Ductility and pseudo-ductility of thin ply angle-ply CFRP laminates under quasi-static cyclic loading

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The quasi-static loading-unloading performance of thin ply carbon-epoxy [±26]s, [±27s], and [±26s/0s] laminates is presented. Consistent experimental results allow the reduction in laminate modulus to be evaluated using three different methods: secant modulus of each loading cycle; a secant modulus up to a constant stress, and the initial tangent modulus of reloading. It is shown, via all three methods, that these layups can undergo multiple cyclic loadings and display excellent retention of stiffness. The [±26s/0s] layup displays a limited modulus reduction, despite the gradual failure of the 0\(^\circ\) layers. The [±26s], specimens do not display any decrease in initial modulus and effectively suppress damage accumulation (shown via X-ray CT imaging), which both suggest that the stress–strain behaviour of these layups is ductile, rather than pseudo-ductile.

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1. Introduction

Thin ply composite materials have consistently been shown to offer considerable improvements in laminate strength under tensile, compressive and fatigue loadings via a delay or even suppression of damage prior to failure [1–7]. Demonstrated previously, the stress at which these damage mechanisms of matrix cracking and delaminations initiate increases as ply thickness is decreased [8–14]. Exploiting this phenomenon, work within the High Performance Ductile Composite Technologies (HiPerDuCT) programme grant has established that thin ply angle-ply carbon fibre reinforced-polymer (CFRP) laminates can effectively suppress damage until final failure [15,16]. This suppression of damage allows some matrix plasticity to occur and coupled with angle-ply rotation leads to a highly non-linear stress–strain response. For example, experimental results for [±25\(^\circ\)]s laminates, with ply thickness, \(t_p = 0.03 \text{ mm}\), show strength, \(\sigma_f\), of 950 MPa, failure strain, \(\varepsilon_f\), of 3.60% and pseudo-ductile strain, \(\varepsilon_d\), of 1.22%. Pseudo-ductility in this case is defined by taking laminate failure strain minus the strain at the same stress on a line of initial modulus. A ‘yield’ stress, akin to a proof stress for metals, is also defined for these responses, which is the intersection between the experimental curve and a straight line of initial modulus offset by 0.1% strain. The ‘yielding’ of these \(±\theta\) laminates is very gradual, but other work within HiPerDuCT [17] has shown that the addition of 0\(^\circ\) plies at the laminate mid-plane, \([±\theta_m/0_s]_s\), leads to a metal-like stress–strain curve with a more distinct ‘yield’ point. An initial, largely linear region continues up to this ‘yield’, which occurs when the fibres in the 0\(^\circ\) plies initially fracture. Gradual failure of these central plies, via multiple fibre breaks (fragmentations) and dispersed delaminations at the 0\(^\circ\)–0\(^\circ\) interface, leads to a stress plateau that progresses until the 0\(^\circ\) plies have fragmented into lengths close to the critical length. This pseudo-ductile behaviour is dependent on the relative thickness of the 0\(^\circ\) and \(±\theta\) plies, as well as the absolute thickness of the 0\(^\circ\) plies. For example, for a \([±\theta_m/0_s]_s\) laminate that displays pseudo-ductility via fragmentation and dispersed delaminations, a small increase in the thickness of the 0\(^\circ\) plies leads to a single, large delamination at the 0\(^\circ\)–0\(^\circ\) interface after the initial fibre fracture in the 0\(^\circ\) plies. Further increase in the thickness of the 0\(^\circ\) plies, without any additional \(±\theta\) layers, results in a catastrophic failure following the initial 0\(^\circ\) break before any non-linear behaviour.

