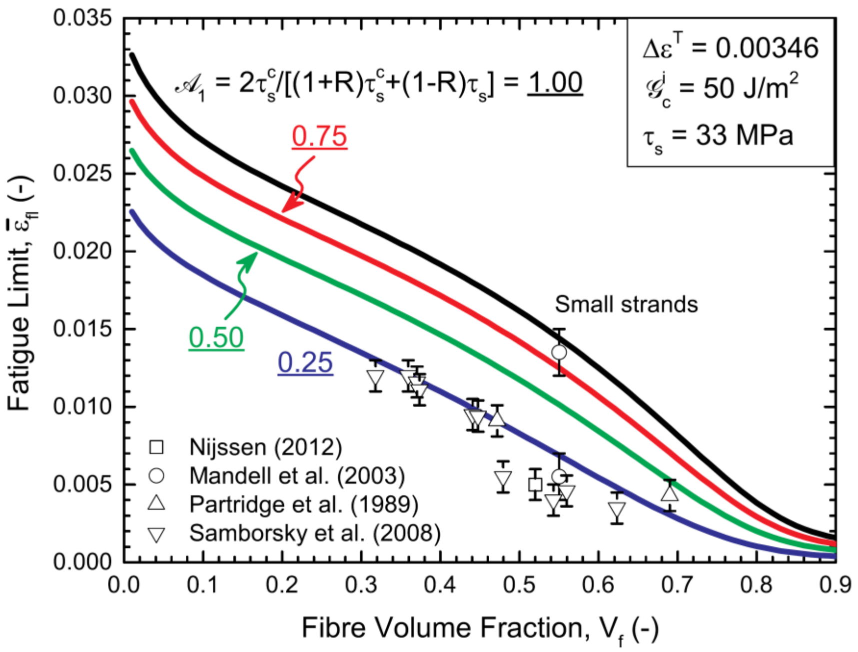

Figure 1: The fatigue limit expressed in terms of strain is shown as a function of fibre volume fraction. Experimental data as shown as symbols.

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COMPRESSIVE FAILURE OF FIBER REINFORCED UNIDIRECTIONAL COMPOSITES: EXPERIMENTAL APPROACH AND FINITE ELEMENT ANALYSIS

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Keywords: composites, compressive strength, bending, Mechanical properties, Finite elements

Session topics: Delamination, Constituent properties, Novel experimental techniques, Structural applications.

ABSTRACT

Composite structures are applied in several engineering fields (aeronautic, renewable energies, marine industry…) due to their high specific properties, especially their stiffness and their strength.
The compressive strength is generally lower than the tensile one; this relative weakness in compression remains a limiting factor in the application of composite materials [1]. Hence, in order to design a reliable composite structure under compressive loading, it is required to accurately identify the compressive strength.

The compressive failure mode of continuous fiber reinforced matrix composites is often attributed to an instability at the micromechanical scale, namely the process of microbuckling [1-5]. This leads to the formation of fibre kink bands which are governed by the composite microstructure (characteristics of the fibre, the matrix, and the interfaces, volume fractions, fibre misalignment etc.) as well as some mesoscale parameters. Indeed, several studies and models [6-8] indicate that this local instability depends on the strain gradient, the ply thickness and the adjacent plies. The compressive strain/stress to failure, as measured during a mechanical test, is therefore no more to be considered as an intrinsic material property, but one should account for the structural effects in the design. Thus, the identification of the compressive strength requires a number of experiments to characterise the governing parameters.

In the past, various test methods have been standardized in order to investigate the compressive behaviour of composite. Usually, the standard test configurations (ASTM D3410, ASTM D695, ASTM D6641 and ASTM D5467) are classified according to the loading mode such as: loading through the ends of the specimen, 4-point bending test, shear loading through frictional contact at the grips, and the combination of axial and shear loading. Yet, a significant dispersion of the measured compressive strength is reported in the literature [2]. In fact, the classical test methods may lead to parasitic failures: local buckling or failures due to stress concentrations. An alternative approach, proposed by Mechin et al. [8], defines a set of elementary experiments at the micro and the meso scales which allow the identification of the parameters to predict the compressive failure.

In this work we investigated the failure modes of composite tested following different compression methods. First, we considered the experiment proposed by Lagunegrand et al. [9]. It is a four point bending test on a sandwich beam with a polyurethane foam core in the pure bending area and two steel tabs allowing the loading of the sample. Experimental and numerical results show that the use of low stiffness core material leads to local buckling of the composite skin (Figure 1), while using high stiffness core material leads to a premature failure controlled by the shear stress concentration in the adhesive layer. To avoid those failure modes, it is recommended to improve the design of the sandwich beam. For this purpose, we designed a sandwich beam with variable core density in order to

reduce the stress concentration. Hence, the sandwich beam was based on the design defined in the ASTM Standard D695. Yet, we reinforced the honeycomb with a resin to locally increase the shear strength of the core and prevent the local buckling of the composite skin.

The following stacking sequences were used: $[0°]_4$, $[0°]_8$, $[(0°/90°)_2]_s$ and $[\pm45°/(0°)_2/\pm45°]$. The specimens were instrumented with strain gauges on the top side and on the bottom side of the sample. A finite-element model was developed to investigate the stress/strain distributions in the specimens to improve their design and to support the interpretation of the experimental results.

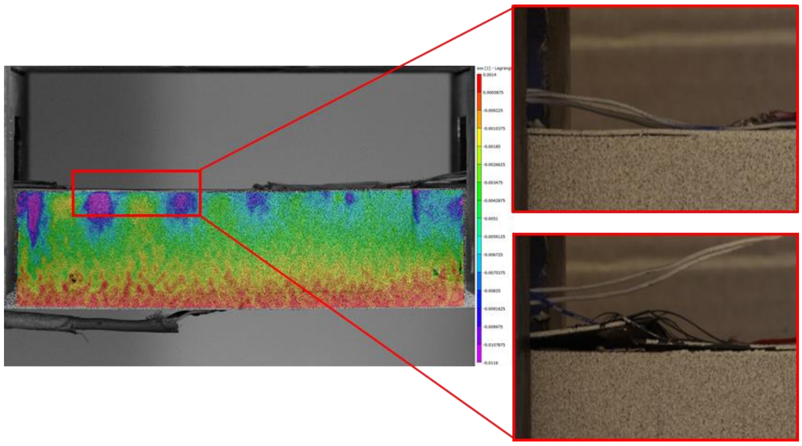

Figure 1: Digital image correlation for longitudinal strain field

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SYNCHROTRON X-RAY COMPUTED TOMOGRAPHY STUDY OF BRAIDED COMPOSITE TUBES UNDER TORSION

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Keywords: Time-lapse in-situ, Damage mechanisms, Braiding, Torsional behaviour, Textile composite

Session topics: Delamination, Computed tomography, Novel experimental techniques, Structural applications

ABSTRACT

In this study, synchrotron radiation X-ray computed tomography (CT) has been performed during the in-situ torsion test of 45° regular (2/2) braided carbon fibre reinforced plastic (CFRP) tubes to investigate the braid microstructure and the damage evolution non-destructively. The three-dimensional (3D) morphology of interlacing braided tows has been extracted from the CT data, which could be potentially employed to establish numerical models from real structures. Local non-uniformity in braid pattern and manufacturing defects such as voids have been observed in the braided composite. We found that the damage modes observed in 45° 2/2 braided composite tube, including inter-tow debonding and intra-tow matrix cracking, and fibre micro-buckling, are similar to those observed in 45° 1/1 braided composite tube; while the extent of damage propagation is different due to fewer tow cross-over points in the 2/2 structure.

Braided CFRP tubes have been increasingly used in industrial applications to replace metal or polymer tubes because of their high specific strength and design flexibility. A number of works have established the relationship between braid geometry and mechanical properties based on mechanical testing results [1,2]. However, because of the challenge in the characterisation of complex 3D structures such as braided tubes, there is still much to be understood in the relationship between braid micro-structure and damage evolution. X-ray CT has been recently employed to study complex-shaped composite structures to provide 3D information non-destructively [3,4]. Time-lapse X-ray CT imaging, which exhibits 3D information along the timescale can provide unrivalled information about the accumulation of damage in composite materials. Previously, we reported [5] the first time-lapse damage evolution study of 45° 1/1 braided composite tube. Here, the microstructure and damage evolution in the 45° 2/2 braided composite has been studied, in order to further the understanding of damage mechanisms with the braid pattern. Further analysis correlating damage evolution with braid microstructure will reveal the key factors affecting torsional performance and aid the future design of torsion-resistant braids, which is the aim of the future work in this project.

T700SC-12K60E (Torayca) carbon fibre tows were braided onto steel mandrel with a braid angle of 45° in the regular pattern and prepared into composite via the vacuum assisted resin infusion method.

The manufactured CFRP tube has an inner diameter of 10 mm and a wall thickness of ~1.3 mm. The CFRP tubes were prepared into test pieces with 15 mm long gauge section between steel end tabs. The in-situ torsion test was performed on a Deben-Manchester Open Frame Rig mounted on the Diamond-Manchester Imaging Beamline I13-2. The sample was loaded under torsion by rotating the top grip relative to the bottom grip. Loading was interrupted for CT imaging at different torsional angles to obtain a time series of 3D images depicting the damage evolution. A polychromatic pink beam (20-24 keV) was used for CT imaging. In each CT scan, 4500 projections were taken at exposure time of 0.12 s over 360° rotation in fly-scan mode. Tomographic projections were reconstructed into 2D slices using N. T. Vo python codes [6]. 3D visualization and data analysis was performed in the Avizo 9.5 software.

The ±45° braided tows can be segmented to obtain the real structure of a 2/2 braided tube (see Figure 1a). Voids can be seen in the matrix-rich regions (see Figure 1b). Damage evolution was monitored in 2D cross-sectional X-ray CT images from the same location in the braided composite tube (see Figure 1b-d). We found that the damage modes observed in 45° 2/2 braided composite tube are similar to those observed in 45° 1/1 braided composite tube; while the extent of damage propagation is more significant, as the density of tow cross-over points is lower in the 2/2 structure. The repeated damage pattern across the cross-section is related to the repeating unit cell of the braid structure. The overall damage distribution can be assessed by segmenting and visualising the damage in 3D and also the correlation between damage and the braid structure can be postulated.

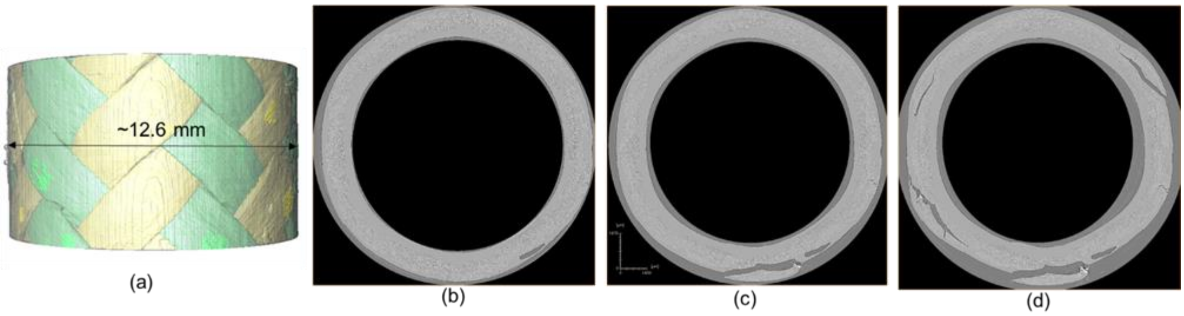

Figure 1: a) 3D volume rendering of extracted ±45° tows, (b-d) X-ray CT virtual cross-sections showing damage evolution in 45°2/2 the braided composite tube

We would like to acknowledge Diamond Light Source for the beamtime granted via the Diamond-Manchester Collaboration. We are grateful to the support from staff at I13-2 and Deben. We acknowledge the Engineering and Physical Science Research Council (EPSRC) for funding the Henry Moseley X-ray Imaging Facility through grants (EP/F007906/1, EP/F001452/1, EP/I02249X, EP/M010619/1, EP/F028431/1, and EP/M022498/1) within the Henry Royce Institute.

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

EFFECT OF INPUT PROPERTIES ON THE PREDICTED FAILURE OF A COMPOSITE PRESSURE VESSEL USING A MULTISCALE MODEL

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Keywords: Weibull parameters, Composite pressure vessels, Multiscale modelling
Session topics: Constituent properties, Multiscale modelling, Structural applications

ABSTRACT

When unidirectional carbon fibre composite specimens are loaded in the axial direction, failure initiation occurs in the form of randomly distributed fibre breaks. This leads to the creation of fibre break clusters and eventually results in the failure of the specimens [1-4]. A similar process also occurs on the load bearing layer of CPVs, i.e. the hoop layer, since the load acting in this case is similar to the tensile load on unidirectional specimens.

A multiscale fibre break model (FBM) developed at Mines ParisTech has shown relatively good comparison to the experimental observations by computed tomography (CT) technique, however, it was mentioned that a better description of T700 fibre properties is necessary as it may affect the accumulation process of the fibre breaks [5]. Therefore, further study on fibre strength characterisation can provide insights on the actual fibre strength variation and thereby also improve the quality of the failure predictions of the FBM. Different attempts have also been carried out to study the effect of loading rate using the FBM which gives a positive indication to further develop the model [7].

$$P(\sigma) = 1\text{- } \exp\left\{-\left(\frac{L}{L_0}\right)\left(\frac{\sigma}{\sigma_0}\right)^m\right\} \tag{1}$$

Eq. 1 shows the standard 2-parameter Weibull distribution equation typically used to represent strength distribution of brittle fibres. Where, $P(\sigma)$ is probability of fibre failure, L being the characteristic gauge length, L_0 the reference gauge length, σ the fibre strength, σ_0 the scale parameter and m the shape parameter or Weibull modulus. In a recent study by Islam et al [8], they have quantified the expected uncertainty in the parameters of the fibre strength distribution arising due to the effect of different causes such as sampling randomness and measurement uncertainty. The sample size used for this study was similar to the sample size popularly used by previous researchers for determining the representative fibre strength Weibull distribution. This knowledge about the uncertainty in input fibre properties can be used to predict the variability in the predictions of composite strength models for the strength and damage behaviour of composite materials and structures. In this paper, an attempt has been made to understand the variability in the predicted failure of CPVs using the FBM, as a result of the expected variability in the input fibre properties.

