Late-stage fatigue damage in a 3D orthogonal non-crimp woven composite: an experimental and numerical study

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Abstract

Late-stage fatigue damage of an E-glass/epoxy 3D orthogonal non-crimp textile composite loaded in the warp direction has been investigated using a combination of mechanical testing, X-ray micro computed tomography (μ CT), optical microscopy and finite element modelling. Stiffness reduction and energy dissipated per cycle were found to be complementary measurements of damage accumulation, occurring in three stages:

a first stage characterised by rapid changes, a more quiescent second stage, followed by a third stage where the (decreasing) stiffness and (increasing) energy dissipation change irregularly and then rapidly, to failure. Microscopy of specimens cycled into the transition between the second and third stages showed macroscopic accumulations of fibre fractures in sections of warp tows which lying adjacent to the surface weft tows which are crowned-over by the Z-tows. At these locations, the warp tow fibres are subjected to stress concentrations both from transverse weft tow matrix cracks and resin pocket cracks.

Keywords: *A. Fabrics/textiles; B. Fatigue; C. Finite element analysis (FEA), D.*

Radiography

1. Introduction

Laminated composite materials based on 2D woven fabric plies having fibre reinforcement in two orthogonal directions have found ready applications in complex engineering structures. However, with only the polymer holding adjacent plies together, delamination is a common cause of failure, and a number of remedies have been proposed to enhance through-thickness properties, such as stitching and z-pinning [1,2]. As reviewed in [3], a further option is to use 3D reinforced fabric preforms which have the advantage of incorporating through-thickness (Z-direction) reinforcing yarns that provide, in a controlled fashion, an extraordinary resistance to delamination unmatched by any other translaminar method of reinforcement. Three variations of 3D warp-interlock woven preforms are available at present: orthogonal, through-thickness angle interlock and layer-to-layer angle interlock. Of these variations, the orthogonal preforms

(where through-thickness yarns pass through the entire preform thickness along a nearly vertical path) generally yield the highest volume fraction for the composite leading to the best in-plane mechanical properties. One particular preform type was termed “non-crimp 3D orthogonal weave” in [3 - 6] and “3D non-crimp orthogonal weave” in a more recent publication [7]; we will use the latter term, and its acronym 3DNCOW, here.

This type of preform has already found numerous civilian and military applications, and many more are envisioned in the future, in areas such as ballistic and blast protection (e.g. in personnel and vehicle armour shields), marine structures, wind energy (wind turbine spar caps), aircraft structures (from small components to primary load-bearing structures) and in civil engineering (decks, beams, columns).

To date, however, there has been limited progress both with understanding the mechanical behaviour of 3DNCOW composite materials and with computational modelling due to their geometrical complexity. Nevertheless, a number of studies have contributed to an increasing understanding of their internal geometry and behaviour for quasi-static loading cases (see, for example [4 - 9]), with a growing number of studies on fatigue loading of 3DNCOW composites and related 3D structures [10 -16]. Several studies of textile composites have used sectioning and microscopy of sacrificial specimens to investigate damage development, with an increasing use of X-ray micro computed tomography (μ CT) to characterise the actual microstructure and damage accumulation behaviour so that comparisons can be made with modelling predictions [17-20].

With regard to modelling, a wide range of strategies for predicting different mechanical properties of 3D woven composites (e.g. elastic properties; impact behaviour) were reviewed in [21]. Fewer publications are available on numerical

simulations of progressive failure and strength predictions of 3D woven composites under quasi-static loading [9, 22]. Of relevance to the current paper, to the best of the authors' knowledge, there are no published works available providing a theoretical basis for understanding progressive damage development in 3D woven composites under fatigue loading.

This work is concerned with contributing to the understanding of the fatigue failure of 3DNCOW composites by focussing on the development of late-stage fatigue damage; interpretation of the experimental results is assisted by a 3D finite-element analysis.

