Evolution of damage during the fatigue of 3D woven glass-fibre reinforced composites subjected to tension–tension loading observed by time-lapse X-ray tomography

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Abstract

The development of fatigue damage in a glass fibre modified layer-to-layer three dimensional (3D) woven composite has been followed by time-lapse X-ray computed tomography (CT). The damage was found to be distributed regularly throughout the composite according to the repeating unit, even at large fractions of the total life. This suggests that the through-thickness constraint provides a high level of stress redistribution and damage tolerance. The different types of damage have been segmented, allowing a quantitative analysis of damage evolution as a function of the number of fatigue cycles. Transverse cracks were found to initiate within the weft after just 0.1% of life, followed soon after (by 1% of life) by longitudinal debonding cracks. The number and extent of these multiplied steadily over the fatigue life, whereas the spacing of transverse cracks along with weft/binder debonding saturated at 60% of life and damage in the resin pockets occurred only just before final failure.

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1. Introduction

Three dimensional (3D) composites were proposed over 40 years ago in an attempt to overcome the shortcomings of 2D laminates, by incorporating fibres into the through-thickness direction. 3D weaving offer significant manufacturing benefits as well as creating versatile textiles having a range of 3D architectures. Unsurprisingly, they have emerged as promising candidates for the load-bearing applications requiring not only high in-plane (x–y) properties, but also some degree of out-of-plane (z) integrity.

In addition, z-reinforcement plays an important role in improving the energy absorption capability of 3D woven composites compared with 2D laminates. Studies [1,2] have demonstrated that a unique energy absorption mechanism is inherent in the 3D woven composites. In the case of glass-fibre reinforced woven composites, energy is often dissipated by means of the extensive straining of z-reinforcement and frictional sliding between z-reinforcement and weft yarns. By contrast a different mechanism was found in carbon fibre composite where higher energy is required for fracture, debonding and pull-out of z-direction yarns [3]. Moreover, z-yarns have also been found to play a significant role in containing local damage [4]. For example, they delay interply delamination and maintain the structural integrity by promoting energy dissipation via tow splitting and fibre breakage. Furthermore, textile composites commonly display load sharing, whereby damage is distributed over the entire specimen [5]. Strain mapping has been used to verify the periodic distribution of meso-strain in the braided composite under tensile loading [6]. Similarly, a periodic pattern of damage was identified in a stitched carbon fabric reinforced composite in tensile tests [7]. In both studies, a Fourier analysis was implemented to evaluate the correlation between the strain and damage distribution with the geometrical features of the textile, showing a good agreement. These results have provided strong support for the hypothesis in this work that the woven nature of a 3D composite along with the through-thickness reinforcement might provide significant advantages in terms of evenly distributing damage under fatigue loading.

Despite the potential benefits there is currently a lack of knowledge about how damage evolves under fatigue loading. The aim of this study is thus to explore the nature and evolution of damage mechanisms in a 3D composite. To date, a range of non-destructive techniques (NDT) have been employed to evaluate the damage behaviour of woven composites. For instance, the acoustic emission (AE) signal has been correlated with characteristic damage events, such as the onset of transverse cracking and...
propagation of delamination \[8,9\]. Digital image correlation (DIC) has been employed to map the surface strain distribution and to identify locations of stress concentration \[6\]. X-ray micro-computed tomography has been used to image glass fibre composites in 3D \[10,11\], to undertake a quantitative analysis of the porosity distribution together with the interaction between damage and microstructure. In situ synchrotron radiation computed tomography (SRCT) has proven to be a useful tool in identifying damage progression in carbon fibre epoxy-notched laminates under tensile loading, involving transverse ply cracks, 0° splits and finally delamination \[12\]. In addition, an in situ SRCT experiment has also been performed to study fatigue damage in the carbon/epoxy cross-ply laminates to investigate the interaction between toughening mechanisms with crack propagation \[13,14\]. However, to date the evolution of fatigue damage for a 3D woven composite has not been followed by non-destructive means.

