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DAMAGE SUPPRESSION IN THIN PLY ANGLE-PLY CARBON/EPOXY LAMINATES

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1 Introduction

Working within the High Performance Ductile Composite Technology (HiPerDuCT) programme, the aim of the current work is to realize some pseudo-ductility through fibre reorientation and damage suppression to fully exploit the potential of angle-ply laminates under uniaxial tension. The term ‘pseudo-ductility’ is used in this case, as the fibre rotation itself, combined with matrix yielding, leads to the large strains exhibited. This concept of ‘excess length’ is made possible by suppressing damage via reduction of ply thickness.

Angle ply laminates loaded in tension have shown promise of high strains to failure, but often fail prematurely due to matrix microcracking and high free-edge stresses that lead to delaminations [1]. These failure mechanisms have been widely studied [2-6]. O’Brien [4] characterised the onset and development of delaminations in angle-ply laminates consisting of [+θ/- θ/90°]. Delaminations were seen to develop at the edges of the specimens, at +θ/0° and –0/90° interfaces. The delaminations at the –0/90° interfaces grew considerably, whereas edge delaminations at the +θ/0° interfaces remained isolated, resembling small triangles. The non-linear stress-strain behaviour was associated with a ‘stiffness loss’ brought about by the accumulation of damage. A strain energy release rate (G) approach was employed to predict the initiation of delamination. The critical value of G was found to depend only on the laminate stacking sequence and location of delaminations. Wang and Crossman [5] and Crossman et al [2] also used a strain energy concept to determine the failure mechanisms of ±25/90° laminates. They presented an analytical model that predicted matrix cracking, edge delaminations and showed the importance of the ply thickness in calculating the value of G. It is suggested in [2] that reduction of ply thickness could suppress microcracking and delaminations. This conclusion is shared by Rodini and Eisenmann [6]. Kim and Soni [3] reported that, for [±30°/90°] and [±30°/90°], laminates, the threshold for delamination decreased with increased ply blocking, effectively thicker plies.

The laminates presented in [2-6] all contain a number of 90° layers – the failure of which via matrix cracking is coupled with the delaminations described. In an angle-ply laminate (+θ/-θ/90°), however, this matrix cracking is not present at such low stresses and the edge effects are more straightforward to isolate. Herakovich [7] investigated this using [(+θ/-θ)2], and [±θ/θ] laminates. It was concluded that, for all angles tested, the strength was inversely proportional to the layer thickness. More recently, Leguillon et al [8] examined edge delamination initiation and the influence ply thickness has on this failure mode. They reported decreases in delamination stress with increased layer thickness.

Despite this knowledge, difficulties and expense of manufacturing have limited work on reducing the ply thickness below the standard 0.125 mm. Recent advances in tow-spreading technology have allowed so called ‘thin’ prepps to be produced. Sasayama et al [9] developed a pneumatic technique, which is described in detail by Sihn et al [10].

Several studies have been undertaken [10-13] to experimentally investigate the general behaviour of thin ply laminates. Sihn et al [10] demonstrated that, from both static and fatigue tension tests of un-notched and notched quasi-isotropic (QI) specimens, impact and compression-after-impact (CAI) tests, thin ply laminates (ply thickness of 0.04 mm) showed less microcracking, delaminations and splitting than specimens with thicker plies (0.2 mm ply thickness). Following fatigue testing of un-
notched specimens, they demonstrated that, after 50000 cycles, the thin ply laminates maintained stiffness and strength. Conversely, in notched tests – both static and fatigue – the thin ply specimens exhibited lower strength. Differences in failure modes were observed: pull-out behaviour from the thick ply laminates and brittle failure from the thin plies. X-ray images taken prior to failure show very little development of damage within the thin ply laminates, especially around the notch itself. Yokozeki et al. [11, 14] investigated the compressive strength and damage resistance of thin ply laminates under both in-plane and out-of-plane loadings. In all cases, the thin ply laminates were shown to be more resistant to damage accumulation. This was particularly noticeable in out-of-plane transverse loading and CAI tests of QI laminates. Thick ply laminates (0.14 mm ply thickness) exhibited considerable delaminations on the back face, whereas the thin ply specimens (ply thickness 0.07 mm) showed only internal delaminations. As above, this suppression of damage led to sudden brittle failure.

Ogihara and Nakatani [13] presented work on carbon/epoxy angle-ply laminates, concentrating on the effect of ply thickness. Specimens of ±45° and ±67.5° both showed increases in strength with ply thicknesses of 0.05 mm rather than 0.15 mm. A mesoscale damage model, devised by Ladeveze and LeDantec [15], was employed to show also that the thin-ply laminates were significantly more damage resistant. The issue of fibre rotation in the thin-ply specimens was not addressed. This principle of fibre reorientation via a scissoring action has been shown to influence the tensile behaviour of angle-ply laminates, particularly calculations of in-plane shear stress [16]. As such, it will be central to the present investigations.