Ductility of a material is defined by its ability to be loaded beyond the elastic limit and then demonstrate some permanent strain upon unloading. Additionally, ductile materials can be reloaded without loss of stiffness. This behaviour is not usually exhibited by composite materials, as brittle failure prevents permanent deformation, or any non-linearity in laminates with off-axis plies is generally governed by shear deformation or the accumulation of damage, which leads to a loss of stiffness. Indeed, quasi-static cyclic loading of \(±45\) laminates has often been used to characterise damage models [18–23]. Ladeveze and Le Dantec [18] established a method of defining a shear damage parameter...
via the cyclic loading of T300/914 and IM6/914 carbon fibre—epoxy ±45° laminates. The secant shear modulus, $G_{12}$, was defined for each cycle by taking a line between the zero load point and the crossover of the unloading with loading paths. Though likely to overstate the reduction of the modulus, this technique was used by Lafarie-Frenot and Touchard [19] in a comparison of the shear behaviour of ±45° laminates of T300/914 (thermoset resin) and AS4/PEEK (thermoplastic resin). Cyclic testing of these laminates showed a large difference between the failure strains, with the AS4/PEEK specimens reaching a strain approximately three times that of the T300/914. This high strain behaviour also allowed large fibre rotations to occur, which were shown to reach a residual value following the unloading of each cycle.

Further investigations of cyclic damage accumulation using similar techniques to [18,19] include studies by Herakovich et al. [20], Van Paepegem et al. [21,22] and Oghara and Nakatani [5]. Herakovich et al. [20] also used a high strain thermoplastic matrix in the damage modelling of IM7/K3B ±45° laminates. Similarly to [19], fibre rotations were accounted for in the analysis and a cyclic loading was once again used to inform the shear damage criterion. Van Paepegem et al. [21,22] applied the method proposed in [19] to glass fibre—epoxy ±45°_s laminates, with cyclic loading showing development of significant hysteresis loops with increasing strain. Oghara and Nakatani [5] performed cyclic loading on ±45° and also ±67.5° laminates to establish shear and transverse damage respectively. A comparison of thick ($t_p = 0.15$ mm) and thin ply ($t_p = 0.05$ mm) laminates concluded that the thin ply specimens were considerably more resistant to damage accumulation. Large strains were reached with the thin ply specimens, but the influence of fibre rotation was not reported.

The work carried out in [15–17] concentrated solely on monotonic responses and it is important to establish a better understanding of the residual load carrying capability of the ±60°_s and ±0°_m/0°_s configurations. Thus, in the present work, quasi-static cyclic loading of thin ply ±26°_s, ±27°_s and ±60°/0°_s laminates is conducted. Cyclic testing has not previously been reported on angle-ply laminates of this type and as such the aim is to present this novel behaviour and an improved understanding of the effect of this loading regime on the damage suppression and fibre rotation demonstrated by these thin ply laminates. The use of ±26°_s, and ±26°/0°_s specimens provides comparison between these laminates. The addition of ±27°_s specimens follows on from [16] and aims to further demonstrate the consistency of the non-linear response. For all layups, the apparent degradation of stiffness with number of cycles is investigated and the capability of the laminates to sustain repeated loadings is discussed.

### 2. Experimental methodology

The material used in this study was a Skyflex USN020 spread tow CFRP, which was consistent with that used in [15–17]. This prepreg material consists of standard modulus carbon fibres in 3K tows (Mitsubishi Rayon TR-30 ($E_f = 235$ GPa)), and a semi-toughened epoxy matrix. The measured cured ply thickness, $t_p$, was 0.03 mm, with a fibre volume fraction, $V_f$, of 0.42. The fibre volume fraction is low in this case but is consistent with previous work conducted on the same batch of material [15–17]. The resulting composite properties are given in Table 1. Specimen size was also maintained from the previous work, at 150 mm gauge length, 15 mm width and GRFP cross-ply end tabs of 40 mm. This sizing was kept to provide direct comparison between monotonic and cyclic responses. Three specimen of each ±0°_s, layup and four of the ±26°/0°_s were manufactured via hand layup and then autoclave cured. Testing was performed on an Instron 8872 hydraulically-actuated machine under displacement control at 2 mm/min for both the loading and unloading stages. Strains were recorded using a Imetrum video extensometer and associated software.

