DAMAGE CHARACTERISATION AND THE ROLE OF VOIDS IN THE FATIGUE OF WIND TURBINE BLADE MATERIALS

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1 Introduction

The technological development of wind turbines and their more widespread use represent an important method towards meeting the growing worldwide energy demand. For increased power generation and a greater efficiency, the general trend is for larger turbines with an increased blade diameter. Individual blade lengths are currently approaching 60m; a significant increase from around 20m 15 years ago. This emergence of much longer blades has resulted in an absence of relevant long-term in-service data. With expected lifetimes of 20 years, the long-term integrity of the blade material has thus become an important area of research. As a result, extensive fatigue databases for wind turbine composites have been compiled in recent years. The non-transferability of these results between different materials and load conditions however implies a large experimental effort. To reduce this, a greater understanding of the fatigue damage modes and their accumulation is required. In addition to this, further insight into certain microstructural features, such as voids, is required, as their role in fatigue damage accumulation remains uncertain. Accordingly, the aim of the present work is to gain insight into the micromechanical damage processes of a wind turbine blade composite material. In order to achieve this, a suitable experimental technique was needed for imaging and analysis. Computed tomography (CT) has been used to this end in the current work, enabling full 3D imaging of a sample's internal volume. The first goal was the observation and quantification of microstructural detail, notably voids, through scanning untested material. The second main objective was the scanning of fatigue-tested coupons to identify the damage mechanisms occurring. Through scanning specimens that had sustained different fractions of their estimated lives, an idea of damage sequence and interaction could also be obtained.

2 Literature

Composites are known to exhibit four main fatigue mechanisms; fibre fracture, fibre/matrix debonding, matrix cracking and delamination. [1] It has been found that the different mechanisms can dominate at different stress levels. At low load levels, matrix cracking is commonly a predominant mechanism. At medium loads a combination of matrix cracking and interface debonding is observed, while at high loads fibre failures may occur. [1] The low load, high cycle fatigue nature of the wind turbine is thus expected to promote matrix cracking. Although matrix cracking has been widely researched, most work has concentrated on cross-ply laminates, and thus transverse (90º) matrix cracking. [2][3] These cracks have been noted to grow across the specimen width from the free edge, with their growth rate dependent on load-direction spacing rather than their own length. [4][5]

Tong et al. [6] have previously investigated crack development in $[0/90/-45/+45]_s$ glass fibre-epoxy composite in static tension and tension-tension (T-T) fatigue. Transverse ply cracking appeared first, followed by -45° cracks from the edges of existing 90° cracks, and finally +45° matrix cracks. Masters and Reifsnider [7] studied the fatigue crack growth of different quasi-isotropic laminates, confirming the presence of 90° ply cracks that spread to form cracks in neighbouring off-axis plies. Two different lay-ups were considered, $[0/90/\pm45]_s$ and $[0/\pm45/90]_s$. Each displayed distinctly different crack saturation patterns, despite initial damage of the transverse plies in both. This highlights the

dependence on laminate lay-up for damage evolution. Gamby et al. [8] have reported on the development of cracks from free edges in laminates containing 0° and ±45° plies, noting that the matrix cracks propagated from the free edges of the off axis layers, and, as with the transverse cracks previously mentioned, were dependent not on their own length but only on the number of cycles and the distance from the free edge.

Matrix cracking alone does not usually cause fatigue failure, however it is often a precursor to more critical mechanisms. Matrix cracks are often reported to be responsible for delamination onset, due to the local stress concentrations at their tips. [9] Delamination is also known to occur at laminate edges where interlaminar stresses are highest, due to the Poisson's ratio mismatch between differently angled plies. O'Brien[10] experimentally compared the difference between local delaminations occurring within the specimen bulk (associated with a matrix crack), and edge delaminations. The edge delaminations were found to propagate in a stable manner, whereas local delaminations from matrix cracks created stress concentrations that led to premature laminate failure strains below that of the primary load bearing plies.