2. Experimental methods

The textile preform used in this study was a single-ply 3DNCOW E-glass designated "3D-78" [4] with an areal density of $2640 \text{ g}\cdot\text{m}^{-2}$; this type of preform has been used previously in experimental studies [4, 5]. The preform was manufactured by 3TEX. The fibre architecture consists of three layers of weft tows interlaced by Z-tows, and two layers of warp tows. The resin used was a Shell Epikote 828 (Bisphenol-A) epoxy resin with a Shell epicure nadic methyl anhydride (MNA) curing agent and Ancamine K61B accelerator. The PPG Hybon 2022 E-glass fibres in each tow have a very similar refractive index to that of the resin matrix, enabling highly transparent specimens (with a very low void content) to be produced using the wet impregnation technique described in [23]. The total fibre volume fraction of the 3DNCOW composite, measured using the matrix burn-off test, was about 0.47. Test coupons were cut parallel to the warp direction with dimensions 200 mm (length) x 2.2 mm (thickness), and nominal widths of 8 mm, 12 mm and 24 mm; for detailed late-stage fatigue damage observations, the

specimen width of 8 mm was chosen to reduce the ratio of the specimen width to thickness to facilitate the μ CT studies, while retaining a reasonable specimen width.

Quasi-static tensile and fatigue testing were carried out using an Instron series 8800 servo-hydraulic test machine using 40 mm GFRP end tabs that were bonded onto each specimen, so that the final gauge length was 120 mm. An extensometer with a gauge length of 96 mm was used since this enhanced the likelihood of the final failure of the specimens occurring within the gauge length of the extensometer. Fatigue testing was carried out at a frequency of 5 Hz and an R-ratio of 0.1. For the damage inspections, a combination of optical microscopy and X-ray μ CT (Nikon XTH 225; voxel size 3 μ m) has been used.

3. Mechanical test and damage observation results

3.1 Mechanical behaviour

Specimens were loaded in tension under quasi-static loading (in the warp direction); a typical stress-strain curve to failure obtained in the present study is shown in Fig. 1 and the inset figure shows a schematic of the fibre architecture and the direction of loading. The Young's modulus (~ 24 GPa), ultimate tensile strength (~ 420 MPa) and the strain to failure ($\sim 2.4\%$) were in good agreement with the respective values reported in [4]; the work reported in [4] used the same type of 3DNCOW preform and a similar fibre volume fraction, although a different resin matrix (vinyl ester) and a different method of specimen manufacture were used.

The fatigue stress-life (S-N) curve for the three specimen widths is shown in Fig. 2; there were no significant differences in the fatigue lifetimes of the 8 mm wide specimens compared to the 12 mm and 24 mm wide specimens, and no noticeable

changes to the mode of failure. A low magnification image of a 12 mm wide coupon cycled close to final failure (with a peak stress of 200 MPa) is shown in Fig. 3. In addition to matrix cracking damage, short splits aligned approximately parallel to the loading direction can be seen, as well as interfacial debonding of the Z-tows. Similar damage has been observed previously for quasi-static and fatigue loading of 3DNCOW composites [5,7,11]. As indicated above, for this investigation of late-stage fatigue damage in “3D-78” specimens, the 8 mm wide specimens were investigated in detail; the lowest peak stress tested (175 MPa) was used which gave a fatigue lifetime in the range 50,000 – 150,000 cycles. Fig. 4 shows a typical plot of the normalised stiffness of the specimens as a function of number of fatigue cycles for a specimen which failed after 148,800 cycles, together with a plot of the energy dissipated per cycle. The (normalised) secant stiffness reduction has been calculated using the peak-to-peak stress and strain values of the hysteresis loops; examples of these loops are shown in Fig. 5. The (normalised) tangent stiffness reduction is also shown in Fig. 4, where the stiffness has been measured from the low-strain region of the loading curve over a strain range of 0.1%. In this low-strain region, the stress-strain curve is nearly linear and the tangent stiffness has been calculated by fitting a straight line to the stress-strain data over the range $(\epsilon_{\min} + 0.03\%)$ to $(\epsilon_{\min} + 0.13\%)$, where $\epsilon_{\min}$ is the minimum strain in the fatigue cycle. As can be seen in Fig. 5, the minimum strain increases with fatigue cycling, presumably due to the release of residual thermal stresses as cracking damage develops. This method of measuring the stiffness follows the minimum strain while ensuring that the stiffness is always measured over the same strain range (0.1%) as the minimum strain increases; this is equivalent to measuring the stiffness always over a stress range

of about 25 MPa to 50 MPa. The energy dissipation per cycle is given by the area of the hysteresis loops.