Here, a time-lapse experiment using X-ray computed tomography (CT) is conducted in an attempt to identify damage evolution of 3D modified layer-to-layer composites during fatigue loading. Previously we have used a combination of 2D and 3D imaging to characterise the nature of damage after fatigue in a modified layer-to-layer composite \[15,16\]. Here, damage evolution is monitored by X-ray tomography at different fractions of the fatigue life. This enables a quantitative analysis of the extent and progress of damage mechanisms along with a 3D visualisation of the damage distribution. Due to the complex 3D morphology of the fibre architecture, it is often not possible to distinguish the different mechanisms of damage merely by 3D visualisation alone. Consequently, an innovative algorithm for the automatic classification of fatigue damage has been proposed, providing new insights into the evolution of different types of fatigue damage in 3D woven composites.

2. Materials and experimental method

2.1. Fabrication of 3D woven composite specimens

The 3D woven modified layer-to-layer preform contains warp yarns, weft yarns and binder yarns, each comprising S-2 high strength glass fibres. The materials specification of the 3D preform is presented in Table 1. By contrast to typical layer-to-layer composites, additional warp yarns (in pink) are interlaced into the top and bottom layers (shown in Fig. 1). It is worth noting that binder yarns do not act solely as binder, but they are fibres simply aligned in the warp direction, with a higher than normal level of crimp to provide through-thickness constraint from layer to layer. As a consequence, the repeating unit is very similar to that of a plain weave structure on the surface, but consists of more yarns in depth. The dimensions of the unit cell are shown in the yellow rectangle in Fig. 1.

The 3D fabric was infused with epoxy resin in a vacuum bag using Vacuum Assisted Resin Infusion process. For a detailed description of the manufacturing process and the sample preparation method refer to \[15\]. Tensile specimens having the dimensions given in Fig. 2 were cut from the 1.7 mm thick composite panel, with the tensile axis parallel to the warp direction.

The sample dimensions were determined by two considerations. On the one hand, they must be large enough to ensure that a sufficient number of unit cells are present in the gauge region so that the sample can exhibit representative fatigue behaviour. On the other hand, they must be small enough to examine the composite cross-section with a resolution of $\sim$10 $\mu$m so as to detect fatigue damage. Consequently the sample width must be less than 20 mm given that the detector is around 2000 pixels wide. In view of this, a smaller test-piece width was used in this study, compared to a standard configuration (20 mm wide by 200 mm total length). As a result, the sample width was set at 16 mm (shown in Fig. 2), wide enough for at least two unit cells across the width, with a 30 mm gauge length (total sample length of 130 mm) so that at least three unit cells are included along the length.

2.2. Static tensile and fatigue testing

Static tensile testing was carried out to determine the ultimate tensile strength, Young’s modulus and failure strain. In order to make sure that the small samples exhibit a behaviour representative of the panel, five small samples were tested to failure in the warp direction. The results were compared with those of the larger
standard sample test-piece employed in [15] in Table 2. There is on average a 9% difference in the ultimate tensile strength between samples of standard size and the smaller ones. This could be attributed to a size effect. Statistically, small samples such as those tested here are likely to have fewer defects than those of standard size and therefore exhibit a higher tensile strength. A sinusoidal waveform was applied in the fatigue tests, with a minimum to maximum stress ratio $R = 0.1$, at a frequency of 5 Hz. The maximum stress level of 224 MPa corresponds to 40% of the ultimate tensile strength of the small samples. Only when final failure occurs in the gauge region is the fatigue test considered to be valid. The results of three valid tests are shown in Table 3. It is worth noting that there are no straight warp fibres in this structure. Tensile loading is essentially carried by axial fibres, which refer to both additional warp yarns in the surface and binder yarns.