For this, further work has been carried out to extend the use of the FBM for predicting the strength of composite pressure vessels. A reduced volume method has been developed to increase the computation speed by reducing the number of degree of freedoms in a simulation [9]. This method has justified the use of the FBM to be assigned only at the hoop layer of a CPV and has also been favourably validated with an improved NOL ring experiment. An investigation of the stacking

sequence and fibre volume fraction of a type IV CPV has also been carried out in collaboration with The University of Southampton, which resulted in another favourable comparison with the experiment [10]. These are very critical parameters for the FBM, as it affects the stress state of the hoop layer. The comparison has been done using the T600S fibre properties determined in 2005 and the T700S fibre properties with corresponding uncertainty that has been determined through an experimental and statistical study recently. Nevertheless, the difference in stiffness in the fibre direction of the different fibre types is not significant; see table 1, but the difference in their failure strength behaviour is significant. The parameters of the fibre strength Weibull distribution for T700S fibres along with corresponding uncertainty obtained by Islam et al are given in Table 2.

Table 1: Material properties for the FBM [5]

Composite	C_{11} (MPa)	C_{22} (MPa)	C_{66} (MPa)	m	σ_0 (GPa)
T600S/epoxy [1]	149080	13974	5470	5.62	4.32
T700S/epoxy [6]	151090	11375	4500	4	5.8

Table 2: Fibre strength distribution parameters with uncertainty

Fibre type	Shape (m)	Scale (σ_0) (GPa)
T700S [8-12]	3.8 ± 1.0	4.4 ± 0.5

The research leading to these results has been done within the framework of the FiBreMoD project and has received funding from the European Union's Horizon 2020 research and innovation programme under the Marie Sklodowska-Curie grant agreement No 722626.

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

DEVELOPMENT OF A MODELING STRATEGY TO IMPROVE THE EMBEDDED ELEMENT METHOD IN COMPOSITES MODELING IN CASE OF MATRIX NONLINEARITY

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Keywords: Composite materials, Finite element, Embedded element method, Volume redundancy
Session topics: Constituent properties, Fibre-hybrid composites, Discontinuous fibre composites, Multiscale modelling

ABSTRACT

In embedded element method, component materials are meshed separately, as embedded and host regions. Constraint equations are defined between these regions such that they behave as a single material. Continuity of meshes is not needed in this method, hence proper and fewer elements can be used. Besides, modeling of mechanical behavior is not limited to assumptions, and stresses and strains on the components can be investigated. However, in order to achieve realistic results, relative size of embedded regions should be smaller than host region, and host materials should show linear-elastic material behavior [1]. In addition to these limitations, the host material in the embedded region still exists in this method and this redundant volume affects the mechanical behavior. Most common composite host (matrix) materials are thermo-elastic and thermo-plastic materials, and also muscles and tissues show non-linear mechanical behavior. When the embedded region ratio increases, this non-linearity and effect of the redundant volume should be taken into consideration.

The aim of this study is to develop a method to use embedded element method for elasto-plastic and inelastic materials. This is realized by including the effect of the redundant volume. Material properties of this redundant volume will be different in each point of the material, and change in each loading step, due to the non-linear material behavior. After each analysis step, resulting material properties of the redundant volume is included in the next step to update the material properties of the embedded region, according to their position. This procedure is conducted by complex user subroutines, and its results are compared with classical continuous element method and also with a model that does not consider plastic material behavior, which is applied by an elastic stiffness correction. The study used different fibre configurations with different fibre volume fractions. A good agreement is reached between classical continuous meshed models and our developed embedded element method in different configurations.

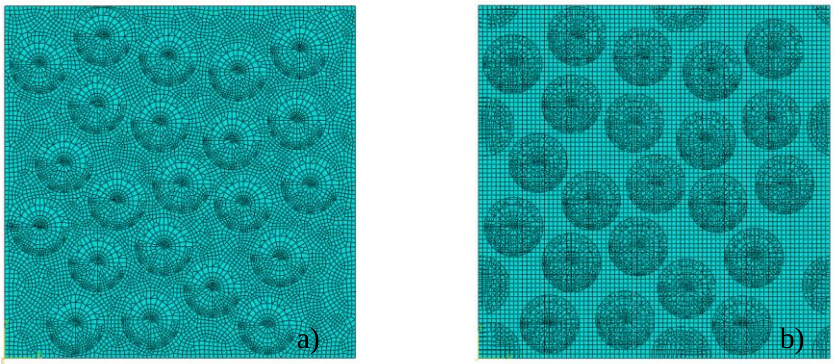

Figure 1 - Mesh configurations of a) random distribution continuously meshed model, b) random distribution embedded element models, both at 0.6 volume fraction

Table 1 - Percent differences of strains on the principal direction (E11) on different volume fractions between the continuously meshed model (Classical), our proposed model (UMAT) and elastic stiffness correction application model (EE Equivalent)

Volume Fraction	Classical v. UMAT	Classical v. EE Equivalent
0.4	1.35	12.85
0.6	0.68	6.82

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AUTOMATED RVE GENERATOR OF REALISTIC VOIDS FOR 3D TEXTILE COMPOSITES MANUFACTURED BY RTM

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Keywords: Effects of voids, Nearest Neighbour Algorithm, Clustering, Realistic shape defects.
Session topics: Constituent properties, Micromechanics, Multi-scale modelling.

ABSTRACT

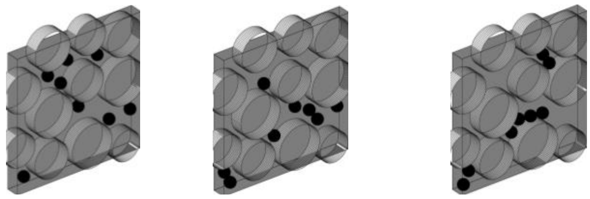
Mechanical behaviour of composite materials could be affected by defects induced during their manufacturing. Voids are one of the common defects in parts manufactured by Resin Transfer Moulding (RTM) and have a negative influence on mechanical properties. Depending on the type of manufacturing process (RTM, autoclave, etc.) or its parameters (pressure, temperature) and architecture of the reinforcement's, voids will have varying shape, size, and distribution [1]. To evaluate the properties of textile composites multi-scale modeling is used, and it involves homogenization of yarns at microscale and architecture with matrix at mesoscale [2].Most of the studies are focused on influence of overall void volume content [3], size [4] and shapes [5], by approximating voids to geometrical shapes and as homogenously distributed in the matrix domain. In reality they might have different spatial distribution, which can lead to clustering of voids. Another important aspect is the realistic shapes of the voids, which have received little attention. In this study we propose a methodology based on Nearest Neighbour Algorithm (NNA) [6] to generate a Representative Volume Element (RVE) taking into account of the realistic void shapes and their different spatial distributions. Voids are approximated as spheres and elongated cylinders with realistic cross-section were compared. The in-house script has been developed for RVE generation and application of periodic boundary conditions application in ABAQUS.

Generation of Microstructure: An extension for the classical NNA algorithm was developed, to generate the microstructure containing fibres and voids with shapes. At first, fibres were generated and later the voids were placed depending on the position of the fibres. To find the nearest neighbor ($d = R_1 + R_2 + IFD$) a search algorithm based on the radius of current fibre R_1, new nearest fibre R_2, interfibre distance IFD_{12} and the angle θ was developed. The angle theta was randomly chosen for each iteration (Figure 1).The interfibre distance was provided from real experimental measurements. A mean fibre radius of 6 µm with a standard deviation of 2 µm was imposed to generate the RVE with glass fibres having normal distribution for fibre diameter.

a) initial guess for the position of new fibre   b) accepted position at end of the iteration

Figure 1: Search Algorithm for Nearest Fiber

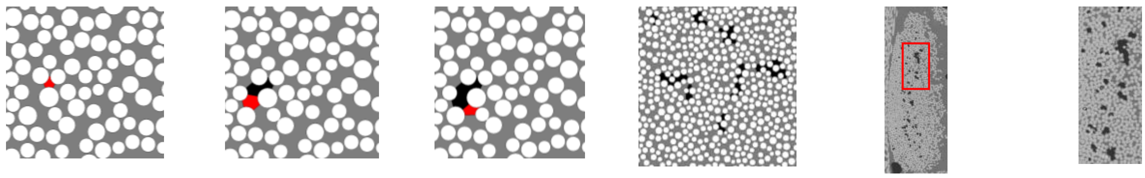
Spherical voids were generated with a fixed radius of 2µm, it is 0.25 times of maximum fibre radius and the spatial distribution was controlled by fixing nearest neighbour distance, following algorithm specified in (Figure 1). Results for RVE with a fibre volume fraction of 0.60 were presented in (Figure 2).



a) homogeneous b)random c)clusters

Figure 2: Results for the NNA, fiber volume fraction 0.60 and void volume fraction 0.02

The search algorithm for identification of the fibre clusters is based on the Voronoi tessellation. A Delaunay triangulation was performed on the fibre centres which results in the triangular mesh and vertices of each triangle is a fibre centre. The properties of each mesh element (area, Nearest Neighbours, triangle type and centroid) were calculated and a most equilateral triangle with smallest area was assigned as an initial void (Figure 3a). Then this initial void was allowed to propagate into the larger one by merging with largest triangle amongst of all the neighbouring triangles of the void (Figure 3b), once void reaches the limiting void volume the propagation is terminated and a new void can be initiated (Figure 3c). This process continues until the prescribed void volume fraction was achieved.



a)void initiation b)propagation c)termination d)large RVE e)real image f)magnified

Figure 3: Initiation, propagation, and termination of the realistic void shapes

Numerical simulation and results: To evaluate the effect of void size, shape and distribution on mechanical properties at the microscale, an RVE of length 50µm with fixed fibre distribution and variable void packings were generated. After a convergence study, a mesh size of 1.8µm was selected. Periodic Boundary Conditions were applied to calculate homogenised properties of composite yarn with the presence of voids. From the numerical simulations it was found that dispersion of the elastic properties increases from mean value as voids have random or clustered spatial distribution when compared to homogeneous distribution. With the increase of void content, the microstructure tends to be highly orthotropic and this effect is significant when voids were clustered.

Conclusions: Extension of the standard NNA algorithm was developed for RVE generation with voids of the spherical shape of controlled spatial distribution. A robust methodology was proposed for investigation of the realistic void shapes based on the local fibre distribution.

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

INFLUENCE OF STITCHING ON THE TENSILE STRENGTH OF UNIDIRECTIONAL NON-CRIMP FABRIC COMPOSITES: INVESTIGATION USING MULTI-SCALE FIBRE BUNDLE STRENGTH MODEL

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Keywords: Finite elements, Tensile Strength, Fibre breaks, Non-crimp fabric
Session topics: Multiscale modelling

ABSTRACT

Non-crimp fabrics (NCFs) are designed to combine the exceptional mechanical properties of unidirectional (UD) fibre-reinforced composites with the ease of handling and draping of conventional textiles [1]. NCFs are made by placing UD fibre tows next to each other in desired orientations and stitching them together with stitching yarns. This process binds the UD tows together without introducing a significant loss in the mechanical properties of the resulting composite in fibre directions and without any crimp of the UD tows. The extent to which the mechanical properties in fibre directions degrade is likely to depend on the diamond-shaped fibre-free zones formed by the stitching yarns piercing the fabric. These zones push the adjacent fibres closer to each other leading to the variation of local fibre volume fraction and fibre orientations in these closely packed bundles. Furthermore, the fibre-free zones can stay separated or merge to form continuous channels depending on the fabric material, yarn size and stitching pattern. During the manufacturing of composites from NCFs, the resin fills fibre-free zones making them resin-rich zones. Having such resin pockets in composites will induce strain concentrations around them depending on the size and shape of the pockets and the load direction. All these features are expected to affect the performance of the composites made from NCFs. Examined studies from the literature support this, for example, different stitching patterns in carbon fibre UD NCF were found to affect their tensile strength [2]. Comparison between NCF and unstitched carbon fibre/epoxy UD composites points towards a reduction in the tensile strength for the NCF composites [3]. To investigate this effect in more detail and to draw own conclusions, we propose a novel multi-scale modelling approach. Its main advantage is in the use of the state-of-the-art strength model which predicts the fibre break development in UD fibre bundles combined with the meso-level finite element (FE) modelling of the resin-rich zones.

To study the influence of stitching, carbon fibre UD NCFs with different stitching yarns and their tension (resulting in different average dimensions of resin-rich zones) as well as unstitched carbon fibre UD fabric were manufactured. The modelling workflow (Figure 1) starts by building geometry models of UD NCF unit cells in WiseTex software according to the investigated materials. Afterwards, the WiseTex geometry models are imported into Siemens Simcenter 3D software with the help of the built-in Virtual Material Characterization (VMC) ToolKit [4]. There, the models are modified into new unit cells of the reoriented fabrics where the resin-rich zone largest dimension (usually along the fibre direction) is oriented along the uniaxial loading direction (e.g. z-axis in Figure 1(3a)). After the geometry models in Simcenter 3D are finalised, FE models are built, periodic boundary conditions and uniaxial loading are applied using the dedicated tools of the VMC ToolKit. The created FE models consider the variation in local fibre volume fraction and fibre orientations around the resin-rich zones. After running the simulations, tensile strains along the fibre directions are extracted from the area around the resin-rich zone as shown in Figure 1(3b). These strains are later imported into the strength model which considers the fibre break development and can model fibre bundles under non-uniform longitudinal tensile strain fields. This model is based on the strength

model of Swolfs et al. for UD composites under uniform tension [5]. Fibre volume fraction variation in the strength model is taken into account by applying different stress concentration factor (SCF) distributions according to the local fibre volume fractions around broken fibres. Local fibre orientations in the strength model are taken into account by adjusting the fibre distributions in each cross-section of the bundle according to the fibre trajectories. When all these features are considered, the strength model can be used to obtain failure strain, tensile strength and fibre break accumulation.