Voids are present in composites either due to the manufacturing process, that leaves air trapped in the laminate during the curing cycle, or by nucleation from volatiles during processing. [11] Critical void levels below which mechanical performance isn't affected have been predicted as between 1-4% in various studies.[12]-[14] Whether void content is more of a concern to static or fatigue properties is an area of disagreement, with some studies concluding that static performance is largely unaffected by voids, [15] while others report the same order in the reduction of both static and fatigue performance. [16] Studies explicitly relating to fatigue include that of Prakash, [17] who performed axial fully-reversed (R=-1) tests on unidirectional carbon fibre composite. It was hypothesised that voids, along with other damage, contributed indirectly to reduced fatigue performance through their poor heat dissipation properties. No quantitative analysis was however performed. Chambers *et al.* [16] performed a controlled analysis; varying the void content through set changes in the cure cycle of the composite. Both static and flexural fatigue performance were found to deteriorate with increasing void content, however the authors argue that this global percentage view is too simplistic and that factors such as void size, shape, and distribution must be considered. Mandell and Tsai[18] also concluded that the overall void content is an inadequate representation of potential michromechanical influence, and that void size and shape must be considered. A critical void size was proposed, above which fracture mechanics based crack growth could be used to assess void influence, while at lower sizes voids simply contributes to a reduction in the cross-sectional area of the specimen. Overall there is a noticeable absence of literature on the effect of voids on mechanical performance, especially when concerned with fatigue. Moreover, the studies that do investigate fatigue are usually limited to unidirectional carbon composites, and the effect of voids in multidirectional laminates remains largely unknown.

3 Experimental Detail

150x25x7mm prepreg glass-epoxy specimens, representative of the material used in wind turbine blade surfaces, with a gauge length of 28mm and a layup of $[0/+45/-45]_{3s}$ were studied. In addition to these directional layers a "fleece" resin rich layer containing randomly orientated fibres, used to improve the surface finish of the composite, was present adjacent to each 3-directional fibre ply. 2mm thick aluminium was used for end tabs, adhesively bonded using Araldite 2011 and post-cured for 30 minutes at 80ºC. Prior to any fatigue testing, 9 specimens were scanned using microfocus X-ray computed tomography (µCT) to enable the quantitative void analysis to be performed and eventually correlated with their fatigue lives. A 25x25x6.5mm volume within the gauge length of the material was scanned using an XTek™ Benchtop 160i µCT system. A voxel resolution of 25µm was achieved, implying the reasonably direct detection of voids >40µm (*i.e* being of the order of 2 or more voxels across). Using the reconstructed µCT volume as shown in Fig. 1, the voids were segmented using their greyscale histogram peak, and the rest of the bulk volume was removed.

Using the binarised volume of voids within a given sample, individual voids were then labeled and sorted in ImageJ [19]. This allowed the measurement of various parameters of each void including size (Feret measurements, equivalent volume spheres *etc.*) centroid location, and sphericity. After these scans were performed, the specimens were tested to failure in fully-reversed axial fatigue using an Instron 8852 servo-hydraulic test machine. A load level corresponding to +/-40% of the static tensile strength of the material was used, and the tests conducted at a frequency of 2Hz. A further five specimens were tested to various proportions of their estimated fatigue life and then µCT scanned in order to gain an insight into the chronology of microdamage development.

4 Results and Discussion

The nine specimens tested to failure sustained between 21,649 and 127,140 cycles with a mean lifetime of 70,995. The final failure was in all cases compressive, and was characterised by multiple total delaminations at each 45º/-45º interface, which permitted global buckling to occur. In order to identify which mechanisms contributed to this final failure, the interrupted-test CT scans were then examined.

4.1 Interrupted-test CT observations

The interrupted-test scans were conducted in order to gain an insight into damage evolution, and also to examine in what capacity the voids were found to be affecting damage propagation. It was hoped that full-cross section scans could be used, allowing the specimens to then continue to be fatigue tested after scanning. The 25µm voxel size associated with the full cross-section specimen scans was however unable to resolve the smaller scale damage, therefore the sectioning of the specimens to a 3x3mm cross section was required. This enabled a resolution of up to 6µm, and was able to resolve individual fibres, matrix microcracking and partial delamination. As the five different coupons' tests were stopped at varying life proportions, ranging from 8000 cycles to just prior to final failure, a detailed insight into damage accumulation throughout life was achieved.