### 2.3. X-ray time-lapse experiments

In order to reveal the overall damage distribution at various fractions of the fatigue life, X-ray time-lapse experiments were performed by scanning the same specimen before fatigue testing and after interrupting fatigue testing at $10^2$, $10^3$, $5 \times 10^3$, $1 \times 10^4$, $5 \times 10^4$, $6 \times 10^4$, $7 \times 10^4$, $8 \times 10^4$ and 80,454 (ultimate life) cycles. The contrast agent and staining procedure employed in [15] was used to increase the contrast of damage in the X-ray images, so that the growth of damage with increasing numbers of fatigue cycles could be effectively detected.

The CT scans of the sample were conducted on a Nikon Metrology 225/320 kV Custom Bay, with an accelerating voltage of 70 kV and current of 90 μA. 3142 projections were taken as the specimen was rotated over $360°$ in equal increments. These projections were collected on a $2000 \times 2000$ pixel 16-bit amorphous silicon digital X-ray detector. The exposure time for each projection was 1 s, resulting in the total acquisition time of 2 h 37 min. A resolution of 11.8 μm, corresponding to the full cross-section scans proved to be sufficient to resolve small scale damage in this type of material. No load was applied during each scan. Two CT scans were performed side-by-side along the gauge length 5.6 mm apart (along y). In each scan, the total volume in the field of view is $23.6 \times 23.6 \times 23.6$ mm$^3$. To stitch two volumes together, there are a number of stitching strategies, as described in [17]. In this study, the projections for each scan were reconstructed separately and thus a composite dataset to dataset even if the same imaging conditions are maintained throughout, which if unaccounted for would affect the segmentation and quantification process. To this end, it is necessary to correct the reconstructed volumes to account for changes in illumination using a histogram normalisation process. This was first performed on the pairs of scans before stitching to form a composite image, followed by normalising the histograms of whole sample volumes acquired at different cycles to ensure a constant threshold throughout the segmentation process. The equation of the linear histogram normalisation is written in Eq. (1). Note that the dataset recorded before cycling (0 cycle) is taken as the reference.

$$D_n = [D_o - I_o \text{ (resin)}] \frac{I_o \text{ (fibre)} - I_o \text{ (resin)}}{I_o \text{ (fibre)} - I_o \text{ (resin)}} + I_o \text{ (resin)}$$

where $D_n$ is the new data as a result of linear normalisation, $D_o$ is the original data, $I_o$ (resin) and $I_o$ (fibre) are the mean grey scale intensity value of fibre and resin of the original dataset respectively. Similarly, $I_o$ (resin) and $I_o$ (fibre) are the mean grey scale intensity value of fibre and resin of the reference dataset, respectively. After the normalisation, the same threshold value was used to select the cracks for all the datasets.

Another challenge is to automatically segment (distinguish) different types of cracks such that their growth during fatigue cycling and interaction with the microstructure can be quantified. In this study, a novel image processing method has been developed and applied to each CT dataset to distinguish the different features in each 3D image. It comprises the following steps:

1. A prerequisite of damage segmentation is the classification of the resin, weft and binder yarns, which has been performed on the original volume (0 cycle). Unsurprisingly, resin voxels can be easily distinguished from fibre voxels due to their low grey scale intensity, whereas voxels representing weft and binder yarns have the same intensity so that the yarns must be characterised by their orientation. To this end, a texture map discriminating weft and binder orientations was designed, combining top-hat transforms [20] with linear structuring elements oriented in the x and y directions respectively. Thresholding was used jointly on the intensity and texture maps to obtain robust markers for the resin, weft and binder pixels, which are then expanded using a watershed-based algorithm to classify the entire volume. The result of the first step is shown in Fig. 3.