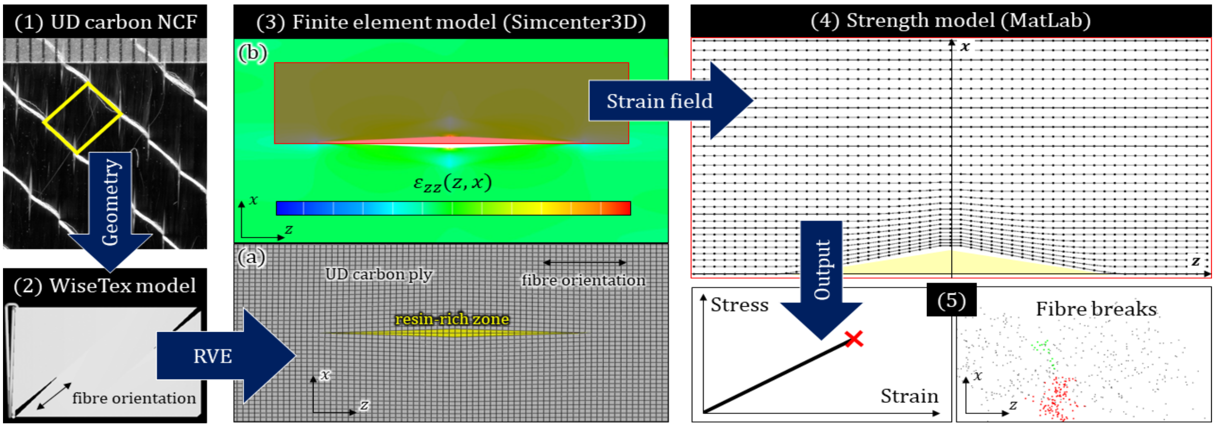

Figure 1: UD NCF modelling workflow. (1) Measuring geometry, (2) building WiseTex model, (3a) building FE model in Simcenter 3D based on the WiseTex model, (3b) obtaining strain fields (4) as input to the strength model, (5) obtaining tensile strength, failure strain and fibre breaks accumulation.

The modelling results show the dependency of the tensile strength of carbon fibre/epoxy UD NCF composites on the geometry of the resin-rich zones. The materials manufactured for this study have variability in the length and width of resin-rich zones with the standard deviation being about 20-30% of the average values. There is about 11% reduction in the composite strength for the material with the largest average resin-rich zone size in comparison to the unstitched composite. By increasing the size of the resin-rich zone we expect to amplify the effect. Such influence of the resin-rich zone size on the tensile strength of the composite can be mostly attributed to higher strain concentrations arising when the resin-rich zone becomes bigger. The developed approach will be experimentally validated in the future on the manufactured materials we used to build the geometry models in this work.

The research leading to these results has been done in the framework of the FiBreMoD project and has received funding from the European Union's Horizon 2020 research and innovation programme under the Marie Skłodowska-Curie grant agreement No722626. SVL acknowledges Toray Chair. AM acknowledges Laszlo Farkas and Oxana Shishkina for the help during his secondment at SISW and Chomarat Textiles Industries for the material. SISW acknowledges SIM (Strategic Initiative Materials in Flanders) and VLAIO (the Agency Flanders Innovation & Entrepreneurship) for their support of the ICON project M3Strength (Grant No140158), part of the research program MacroModelMat (M3).

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Abstracts

For the poster programme

DISCRETE ELEMENT APPROACH FOR MODELLING MECHANICAL BEHAVIOUR AND DAMAGE OF PA6/GF30

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Keywords: Composite materials, Mechanical properties, Experimental study, Discrete Element Method
Session topics: Constituent properties, Micromechanics, Discontinuous fibre composites, Multiscale modelling,

ABSTRACT

The transport sector, particularly the automotive industry, is in a deep change as regards the environmental and social challenges it poses. The concept of sustainable mobility has become a major issue for everyone, car manufacturers as well as economic actors. Recent European Union legislations restrict pollutant emissions of light vehicules. As a result, working on vehicle weight reduction has become necessary in order to meet the new European standards. Recently, engineers and scientists are committed to discovering and developing lighter-weight alternatives to metal, materials that can withstand the intense heat, the aggressive chemicals, and the high pressures in constant play within automotive engines. Vehicle weight reduction saves energy, minimizes brake and tire wear, and, perhaps most welcome, it cuts down emissions. Lightweighting vehicles is directly linked to lower CO2 emissions and improved fuel economy. PA6/GF30 (Polyamide 6 reinforced with 30% of glass fibers) is an example of such materials used in the automotive sector. It represents an alternative to metal structures as part of the engine compartment and interior equipment of vehicles thanks to its duality: mechanical strength-lightness. However, this material is subjected to several environmental conditions that affect its mechanical behaviour and durability [1]. Fortunately, numerical simulation combined with experimental tests enables to evaluate the mechanical behaviour of composite and the phenomena arising in the material such as crack initiations, interfacial debonding, defect effects, local variability and heterogeneity.
This contribution is part of a process aiming at establishing a numerical model to predict the mechanical behaviour and long-term damage of a composite material used in the automotive sector, the PA6 / GF30. Several numerical methods have been developed to investigate the failure of materials caused by the crack propagation. The most famous of these methods are based on Finite Element Method (FEM) which use the principles of fracture mechanics in which the failure condition is defined either by the stress intensity factor or the energy release rate in the specimen. Due to the limitations of FEM to handle these fracture mechanic issues accurately, FEM has been extended to become XFEM (eXtended Finite Element Method) which was first introduced by Belytschko and Black [2] to overcome the discontinuity difficulties. The key advantage of XFEM compared to the conventional FEM is related to the FE mesh. It does not need to be updated in order to track the crack path. However, dealing with multiple fractures with joining and bifurcation cracks makes this method costly to track crack path enforcement [3]. The Discrete Element Method (DEM) is a flexible numerical tool which naturally treats damaging and contact effects whatever the heterogeneity of the medium. The DEM was originally used in the field of rock mechanics [4] and has been since widely developed for becoming an efficient tool for solving mechanical problems for which multiple scales and discontinuities arise [3]. Based on the works of Schlangen et al. [5], it was established that the DEM based on a cohesive beam model [6] is more suitable than spring elements to model continuous media and its damaging under several mechanical loadings. Furthermore, recent works exhibited the

ability of the cohesive beam model to simulate the mechanical behaviour of heterogeneous materials throughout the example of a unidirectional fiber composite [7].
In the present work, we consider the simulation of 3D heterogeneous media (PA6/GF30) using DEM based on a cohesive beam element model [8]. In a first step, we characterize the macroscopic elastic behaviour of PA6/GF30 while taking into account the influence of different geometric parameters such as fiber dispersion and shape factor. For this purpose, a set of cubic elementary volumes describing the microstructure of the material is generated (Figure 1) and simple mechanical tests (tensile, bending) are performed to evaluate the elastic properties of the medium.

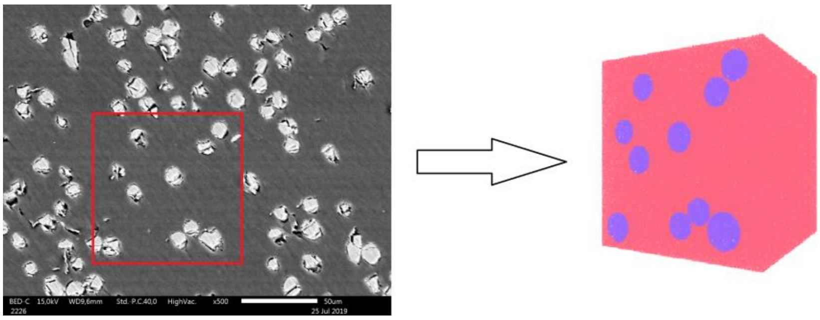

Figure 1 : MEB microstructure and its numerical model

Comparisons are then made with experimental tests to validate the choice of an adequate elementary volume. In a second step, the study is extended to the evaluation of stress fields and the simulation of PA6/GF30 damage. For this purpose, we consider the Halo approach [9], which makes it possible to regularize the typically heterogeneous field of stress in the context of the DEM. Two damage models are then considered, the first one is introduced at the particle's scale to simulate fiber failure and the second one is introduced at the fiber/matrix interface to take into account the interfacial debonding in our simulations.

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MULTISCALE SIMULATION FRAMEWORK FOR INTERLACED LAMINATES

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Keywords: Polymer composites, Mechanical properties, Finite elements
Session topics: Delamination, Multiscale modelling, Fracture toughness, (Pseudo-)ductile composites, Impact resistance

ABSTRACT

The pursuit of weight savings in the aerospace and automotive industries has driven the adoption of composites for structural applications. Despite exceptional in-plane strength and stiffness, further adoption of composite materials is limited by their poor response to low velocity out-of-plane impact loading [1]. Low-velocity impacts cause delamination - leading to significant reductions in laminate properties, most notably, compressive strength. Furthermore, because damage occurs at the interfaces between plies, rather than externally, delamination damage can be difficult to detect (not visible to the human eye) and challenging to repair.

Improving the impact tolerance of composite laminates is achieved primarily by reducing the incidence and propagation of delamination. To this end, existing research has focussed on either the development of high toughness constituents e.g. thermoplastic matrices, or on the modification of laminate's internal architecture. The issue with existing architectural methods such as z-pinning, stitching etc. – is that despite their proven efficacy in improving impact tolerance, they also negatively affect a laminate's undamaged strength and stiffness [2].

A novel laminate design concept; interlacing (also known as AP-PLY or pseudo-weaving) has the potential to provide improved impact tolerance while retaining the excellent stiffness and strength of conventional angle-ply laminates. Interlaced laminates are produced using automated tape laying (ATL) machines. Conventionally, ATL machines form laminates in a process emulating manual layups; tapes are placed in unbroken layers resulting in angle ply laminates. In the production of interlaced laminates gaps are left between tapes placed in a single layer, see Figure 1. Later passes of the tape placement machine "fill in" these gaps, causing tapes to undulate between layers in the laminate. In doing so, interlaced laminates form a woven-like architecture which effectively arrests delamination.

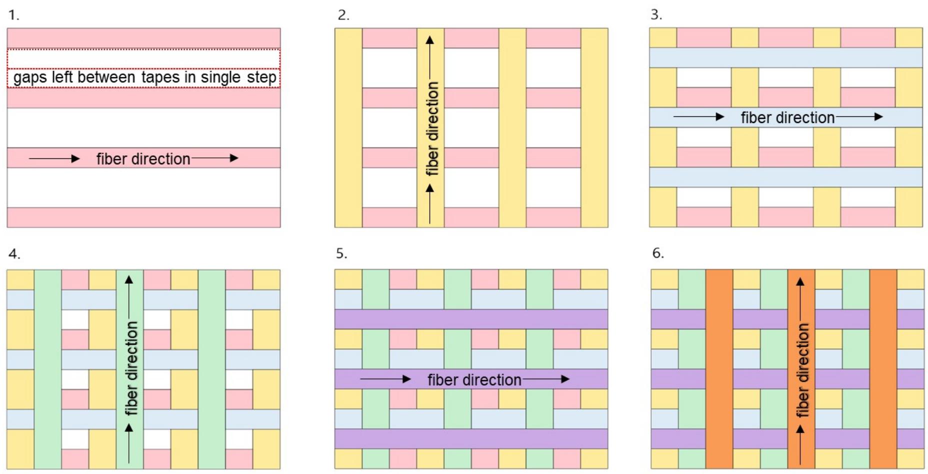

Figure 1: Interlaced laminate manufacturing process.

A multiscale numerical framework was developed using Python and Abaqus to simulate the behavior of interlaced laminates and to determine how the specific internal architecture of a panel affects its behavior. Numerical and experimental tensile tests were conducted to compare the in-plane stiffness, and strength of interlaced and angle ply laminates. Figure 2 illustrates the tensile stiffness of three interlaced panels relative to a baseline conventional cross-ply laminate. Figure 3 illustrates the interlaced configurations of the aforementioned panels. Clearly, the interlaced architecture had a negligible effect on the in-plane stiffness of the laminates. Other researchers have reported similar losses of between 1.5% to 4.7% (depending on the interlacing pattern) [3][4]. There is good agreement between the numerically calculated stiffnesses and their experimentally determined values.