Fig. 4 shows that changes to the normalised stiffness and to the energy dissipation per cycle are complementary. In the first stage, a rapid stiffness reduction is accompanied by a high, but rapidly reducing energy dissipation per cycle; probably this effect is the consequence of the development of rapidly accumulating matrix cracking in the specimen. The second stage, beginning at about 15,000 cycles for this load level, is characterised by a gradual loss of stiffness accompanied by a slow increase in the energy dissipation per cycle. In the third stage, rapid and irregular reductions in the specimen stiffness occur, which are mirrored by the rapid and irregular increases in the energy dissipation. This leads to a final, rapid reduction in stiffness and a sharp increase in the energy dissipation, culminating in the complete failure of the specimen. A distinguishing feature of the final few cycles of the specimen life is the audible sound of fibre fractures.

Interestingly, close inspection of the hysteresis loops of Fig. 5 shows why the secant stiffness reduction is a better characteristic than the tangent stiffness reduction for capturing the changes in the behaviour of the specimen, especially in the later stages of the fatigue life. This is because after extensive cycling, both the loading and unloading parts of the hysteresis loops at higher strains become increasingly non-linear with increasing damage, whereas the low-strain portions of the loops change more slowly. Similar hysteresis loop shape changes have been noted during the fatigue cycling of off-axis glass/carbon hybrid laminates with fibre angles of $\pm 10^\circ$ [24], where the loop shape changes were attributed to viscoelastic deformation of the matrix in the

early stage of fatigue, and stick-slip frictional sliding in the latter stages, near failure; it is possible that a similar explanation applies here.

Using the energy dissipation curves as an indication that specimens had entered the third stage of their fatigue life, fatigue tests were interrupted prior to ultimate failure. This was achieved by a real-time analysis of the energy dissipated per cycle in the fatigue test (i.e. area of the hysteresis loops) and noting that the energy dissipation increased rapidly just prior to failure, as can be seen in Figure 4. The accumulated damage was then investigated using optical microscopy and μ CT and these results are presented in the next section.

3.2 Damage observations

The focus of this work is late-stage fatigue damage. However, some details are provided here on the development of matrix cracking damage in the early stages (as discussed in [25]) since, as will be seen later, these cracks play an important role in final failure. Figure 6 shows the development of damage in a section of a 24 mm wide specimen cycled with a peak stress of 200 MPa for the first 8000 cycles during which the majority of the stiffness reduction occurs; the normalised stiffness and energy dissipation behaviour for this specimen over the first 8,000 cycles, and for the complete life of the specimen (173,098 cycles), is shown in Figure 7. Figure 6 shows a gradual accumulation of weft tow matrix cracks, both within the weft tows and in the resin channels between weft tows; there is also Z-tow interfacial debonding at the locations where the resin channel cracks intersect the Z-tows, as previously observed [11]. In addition to the matrix cracks, tow-level interfacial fractures (i.e. micro-delaminations)

can be seen to have initiated from about 200 cycles (Figure 6(a)), becoming a significant feature of the damage from about 500 cycles (corresponding to matrix crack densities above about 0.3 mm^{-1} (Figures 6(b) to 6(d)). The normalised stiffness as a function of matrix crack density is shown in Figure 8 for two specimens cycled with a peak stress of 200 MPa, where the crack density (the cracked area per unit volume) was measured over a specimen length of 85 mm; the large error bars reflect the difficulty of estimating the crack density from plan view images. In Figure 8, the data points labelled (a) to (d) correspond to the images in Figures 6(a) to 6(d).

Figure 9(a) shows a low-magnification image of one of the transparent 3DNCOW specimens which was cycled at a maximum stress of 175 MPa into the third stage of the fatigue lifetime. To aid the understanding of the microscopy, a schematic of the architecture of the “3D-78” composite specimens is shown in Fig. 6(b), with two vertical through-thickness planes (1 and 2) shown in the vicinity of a Z-tow; these vertical planes are useful in understanding the locations of the micrographs discussed below.

An X-ray μ CT image of the plane labelled as “1” in Fig. 9(b) is shown in Fig. 10(a). The presence of extensive fibre fractures can be seen in the warp tow closest to the Z-crown (the Z-crown is that portion of a Z-tow which, in passing over a weft tow, lies parallel to the warp direction). These fibre fractures are located towards the edge of the warp tow as can be seen in Fig 10(b) which shows the X-Y plane at the location of the fibre fractures. Such accumulations of warp tow fibre fractures were observed at similar locations in other parts of the same specimen, and different specimens, and appear to be a characteristic feature of late-stage fatigue damage in this material (no similar accumulations of fibre fractures were observed in specimens cycled into stages 1 or 2).