2. The following procedures are repeated for each dataset acquired at various fractions of fatigue life. Firstly, a coarse crack classification is performed according to the orientation of cracks, as a result of which, transverse vs debonding cracks are distinguished (shown in Fig. 4(b)). This is done by top-hat segmentation using linear structuring elements. Cracks are classified as transverse cracks if the top-hat transform has the highest response in the y direction, otherwise they are assigned to debonding cracks. Subsequently, a fine classification is achieved according to the location of cracks. For each crack voxel, the proportion of voxels classified as resin, weft or binder in a local neighbourhood are examined. Consequently, as shown in Fig. 4(c), six types of cracks are distinguished, namely transverse resin cracks, transverse cracks within weft yarns, debonds

### Table 2

<table>
<thead>
<tr>
<th>Sample number</th>
<th>Fatigue life, $N_f$ (cycles)</th>
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<tr>
<td>I</td>
<td>98,856</td>
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<tr>
<td>II</td>
<td>97,278</td>
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<tr>
<td>III</td>
<td>81,464</td>
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### Table 3

Cycles to failure of small sample test-pieces in valid tension–tension fatigue tests at a maximum stress of 224 MPa.

<table>
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at the binder/binder (B/B) interface, debonds at the weft/binder (W/B) interface, debonds at the binder/resin (B/R) interface and debonds at the weft/resin (W/R) interface. The W/B and B/B debonds are also termed in the rest of the paper as intertow and interply debonds, respectively. It must be noted that this algorithm has simply classified cracks according to their locations. Therefore, longitudinal cracks within binder tows are classified as B/B debond (red), because of having the same voxel environment. For the same reason off-axial resin crack on the surface are classified as transverse resin crack (light blue).

3. Results and discussion

3.1. Damage distribution

The 3D renderings of the fatigue damage accumulated in the specimen at different states (0.1%, 1%, 60% and 99% of fatigue life) are illustrated in Fig. 5. It is evident that the fatigue damage develops progressively from the very beginning until the composite approaches the end of life. Moreover, damage is found to be evenly distributed over the entire volume, with transverse cracks and interfacial debonding appearing in a regular periodic pattern. It can be seen that a regular pattern of transverse cracking has begun to appear after just 0.1% of the fatigue life. The debonding cracks are evident after 1% of the life, shown in Fig. 5(b). At 99% of life, the damage is very extensive and has propagated over the whole x–y plane indicative of a high level of damage tolerance. The binder yarns may confer a significant capacity for the re-distribution of load as cracking progresses; which may also give significant energy absorption. A similar finding was reported in [5]. The repeating woven pattern has a regular array of structural defects, such as fibre crimp and resin rich regions. These defects make up a network of stress concentrations where energy is preferentially released via the formation of new crack thereby redistributing the load elsewhere. This is consistent with a model developed for a 5H Satin woven composite, which simulates the sequence of cracking in the transverse yarns [21]. It suggests that after the appearance of the first transverse crack, the strain energy release rate is higher for a crack to form in another yarn than that for a
crack to form in an already cracked yarn. Therefore, new cracks are most likely to form in the non-cracked yarn. The occurrence of multiple cracking within each transverse yarn was found with increasing numbers of fatigue cycles and is evident in Fig. 5. This is in line with the shear-lag model [22,23], which determines the redistribution of stress in the transverse yarn arising from the presence of a crack. The stress is zero at the crack surface, whereas away from the crack, load is transferred back to transverse yarns by shear-lag in the vicinity of interface between transverse yarns and the adjacent yarns. Thus another crack can occur at a distance from the first crack over which the elastic strain energy essentially reaches a critical value. The 3D nature of the composite appears to suppress the formation of a localised damage zone by driving subsequent damage to initiate elsewhere. Consequently, failure occurs progressively until the whole composite is heavily damaged and final failure occurs by axial fibre breakage. Moreover, from the material design point of view, the ability to distribute damage and structural defects across many sites can be useful, in terms of achieving a high level of energy absorption.