Results of the monotonic testing conducted on ±26°_s, ±27°_s and ±26°/0°_s specimens, as reported in [16,17], have been used to determine the average maximum displacement of each layup. This has allowed the unloading points and number of cycles to be defined. For monotonic testing of the ±26°_s and ±27°_s layups, the average maximum displacements were very similar at 9.35 mm ($\epsilon_s \approx 4.3\%$) and 9.60 mm (4.5%) respectively. This similarity allowed a single set of unloading displacements to be selected, which were: 3 mm, 6 mm, 7.5 mm and 8.75 mm. Expressed as strains these are approximately: 1.0%, 2.5%, 3.25% and 4.0%. This led to three cycles containing both an unloading and subsequent reloading of the specimens. Specimens were not unloaded following unloading from the final displacement. The ±26°/0°_s specimens tested monotonically reached a mean maximum displacement of 8.50 mm. It was desired to have unloading points distributed over the entire range of the response, both pre and post-fragmentation. This gives insight into the effect of that damage on the overall behaviour of the specimens. The first two ±26°/0°_s specimens were unloaded at: 3 mm, 4.5 mm, 5 mm, 5.5 mm and 6 mm (approximately: 0.75%, 2.25%, 3.0%, 3.25% and 3.5%), and the last two at: 4.5 mm, 5 mm, 5.5 mm, 6 mm and 6.5 mm (giving five cycles at approximately: 2.25–2.5%, 2.75–3.0%, 3.25–3.5%, 4.0% and 4.25%). These specimens were cycled at different displacements to investigate the behaviour over the whole range of the response.

### 3. Experimental results

**Stress-strain curves are included in each results plot for ease of comparison between the types of loading.**

#### 3.1. Results: ±26°_s and ±27°_s

The experimental results for the ±26°_s and ±27°_s specimens are presented in Figs. 1 and 2 respectively. An example of the monotonic response is also included in both figures. The cyclic responses are well-matched to the monotonic, undergoing yielding followed by a final region of stiffening at high strain. Hysteresis develops in the loading-unloading cycles even at the lowest strain and the unloading does not return to the origin, showing that there is some permanent strain developed.

#### 3.2. Results: ±26°/0°_s

**For clarity, the stress-strain responses of just two specimens are presented on each plot in Figs. 3 and 4. There are three distinct regions to these stress-strain curves. An initial, largely linear loading up to the point of the first 0° fibre fracture. This is followed by a**

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**Table 1**

| $E_{11}$     | 101.7 GPa | $E_{22}$     | 6.0 GPa | $v_{12}$ | 0.3 | $t_p$ | 0.03 mm |
|--------------|-----------|--------------|---------|----------|-----|-------|--------|
| $G_{12}$     | 2.4 GPa   |             |         |          |     |       |        |
| $v_{12}$     | 0.3       |             |         |          |     |       |        |
stress plateau, as the 0° fibres continue to fail gradually, fragmenting into increasingly shorter lengths. In this case, fragmentations are coupled with dispersed, local delaminations at the 0°/C176 interfaces. Once a critical length is reached and fragmentations have saturated, the laminate continues to be loaded further, with the intact ±26° layers carrying the majority of the load up to failure. Fig. 3 shows the responses of the first two ±26°/0° specimens – an initial cycle was performed in the initial, virtually linear region and the final unloading was considerably below the monotonic failure strain. This initial cycle was removed for the third and fourth specimens (Fig. 4) and the final cycle was programmed to occur closer to failure, to investigate more of the final region of the stress-strain curve.

Fig. 3 shows that a stress plateau can be achieved under cyclic loading. The maximum stress of the first unloading cycle is approximately half of the fragmentation stress, \( \sigma_{\text{frag}} \) and as such there is almost no hysteresis, the strain returning to the origin. This is because the cycle has been performed prior to fragmentation initiation and fibre rotation of the ±26° plies is minimal. There is considerable hysteresis once the unloading takes place in the non-linear portion of the curve, post-fragmentation initiation. Each reloading cycle of the specimens reaches a similar stress to the monotonic case, following the trend of the plateau and then the increase in stress that follows. The loading of the first two specimens is stopped well short of laminate failure and so does not extend very far into the final increase of stress following the fragmentation plateau (Fig. 3).