The initial fatigue damage mechanism was found to be transverse matrix cracking in the surface resin ply. This was found to occur along the entire specimen gauge length, however not at regular intervals. Some of the transverse matrix cracks interacted with voids in the resin ply, however others were completely independent of them, as shown in Fig 4.1.

It can also be seen from Fig 4.1 that the crack path remains relatively unaffected by the randomly orientated fibres present in the resin rich "fleece" layer. There was a prevalence of surface layer transverse matrix cracking at the edges of the specimens. These edge cracks had usually grown to a longer length than the other surface cracks, and some were also found turn to propagate at +45º as they grew inwards, allowing them to run parallel to the fibres between the tows of the +45º layer. These "turned" matrix cracks, combined with the specimen edge, provided a favourable stress state facilitating the next stage of damage. This further damage was found in all instances to be a delamination propagating along the specimen length between outer +45º and -45º fibre layers. This can be seen to support the experimental evidence of two previous studies: delamination propagation from the edge of the specimen, as documented by O'Brien[10], and also from an existing matrix crack, as identified by Kashtalyan [9]. Figs. 4.2 and 4.3 illustrate this interaction of damage, showing the same specimen with the outer material sequentially cropped away to reveal the sub-surface damage. The voids and cracks have been enhanced in yellow and red respectively by digital segmentation.

The role of the voids in allowing the crack to propagate more freely is visible in Fig 4.3 (a), where the majority of the cracked length is occupied by large inter-tow voids. Fig 4.3 (b) shows the partial edge delamination (in red) that has formed at the final crack depth, corresponding to the +45º/-45º interface. It is important to note that these edge delaminations were stable in growth rate and occurred well before coupon failure; at roughly 50-70% of the nominal estimated life. It was only when the damage progressed to the inner plies that critical accelerating degradation occurred. This was found to occur due to some critical

"damage event" that allowed crack propagation inwards to the 2nd (or 5th, from the opposite face) ply, usually resulting in its immediate delamination. This damage event was in some cases concentrated tensile or compressive fibre failure in the 0º tow nearest the edge, and in others a delamination that was able to turn inwards. Fig 4.4 shows a cross sectional view of fibre failure in the edge 0º tow that has caused a crack to pass through the 2nd ply resin rich layer, by means of a void, to form a large total-width delamination at the 2nd ply 45º/-45º interface.

Once delamination occurred in this inner ply, failure by buckling ensued soon afterwards due to the large reduction in compressive strength. It is of note that prior to the "damage event" causing damage propagation to the neighbouring inner ply, the only damage mechanisms present were in the surface plies. This absence of interior damage, be it inner ply matrix cracking, distributed tensile or compressive fibre failure, or any local delamination, suggests a damage process strongly dependent on coupon architecture. This in turn casts doubt on the validity of this form of testing for material characterisation, particularly in wind turbine applications, where large bulk composites without many free edges are prevalent.

4.2 Void Analysis

The nine specimens that were tested to failure were analysed using the void segmentation procedure described in Section 2. The measured global parameters are described in Table 1., containing the extreme values of individual specimens, and the overall mean. Although many further individual specimen parameters were analysed, these have been omitted as no positive correlation with fatigue life was identified. Spatial distribution was also examined, looking for regional concentrations in void population that may have contributed to a raised local stress state and thus a shorter fatigue life. As this analysis also provided no positive result, a more local approach was adopted. As explained in Section 4.1, it is believed that the critical damage stage occurs when delamination propagates from the outermost ply to the inner plies. As shown in Fig. 4.4, this has been experimentally observed to occur through voids in the 2nd/5th ply resin rich layer. These layers were thus extracted and analysed explicitly within each specimen. In this instance the size of the largest void within these resin layers was found to be inversely proportional to the fatigue life, with reasonable correlation, as shown in Fig 4.5.

The absence of any global or spatial distribution trends further emphasizes the complexity of void-failure interactions, corroborating previous findings. [16][18] In addition, however, more detailed parameters such as the largest void and average void sizes have also proven to be largely independent of fatigue life. It is possible that the complexity of the material studied, which contains not only multidirectional fibre plies but also the resin rich "fleece" layer, inhibits the extraction of meaningful global void parameters.