3.2. Damage growth and interaction

3.2.1. Progressive damage observation

As mentioned above, fatigue damage is evenly distributed over the entire gauge section, according to the periodic pattern of geometrical features. Thus, a representative unit cell was digitally cropped from the original volume for more detailed investigation, as illustrated in Fig. 6. In order to identify and follow the damage as a function of the increasing number of fatigue cycles, the datasets obtained at different fatigue cycles were aligned using the registration module in Avizo 8.0. Subsequently a series of co-registered cross-sections parallel and perpendicular to the loading direction have been taken from each dataset to investigate the evolution of damage and the interaction between the different types of damage.

3.2.1.1. Regime I (transverse crack dominated). Regime I is characterised by the development of transverse cracking within the weft tows very early in the fatigue life. It is evident from Fig. 7(a) that no significant manufacturing damage is present prior to fatigue.
The different types of damage are shown in Fig. 7 based on the segmentation procedure described above. Fig. 7(b) confirms that transverse cracks have initiated at just 0.1% of life. This interesting finding might be explained by the deformation of surface warp and binder yarns upon consolidation, which may cause a stretching effect on the weft yarns. As a result, the weft yarns are more susceptible to tensile fatigue loading, leading to the early failure. Slice $Y_1$ shows that these transverse cracks arise in the weft yarns, with one of these cracks propagating a long distance (>than a unit cell) through the weft yarn. It is noteworthy that the location of these early transverse cracks is at the centre of weft yarns where the surface warp and binder yarns are heavily crimped. This might be explained by the observation in [24] that the highest stress usually appears at the centre of yarn crossing point where the longitudinal fibres experience significant straightening, causing the onset of transverse cracks. Moreover, another interesting feature is that these transverse cracks are mainly developed within the weft yarns in the surface layer while no such kind of damage is found within the interior of the composite. This may be because the surface layer undulates more than the inner layers, because extra crimp is induced by the existence of the surface warp yarns (shown in pink in Fig. 1). In addition to the increased crimp, the surface warp yarns are probably more free to deform during tensile loading, due to less constraint from the surrounding plies. Therefore, the weft yarns close to surface were more easily subjected to local bending, causing transverse cracks to first propagate through these weft yarns.
After fatigue loading to 1000 cycles (1% of life), it can be seen in Fig. 7(c) that a few transverse cracks have now started to develop in the weft yarns within the interior. The 3D rendering illustrates that not only did the transverse cracks propagate along the weft yarns, but also in the $z$ (through-thickness) direction; when they reach the interface between the weft yarn and the binder yarns, their further propagation is arrested by the axial fibres. Instead, they split at the interface and grow as debonding cracks (W/B debonds). In addition to the transverse cracks, longitudinal splitting is observed, as indicated in Fig. 7(c) Y1. It seems to cause debonding cracks to initiate at weft/binder interface, in a similar way that the transverse cracks do. Perhaps, the most striking observation (indicated by the yellow arrows in Fig. 7(c) Y2) is the interaction between the debond at the binder/binder interface (interply debond) and the debond at the weft/binder interface (intertow debond). One could hypothesise that the intertow debond grows along the interface between the weft and binder yarns until a point where the two interfaces converge, at which point it can deflect into the interface between binder and binder yarns, causing an interply debond.

3.2.1.2. Regime II (multiple transverse cracking and debonding). In regime II, further transverse cracks appear and debonding cracks form. The extent of damage after $5 \times 10^3$ cycles (6% of life) is evident in Fig. 8(a) where new transverse cracks have been generated within the weft yarns and many transverse cracks that formed previously have started to generate debonding cracks along the weft/binder interface. By $10^4$ cycles (10% of life) many yarns have more than 1 transverse crack (see Fig. 11(a)), and many transverse cracks have grown laterally through the weft yarns. This implies the dominant growth direction of transverse crack is along the weft yarns (in $x$) rather than through the thickness of the weft yarns (in $z$). Most of the transverse cracks are confined within one ply/yarn. Only in regions where two weft yarns are stacked above one another can a transverse crack propagate from one yarn to another (see the LHS of X1 in Fig. 8(a)). A previous study into glass fibre composites has reported that the maximum strain build-up in the resin rich region is higher than the bulk strain, leading to transverse cracks initiating there [25]. It may be that in our case the resin pocket is much more tolerant to strain than the fibre rich regions (weft yarns) because the transverse cracks tend to propagate within the weft yarns. Unsurprisingly, the yellow arrows in (a) and (b) show more debonds occurring at the binder/binder interface in section Y1, which are often associated with weft/binder debonding. It is likely that weft/binder debonding acts as a driving force for initiation of binder/binder debonds.