To explore the behaviour in the final region of the stress-strain response, the second pair of specimens were subjected to more cycles and larger strains. Fig. 4 shows that specimen #3 completed four of the five cycles but then failed before the fifth at a displacement of 6.5 mm (4.3%). The final specimen, however, failed before the fourth cycle up to 8.5 mm, meaning that only three cycles of loading-unloading were captured. This failure was unexpected and considered to be premature, as the failure strain of 3.85% is somewhat lower than the monotonic mean value of 4.20%.

4. Analysis of modulus reduction

The stress-strain responses of the ±30°, and ±26°/0° specimens are highly non-linear. As such, it is expected that there will be considerable deviation from the initial laminate modulus, \( E_0 \), as the cycles progress. Any change in the laminate modulus can be discerned from the slope of the stress-strain curve upon reloading.
The value of $E_x$ is the only effective elastic property monitored in the current work, with no account taken of the possible variation in $E_y$, $G_{xy}$ or $v_{xy}$. This is due to the uniaxial loading of these laminates, where the value of $E_x$ is of primary concern.

The degree of change of $E_x$ for the $[\pm \theta_s]_t$ specimens is expected to be influenced by the fibre scissoring that is manifested as the stiffening of the stress-strain curve. In the case of the $[\pm 26s/0]_s$ specimens, it is anticipated that, after damage begins to evolve within the UD plies, there will be a discernible loss of modulus at each cycle of loading.

Commonly, the variation in $E_x$ over the course of the loading-unloading cycles is estimated by taking a line between the zero load point (effectively the minimum strain for the cycle) and the crossover of the unloading with loading paths (the maximum strain for that cycle). This technique, shown in Fig. 5a, gives the secant modulus of each cycle. This technique has been employed in many previous studies [18,21,22,19,23] on modelling damage from constant stress secant method has the advantage that any deviation from $E_x$ can be defined in the same way. This method, however, has the potential to overestimate the decrease in $E_x$, as the secant modulus of each cycle is calculated from a higher applied stress than the previous one. This overestimation of the reduction gives an apparent modulus for each cycle that is misleading. It is possible to avoid this through the selection of a more appropriate modulus evaluation technique. Two such methods are: a secant modulus up to a constant value of applied stress, $\sigma_{s}$, (Fig. 5b) and the constant modulus of reloading for each cycle (Fig. 5c). A constant stress secant method has the advantage that any deviation from $E_x$ is easily represented, without any uncertainty over the influence that the value of stress at the maximum strain of each cycle may have on the results. Taking the tangential modulus of reloading also has this advantage, as there is no need to evaluate more than the initial slope of each cycle. This method, however, is slightly less reliable, as there can be some subjectivity in locating the correct point of reloading.

For the reasons discussed above, the constant stress secant method shown in Fig. 5b has been selected to evaluate the effect of cyclic loading on these thin ply angle-ply laminates. Values of 360 MPa and 400 MPa have been selected for the $[\pm \theta_s]_t$ and $[\pm 26s/0]_s$ specimens respectively. These values of $\sigma_{s}$ have been selected to coincide with the maximum stress of the first cycles in each case.

4.1. $[\pm 26s]_t$ and $[\pm 27s]_t$

The laminate modulus of every cycle has been measured using the constant stress secant modulus method for all $[\pm 26s]_t$ and $[\pm 27s]_t$ specimens and the results are presented in Fig. 6. The different fibre angles have been presented together, as it was found that they displayed very similar characteristics in each case. The value of $E_x$ has been normalised against the initial value and plotted against global applied strain, $\epsilon_s$. It should be noted that $\epsilon_s$ for each $E_x$ is the strain at zero stress.

Fig. 6 shows that there is very minimal reduction in $E_x$ when the secant modulus is taken between zero stress and 360 MPa. There is a <10% loss in $E_x$ that coincides with the point of laminate softening seen in the stress-strain plots of Figs. 1 and 2. At higher strains, however, $E_x$ is actually seen to increase above the initial value. Once again, the stress-strain plots show that this matches with the stiffening of the laminate due to fibre rotations.

These results show that, when subjected to cyclic loading, thin ply angle-ply laminates do not display any real decrease in stiffness and can even become stiffer at high strains. This behaviour supports the authors’ previous findings of damage suppression and fibre rotation, presented in [15,16].