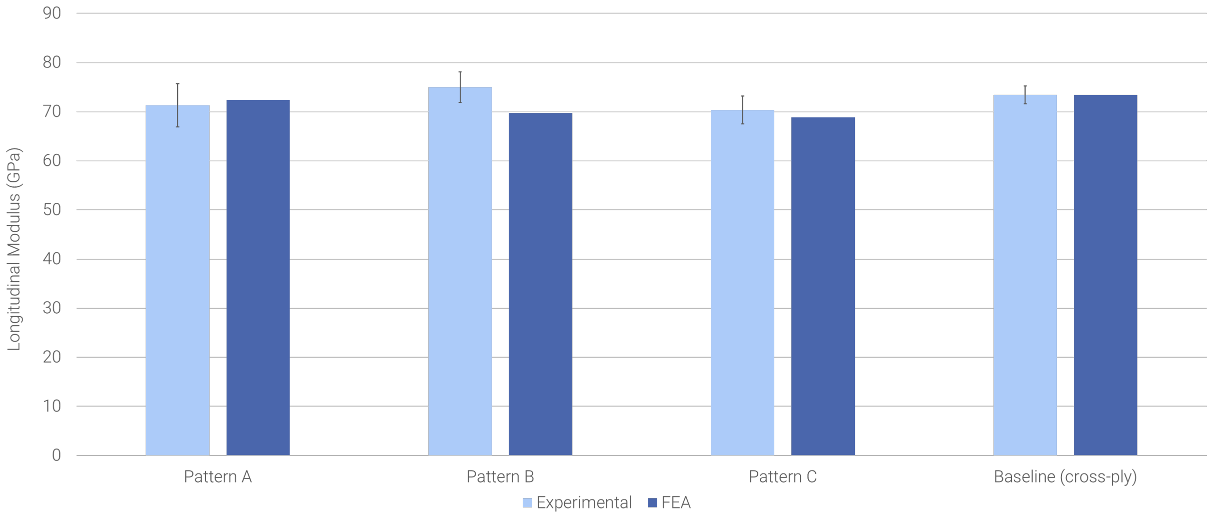

Figure 2: Longitudinal stiffness of various interlaced architectures compared with a baseline cross-ply laminate. Experimental results and numerical simulations.

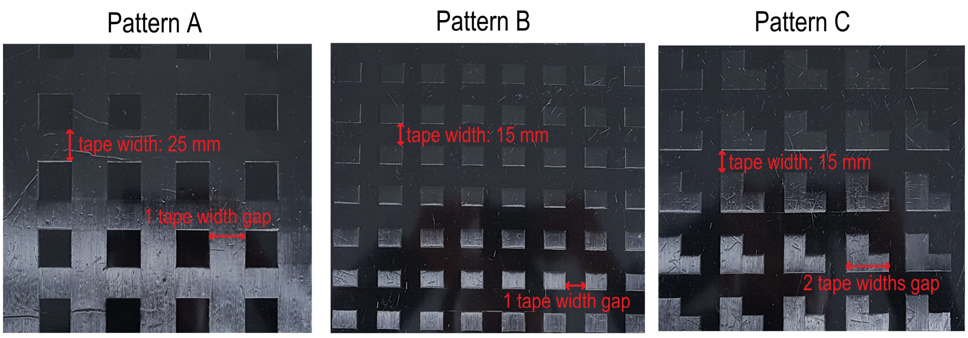

Figure 3: Images of the three different interlaced laminate configurations presented in Figure 2.

Future work will focus on impact testing of various interlaced laminate architectures and their numerical simulation. Subsequent work will aim to implement optimization routines to determine the best placement of tapes for improved impact tolerance and in-plane strength and stiffness.

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MICRO SCALE FIBER-MATRIX MODEL BASED ON THE DISCRETE ELEMENT METHOD FOR A MULTI-SCALE MODELING APPROACH

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Keywords: Micro-Scale Model, Discrete Element Method, Material Modelling, Finite elements
Session topics: Constituent properties, Micromechanics, Multiscale modelling

ABSTRACT

The interest in fiber reinforced composites (FRC) is rising due to the high potential in lightweight applications. Even though thermoplastic fiber reinforced composites (TPFRCs) have slightly inferior specific mechanical properties compared to thermoset fiber reinforced composites, the demand for TPFRCs is increasing because of the easy storaging and handling, the potential of short cycle times, an increased toughness and the hot-weldability of TPFRC [1]. TPFRCs in structural applications come in the form of blanks. These blanks are processed through heat and pressure in the thermoforming process. The main challenge in thermoforming is the process stability. Several effects on different scales cause residual stresses, which cause unwanted deformations [2]. State of the art method to reduce these deformations and stresses is a time-consuming and cost-intensive try and error method. Computational models could be used to reduce or eliminate the trial and error method. These models could predict the material and structural response of the part during forming. A multi-scale modeling approach is developed to capture the multi-scale nature of the residual stresses during thermoforming. The general approach is shown in Figure 1. Typical scales for modeling of FRC are micro- (matrix and fiber), meso- (textile and matrix) and macro-scale (laminate layer). The presented research is about a micro-scale fiber-matrix model based on the discrete element method. For this, a novel matrix-model is developed based on testing data. This model is implemented in a discrete element model based on work done by Stapleton et al [3] and compared to an equal abaqus model.

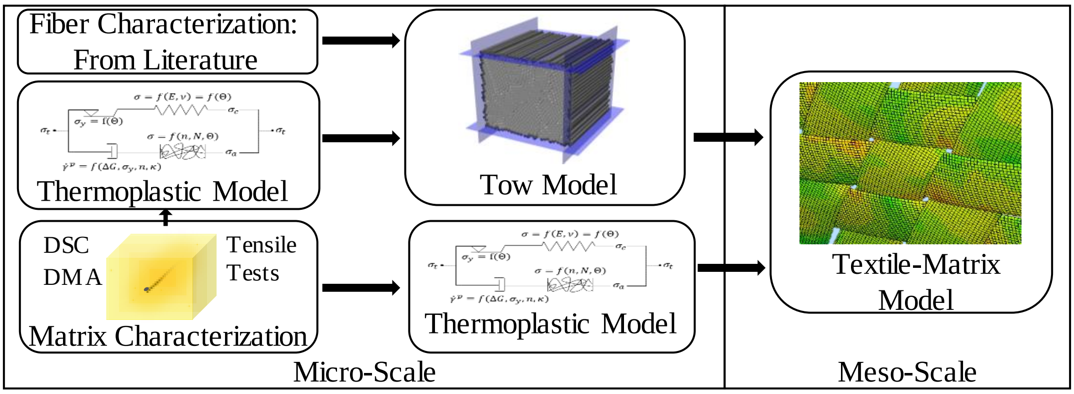

Figure 1: Multi-scale modeling approach

2. Material Characterization

Based on the timescale and temperature of the observation, the matrix polymers in TPFRCs show typical behavior of ideal solid and also ideal liquid behavior [4]. These polymers are usually viscoplastic materials. In general, thermoplastic polymers can be sorted as semi-crystalline or amorphous. Polyamid 6 (PA6) which is a semi-crystalline polymer shows good wear behavior, good chemical resistance and good strength. Some material properties like the Young’s modulus depend highly on the degree of crystallinity in semi-crystalline polymers. The influence on these material properties is analyzed. For this, samples with different degrees of crystallinity are produced through a variation in cooling rates. Afterwards, the degreed of crystallinity is measured utilizing the differential scanning calorimetry (DSC) and tensile tests for different temperatures and crystallinities are conducted.

3. Matrix Material Model

A material model formulation for semi-crystalline polymers, which is valid for large deformations in non-isothermal processes will be introduced. Essential for the biphasic nature of the polymer are different densities of amorphous and crystalline regimes. Considering these regimes is mandatory to apprehend the evolution of the residual stresses on micro scale. In accordance with the research of Dusunceli and Colak [5], a parallel alignment of crystalline and amorphous phase is selected in a phenomenological modelling approach. Thus, the overall free Helmholtz energy is obtained by a rule of mixture of the energy contributions of both phases, taking into account the degree of crystallinity. The crystallinity is predicted by a non-isothermal crystallization kinetics model. A modelling of the crystalline regime is based on a thermodynamically consistent, finite strain elasto-plastic material model, seen in [6], is presented. The amorphous phase is captured by means of a thermodynamically consistent, visco-hyperelastic material formulation in line with [7], which is valid for large deformations and large deviations away from the thermodynamic equilibrium.

4. Fiber-matrix model

During thermoforming a semi-liquid to solid state of the polymer needs to be considered. Therefore, fiber movement and contact mechanics between fibers need to be considered. The discrete element method (DEM) will be utilized to represent the semi-liquid state as well as the solid state of TPFRC on micro scale. The DEM model presented in [3] will be further developed including a modification of the developed matrix material model. Hereby, an algorithm to define matrix areas between fibers is introduced and utilized to implement the novel material model. In the end, the model will be validated by a comparison to an equal FEM-Model.

Acknowledgments

Financial support of the project RE 1057/41 by the German Science Foundation (DFG) is gratefully acknowledged.

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A UNIQUE APPROACH TO ADHESIVE JOINT DESIGN: COMPOSITE LAYUP TAILORING FOR BONDLINE STRESS OPTIMISATION

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Keywords: Polymer composites, Finite element analysis, Adhesive Joints, Optimisation
Session topics: Thin ply composites, Constituent properties, Structural applications, Fatigue

ABSTRACT

Adhesive Joints have numerous advantages over mechanical joints such as improved damage tolerance, reduced weight (in comparison to mechanical joints) and lower machining costs. Single-lap joints (SLJs) are perhaps the simplest and most common configuration of adhesive joint between composite materials, but they are often structurally inefficient due to the formation of large shear/peel stress peaks near the edges of the adhesive region. This commonly results from loading eccentricity which leads to the formation of an internal bending moment (in the case of SLJs under tension). The capability to tailor the layup of laminate composites provides a unique means of reducing these critical stress peaks along the bond-line. The potential to enhance joint design via layup tailoring remains relatively unstudied; however, this work identifies the joint design parameter combinations that decrease the critical stresses within the adhesive layer (under quasi-static loading conditions). Parameters include lengths (overlap, grip, free), width, ply-angles, thicknesses and stiffnesses.

The computational study was approximately based upon the specimen geometry and loading setup provided in test standard, BS ISO 4587:2003 (although the specimens most frequently used in this study were shortened slightly). A multi-island genetic algorithm (MIGA) [1] was selected as the underlying methodology for the optimisation process. This type of metaheuristic is inspired by the process of natural selection in evolutionary biology and provides a robust framework for optimization. A combination of Python 2.7, cmd.exe, Abaqus/CAE and Isight programs were utilised. The Analysis automatically generated, processed and post-processed finite element (FE) models as a means of identifying the layup, at which the joint had the highest load bearing capacity (determined from cohesive element QUADSCRT and MAXSCRT criteria). 3D FE models were created where the adherends were partitioned and assigned material orientations (C3D8 elements). The adhesive regions were modelled as solid-elastic cuboids (C3D8 elements) and connected to the adherends on both sides via cohesive elements (COH3D8); see *Figure 1* for images of an example finite element analysis.

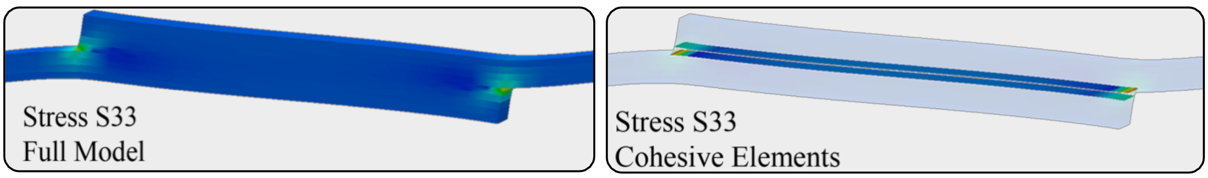

Figure 1: Images of a sample FEA, which the optimization utilizes. Left: S_{33} stress distribution for an SLJ with a [0/0/90/0] layup. Right: S33 stress distribution of the cohesive layers.

In order for the MIGA to efficiently find an optimal solution, a number of test runs were performed to gauge the effects of the number of islands, rate of mutation/migration, elite size, interval of migration etc. (to correctly calibrate future simulations). Figure *2* provides a visual comparison showing how the mutation rate or 'likelihood of mutation' (analogous to biological mutation) alters how the algorithm converges towards a set of solutions. In this example, decreasing this value caused the simulation to converge more readily towards a singular solution.

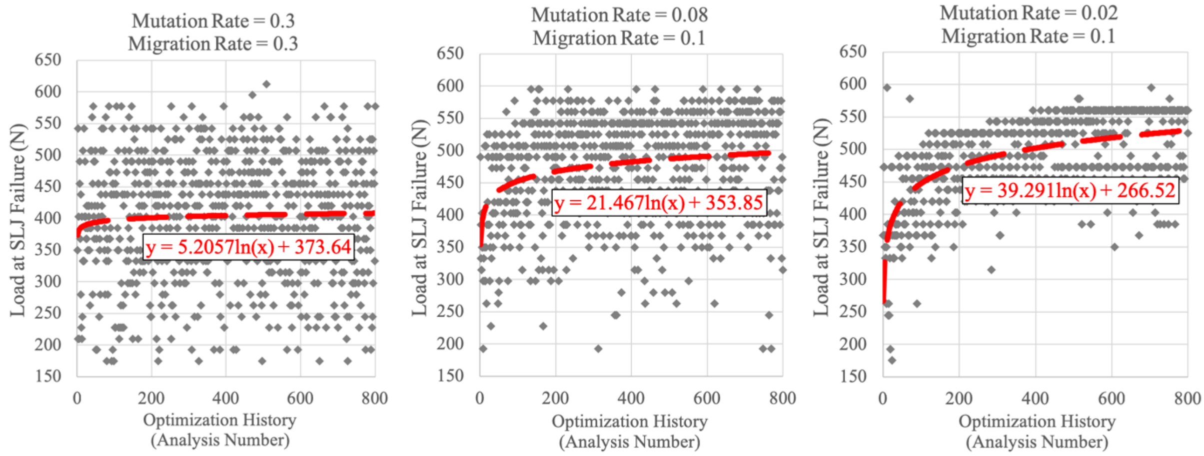

Figure 2: Graphs showing the effects of the MIGA's mutation rate on the optimization history (proportional to elapsed time of the MIGA)

The results produced from this work provide a new methodology for designing adhesive joints for static loading (under any user defined loading condition) but still require further development to consider fatigue loading cases. Additionally, intra-composite failure (i.e., that within the composite laminates themselves) is not factored here due to dramatically increased analysis times. Thus, it is assumed that the SLJ reaches first failure along the bond line. The results produced from this work afford a profound insight into which composite layups will make an adhesive joint as strong as possible and also which parameters from the CLPT ABD matrix are most beneficial for creating a strong adhesive joint [2]. Figure *3* shows the influence of the global x-direction stiffness (ABD[0,0]) and relevant bending-extension coupling term (ABD[0,3]) on the failure load (where an index of [2,4] would represent the third row and fifth column of the 6x6 ABD matrix).