3.2.1.3. Regime III (debonding crack dominated). By 60% of fatigue life ($5 \times 10^4$ cycles) multiple transverse cracking is extensive and approximately evenly spaced at around 3 cracks per yarn and has started to saturate (see Fig. 11(a)). At this point debonding cracks have spread along the weft/binder interface and binder/binder interface. It can be seen from the interaction between the red and light blue regions that, in most of the cases, the B/B debond (interply debond) appears to have originated from the W/B debond (intertow debond), as indicated by the yellow arrows in (a) Y1,

Fig. 8. Representations of the damage after (a) 6% and (b) 10% of the ultimate life. On the left hand side, a 3D rendering of fatigue damage, with different types of cracks in different colour; on the right hand side, selected virtual sections perpendicular to the $x$ and $y$ directions labelled X1 and Y1 respectively. The colour map of different types of damage refers to Fig. 4(d). (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)
However, the further development of interply debonding is probably due to the straightening of binder yarns during tensile loading. This has been proposed in our previous post-mortem study [15].

Fig. 9(b) and (c) shows that fatigue damage continues to propagate steadily with increased cycling. It should be noted that the debonding cracks propagate reasonably uniformly, advancing slightly from the previous crack tip. Moreover, it is clearly observed that after $6 \times 10^4$ cycles (75% of life), the rate of increase in the number of transverse cracks has diminished, with just a few new transverse cracks appearing towards the edge of weft yarns (the tip of lenticular-like yarn cross-section), as denoted by the yellow ellipses in (b) $X_1$ and (c) $X_1$. This contrasts with previous studies [26,27] where the FE analysis suggests the yarn edges to be the site of local stress concentration. It might be due to the fact that their models assume a uniform distribution of fibres over the yarn cross-section. However, the uneven distribution of fibres is a very important factor for determining damage initiation and development [28]. In reality the fibre content decreases towards the edge of the yarn. The dispersed fibres at the edge of the yarn are held together by resin continuously spreading from the volume of yarn to the volume of matrix pocket. As a result, there is no physical edge boundary and the stress concentration due to the yarn/matrix boundary is not realistic [29]. Thus, the initial damage takes place at the yarn centre rather than the edge. In Fig. 9(c), the white

![Fig. 9. Representations of the damage after (a) 60%, (b) 75% and (c) 90% of the ultimate life. On the left hand side, a 3D rendering of fatigue damage, with different types of cracks distinguished by different colours; on the right hand side, selected virtual sections perpendicular to the $x$ and $y$ directions labelled $X_1$ and $Y_1$ respectively. The colours corresponding to different types of damage are explained in Fig. 4. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)](image-url)
ellipse denotes the formation of a surface resin crack on cycling from $6 \times 10^4$ to $7 \times 10^4$ cycles, which is oriented at $45^\circ$ to the loading direction. It is possible that the debonding cracks between surface resin and binder yarns propagate further into the surface resin rich region where local shear stress dominates, resulting in the $45^\circ$ crack. What’s more, longitudinal splitting (see the yellow arrow in (c) Y1) is found in the binder yarn, indicating this type of damage is present throughout the fatigue life. By $7 \times 10^4$ fatigue cycles (90% of life), damage has continued to grow in the form of interfacial debonding, while the creation of new transverse cracks has largely ceased, probably because the extensive cracking means that it is no longer possible to generate enough strain for the creation of any more cracks.