Fig. 7 shows the amount of fibre angle change with applied strain in the $x$-direction for $[\pm 26s]_t$ and $[\pm 27s]_t$ specimens. The stress-strain responses shown in Figs. 1 and 2 are consistent, which allows the fibre rotation to be plotted for a single, representative specimen from each batch. The rotation, $\theta^*$, is assumed to act as a fibre scissoring mechanism and is calculated using the experimentally measured $\epsilon_x$ and $\epsilon_y$, as shown in Eq. (1) [24].
in $E_i$ observed are a consequence of the damage-free state of the laminate and the subsequent fibre rotations that take place at high strains [15]. In addition to the above analysis, these loading-unloading results indicate that the integrity of the specimens is maintained throughout the loading and they are capable of withstanding multiple loadings without significant loss of properties.

4.2. $[\pm 26/0]_i$

The constant stress secant modulus method of determining the reduction in $E_i$ has been applied to the $[\pm 26/0]_i$ stress-strain results and is shown in Fig. 8. Once again, the value of $E_i$ for each cycle is normalised against the initial laminate modulus, $E_{in}$, and plotted against $\epsilon_i$ at the zero stress point for each cycle.

An initial decrease in $E_i$ of 10% is exhibited in Fig. 8 when the secant modulus of each cycle is taken between zero and a constant stress of 400 MPa. This reduction coincides with the point of unloading for the first cycle following the start of the stress plateau. This signifies that this loss of stiffness occurs immediately after the onset of fragmentation in the $0^\circ$ plies. The modulus continues to decrease to a point where $E_i/E_{in} \approx 0.8$. This reduction coincides with the initial modulus of a $[\pm 26/0]_i$ laminate, which is shown as the blue line on Fig. 8. There is no further loss of stiffness, and at the highest strains, following the saturation of fragmentations and delaminations, there is a small indication of an increase in $E_i$ near to failure of the specimen. This, much like the $[\pm \theta_i]_s$ laminates above, shows the influence of the $\pm 26^\circ$ fibre rotation.

The overall loss of stiffness of these $[\pm 26/0]_i$ specimens is relatively low, with no more than 20% reduction for any of the tested specimens. This is primarily due to the stress plateau caused by the fragmentation and delamination of the specimens, which means the response is somewhat different to that seen from angle-ply laminates without UD plies.

As demonstrated for the $[\pm \theta_i]_s$ laminates, there is also some reorientation of the $\pm 26^\circ$ fibres in the $[\pm 26/0]_i$ specimens, shown in Fig. 9. Each specimen, plotted separately for clarity, shows a development of fibre rotation up to a peak rotation of $2.0–2.5^\circ$. On plots Fig. 9a,b and d a slight increase in the rate of reorientation at $\epsilon = 1.9\%$ can be seen. This discrete change coincides with the start of the stress plateau for each specimen and the initiation of fragmentation in the $0^\circ$ plies.

The cyclic tests have shown that these $[\pm \theta_i/0]_i$ configurations can be reloaded following the initiation of fragmentation and still retain much of their initial stiffness. This is an important aspect

It is clear from Fig. 7 that the rotations for the $[\pm 26/0]_i$ and $[\pm 27/0]_s$ specimens, reaching $2.8^\circ$ and $3.2^\circ$ respectively, are significant. Rotations of this magnitude have a discernible stiffening effect on the laminate at high strains. Whilst not immediately obvious from the stress-strain behaviour of the cyclically loaded specimens in Figs. 1 and 2, the quasi-static responses (solid black lines) in these plots clearly show a stiffening at strains in excess of 3%. Additionally, the effect of the matrix on the fibre rotations can be seen in these plots. The fibre angle of each specimen shown in Fig. 7 returns to the original value after the first cycle – prior to any ‘yielding’. Over the subsequent cycles above $\epsilon_1 = 1\%$, the fibres do not return to the original angle at the minimum strain (zero stress) point of each cycle, further indicating that the laminates develop some permanent deformation.