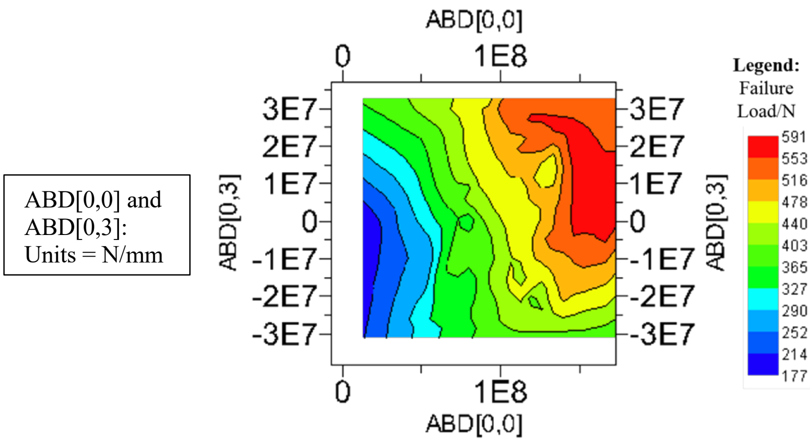

Figure 3: Surface plot showing the effects of ABD[0,0] and ABD[0,3] on the load bearing capacity of SLJs (with specs=Laminates: 4ply-thickness=1.6mm, Adhesive: thickness=0.2mm).

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QUANTITATIVE COMPARISON BETWEEN FAST FOURIER TRANSFORM AND FINITE ELEMENT METHOD FOR MICROMECHANICAL MODELING OF COMPOSITE

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Keywords: Micromechanics, Fast Fourier Transform Method, Neighbour Voxels Average algorithm

Session topics: Micromechanics, Multi-scale modeling

ABSTRACT

A variety of numerical methods can be applied for multi-scale simulation of composite materials in general and textile reinforced composites in particular. Among numerical methods, the Finite Element Method (FEM) is the main tool for modeling textile composites [1]. Recent developments have brought increased interest in the Fast Fourier Transform (FFT) based method for multi-scale material modeling. This method uses image-based techniques and also gives accurate results as FEM voxel-based models do [2]. Backing to 1994, the FFT method was proposed initially by P.Suquet and H.Moulinec [3], as a voxel-based methodology that does not need stiffness matrix assembling like FEM. It can thus be very efficient in the field of digital materials and easily parallelized. The main drawback of voxel-based models is the presence of strong oscillations due to the non-smooth interface [4]. From the best of our knowledge, the FFT and FEM are often compared in a general way. In this work, specific problems of the micromechanics of composite materials were addressed in order to compare quantitatively FFT and FEM solutions of the stress field at the interface, based on the direct output and with the introduction of a smoothing method. The open-source software AMITEX [5] is applied for all FFT calculations and ABAQUS is applied for all FEM calculations.

Model: The study case is a cubic unit cell model consisting of cylindrical fibre (V_f=0.55, E_f=72GPa, and ν_f=0.22) and matrix (E_m=3.3GPa, υ_m=0.375) subjected to macroscopic deformation of 0.1% with periodic boundary conditions. Three numerical models have been investigated: 1). Conformal mesh (FEM) with x25 resolution and 2). Voxel-based mesh (FEM & FFT) with x101 resolution.

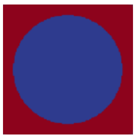
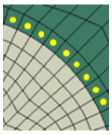
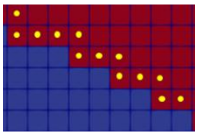
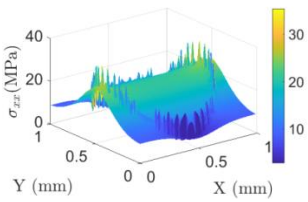
a)  b)  c) 

Figure 1: General presentation of the models: a). Unit Cell (Blue: Fibre; Red: Matrix); b). Conformal mesh (Green: Matrix; Grey: Fibre) and c). FFT and FEM Voxel-based mesh (Blue: Fibre; Red: Matrix). (Yellow points: Centroid points of interface zone)



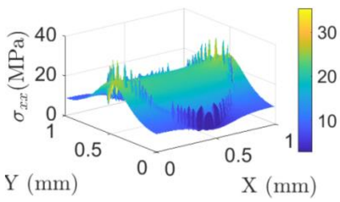


Figure 2: $\boldsymbol{\sigma_{xx}}$ Stress field (Left: FEM Voxel Mesh; Right: FFT Voxel Mesh)

σ_{xx} stress fields for voxel mesh models: As the direct output results are shown in Figure 2, strong oscillations only occur along with the interface. Which are almost the same between FEM Voxel mesh and FFT considering shape, localisation, and values. These oscillations will affect the accuracy of the final results, which need to be eliminated by introducing the smoothing method.

Smoothing method: In this work, "Neighbour Voxels Average" was implemented based on usage of linear weight function [6]: $WF\_i = 1 - L\_i/(L\_max + 1)$, as shown in Figure 3: the **red voxel** is the interface voxel on the matrix side where we want to eliminate oscillations; the **orange voxels** belong to matrix and **the blue voxels** are fibre.

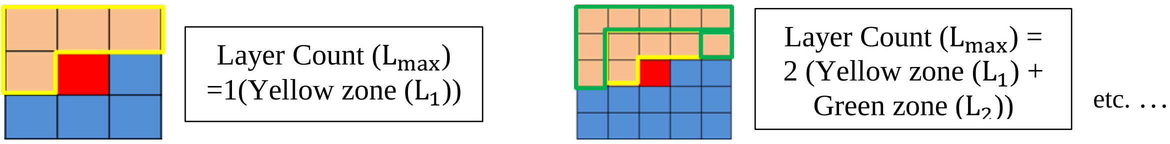
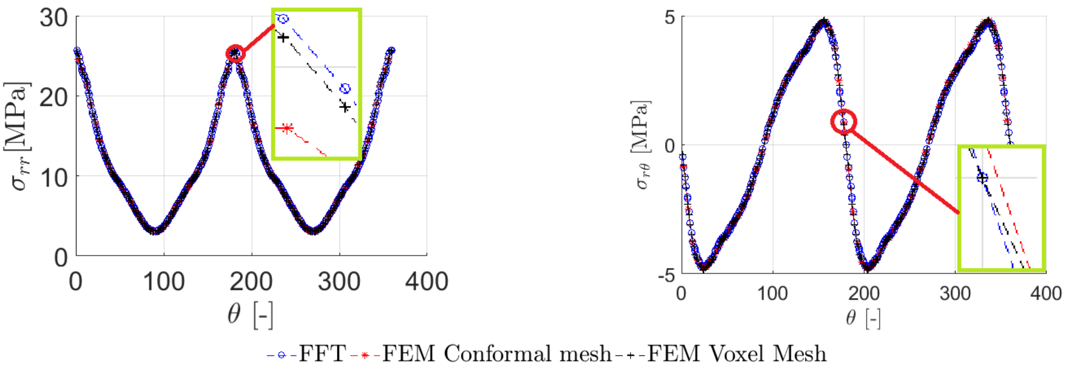


Figure 3: Neighbour Voxels Average method

Quantitative comparison of interface stresses (FEM vs FFT): The comparison of normal σ_{rr} and tangential $\sigma_{r\theta}$ stresses which were obtained from FEM conformal, FEM Voxel-based and FFT model are shown in Figure 4. The difference in numerical values for FEM voxel-based and FFT is less than 0.2%. The differences between FFT and FEM conformal mesh in σ_{rr}, which are also negligible, are mainly due to discretization type.



a) Normal stress b) tangent stress

Figure 4: Comparison between FFT and FEM using local smoothing method

Conclusion: It was confirmed that FFT and FEM Voxel based give almost the same results. Moreover, when the local smoothing method is applied FFT gives very close results for the interface stress field as FEM with conformal mesh. Considering that FFT has much more advantages in computation time, FFT is a competitive method in simulating complex multi-scale problems for textile composites.

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ASSESSMENT OF EFFECTIVE ELASTIC PROPERTIES OF SHORT-FIBER REINFORCED POLYMER BY MICROMECHANICAL PROBABILISTIC MODEL

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Keywords: Polymer composites, Mechanical properties
Session topics: Constituent properties, Micromechanics, Discontinuous fibre composites, Structural applications

ABSTRACT

Short fiber reinforced polymer (SFRP) are used for many engineering application namely for automotive industry due to their attractive properties such as mechanical strength, corrosion resistance, easy shaped of complex geometry and to their low cost of manufacturing .

In this study a probabilistic approach is adopted to estimate the effective properties of short fiber reinforced polymer (SFRP) based on micromechanical modified Cox's model using Monte Carlo simulation where longitudinal young's modulus (Figure 1) and in-plan shear modulus are estimated . To conform the results obtained by Monte Carlo simulation, Mori Tanaka scheme is used which can relate the microstructure with the mechanical properties, then the two approaches (Figure 2) was validated by experimental data result from literature [1].
Further, the model have been used to investigate the effects of aspect ratio and volume fraction on longitudinal young's modulus (Figure 3) where is found that more the fraction volume of fiber increase more the young's modulus increase. Is found also, that young's modulus increase gradually with increasing aspect ratio until critical length where after the aspect ratio have no effects on it.
Using the obtained engineering constants, a reliability based design optimization have been adopted which can provide optimum designs in the presence of uncertainty. A single loop approach algorithm (SLP) [2] is used which can be a powerful tool to assist design under uncertainty with accuracy and low computation effort.
An example of cantilever beam subjected to bending load is studied [3]. The beam is loaded at its tip by concentric load in y direction and z direction fig the objective is to minimize the weight of beam subject to non linear constraint and satisfy the target reliability index $\beta = 2.3263$, thus the problem is formulated as

$$\begin{gathered} min f = w * t \\ s.t \;\; P(G_j(d,P) \geq 0) \geq R \qquad j = 1 \sim 2 \\ G(E,Y,Z,w,t) = D_0 - \frac{4L^3}{Ewt}\sqrt{\left(\frac{Y}{t^2}\right)^2 + \left(\frac{Z}{w^2}\right)^2} \qquad 0 \leq w, \;\; t \leq 15 \end{gathered} \tag{1}$$

To assist damage evolution of structure, a reliable predictive model by neural network algorithm is used for structural health monitoring [4]. The structure default is detected through young's modulus degradation. The input vector contain three parameters, the loading in vertical , horizontal directions and young's modulus . The output is the value of objective function , 80% of data is used to train the network where the other remained of 20% is used to validate the model. A threshold value is fixed corresponding to 5% of allowable damage , during damage evolution the program compare the estimated value from neural network algorithm with fixed threshold value. The graphical output (Figure 4) of results is given below.

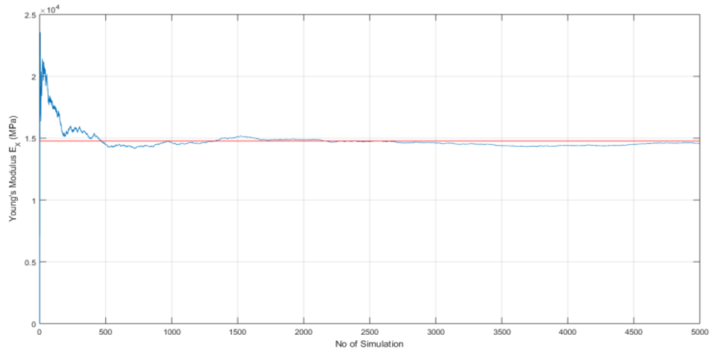

Figure 1 : Mean of young's modulus as function of number of simulation

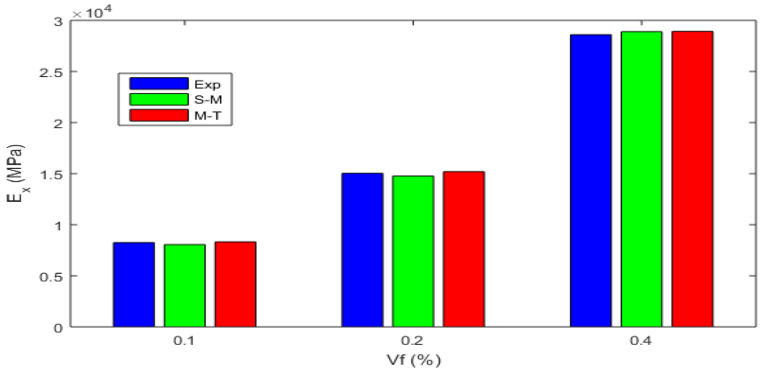

Figure 2 : Comparison between experimental [1] , Simulation Method (S-M), Mori-Tanaka method (M-T)