Interestingly, no fibre fractures were found within the equivalent section of the lower warp tow at the same location, and optical microscopy provides an indication of the reason for this. Fig. 11(a) shows a reflected light photomicrograph of the same specimen as the μ CT image of Fig. 10(a), but about 0.5 mm further from the edge of the warp tow; a similar cluster of warp tow fibre fractures can be seen. Immediately above the cluster of fibre fractures, there is a matrix crack in the surface weft tow. Fig. 11(b) shows the same X-Z location as Fig. 11(a), but at position $Y = 0$ (i.e. in the plane labelled “2” in Fig. 9(b)). In this plane, the warp tows cannot be seen and the gap between the weft tows is filled by matrix alone (a resin pocket). It is evident that the matrix crack in the resin pocket is an extension of the matrix crack in the surface weft tow, and that the crack continues through the middle weft tow, through the lower resin pocket, and into the lower weft tow (a schematic of the relationship between the matrix cracks and the warp tow fibre fractures is shown in Fig. 12). In addition, extensions of the same crack continue to both outer surfaces at those locations where there are no Z-tows on the surface. The location of the fibre fractures at the edge of the warp tow is likely to be a consequence of the combined effect of the weft tow matrix cracks and the resin pocket cracks. Indeed, the fibres towards the edge of the warp tow at these locations will experience a stress concentration from the weft tow matrix cracks and a stress concentration from the resin pocket cracks.

The cracking at the lower warp tow/weft tow interface visible in Fig. 11(a) is associated with a matrix crack. This interfacial cracking (highlighted in the figure with an ellipse) suggests a reason why fibre fractures are not seen in the lower warp tow. Such interfacial cracking reduces the stress concentration associated with the matrix crack, thus making fibre fractures at this location less likely. Explanations, both for the

location of the interfacial cracking and for the 20-25% stiffness reduction of stages 1 and 2 of the fatigue life, have been explored using the finite element analysis described in the next section.

4. Finite Element Analysis

A voxel (i.e. the volumetric equivalent of a pixel) FE model has been constructed as a first approximation of the complex 3D architecture, in order to investigate the effect of the matrix cracking on the properties of the composite and to try to understand the location of the warp tow fibre fractures. Voxel models enable a 3D geometry to be generated with prismatic hexagonal (brick) finite elements. An advantage of this type of mesh is that, unlike general 3D meshed models, the elements are always undistorted. However, a drawback is that the surfaces formed are not smooth, which may cause loss of material for the cases of features with curved edges (in this case, the Z-tows and their crowns). This topological problem is minimised by using small voxels, although there is always a compromise between model size and discretisation errors. In this work a model with 1.5 million elements was used (the reference mesh) which gave a practical run-time of 6 hours on the largest multi-processor unit available. In addition a coarser and finer mesh were used in a mesh refinement study.

An approximate representation of the microstructural arrangement of yarns in the real 3DNCOW composite was provided by developing a representative volume element (RVE; **Figure 13**) using Texgen [26]; the same element has been used in previous representations of the 3D architecture [e.g. 4]. A significant complication that arises in the construction of any FE model of the 3DNCOW composite fibre architecture is that

the yarn cross-sections are not constant throughout the structure, as has been noted by others [e.g. 10], and that the cross-sections are not regular in shape [27] (see Figs. 10 and 11, for example). In the voxel model, the cross-sections of the fibre yarns have been created as elliptical, with a narrower cross-section for the Z-tows. All materials have been modelled as linear elastic, the tows having orthotropic, and the resin isotropic, properties. The set of elastic properties of the isotropic epoxy matrix and the orthotropic resin-impregnated E-glass fibre tows are shown in Table 1, with the data taken from [28]. The total specimen thickness has been taken in the model to be the same as the experimental thickness (2.2 mm). As Table 2 shows, the use of constant elliptical cross-sections for the yarns produces fibre volume fractions which are not the same as the actual values **and** recent work [18] has shown that simulated geometry models making use of CT scan images can improve the representation of the fibre architecture. Nonetheless, the current model retains the essential features of the fibre architecture required to investigate the effect of increasing matrix cracking damage as well as the location of the warp tow fibre fractures in late-stage fatigue.