4 Conclusions

• A detailed understanding of the material's fatigue damage accumulation has been achieved through using µCT analysis. Transverse matrix cracking of the surface resin layer was the initial mechanism observed. The transverse cracks propagating from the specimen edge then appeared to facilitate a partial edge delamination, which then propagated along the specimen gauge length. This delamination in turn degraded the material so as to allow a crack to pass into the inner plies, allowing them to delaminate and cause overall compressive failure due to buckling.

• A first-of-its-kind 3-dimensional void analysis, measuring parameters from over 10,000 voids within a specimen's gauge length was achieved using µCT. The global parameter and spatial distribution studies showed no correlation to specimen fatigue performance, however a local void population study of critical areas showed a relationship between the largest void present and fatigue life.

5 References

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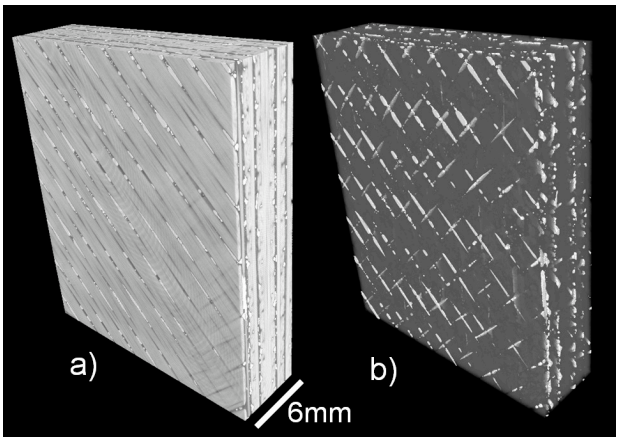

Fig 3.1.a) Reconstructed CT volume showing the voids segmented and highlighted in white. b) Same volume with the near-removal of the fibre and matrix material, highlighting the voids.

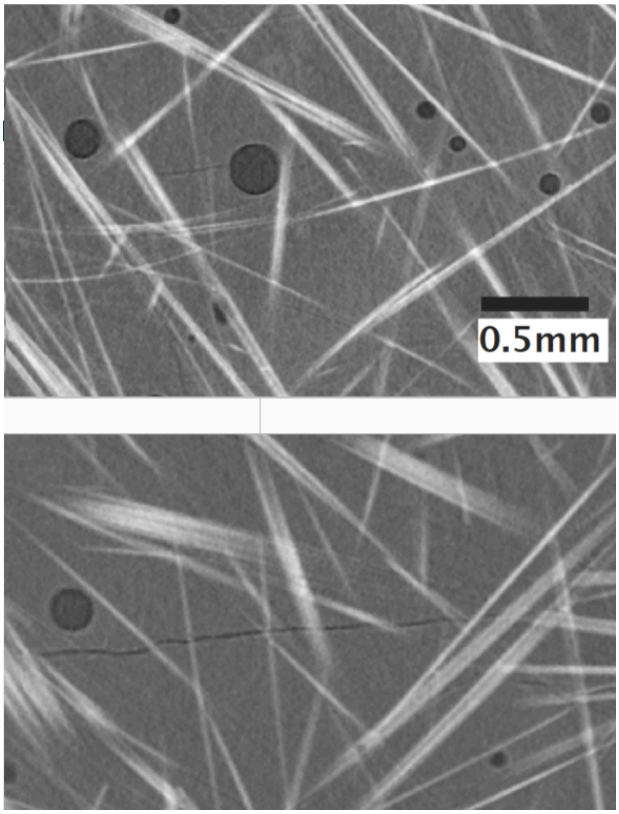

Fig 4.1 In-plane 2D CT slices, parallel to fibre plies,of the outer resin rich ply showing a) a transverse matrix crack interacting with a spherical matrix void b) a transverse matrix crack propagating independently of any voids