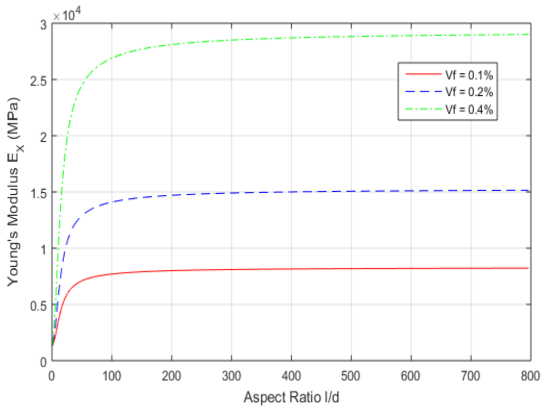

Figure 3 : Influence of aspect ratio and volume fraction on young's modulus

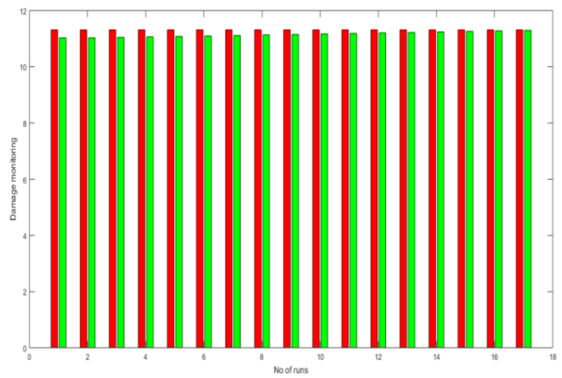

Figure 4 : Damage monitoring estimated by neural network

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

MODE I TRANSLAMINAR FRACTURE TOUGHNESS OF CARBON-CARBON HYBRID THIN-PLY COMPOSITES: MODELLING AND CHARACTERIZATION

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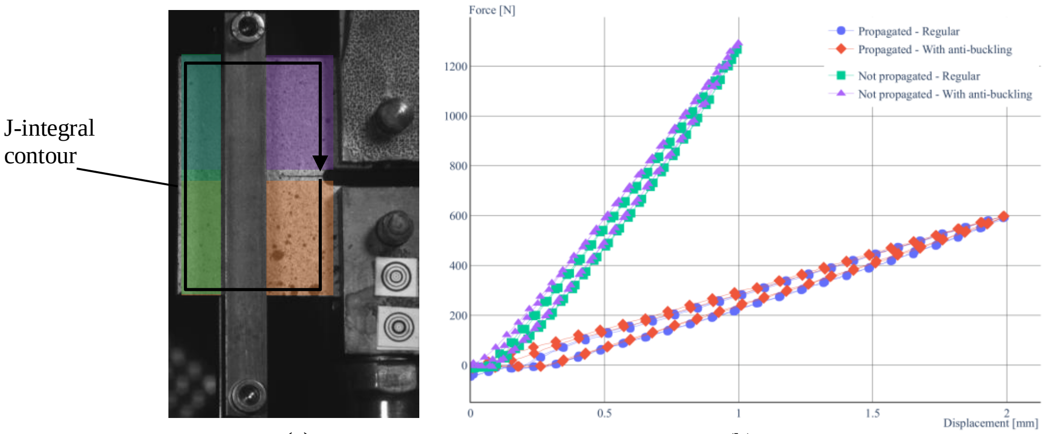
Keywords: Thin-ply, Fiber-hybrid, Translaminar toughness, Pull-out, Digital Image Correlation
Session topics: Fibre-hybrid composites, Thin ply composites, Fracture toughness, Novel experimental techniques

ABSTRACT

Thin-ply composites are known to exhibit unique mechanical performances and are regarded as a key technology to further improve mass reduction in composites structures [1-3]. However, recent studies have shown that this improvement is done at the expense of pull-out length and translaminar toughness, rising concerns about thin-ply reliability [4, 5]. Introducing a hierarchical microstructure in thin-ply composites during processing has been proven to restore pull-out length, being an effective toughening strategy, but brings drawbacks such as a loss of unnotched strength and a poor manufacturability [6,7]. Besides, fiber-hybrids, if carefully harnessed, can be used to introduce fragmentation in glass-carbon hybrid thin-ply composites without incurring delamination nor catastrophic failure thanks to synergistic effects [8,9]. It has been experimentally demonstrated that this property of thin-ply hybrids can be used to improve pull-out length and translaminar fracture toughness in carbon-carbon hybrid composites, using one carbon grade as a low strain fiber and the other as a high strain fiber [10].



(a) (b)

Figure 1: (a) Anti-buckling device and representative zones used to fit displacement fields obtained by DIC (purple, orange, green and superposed green); (b) Influence of anti-buckling device on compact tension loading and unloading of 34-700-TP415 cross-ply laminates with a ply thickness of 60 µm for different crack lengths.

In this work, an analytical model based on pseudo-ductility framework [11] is proposed to predict failure mechanisms, pull-out length and strength in interlayer carbon-carbon hybrid thin-ply composites. This model is currently being developed in order to predict translaminar fracture toughness of fiber-hybrids. The input parameters are characterized for thin-ply prepregs supplied by North Thin Ply Technology Sàrl, as specified in Table 1 and the model is used to design and produce cross-ply fiber-hybrids with these materials. To validate the model, a reliable mode I translaminar toughness characterization is needed. To this end, compact tension tests are performed using digital image correlation (DIC) and J-integral data reduction. To avoid buckling with tough laminates without increasing specimen thickness, the use of an anti-buckling device that spans across the compact tension specimen is introduced. A data reduction algorithm is developed to take into account data points hidden by this anti-buckling device and is found to correlate well with both experimental and simulated data. The influence of the anti-buckling device on the experimental setup is validated by loading and unloading cycles as shown in Figure 1.

Table 1: Materials used in this study.

Fiber	Resin	Usage	Strain [%]	Longitudinal modulus [GPa]	Longitudinal strength [MPa]	Prepreg thickness [µm]
34-700	TP415	High strain	2.1	234	4830	60
HR40	TP415	Low strain	1.2	390	4610	29
HS40	TP415	Low strain	1	455	4610	29

The research leading to these results has been performed in the framework of the HyFiSyn project and has received funding from the European Union’s Horizon 2020 research and innovation programme under the Marie Skłodowska-Curie grant agreement No 765881.

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PROPERTIES OF HYBRID STAINLESS STEEL WOVEN WIRE MESH/GLASS FIBER-REINFORCED EPOXY COMPOSITES UNDER QUASI-STATIC TENSILE LOAD

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Keywords: Polymer composites, Digital Image Correlation, Hybrids, Metal Mesh
Session topics: Fibre-hybrid composites

ABSTRACT

There are two main issues that composites face which hinder their wider uptake in the automotive industrial applications such as crash or impact components. The first issue is their inherent brittle nature and their poor resistance to crack initiation and growth, which results from the low failure strain of used fiber and matrix system. For example, carbon fiber composites have failure strain in the range of 0.4 to 2.1% [1]. The design ultimate allowable strain of carbon fibers is nearly 0.4% [1]. An improvement in the damage tolerance would result in 50% improvement in the allowable design strain [1]. The second issue is the high cost of fibers with high specific strength and stiffness. One possible solution to these problems is the fibre-hybridization of brittle fiber composites with tough and low cost fibers. Fibre-hybridization is defined as the combination of two or more fibres in a particular configuration and relative fiber volume fraction. The resultant hybrid material makes it possible to achieve properties that cannot be achieved with non-hybrid composites [2].

The goal of this study is to hybridize glass fiber composites with three different stainless steel (AISI 316) woven wire meshes while keeping a constant hybrid ratio and then comparing their tensile properties with pure glass and steel mesh reinforced composites. The three meshes have different wire diameters and yarn spacing as shown in Table 1. The aim here is to select the stainless steel mesh for future experiments that performs best in terms of ease of processing i.e. can be processed via high volume production industrial processes such as RTM and wet compression molding (also known as wet pressing), can achieve high fiber volume fraction (up to 50%) and provides improvement in energy absorption capability when combined with low failure strain fibers to produce a hybrid composite.

Experimental testing

Commercially available non-crimp E-glass fiber textile produced by SGL was hybridized with three different stainless steel woven wire meshes (details shown in Table 1 and 2) in epoxy* matrix. All the samples were produced using wet compression molding method available at BMW Group using 125 bar of pressure. All the specimens were tensile tested (ASTM D3039) using the same overall dimensions. Digital image correlation technique (DIC) was used for accurate strain measurement.

We have observed that, for the hybrid specimens with fine woven wire mesh a positive hybrid effect in the final failure strain was achieved while hybridization with medium and coarse meshes resulted in negative hybrid effect in the final failure strain. The microscopic analysis of the longitudinal crossection of the hybrids showed that under same processing parameters the use of medium and coarse wire meshes resulted in the disturbance of fiber architecture along with cavities formation. The microstructure disturbance shown in Figure 3 resulted in a negative hybrid effect as the final failure strain of coarse and medium mesh hybrids is less than the pure glass fibre composites as shown in figure 2.

Table 1 Specifications of stainless steel wire meshes used for experiments.

	Wire diameter (mm)	Yarn Spacing (mm)	Laminate	Thickness (mm)	Weave Type
Fine Mesh (F)	0.03	0.036	F_{21}	1.32±0.011	Twill

Medium Mesh (M)	0.16	0.263	M_7	1.51±0.014	Plain
Coarse Mesh (C)	0.28	0.567	C_4	1.35±0.012	Plain

Table 2 Stainless steel mesh/Glass fiber hybrids.

Hybrids	Layup	Thickness (mm)	Hybrid Ratio*	Fiber Volume Fraction (%)
F with Glass Fiber (FG)	F_7(GF)[0]F_7(GF)[0]F_7	1.37±0.013	0.73	47.7
M with Glass Fiber (MG)	M_2(GF)[0]M_2(GF)[0]M_2	1.57±0.012	0.76	47.6
C with Glass Fiber (CG)	C_1(GF)[0]C_1(GF)[0]C_1	1.36±0.011	0.71	45
E-glass (G)	$(GF)_8$[0]	1.52±0.011	-	48.4

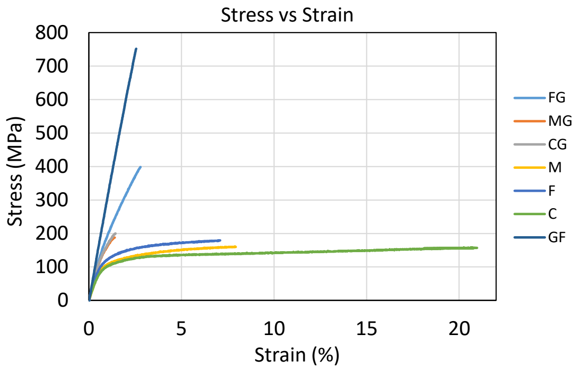

Figure 1 Representative stress vs strain plots of tested specimens.

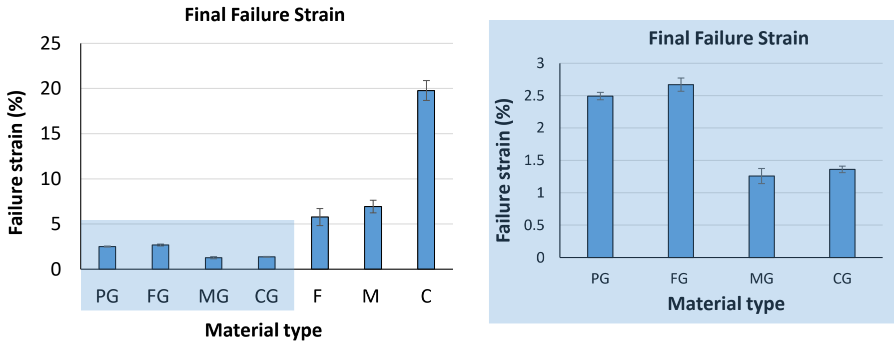

Figure 2 Failure strain comparison of glass fiber/Steel mesh epoxy hybrids with mono-material.

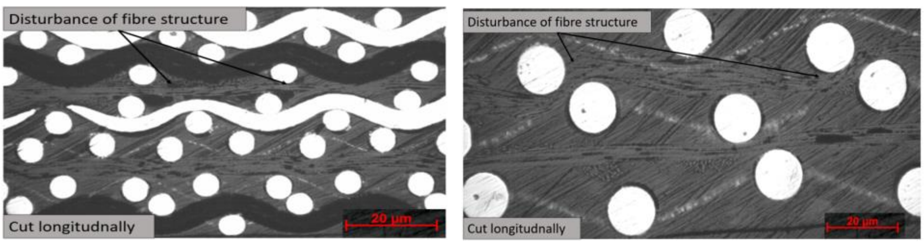

Figure 3 Microscopic analysis of MMG (left) and CMG (right).

This project has received funding from the European Union's Horizon 2020 research and innovation programme under the Marie Sklodowska-Curie grant agreement No 765881.

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*Hybrid ratio = (V_f of High strain fibres)/ (Total V_f)   *Epoxy = Epikote Resin 06000

THE INFLUENCE OF CONTAMINATION ON THE RECYCLED AND REUSE CARBON FIBRE PRODUCT

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Keywords: Polymer composites, Carbon Fibre, Fibre-hybrid composites, Recycling, Short fibre composites

Session topics: Recycling

ABSTRACT

Carbon fibre reinforced plastics (CFRP) composites provide weight saving advantages in various industries, thanks to their high specific mechanical properties. However, uptake of these materials is limited in many applications due to their high cost. As an alternative, recycled carbon fibre (rCF) offers the potential to lower both the cost and environmental impact associated with the use of CFRP. Recycled product quality depends upon the mechanical properties of the individual recyclates such as the fibres as well as the contamination level of other materials used in the fabric such as stitching thread. In addition, the use of hybrid composite configurations is gaining popularity but comes with an increased level of contamination in the manufacturing waste that needs to be recycled. This work aims to assess the properties of the recycled product by varying the amount of contamination and comparing the results with a 100% recycled carbon fibre product. The fibre materials used in the non-woven mat are recovered via a pyrolysis process (with an interlayer configuration of cured hybrid fibre carbon/glass composite laminate waste). The strength of the recycled product being determined by experimental testing. Finally, the finding of this work will contribute to efficient reuse of recycling carbon fibre (rCF) product.

INFLUENCE OF BUILD PLATE TEMPERATURE ON THE TENSILE STRENGTH AND STIFFNESS OF 3D PRINTED RCF/RPET PARTS

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Keywords: Polymer composites, Additive manufacturing, Mechanical properties
Session topics: Discontinuous fibre composites

ABSTRACT

Since the world's interest in the transition to a circular economy has started to grow, researchers have made an effort to create and characterise recycled materials. As the use of carbon fibre reinforced polymers increases over the years, recycling and re-manufacturing processes are essential to limit waste and minimise their environmental impact. The reclamation, recycling and remanufacturing of carbon fibre reinforced polymer composites has already been studied and documented in literature [1]. In combination with a recycled polymer, one can produce an environmentally and economically interesting product. The recycling of poly(ethylene terephthalate) in particular is already a soundly established process in the polymer recycling industry.