The dimensions of the RVE were 8.0 mm as measured in the warp direction and 8.35 mm as measured in the weft direction; it should be noted that the use of Texgen leads to a slightly larger RVE than the real geometry because of the way the programme generates the 3D structure. The voxel FE model produced a Young's modulus for the composite loaded in the warp direction of 16.8 GPa, which is significantly below the 24 GPa measured experimentally. The authors believe that the discrepancy between the predicted modulus and the experimentally-measured value is primarily due to a much lower total fibre volume fraction in the FE 3DNCOW composite model (36%) compared to the actual 3DNCOW composite material (48%). Scaling the predicted modulus to

account for the difference between the modelled and actual fibre volume fraction results in a modulus that is close to the measured modulus. This correlation between the experimental and the predicted modulus, and the good agreement between the experimental and predicted modulus degradation with damage (see below), lends confidence to the FE modelling.

A study of the effect of mesh size on the predicted modulus was also undertaken and the results are summarised in Table 3 for both a coarser and a finer mesh than the reference mesh. It can be seen that the modulus decreases only marginally (1%) from the coarsest (16.83 GPa) to the finest mesh (16.65 GPa). In summary, although the experimental and modelled fibre volume fractions are different, the model does reproduce the key features of the fibre architecture (i.e. representing the warp, weft and Z-tows) and the mechanisms identified later for the location of warp fibre damage will be applicable.

The voxel model has been used to predict the effect of an increase in the density of matrix cracks and resin pocket cracks on the modulus of the 3DNCOW composite material. The cracks have been generated by deleting the voxel elements permanently from the model and the locations for the cracks have been informed by the optical microscopy images (see Fig. 11). As found in the microstructural investigations of the specimens, the cracks in the model extend through the weft tows and around the warp tows as resin pocket cracks. Examples of the crack distributions used in the model are shown as insets to Fig. 14 which also shows a comparison of the predictions of the model for the normalised stiffness as a function of crack density compared with the experimental measurements of Figure 8 (in Figure 14, only the average crack density measurements are shown). Figure 14 shows that there is good agreement between

prediction and experiment up to a normalised stiffness of about 0.9 (i.e. a matrix crack density of about 0.3 mm^{-1}). As Figure 6 shows, a crack density of 0.3 mm^{-1} corresponds approximately to the onset of significant micro-delaminations in the composite which can be expected to contribute increasingly to the stiffness reduction as they grow in extent; the contribution of the micro-delaminations to the stiffness reduction has not been modelled here.

In the FE model, the crack has a small but finite width (one element). The effect of this finite width and model mesh density on the predicted stiffness reduction with cracking has also been considered. It was found that a smaller crack width resulted in a smaller stiffness reduction, whilst increasing the mesh refinement resulted in a larger stiffness reduction. The net result of these competing factors is that the stiffness reduction presented in Fig 14 is a reasonable approximation and this is reinforced by the good agreement with the experimental data.

Of most interest, the voxel model also provides an insight into the location of the clusters of warp tow fibre fractures. Previous work has shown that matrix cracking can lead to fibre fracture in adjacent 0° plies in cross-ply laminates and in non-crimp fabrics due to the stress concentrations associated with the matrix cracks [29, 30]; the occurrence of delaminations at ply interfaces would reduce the stress concentrations caused by the matrix cracks. In the 3D composites, the voxel model suggests that the through-thickness (σ_{zz}) peel stresses at the interface between the faceted warp tows and the surrounding (matrix) material are lower for the surface weft tow close to the Z-tow crown (by about 3% towards the middle of the tow, and by about 50% towards the edge of the tow; see Fig. 15). This suggests cracking at the tow interface, which would reduce the stress concentration caused by an adjacent matrix crack, is less likely to

occur for the surface weft tow close to the Z-crown; such interface cracking is more likely to occur for the opposite surface weft tow at this location, as observed in Fig. 11(a). Although mesh refinement affects the values of these interfacial peel stresses, a study of the results from the reference and refined meshes showed that the asymmetric distribution of peel stresses was maintained with increasing mesh refinement, such that delamination is always more likely to occur on the weft tow/warp tow interface furthest from the Z-crown.