2 Experimental Methods

Skyflex USN020A, a newly available spread tow carbon/epoxy prepreg, produced by SK Chemicals, with a cured ply thickness of only 0.03 mm has been chosen as a model material. The prepreg consists of Mitsubishi Rayon TR30 carbon fibres [17] and K50 resin, a semi-toughened epoxy. To fully characterize the material elastic properties, quasi-static tensile testing of [0]₁₆, [90]₁₆ and [±45]₁₆ laminates has been performed, the results of which are given in Table 1. Unidirectional (UD) specimens had a gauge length of 100 mm, width of 10 mm and thickness of 0.48 mm. [±45]₁₆ laminates had a gauge length of 150 mm, width of 15 mm and thickness of 0.6 mm. From knowledge of the fibre and resin properties, a fibre volume fraction (V_f) of 42% has been calculated. Mechanical testing of unidirectional specimens was conducted at 1 mm/min and all angle-ply laminates at 2 mm/min.

To validate modelling, tensile testing has also been performed with angle-ply laminates covering a range of layup angles: ±15°, ±20°, ±25°, ±26°, ±27° and ±30°. All specimen dimensions were as for the ±45° layup described above.

Strain data was captured using an Imetrum Video Extensometer and associated software. In all cases, the true stress and strain have been computed to account for the change in cross-sectional area at high strains.

3 Modelling

X-ray Computed Tomography (CT) imaging performed on a selection of the specimens following tensile testing of [±30]₁₆ and [±45]₁₆ laminates (Fig. 1), shows that there is very limited damage within the gauge length. These images show that failure was a local event, also demonstrated by the micrograph of a section of [±45]₁₆ laminate in Fig. 2. A failure surface similar to that of woven fabric composites is seen, with any damage limited to the immediate area. This suppression of damage allows a modelling approach based on plasticity, rather than damage mechanics, to be undertaken.

The selected methodology is based on that presented by Sun and Chen [18, 19]. Initially a micromechanical model [19] that treats the constituent fibre and matrix as an elastic-plastic material, is used to identify the plastic behaviour of the matrix. Due to the complexity of the micromechanical approach, however, it is solely used as a tool to define parameters for the orthotropic plasticity model [18]. The next stage of the process models the tensile behaviour of an angle-ply laminate, incorporating the previously determined matrix properties. The plasticity modelling is described in detail within [18] and [19], so only the key elements and assumptions of each stage of the method shall be highlighted at this point.
All modelling has been conducted using the Matlab programming language.

### 3.1 Micromechanical Model

The micromechanical model is utilised to define the plastic properties of the matrix material. Incremental solutions of the fibre and resin compliance matrices are found and then combined to give the tensile response of the unit cell of composite. The unit cell of material is idealised, as shown in Fig. 3, as a square cross-section of fibre (region AF), bounded by two regions of matrix (AM and B). This area is quarter-model of the fibre and matrix, assuming a rectangular distribution of fibres in the composite. The fibre region is assumed to be orthotropic linear-elastic and the matrix isotropic elastic-plastic, following the von Mises J2-flow rule. The respective material properties used in the model are set out in Table 2. The transverse, E_{22}, and shear, G_{12}, moduli have been reached via knowledge of the longitudinal fibre modulus, E_{11}, (from Skyflex data [17]) and unidirectional material stiffness matrix, [Q]. The values of E_{22} and G_{12} were adjusted until the values of [Q] matched those already calculated for the Skyflex material from material characterisation testing. A state of plane stress is also assumed to exist perpendicular to the x – y (1 – 2) plane, leading to \( \sigma_{33} = \sigma_{55} = \sigma_{13} = 0 \). The plastic behaviour of the matrix is described by the effective plastic strain, \( \varepsilon^{plM} \), and effective stress, \( \sigma^{M} \), which are related via a power law,

\[
\varepsilon^{plM} = \beta(\sigma^{M})^n
\]  (1)

Initial values of \( \beta \) and \( n \) are found, using (1), and provide starting points for describing the plastic strain increments in the matrix regions (AM and B) within the model. The values of \( \beta \) and \( n \) are then adjusted to give an axial stress – strain response that closely matches the experimental results, as shown in Fig. 4. These values are stated in Table 3.