3.2.1.4. Regime IV (final failure). Fig. 10 shows the damage in the unit cell after final failure. In contrast to Fig. 9(c) cracks have finally developed within the resin rich region, as indicated by the yellow ellipse in Y1. This suggests that resin rich areas are less prone to damage than other areas until the final stage of fatigue life. A possible explanation might be that the local thickness of resin rich region has an effect on the local toughness. Studies have been reported that toughness increases with thickness [30,31]. In general the size of plastic zone of crack tip is strongly dependent on local microstructure. A thick resin rich region may result in a larger amount of matrix sustaining the plastic deformation upon crack extension. This plastic deformation absorbs more energy and therefore increases local toughness. It is worth noting that Z-yarns have a significant influence on delaying final fracture in 3D composites by promoting energy dissipation via extensive interfacial debonding, whereas studies [32–34] have shown that in the absence of the restraint imposed by Z-yarns, cracks can easily grow and coalesce in 2D laminates, resulting in ply separation and failure. This specimen eventually failed due to tensile fibre fracture. An overall failure of the specimen is shown in Fig. 10(b), where the fracture surface is almost perpendicular to the loading axis, indicating a catastrophic failure. This is consistent with our previous post-failure study [15] which has found that the broken fibres are localised in a narrow band, rather than randomly distributed throughout the specimen.

The normalised stiffness was plotted against the number of fatigue cycles, shown in Fig. 11. $E$ and $E_0$ are residual and original stiffness, respectively. It can be seen that there is no significant stiffness reduction until it reaches the second regime, where the stiffness has reduced by $\sim8\%$ mainly due to the onset of debonding cracks. It was measured that the stiffness has dropped by $\sim20\%$ until just before the final failure, indicating the material is softened by extensive straightening of binder yarns. Other mechanical properties such as shear modulus may suffer a higher drop, but these have not been measured in the present work.

3.2.2. Damage interaction

The microstructure of a 3D woven composite gives rise to complex interactions between the different modes of damage. The modes can be broadly aggregated into three types, namely interactions between transverse cracks and weft/binder debonds, interactions between weft/binder debonds and binder/binder debonds and the interaction between longitudinal cracks and weft/binder debonds. As mentioned above, transverse cracks start to appear early in the fatigue life, their growth giving rise to debonds at
the weft/binder interface. However, as these debonds grow in the lateral direction, they can be deflected into the binder/binder interface and continue to propagate in a benign manner. Strong evidence has shown that in most cases, the binder/binder debonds originate from the crack tip of weft/binder debonding. Similar to transverse cracking, longitudinal cracks within the binder yarns also initiate debonds at weft/binder interface. All three types of damage interaction absorb energy via crack deflection, distributing the damage and improving the fatigue life. Taken together, these mechanisms can be considered as beneficial behaviour in the sense of decoupling the damaged part and providing alternative crack path.

3.3. Quantification of damage evolution

A quantitative analysis on the extent of each damage mode has been performed to examine the growth rates and relative extents of the different types of damage as a function of the number of fatigue cycles. These have been quantified by analysing the damage fractions in 10 equally spaced 2D cross-sections taken from the 3D unit cell as a function of fatigue cycling. The six types of fatigue damage were grouped into two categories (i) transverse cracks and (ii) debonding cracks. Although the transverse cracks include those within the resin and those within the tows, the former are small in number. The total number of transverse cracking can be represented in terms of the number of cracks within the tows and this is plotted in Fig. 12(a). As regards debonding cracks, the total length of debonding was divided by the total interface length to determine the debonded fraction as a function of life (shown in Fig. 12(b)). Unsurprisingly, the number of transverse cracks grows rapidly at first, accompanied soon after by an extensive development of interfacial debond indicating the damage propagation is profuse but stable. It is clear that the number of transverse cracks largely saturates by around 60% of life at around three parallel transverse cracks per yarn. By contrast debonding increases throughout life reaching 30% of the available interface by the end of life, suggesting the system has a high fatigue resistance.