5. Concluding remarks

Late-stage fatigue damage in “3D-78” 3D non-crimp orthogonal woven composite loaded in the (longitudinal) warp direction is characterised by the prior development of extensive matrix cracking, which includes matrix cracking within the (transverse) weft-directional resin impregnated tows, matrix pocket cracking and short longitudinal splitting cracks (parallel to the loading direction), as well as local separation between different tows at their mutual interfaces. In terms of the lifetime of the specimen, the most significant feature of the third stage of damage is an accumulation and localisation of fibre fractures, developing from the edges of warp tows. Overall, stiffness reduction curves for this textile composite have a strong similarity to stiffness reduction curves for a cross-ply laminate, following the classic pattern of a rapid reduction in Young’s modulus initially followed by a gradual reduction, and then a rapid loss of modulus in the third stage. Energy dissipation monitored through the area of the hysteresis loops shows a complementary progression, with a stage of rapid energy dissipation being followed by a stage where energy

dissipation increases slowly, and a final stage where energy dissipation rises irregularly and then rapidly to failure.

Optical microscopy and μ CT observations have shown that clusters of fibre fractures occur in late-stage fatigue at the edge of warp tows closest to the Z-tow cross-over (the Z-crown) at the location where the fibres in the tow are subjected to stress concentrations from weft tow matrix cracks and resin pocket cracks; by contrast, cracking at the tow interface occurred at the edge of the warp tow furthest from the Z-crown, with no clusters of warp tow fibre fractures at this location.

An FE model (voxel) constructed and applied for the purpose of gaining some insight into the consequence of the damage development, showed a decreasing normalised stiffness with increasing density of transverse matrix cracks in weft tows and in matrix (resin) pockets. The model also suggests a possible explanation for the development of cracking at the warp tow/weft tow interface furthest from the Z-crown, thus reducing the likelihood of fibre fractures at this location.

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References

- [1] Dransfield KA, Baillie C, Mai YW. Improving the delamination resistance of CFRP by stitching - A Review. *Comp Sci Technol* 1994;50(3):305-17.

- [2] Mouritz AP. Review of z-pinned composite laminates. *Compos Part A* 2007;38(12):2383-97.
- [3] Bogdanovich AE, Mohamed MH. Three-dimensional reinforcements for composites. *SAMPE J* 2009;45(6):8-28.
- [4] Lomov SV, Bogdanovich AE, Ivanov DS, Mungalov DD, Karahan M and Verpoest I. A comparative study of tensile properties of non-crimp 3D orthogonal weave and multi-layer plain weave E-glass composites. Part 1: Materials, methods and principle results. *Compos Part A* 2009;40:1134-43.
- [5] Ivanov DS, Lomov SV, Bogdanovich AE, Karahan M and Verpoest I. A comparative study of tensile properties of non-crimp 3D orthogonal weave and multi-layer plain weave E-glass composites. Part 2: Comprehensive experimental results. *Compos Part A* 2009;40:1144-57.
- [6] Karahan M, Lomov SV, Bogdanovich AE, Mungalov D, Verpoest I. Internal geometry evaluation of non-crimp 3D orthogonal woven carbon fabric composite. *Compos A* 2010; 41(9):1301-11.
- [7] Bogdanovich AE, Karahan M, Lomov SV and Verpoest I. Quasi-static tensile behavior and damage of carbon/epoxy composite reinforced with 3D non-crimp orthogonal woven fabric. *Mech Mater* 2013;62:14-31.
- [8] Kuo W-S, Ko T-H, Chen C-P. Effect of weaving processes on compressive behaviour of 3D woven composites. *Compos Part A* 2007;38:555-65.
- [9] Bogdanovich AE. Multi-scale modelling, stress and failure analyses of 3-D woven composites. *J Mater Sci* 2006;41(20):6547-90.
- [10] Mouritz AP. Tensile fatigue properties of 3D composites with through-thickness reinforcement. *Compos Sci Technol* 2008;68(12):2503-10.