The next step is to produce off-axis stress – strain curves over a range of fibre angles (Fig. 5). The unit cell of material is maintained, whilst the fibre is oriented at an angle, \( \theta \), to the loading direction. Employing (2), and arbitrarily setting \( a_{66} = 1 \), the \( \varepsilon^{plM} \) and \( \sigma^{M} \) for each fibre angle can be calculated. It is necessary to determine another value of \( a_{66} \) for the laminate-level plasticity model. Sun and Chen [19] show, using a composite of boron – aluminium, that one value of \( a_{66} \) can collapse the effective stress – plastic strain curves of each off-axis angle on to the 90° data. Fig. 6 demonstrates that the same is applicable for the Skyflex USN020A material.

### 3.2 One-Parameter Orthotropic Plasticity Model

The simple one-parameter plasticity model presented by Sun and Chen [18] assumes that non-linearity in the fibre direction is negligible; with all the plasticity originating from the transverse and in-plane shear stresses. In this respect, the model is well suited to implementation with angle-ply laminates, as the transverse and shear stresses interact and can be modelled with the plastic potential function,

\[
f = \frac{1}{2} \left( \sigma_{22}^2 + 2a_{66}\sigma_{12}^2 \right)
\]  (6)

A state of plane stress is deemed to exist and the material is assumed to remain orthotropic throughout – allowing the use of a constant value of \( a_{66} \).

The plastic behaviour of the composite is again defined by a power law, relating effective plastic strain, \( \varepsilon^{p} \), and effective stress, \( \sigma^{e} \),

\[
\varepsilon^{p} = A(\sigma^{e})^c
\]  (7)
Where $A$ and $r$ are calculated from performing a regression analysis in order to fit a power law curve to the data shown in Fig. 6. The values of $A$, $r$ and $a_{66}$ used in the model are presented in Table 3. Within the model, the effective stress is defined as:

$$\bar{\sigma} = \sqrt{3f}$$

(8)

The solution is found by relating the incremental strains and stresses for the laminate,

$$[d\epsilon] = [S][d\sigma]$$

(9)

The compliance matrix, $[S]$, allows computation of the overall behaviour by including the elastic and plastic contributions to the stiffness of each ply in the laminate.

Fibre rotation is assumed to take place as a scissoring action, as described in [16, 20, 21]. The fibres are treated as inextensible and modelled as rotating towards the longitudinal axis of the specimen, as shown in Fig. 7, where the ‘rotated’ fibre angle, $\theta'$, is defined as:

$$\theta' = 90 - \arctan \left( \frac{1 + \epsilon_x}{\tan(\theta) + \epsilon_y} \right)$$

(10)

The plasticity model is contained within an iterative, non-linear classical laminate analysis (CLA) solution that incorporates the fibre rotation. The compliance matrix, $[S]$, is recalculated at each loading increment, allowing the change in ply stiffness, caused by the reorientation of fibres and matrix plasticity, to be accounted for.

Failure of the laminate is based on a maximum strain criterion. Using experimentally obtained values for both tensile and compressive strains to failure in the three principal material directions ($\epsilon_{11}$, $\epsilon_{22}$, $\gamma_{12}$), each ply is checked for failure at the end of every loading increment. If a failure is recorded, the loading stops and the stress – strain data for that laminate is stored.

4. Model Validation

Quasi-static tensile testing of various angle-ply laminates has been conducted in order to validate the modelling described above.

Table 4 presents strengths, failure strains and final fibre angles for all the angle-ply laminates experimentally tested and the respective model results in each case.

The $[\pm 15]_s$, and $[\pm 20]_s$, specimens in particular were dominated by the fibre direction material properties. This can be seen, in Fig. 8, from the low level of non-linearity developed prior to failure. The relatively low initial value of $G_{12}$ (2.4 GPa) for the Skyflex material, meant that the shear yield point was not reached and only a small degree of non-linearity was developed. Failures in these cases were explosive and fibre dominated. The model provides a very good match to the experimental results. At strains close to 2%, a slightly stiffer response is predicted for the $[\pm 20]_s$ layup than seen in testing.

Tests of $[\pm 25]_s$, $[\pm 26]_s$, and $[\pm 27]_s$, specimens were performed in order to evaluate a range of layups that were predicted to exhibit relatively high stiffness and promising levels of non-linearity in strain. Fig. 9 shows that the predicted strains to failure are surpassed by the experimental results. At low strains the correlation is excellent. Beyond 2% strain, all three specimens show divergence from the predictions. The strength predicted for both $[\pm 26]_s$ and $[\pm 27]_s$ laminates matches closely with the experimental values. The experimentally observed sudden, brittle failures, exhibited by each specimen in the $[\pm 25]_s$, $[\pm 26]_s$, and $[\pm 27]_s$, series of tests were all initiated within the gauge length. Inspection of the failed specimens showed that any damage was local to the failure surface. In the case of specimens that failed at near to mid-gauge length (i.e. far enough from the end tabs to diminish their influence), failure occurred perpendicular to the specimen edge – indicating fibre failure across all plies.