Considering that the initiation of intertow debonding is often associated with the propagation of transverse cracks, the fraction of intertow debond were plotted together with number of transverse cracks within the tows against the life fractions in Fig. 12(a). The results show that transverse cracks form initially, however, the growth rate of two types of damage gradually becomes equivalent, suggesting that the development of intertow debond is primarily governed by the transverse cracks.

In order to look in detail at the different types of damage, in terms of their role at various life fractions, their growth through life has been plotted separately. Debonding cracks are classified into four types, namely binder/binder debond (interply debond), weft/binder debond (intertow debond), weft/resin debond and binder/resin debond. Fig. 13 shows the growth rate of different types of debonds. As the number of transverse cracks saturate after around 60% of life, the growth of intertow debonding also slows. In contrast, the interply debond continues to grow throughout life, which demonstrates that the degradation of the material in the later stage of life is mainly due to this type of damage. It is also noteworthy that debonds at the resin interface are less common suggesting that they play a peripheral role in damage development. From the data in Fig. 13, it is apparent that the interfaces in this structure are relatively weak, but that damage grows stably. This is probably due to the 3D interconnectivity between neighbouring yarns, which effectively disperses debonding and prevents damage localisation.

4. Concluding remarks

Fatigue damage evolution in a 3D woven composite reinforced by modified layer-to-layer fabric has been monitored by time lapse X-ray computed tomography. The governing damage mechanisms at different stages of fatigue life have been identified, highlighting
the interaction between damage accumulation and the local microstructure. The following conclusions can be drawn from the present investigation.

a. Damage occurs initially by transverse cracking in the centre of weft yarns at very early in the fatigue life (0.1%). This is followed shortly (1% of life) by longitudinal cracking within the binder yarns. Midway through the fatigue life, the number of transverse cracks increases with debonding cracks spreading along the interface between the binder yarns and the surroundings. Just prior to the final failure, damage started to occur in the resin pockets, suggesting resin pockets are more tolerant to strain in this material.

b. Based on virtual cross-section observations, the initiation of binder/binder debonding is strongly correlated with the crack tip of the weft/binder debonds. Further development of binder/binder debonding might be due to the straightening of binder yarns under tensile loading.

c. The growth of different type of damage was quantified as a function of fatigue cycles. It is suggested that fatigue damage progression in 3D woven materials is a fairly stable process, with a gradually decreasing growth rate. Moreover, the growth of weft/binder debonding slows as the number of transverse cracks becomes saturated, whereas the growth of binder/binder debonding contributes significantly to the extensive debonding in the later stages.

d. This study has shown that damage is evenly distributed, having a periodicity that can be correlated with the composite unit cell. This suggests a high level of damage tolerance and load redistribution. With the resolution obtained in the present work, a full picture of damage pattern across the composite has been acquired but at a resolution too coarse to capture finer features such as fibre breakage. In view of the periodicity there is significant potential in future work to focus on the response of a single unit cell (representative volume element (RVE)) using micro-computed tomography (micro-CT) to obtain a high resolution image thereby allowing a more accurate quantification of all damage modes involved without excessive image stitching.

In conclusion the work has confirmed that damage can arise very early during fatigue partly due to the crimp, but that the periodic through thickness constraint means that load is redistributed around damage sites providing a highly damage tolerant structure. This propensity for widespread damage that accumulates steadily during the fatigue life is reflected in the decrease in composite stiffness with cycling. It has also shown that the evolution of fatigue damage results from the competition and interaction between different mechanisms. Transverse cracks, occur almost at the beginning but don’t compromise fatigue life, whereas other type of damage, such as interply debonding, has a large growth towards the end of life and leads to the final failure. Finally, the nature of damage evolution captured in this study could guide the design of a 3D composite with more resistance to fatigue damage.

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