Any reduction in $E_i$ is often cited as the formation of damage in the form of matrix cracking or delaminations [9,25–27]. The above analysis suggests that these damage mechanisms do not occur in these thin ply laminates and so there is a complete retention of modulus after all cycles have been completed. The small increases...
of the behaviour, as maintaining stiffness is key if these laminates are to prove useful in the non-linear regime once gradual failure of the central UD plies has commenced.

5. X-ray computed tomography analysis

Following testing, specimens (#2) from each \([\pm 26_s/0_s]\) and \([\pm 27_s/0_s]\) batch have been analysed using X-ray computed tomography (CT). This method of imaging the specimens gives an insight into the internal condition and is a robust way to evaluate if matrix cracking and delaminations are present. The decision was taken not to X-ray the \([\pm 26_s/0_s]\) specimens, as a detailed CT assessment of the formation and location of damage has been made previously and is reported in [17].

Both specimens were submerged in zinc iodide dye penetrant for 24 h, in order to highlight any damage connected to the specimen edge. They were subsequently removed, wiped clean and scanned in a Nikon XT H 320LC X-ray machine. An overview scan, covering the entire gauge length, and a further three, higher resolution scans (covering the lower, middle and upper sections of the gauge length) were conducted. All X-rays were captured at 50 kV, 138 \(\mu\)A with 3141 projections. The overview and detailed scan voxel sizes were 0.084 mm and 0.026 mm respectively, (similar to the resolutions achieved in [17], which plainly showed damage in the angle-ply layers and so are deemed sufficient in this case).

Fig. 10a and b show the results of the scans of the \([\pm 26_s/0_s]\) and \([\pm 27_s/0_s]\) respectively. A view of the specimens was taken at a depth halfway between the outer surface and mid-plane, as indicated by the green dashed line on the specimen cross-section in Fig. 10. This point in the laminate thickness has been deemed to be representative of the condition of the specimens after testing. The inset images in each figure show the higher resolution scans (0.026 mm) of the central sections of the specimens. Note that the white
The areas indicated in Fig. 10 are shown in detail. The gradual reduction in specimen widths from the end tabs upwards is clear. Following the cyclic loading undergone by these specimens there is a permanent deformation that resembles the necking often observed in metals. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

Fig. 11. The areas indicated in Fig. 10 are shown in detail. The gradual reduction in specimen widths from the end tabs upwards is clear. Following the cyclic loading undergone by these specimens there is a permanent deformation that resembles the necking often observed in metals. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

dots visible in the overview images of Fig. 10 are the patterns used to track strain with the video extensometer and are not damage.

The overview images give an initial indication of the overall damage state of the specimens. For both the $[\pm26_5]_s$ and $[\pm27_5]_s$, the images show no sign of dye penetrant (a high intensity white in the X-ray) within the specimens. This confirms that there is no large scale damage such as free-edge delaminations, but there is insufficient resolution to determine the presence of other damage, such as matrix cracking. The extent of permanent deformation of these specimens is also visible: the regions near the ends show a change in width akin to the necking displayed by metals. This area is highlighted by dashed lines on each specimen in Fig. 10 and larger images of these areas are presented in Fig. 11. Whilst the deformation is slight, there is a clear reduction in width, which is highlighted using vertical dotted lines.

The uppermost inset images, which have a sufficiently high resolution (finer than the cured ply thickness) to show the fibre directions, also show no infiltration of dye into the specimens. The only dye highlights visible are on the machined edges of the specimens. Similar resolution X-ray CT scans of $[\pm26_5]_0$ specimens conducted by the authors in [17] clearly showed damage such as fibre fractures and cracking, with obvious areas of dye penetration. There are no such areas in Fig. 10, which, after multiple cycles of loading and unloading would be visible if present. The pristine condition of the tested specimens leads to the conclusion that there neither matrix cracks nor delaminations at any location in the laminate.