Re-manufacturing through additive manufacturing could improve mechanical properties compared to conventional re-manufacturing methods, as the degrees of freedom provided by a 3D printer can be exploited. The mechanical performance of printed parts however is dependent on the adhesion between adjacent deposited filaments. Since the level of entanglement of polymer chains from one filament to the other depends on temperature, the combination of the extrusion temperature and build plate temperature are important processing parameters. For example, Hertle et al. have reported that the contact temperature between substrate and newly extruded filament should be in the vicinity of the melting temperature of the semi-crystalline polymer [2]. The melting of the polymer in the area of contact between substrate and extruded filament allows for a higher level of interdiffusion, leading to a higher level of load transfer, and thereby increasing mechanical performance.

This work investigates the influence of the build plate temperature on the tensile properties of recycled carbon fibre-reinforced recycled poly(ethylene terephthalate) composite parts, printed using the extrusion-based printing technique fused filament fabrication, also known as fused deposition modelling.

The influence of the build plate temperature on the tensile strength and stiffness of the material was determined by comparing specimens printed on a plate at a temperature above and below the glass transition temperature of the polymer matrix, holding all other printing and material parameters constant. It was found that samples printed using a build plate temperature of 100°C showed no significant difference in stiffness, nor failure stress, compared to the samples printed using a build plate temperature of 40°C. Raising the build plate temperature above the glass transition temperature of the polymer thus had no influence on the investigated tensile properties of the samples.

The authors wish to thank SIM for their support of the SBO project RELFICOM, running in the NANOFORCE Program.

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THREE DIMENSIONAL FIBRE AND DAMAGE CHARACTERIZATION OF INJECTION MOULDED FIBRE REINFORCED THERMOPLASTICS BY X-RAY COMPUTED TOMOGRAPHY

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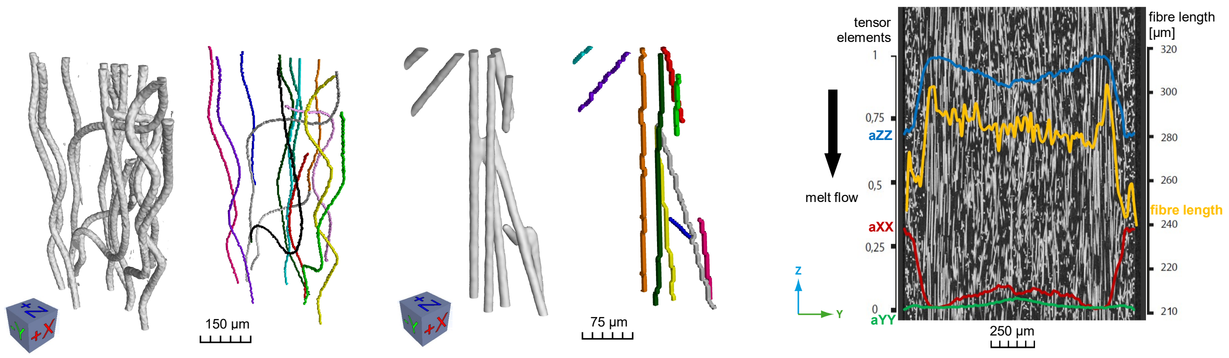
Keywords: Polymer composites, Fibre characterization, Damage characterization, X-ray computed tomography, In-situ measurements
Session topics: Micromechanics, Discontinuous fibre composites, Computed tomography

ABSTRACT

To improve the mechanical performance of fibre reinforced polymers the material behavior in the microscale range needs to be understood. Therefore a closer look on the microstructure and furthermore a microstructural investigation of the damage type and propagation is necessary. By the use of X-ray computed tomography (XCT) in combination with in-situ tensile testing, a lot of information about the material and its behavior can be gained.

In this work a methology for the characterization of fibre reinforced polymer samples produced by injection moulding is presented. The XCT measurements were performed on a laboratory device (GE Phönix|x-ray Nanotom 180 NF) and analyzed with VG Studio Max (Volume Graphics GmbH) as well as with open_iA [1]. The latter allows a fibre characterization by the extraction of the middle axes of each fibre [2, 3] as shown exemplary for PET and glass fibre (GF) filled polypropylene samples in figure 1a respectively 1b. Furthermore, the knowledge of start and end point of each fibre enables the investigation of the fibre orientation and fibre length distribution as shown for a glass fibre reinforced material in figure 1c.



(a) (b) (c)

Figure 1: 3D views of a cut out of different fibre types, with according results of the extracted middle axes of each fibre shown for PET fibres (a) and glass fibres (b). Figure (c) shows the fibre orientation and fibre length distribution for a glass fibre reinforced polymer over thickness, whereas aXX, aYY and aZZ describe the main components of the orientation tensor.

So called interrupted in-situ tensile tests can be used for the investigation of the damage evolution [4]. Those quasi-static tensile tests are performed until fracture. After each loading step a scan is performed and the defects can be characterized. The defect characterization was carried out with an in-

house developed software tool [5] and is based on the three dimensional XCT data and the results of the fibre characterization. This tool enables the visualization and quantification of four different defect types, namely fibre fracture, fibre pull-out, fibre/matrix debonding and matrix fracture. Despite the total number of defects, also the defect volume can be determined.
Results are presented for glass fibre filled polyamide 6.6 test samples (PAGF) at the last loading step before breakage (figure 2). Clear differences in damage induction for different main fibre orientations (FO), as well as differences in the number of defects depending on fibre volume fraction, can be observed.

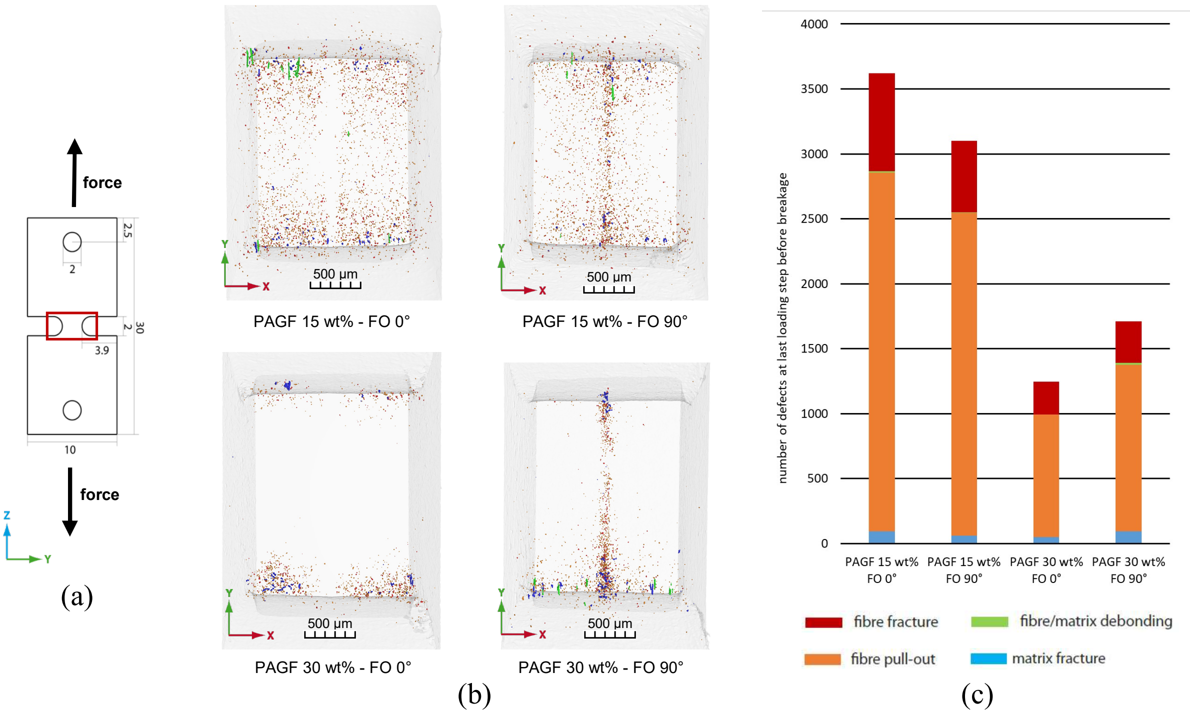


Figure 2: (a) test specimen geometry (red rectangle indicating the measurement area), (b) rendered 3D images (top view) with color coded defect types for the investigated materials differing in glass fibre content (15 wt% and 30 wt%) and expected main fibre orientation (FO 0° - fibres oriented mainly in direction of applied force and FO 90° - fibres oriented perpendicular) and (c) shows a diagram with the total number of defects at the last loading step before breakage.

This work was supported by the K-Project "Photonic Sensing for Smarter Processes" (PSSP), the COMET program of FFG and the federal government of Upper Austria and Styria.

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

DYNAMIC RESPONSES OF SANDWICH COMPOSITE BEAM WITH PVB VISCOELASTIC CORE UNDER MOVING LOAD

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Keywords: Vibration analysis, Sandwich beam, Viscoelastic material, Moving load, Finite element
Session topics: Impact resistance

ABSTRACT

In this present study, a numerical approach with a high-order theory is adopted to study the vibratory behavior of the symmetrical sandwich beams with a viscoelastic core under moving load by considering the frequency dependence of viscoelastic properties. The longitudinal and rotational are considered by applying the theory of Euler-Bernoulli to the faces and Timoshenko's theory to the viscoelastic core. The numerical asymptotic method is employed in order to solve the complex eigenvalue problem.
The linear vibration of viscoelastic beams under a moving load is conducted in the context of small deformations. The hypotheses considered by Bilasse [1] are modified in order to take into account the effect of longitudinal and rotational inertia as well as the sandwich asymmetry.

The global matrix system describing the vibratory behavior of sandwich beam after assembly of the elementary matrices is written in the form:

$$[M]\ddot{U} + [K(\omega)]U = \{F(vt)\} \tag{1}$$

where **[M]** and **[K]** are respectively the global mass and frequency dependent stiffness matrices, **{F}** is the nodal force vector, **U** is the nodal displacement vector.
The motion equation that describe the vibratory behavior in the time domain can be obtained by decomposing the stiffness matrix into two parts $\boldsymbol{K(\omega) = K^R + iK^I}$ with $\mathbf{K^R}$ and $\mathbf{K^I}$ are real and imaginary parts of the stiffness matrix. By replacing the new matrix **K(ω)** in (1), the following expression is obtained:

$$[M]\ddot{U} + ([K]^R + i[K]^I)U = \{F(vt)\} \tag{2}$$

Using the property $\dot{U} = i\omega U$, Eq. (16) becomes:

$$[M]\ddot{U} + [C]\dot{U} + [K]^R U = \{F(vt)\} \tag{3}$$

where $[C] = \frac{[K]^I}{\omega}$ is the equivalent damping matrix [2].
The solution of the Eq. (3) can be determined by applying the Newmark integration method.
The resolution of the complex eigenvalues problem Eq. (4) can be found by applying the set of techniques of the asymptotic numerical method.

$$([K(\omega)] - \omega^2[M])U = 0 \tag{4}$$

The viscoelastic model in the present work is considered dependent on frequency, which fractional derivative model describes the operational modulus of the Butyral polyvinyl viscoelastic material (PVB) considered at 20° [3]. The configuration and geometric properties of the sandwich beam considered are given in Figure 1 and Table1.

Table 1. Mechanical and geometrical properties of the viscoelastic sandwich

	Composite face	Viscoelastic Core
Young's modulus (Pa)	$E_{11} = 14.7 \times 10^{10}$ $E_{22} = 9 \times 10^{9}$ $G_{12} = 5 \times 10^{9}$	Eq.(5)
Poisson's ratio υ	$\upsilon_1 = 0.3$	$\upsilon_2 = 0.49$
density(Kg/m^3)	$\rho_1 = 1580$	$\rho_2 = 970$
Thickness (m)	$h_1 = (H - h)$	$h_2 = 2h$
	H=0.012; h=0.0012	
Length(m)	L=0.8	
Width (m)	b=0.02	

The shear modulus of the PVB viscoelastic material is given by:

$$G_c{}^*(\omega) = [G_\infty + (G_0 - G_\infty)[1 + (i\omega\tau)^{1-\alpha}]^{-\beta} \tag{5}$$

with: $G_0 = 479 \times 10^3 Pa$;$G_{inf} = 2.35 \times 10^8 Pa$; $\tau = 0.3979$; $\alpha = 0.46$; $\beta = 0.1946$;

The dynamic responses of the sandwich beam for different thickness ratio values are shown in figure (1.a). The dynamic response is composed of two responses, forced for 0≤x/L≤1 and free for t> x/L, where x/L represents the ratio of the position of the load, where x/L=1 corresponds to the time required for the load passes completely through the beam. It is observed that the reduction in the thickness of the composite face layers caused the increase in the amplitudes of the deflection responses in the forced vibration region corresponding to x/L≤1. This evolution is due to the simultaneous variations of the stiffness and the equivalent mass very especially for the face layers which aim to provide reinforcement to the central viscoelastic layer. However, it can be seen that the structure damping improves as the viscoelastic layer becomes thinner. That is validated in Figure (1.a) which free vibrations for h/H = 0.1 are disappeared faster compared to other thickness values. The dynamic amplification factor (DAF) responses for different values of h/H = 0.1, 0.3, 0.6 and 0.8 are presented in Figure (2.a). This factor is defined as the ratio between the maximum value of the dynamic displacement under moving load and the static displacement under concentrated force. The maximum values of DAF are reached at critical speeds at V_c=150,140,120 and 110 (m/s). It is observed that the critical speed decreases when the ratio h/H increases. This is relevant to the proportionality of critical speed with the natural frequencies which has a considerable impact on the vibratory responses because the low natural frequencies generally generate high amplitudes of displacement.