- [11] Carvelli V, Gramellini G, Lomov SV, Bogdanovich AE, Mungalov DD, Verpoest I. Fatigue behaviour of non-crimp 3D orthogonal weave and multi-layer plain weave E-glass reinforced composites. *Compos Sci Technol* 2010;70(14):2068-76.
- [12] Karahan M, Lomov SV, Bogdanovich AE, Verpoest I. Fatigue tensile behavior of carbon/epoxy composite reinforced with non-crimp 3D orthogonal woven fabric. *Compos Sci Technol* 2011;71(16):1961-72.
- [13] Rudov-Clark S, Mouritz AP. Tensile fatigue properties of a 3D orthogonal woven composite. *Compos Part A* 2008;39(6):1018-24.
- [14] Dadkhah MS, Cox BN, Morris WL. Compression-compression fatigue of 3D woven composites. *Acta Metall Mater* 1995;43(12):4235-45.
- [15] Tsai K-H, Chiu C-H, Wu T-H. Fatigue behaviour of 3D multi-layer angle interlock woven composite plates. *Compos Sci Technol* 2000;60(2), 241-8.
- [16] Gude M, Hufenbach W, Koch I. Damage evolution of novel 3D textile-reinforced composites under fatigue loading conditions. *Compos Sci Technol* 2010;70(1):186-92.
- [17] Desplentere F, Lomov SV, Woerdeman DL, Verpoest I, Wevers M, Bogdanovich A. Micro-CT characterization of variability in 3D textile architecture. *Compos Sci Technol* 2005;65:1920-30.
- [18] Green SD, Matveev MY, Long AC, Ivanov D, Hallett SR. Mechanical modelling of 3D woven composites considering realistic unit cell geometry. *Compos Struct* 2014;118:284-293.
- [19] Shigang A, Xiaolei Z, Yiqi M, Yongmao P, Daining F. Finite element modeling of 3D orthogonal woven C/C composite based on micro-computed tomography experiment. *Appl Compos Mater* 2013;21(4):603-614.

- [20] Yu B, Bradley RS, Soutis C, Hogg PJ and Withers PJ. 2D and 3D imaging of failure mechanisms of 3D woven composites. *Compos Part A* 2015;77:37-49
- [21] Ansar M, Xinwei W, Chouwei Z. Modeling strategies of 3D woven composites: A review. *Compos Struct* 2011;93(8):1947-63.
- [22] Bogdanovich AE. Progressive failure modeling and strength predictions of 3-D woven composites. In: *Proceedings of 50th AIAA/ASME/ASCE/AHS/ASC Structures, Structural Dynamics, and Materials Conference*, Palm Springs, CA, May, 2009, AIAA Paper 2009-2658, p. 1-20.
- [23] Belmonte HMS, Manger CIC, Ogin SL, Smith PA, Lewin R. Characterisation and modelling of the notched tensile fracture of woven quasi-isotropic GFRP laminates. *Compos Sci Technol* 2001;61(4):585-597.
- [24] Gamstedt EK, Redon O, Brondsted P. Fatigue dissipation and failure in unidirectional and angle-ply glass fibre/carbon fibre hybrid laminates. *Key Eng Mater* 2002;221/222:35-47.
- [25] Poole MC, Ogin SL, Smith PA, Wells GM, Potluri P and Withers PJ. Early-stage fatigue damage development of a non-crimp 3D orthogonal weave composite loaded in the warp and weft directions. In *Proceedings of the 20th International Conference on Composite Materials, ICCM20*, Copenhagen, July 2015.
- [26] Texgen; http://texgen.sourceforge.net/index.php/Main_Page
- [27] Green SD, Long AC, El Said BSF, Hallett SR. Numerical modelling of 3D woven preform deformations. *Compos Struct* 2014;108:747-756.
- [28] Kyriazoglou C, Guild FJ. Quantifying the effect of homogeneous and localized damage mechanisms on the damping properties of damaged GFRP and CFRP

continuous and woven composites – an FEA approach. *Compos Part A* 2005;36(3):367-79.

- [29] Jamison RD. On the interrelationship between fiber fracture and ply cracking in graphite/epoxy laminates. *Composite Materials Fatigue and Fracture*, ASTM STP 907, HT Hahn, Ed, American Society for Testing and Materials, 1986, pp252-273
- [30] Zangenberg J, Brondsted P, Gillespie JW. Fatigue crack propagation in unidirectional glass fibre reinforced composites made of a non-crimp fabric. *J Compos Mater* 2014;48(22):2711-27.

Fig. 1. Typical stress-strain curve for a 3DNCOW composite specimen loaded in the warp direction; the inset shows a schematic [4] of the fibre architecture of the warp tows (red), weft tows (blue) and the Z-tows (green) and the loading direction.

Fig. 2. S-N curve for the “3D-78” 3DNCOW composite specimens having different widths.

Fig. 3. Low magnification image of a section of a 3DNCOW composite specimen under load, which has been cycled close to failure, showing examples of matrix cracking (A), short splits (B; small ellipse) and Z-tow interfacial debonding (C; large ellipse); the loading direction is indicated by the arrows.