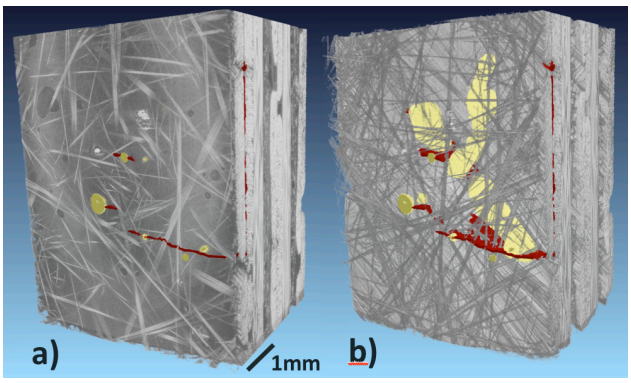

Figure 4.2 a) 3D reconstructed view of a tested 3x3.8mm cross-section "matchstick" specimen 5 mm in length containing a transverse matrix crack. b) The same scan with the matrix material revealed to show defect extension into specimen bulk

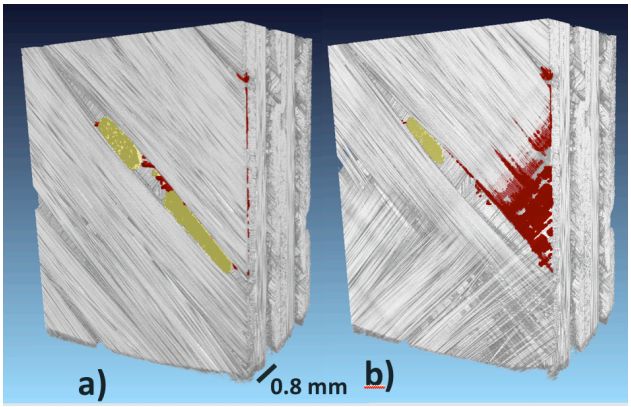

Figure 4.3. a) 3D reconstructed view taken 0.2mm from the front face of the specimen. b) 3D reconstructed view taken 0.6mm from the front face of the specimen.

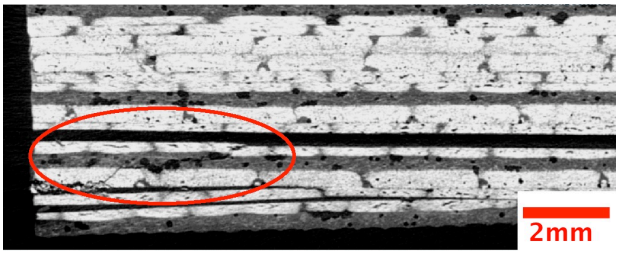

Fig 4.4 Cross-sectional 2D CT slice showing a crack passing through a void in order to allow damage to progress from an outer ply edge delamination to a 2nd ply total delamination

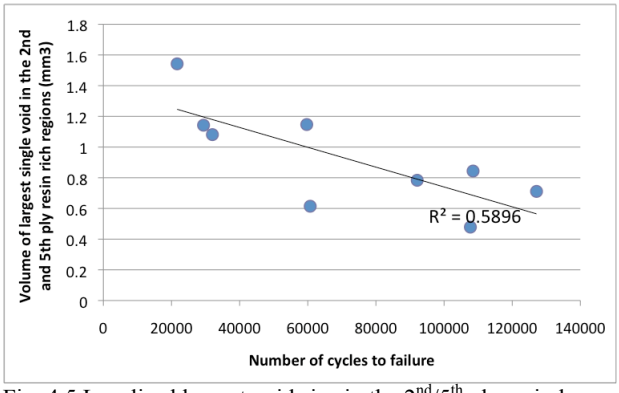

Fig. 4.5 Localised largest void size in the 2nd/5th ply resin layer compared to the specimen's fatigue life

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Parameter	Number of voids in analysis volume	Void content (%)	Average void volume (mm^3)	Volume of the largest single void (mm^3)	Total volume of largest 1% of voids (mm^3)
Minimum	9,831	2.47	0.0073	0.70	31.32
Maximum	13,230	3.70	0.0096	2.10	47.44
Mean	11,807	3.02	0.0081	1.53	34.99

Table 4.1 Overview of the void parameters measured, their extreme values, and averages