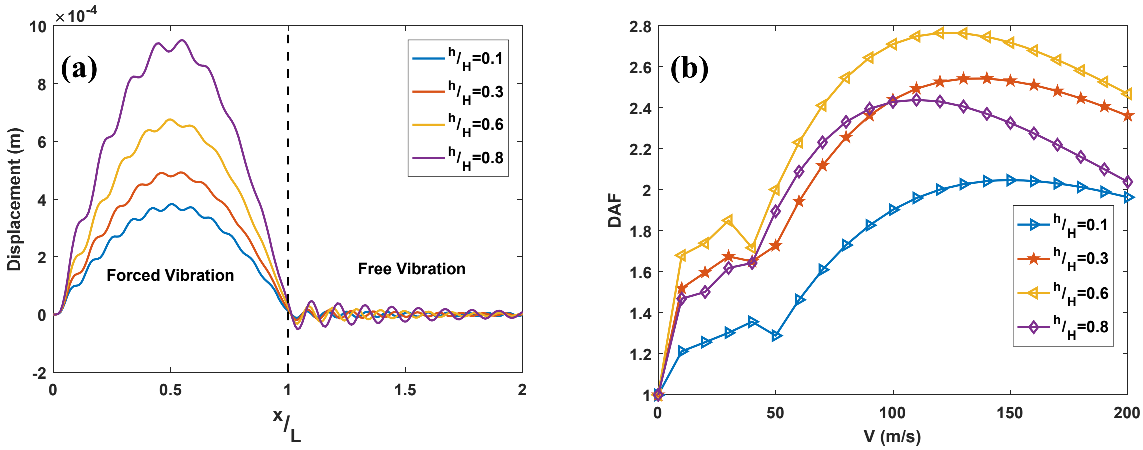

Figure 1. Dynamic responses of the viscoelastic sandwich beam under moving load for different values of h/H. ((a). transverse displacement, (b). dynamic amplification factor)

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FIRE PROPERTIES OF CFRP AND GFRP

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Keywords: Fibre-reinforced polymers, thermal behaviour, properties, cone calorimeter
Session topics: Delamination, Discontinuous fibre composites, Novel experimental techniques, Constituent properties, Thin ply composites.

ABSTRACT

The exhibited results are a part of an evaluation study to recover fibres from waste CFRP and GFRP by incineration. This paper aims to presents the heat release characteristics, char formation and fire properties of the waste CFRP and GFRP composites. The materials used in the experiment were analysed using a cone calorimeter in line with the ISO 5660 standard. All the measured data were processed using ConeCalc software provided by Fire Testing Technology UK. The materials for the study were provided by two companies in Finland, namely Exel composites and Patria aerostructures Oy. The material details are tabulated in Table 1. The samples were prepared according to the standard dimension (100X100X10 mm^3) using a cutting machine. A uniform sample size of 100mm length and 100mm height was maintained throughout the process. However, the height of the sample varied from specimen to specimen, not exceeding 10mm. Three samples from carbon fibre A, two from carbon fibre B and a single sample from glass fibre were taken for the study.

Table 1: Specimen information.

Name	Type/ used in	Composition	Nature	Suppliers
Carbon fibre A (CFRP)	Domestic applications	Carbon fibre + epoxy	Unidirectional	Exel composites
Carbon fibre B (CFRP)	Aeronautical grade	Carbon fibre + epoxy	Laminated thin ply	Patria Oy
Glass fibre (GFRP)	Domestic applications	Glass fibre + polyester	Laminated thin ply	Exel composites

All the samples were burnt horizontally using a cone calorimeter equipment according to ISO 5660-1 standard. The samples were wrapped utilising an aluminium foil exposing only the top surface to the heat flux placing in a retainer frame. The distance between the cone heater and the samples was 25mm. Initially, irradiation of 11 kW/m^2 was used and later increased to 50 kW/m^2. The exhaust system flow rate was 24 L/s. The test result information of heat flux, peak heat release rate (pHRR), total heat release (THR), mass loss rate (MLR), effective heat of combustion (EHC) are in Table 2.

Table 2: ConeCalc software results

type	Sample type	Heat flux (kW/m^2)	pHRR (kW/m^2)	THR (MJ/m^2)	MLR (g/s·m^2)	EHC (MJ/Kg)
Carbon fibre A	sample 1	50	225.23	73.63	2.14	21.35
	sample 2	50	213.75	86.47	2.46	22.01
	sample 3	50	248.14	90.32	2.52	22.20
Carbon fibre B	sample 1	50	224.53	53.85	4.18	7.98
	sample 2	11	6.85	3.56	0.63	1.38
Glass fibre	sample	11	5.97	3.77	2.59	5.21

The glass fibre sample and carbon fibre B sample 2 was tested in a lower heat flex environment to have an initial observation. As a result, the versatile compounds evaporated leaving a minor char

formation on the edges and bottom surface of the glass fibre sample. However, the carbon fibre B sample 2 showed no reaction; the aeronautical grade of the material can be a reason. Later, once the heat flux was increased and stabilised at 50 kW/m^2 (as per the standard), all the samples showed a significant change resulting in a delaminated structure. Also, the unexposed edges and bottom surface of the samples had minor char formation. The HRR curve profiles of the tested samples as a function of time are shown in figure 1. Excluding the samples tested in low irradiation, the four HRR curve profiles have nearly identical shapes. The initial sharp peaks are because of combustion of the volatiles released by the samples. The sudden drop in the peaks is because of the char layer formation on the material surface, which can slow down the decomposition under the newly formed char layer. The second peak indicates the cracking of the char layer and the burning of the remaining material or the sample materials thermal feedback effect [1]. Higher the thermal conductivity higher the peak [2], carbon fibre A has a higher peak compared to carbon fibre B.

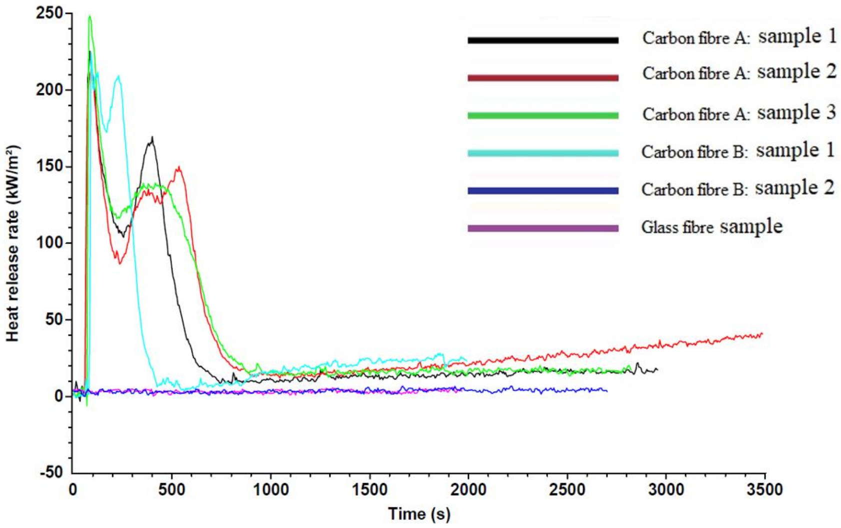

Figure 1: Heat release rate for the samples

The HRR curve profile of carbon fibre B sample 1 vaguely deviates from the rest of the carbon fibre samples. The aeronautical grade carbon fibre seems to have achieved an early second peak. A thinner char layer formation can be a reason for such an effect. Also, THR for carbon fibre B sample was lower compared to carbon fibre A and glass fibre, but MLR was vice-versa. Overall, the glass fibre composite was highly reactive to fire compared to both the carbon fibre composites. The carbon fibre B composite was slightly superior by reacting the least to fire compared to carbon fibre A material. At the end of the test, visually appealing fibres were recovered from the retainer frame shown in figure 2. Our future interests are in testing the recovered fibres, recycle the fibres and evaluate their physical properties. The research is aimed to be published in a scientific journal.

Figure 2: recovered fibres after the burning

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FiBreMoD Conference
L. Gorbatikh, S.V. Lomov, Y. Swolfs (Editors)
Leuven, 11-12 December 2019

IMPROVING THE TRANSLAMINAR FRACTURE TOUGHNESS OF CARBON FIBRE COMPOSITES BY HYBRIDISATION WITH HIGH-PERFORMANCE POLYMER FIBRES

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Keywords: Polymer composites, High-performance polymer fibres, Damage tolerance, Translaminar fracture toughness, Compact tension
Session topics: Constituent properties, Fibre-hybrid composites, Fracture toughness, Notch sensitivity, Novel experimental techniques

ABSTRACT

Carbon fibre composites are gaining more and more interest due to their excellent mechanical properties. Their high specific strength and stiffness are ideal for applications in motion, such as the aviation, aerospace and automotive industry, where weight reductions lead to less fuel consumption, and, as a result, a decrease in the emission of greenhouse gasses. This can help to slow down global warming. However, these applications are also damage critical, restricting the weight reductions achievable with conventional brittle carbon fibre composites. In recent years, the translaminar fracture toughness has attracted interest as it is an important parameter governing the notch sensitivity and damage tolerance of composites [1]. One very promising method of improving the translaminar fracture toughness of carbon fibre composites is by hybridisation with tough fibres. Ductile metal fibres can plastically deform [2] while high-performance polymer fibres fibrillate upon failure [3], both offering additional energy dissipating mechanisms to the material, and therefore potentially strongly increasing the translaminar toughness of the composite. However, reports of the potential improvements on the translaminar fracture toughness of these tough fibres are very scarce in the literature.

The translaminar fracture toughness of three types of high-performance polymer fibre composites has been investigated; PBO fibres (tradename Zylon, of Toyobo), PAR fibres (tradename Vectran, of Kuraray) and aramid fibres (tradename Twaron, of Teijin). T300 carbon fibres (of Toray) were used as the reference, and to examine carbon/high-performance polymer fibre hybrids. Compact tension tests were performed on the in-house manufactured composites. The in-house manufacturing of the composites allowed for optimal control over the epoxy matrix system and the yarn placement of the different designs that were investigated. Different hybridisation designs were examined, such as a yarn-by-yarn intralayer hybrid and a single strip of one fibre type in a ply of another fibre type. The compliance calibration method with finite elements was used for data reduction.

The propagation translaminar fracture toughness (see Figure 1) of PBO (564 kJ/m²) and PAR (439 kJ/m²) fibre composites reported is, to the best knowledge of the authors, the highest value ever reported in the literature. The third type of high-performance polymer fibre composite, the aramid fibre composite, does not show this high toughness value. However, all three high-performance polymer fibre composites show a more gradual crack propagation, as opposed to the sudden, catastrophic failure experienced with the carbon fibre reference. The strip samples display the increase in translaminar fracture toughness achievable by strategically placing high-performance polymer fibre strips in carbon fibre composites. The strip still contributes to resisting the crack from growing due to fibre bridging, even after the crack tip propagates beyond the end of the strip, see Figure 2. Nevertheless, the width of these strips is not wide enough to reach the full potential the high-

performance polymer fibres offer, as the steady state performance is not reached when the crack tip reaches the end of the strip.

These compact tension test results of composites containing tough fibres display the improvements on the translaminar fracture toughness that can be attained. These results can be used to further optimise the application of composite materials in weight and damage sensitive applications, leading to safer materials.

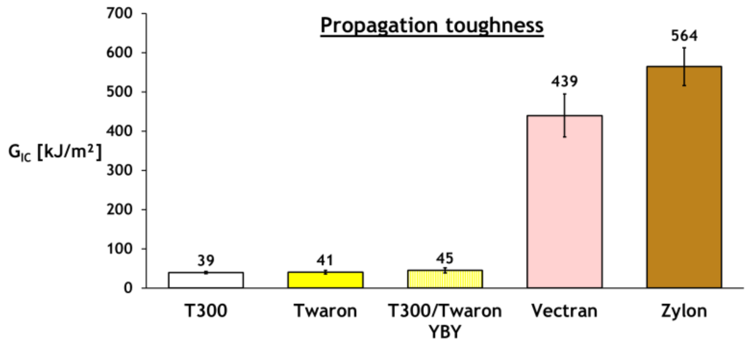

Figure 1: The propagation translaminar toughness values of the non-strip samples.

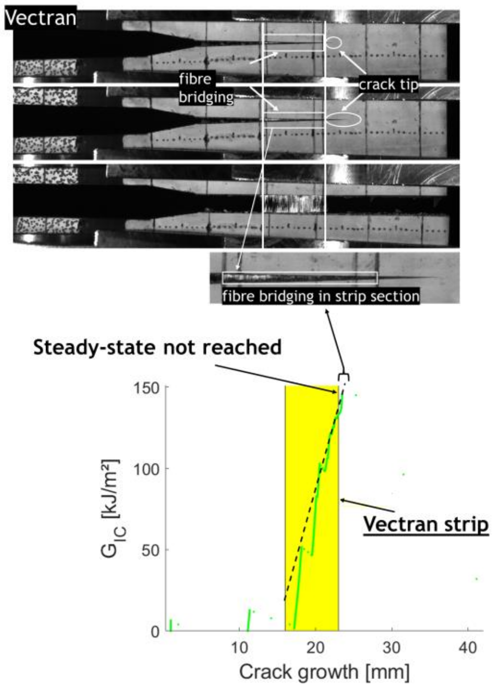

Figure 2: Example of the influence of a Vectran fibre strip in a carbon fibre ply on the translaminar fracture toughness. The strip still contributes to resist crack growth after the crack tip propagated beyond the strip region.

The authors gratefully acknowledge SIM and VLAIO for their support of the SBO project RELFICOM, running in the NANOFORCE Program. YS acknowledges FWO Flanders for his postdoctoral fellowship.

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