Fig. 4. Normalized tangent and secant stiffness (right-hand side axis) and energy dissipated per cycle (left-hand side axis) versus number of cycles for a 3DNCOW composite specimen cycled with a maximum stress of 175 MPa

Fig. 5. Examples of hysteresis loops for a 3DNCOW composite specimen tested with a peak stress of 175 MPa which failed after 148,800 cycles.

Fig. 6. Transmitted light optical micrographs of damage development in a section of a 24 mm wide specimen cycled at a peak stress of 200 MPa, after a) 200 cycles, b) 500 cycles, c) 1000 cycles, d) 8000 cycles.

Figure 7. Normalised stiffness and energy dissipation as a function of cycle number for the specimen of Figure 6. (a) First 8,000 cycles; (b) complete life of the specimen.

Figure 8. Normalised stiffness versus crack density for two specimens cycled with a peak stress of 200 MPa; (a) to (d) correspond to the images of Figure 6.

Fig. 9. (a) Low magnification optical micrograph (X-Y plane) of a section of a 3DNCOW composite specimen cycled with a peak stress of 175 MPa to the beginning of stage 3 (after 66,450 cycles); (b) schematic of the fabric architecture of “3D-78”; reference planes (1, 2) for the micrographs; the loading direction is indicated by the arrows.

Fig. 10. μ CT sections showing extensive fibre fractures towards the edge of warp tows. (a) Dispersed fibre fractures (observed close to the X-Z plane) located at the edge of a warp tow; the edge of the Z-tow can just be seen. (b) Image of the X-Y plane at the same location, showing fibre fractures at the edge of the warp tows adjacent to a resin pocket; the loading direction is indicated by the arrows.

Fig. 11. (a) Optical micrograph of sectioned and polished 3DNCOW composite specimen showing the same tow fracture as seen in Fig. 6 (i.e. the X-Z plane); the loading direction is indicated by the arrows. (b) Optical micrograph at the same X-Z location but for the plane $Y = 0$ (i.e. the plane of the z-tow). The arrow indicates the matrix crack in the surface weft tow which continues through the resin pocket; the ellipse indicates cracking at the warp tow/weft tow interface of the lower warp tow associated with a matrix crack in the adjacent weft tow.

Fig. 12. Schematic diagram illustrating the relationship between the matrix cracks (at the surface, within the weft tows and within resin pockets), and the fibre fractures at the edges of warp tows, adjacent to resin pockets.

Fig. 13. The structure of fibre tows in the representative volume element (RVE).

Fig. 14. Normalised stiffness as a function of crack density for the 3DNCOW composite predicted by the FE model compared with experimental measurements.

Fig.15. Peel stresses at the tow interfaces for the upper warp tow near the Z-crown and the lower warp tow.

Tables

Table 1

The elastic material properties of the epoxy matrix and resin impregnated E-glass fibre reinforcement [28]

<i>Matrix material</i>	<i>Resin impregnated fibre material</i>		
$E = 2 \text{ GPa}$	$E_x = 46 \text{ GPa}$	$G_{xy} = 5 \text{ GPa}$	$\nu_{xy} = 0.3$
$\nu = 0.4$	$E_y = 13 \text{ GPa}$	$G_{xz} = 5 \text{ GPa}$	$\nu_{xz} = 0.3$
	$E_z = 13 \text{ GPa}$	$G_{yz} = 4.6 \text{ GPa}$	$\nu_{yz} = 0.42$

Table 2

Summary of the principal parameters of the 3DNCOW composite used in the voxel FEA model and from literature [4].

<i>Parameter</i>	<i>3D-78 FEA</i>	<i>3D-78 [4]</i>
Composite thickness (mm)	2.20	2.19 ± 0.05
Total fibre volume fraction in composite (%)	35.6	47.5
Fibre volume contents in a fabric ply: warp/weft/Z (%)	42.17/55.57/2.25	47.6/47.8/4.6

Table 3

Summary of the main properties of the three different meshes considered.

	<i>Fine mesh</i>	<i>Reference mesh</i>	<i>Coarse mesh</i>
Voxel counts	X = 120 Y = 120 Z = 180	X = 100 Y = 100 Z = 150	X = 50 Y = 50 Z = 75
Number of elements	2.59×10^6	1.50×10^6	187500
Young's modulus prediction	$E_x = 16.65 \text{ GPa}$	$E_x = 16.75 \text{ GPa}$	$E_x = 16.83 \text{ GPa}$