6. Discussion

The results presented in Figs. 6 indicate that, after quasi-static cyclic loading, the $[\pm26_5]_s$ and $[\pm27_5]_s$ laminates retain their initial modulus. The load carrying capability of these specimens is unaffected by the multiple loadings and there is even some stiffening at higher strains. This retention of modulus, after all cycles are completed, is a consequence of damage suppression, confirmed by the CT scans in Figs. 10a and b, and the slight stiffening behaviour down to the reorienting of the fibres towards the loading direction at higher strains. Using thin ply spread tow prepreg and a dispersed stacking sequence, the damage mechanisms of matrix cracking and delaminations, that normally lead to angle-ply premature failure with laminates of conventional ply thickness, are suppressed. With no discernible damage accumulation, the specimens reach strains high enough to cause a yielding of the matrix, which in turn allows an increased amount of fibre rotation to occur – as shown in Fig. 7. The resulting non-linear stress-strain response poses an interesting question: one of ductility versus pseudo-ductility. Ductility is defined by a material’s ability to be loaded beyond the yield point and then retain the initial modulus upon reloading. It has been demonstrated that there is a negligible drop in modulus and that damage is suppressed. Hence, it can be said that the response of these specimens is ductile rather than pseudo-ductile.

The stress-strain behaviour of the $[\pm26_5/0]_s$ specimens can certainly be considered to be pseudo-ductile, as the metal-like non-linearity arises primarily from the fragmentation of the $0^\circ$ plies. Fig. 8 shows also that there is some loss of modulus. Overlaying a line of the mean initial modulus for a $\pm26^\circ$ layup (normalised to the initial modulus of the $[\pm26_5/0]_1$), the effective contribution of the $0^\circ$ plies is clear. As the number of cycles and the density of $0^\circ$ fibre breaks increase, the modulus reduces until close to the initial value of a $\pm26^\circ$ laminate without any zero plies. Following fragmentation and delaminations, which are dispersed and do not necessarily lead to the complete delamination of the $26^\circ/0$ interface, there is still a low level of load carried by the $0^\circ$ plies. This, as well as the damage suppression and any fibre rotations in the $\pm26^\circ$ layers, prevents any further decrease in $E_s$. The stress-strain responses of the $[\pm26_5/0]_s$ specimens and the limited loss of modulus they exhibit, show that there is a defined lower bound for the modulus following fragmentation initiation. This predictability greatly improves the usefulness of these $[\pm26_5/0]_s$ configurations.

7. Conclusions

Analysis of thin ply $[\pm26_5]_s$ and $[\pm27_5]_s$ specimens after quasi-static cyclic loading has demonstrated that these layups retain their initial stiffness after multiple reloading. Under the same cyclic loading conditions the $[\pm26_5/0]_s$ specimens have been shown to retain ~80% of their initial stiffness.

In both cases, the overall cyclic stress-strain responses were shown to be well-matched to the monotonic results. The $[\pm26_5]_s$ and $[\pm27_5]_s$ specimens followed the trend of a ‘yielding’ and then stiffening at higher strains. The three regions of linear response, stress plateau and further loading to failure were displayed by the $[\pm26_5/0]_s$ cyclic specimens. The analysis of the modulus reduction over each cycle was performed. Taking the secant modulus of each full cycle is known to give erroneous measurements of modulus, as the measurements are taken between zero stress and a stress that increases for each cycle. A more robust method of measuring the secant modulus to a constant stress level was selected.

All specimens of both $[\pm26_5]_s$ and $[\pm26_5/0]_s$ showed some decrease of modulus at low strains, but then, after multiple loading cycles, increased or stabilised respectively. All specimens showed this excellent retention of $E_s$, which was, in part, due to the effect of the reorientation of the $\pm\theta$ fibres.

The $[\pm26_5/0]_s$ specimens, display a reduction in $E_s$ that is equivalent to a heavily reduced contribution from the zero plies; giving a controlled, predictable pseudo-ductility. Significantly, there is a negligible reduction in $E_s$ for $[\pm26_5]_s$ and $[\pm27_5]_s$ specimens. X-ray CT scans confirm that these thin ply angle-ply laminates suppressed any damage accumulation in the form of matrix cracking and free-edge delaminations. Both these aspects demonstrate that thin ply angle-ply laminates can be reloaded and still carry load effectively – showing a ductile behaviour.
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