Most notably, there is very little loss in strength between the experimental $[\pm 25]_s$, and $[\pm 26]_s$, specimens. This occurs, however, with almost a 20% increase in strain to failure for the $[\pm 26]_s$.

Results for the $[\pm 30]_s$ laminates show more non-linear behaviour and, similar to the $[\pm 45]_s$, a shear dominated failure mode. As can be seen in Fig. 9 and Fig. 10, the predictions for both layups are closely matched to the experimental behaviour. Whilst the $[\pm 30]_s$ prediction gives, overall, a stiffer response than the experimental, the gradients of the two curves can be seen to be similar at higher strains. This indicates that the model is accurately predicting the stiffness of the laminate, as the amount of fibre rotation and plasticity develops.
Table 4 shows that the predicted fibre rotation is within 0.4° of the experimentally observed mean value. For both layups, little damage was in evidence, with no obvious signs of edge delaminations.

The [±45]s specimens show strains to failure in the region of 20% and fibre reorientation in excess of 10°, leading to a large offloading of stress onto the fibres. This is manifested, as shown in Fig. 10, by a stiffening of the laminate at high strains. The overall predicted response is reasonably well matched to the experimental data. The experimental curve shows a much more pronounced yield point at about 120 MPa, leading to a large softening of the laminate. The experimental response begins to stiffen sharply around 200 MPa, whereas the model predicts that the laminate will only stiffen at over 250 MPa. In excess of 16%, the failure strain from modelling is only slightly lower than the 17% seen in experiments. The model has been unable to accurately predict the strength, predominantly due to the lower degree of stiffening at higher strains.

It is important to assess the accuracy of the fibre rotation predicted by the modelling. Table 4 also contains the predicted and mean experimental fibre angle at failure for each angle-ply laminate. A good agreement is in evidence for all specimens, giving confidence that the approach is accurately predicting the stress state in the laminates. Fig. 11 shows also that, with the [±27]s specimen as an example, the predicted reorientation of fibres follows the experimental trend. There is an acceleration of fibre rotation with increasing strain. This indicates that the matrix, in agreement with modelling, has undergone some yielding and there is plastic flow, allowing the fibres to reorient.

5. Conclusions

This study has examined the monotonic tensile behaviour of thin ply angle-ply laminates and incorporated a one-parameter plasticity model [18] into analyses of fibre rotations to provide a predictive tool. The very thin, 0.03 mm, ply thickness of the Skyflex USN020A carbon/epoxy prepreg material has effectively suppressed the principal mechanism of failure – edge delaminations – and allowed substantial non-linear strains to develop. Considerable fibre reorientation has demonstrated that the concept of ‘excess length’ is able to produce some pseudo-ductile behaviour in a carbon/epoxy laminate. Angles between 25° – 27° have been shown to hold the most promise in terms of this concept. Laminate strengths in the region of 60% of the unidirectional material strength have been exhibited, in conjunction with failure strains in the range of 3.5 – 4.5 %.

The simple modelling technique employed to predict the overall stress-strain response has been shown to be sufficient. At low strains, the non-linear behaviour of each laminate has been captured well. There is, however, less accuracy at higher strains. This is partly due to the generality and simplicity of the plasticity model – based on a power law to describe the matrix yielding. This has been shown to be accurate for the micromechanical unit cell of unidirectional material, though the modelling of a ±θ laminate is somewhat more complicated. Other factors include the difficulty in accurately representing the fibre reorientations taking place and the effect on the laminate during this process. Despite this, the model has been shown to produce effective predictions of strength and axial failure strain.

Acknowledgement

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Table 1
Elastic properties of Skyflex USN020A

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<table>
<thead>
<tr>
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<tr>
<td>( E_{11} )</td>
<td>101.7 GPa</td>
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<tr>
<td>( E_{22} )</td>
<td>6.0 GPa</td>
</tr>
<tr>
<td>( G_{12} )</td>
<td>2.4 GPa</td>
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<td>( v_{12} )</td>
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Table 2
Elastic properties of TR30 fibre and K50 matrix, as used in micromechanical model. [17]

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<th>TR30</th>
<th>K50</th>
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<tr>
<td>( E_{11} )</td>
<td>234.8 GPa</td>
<td>3.35 GPa</td>
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<tr>
<td>( E_{22} )</td>
<td>13.0 GPa</td>
<td>3.35 GPa</td>
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<tr>
<td>( G_{12} )</td>
<td>15.0 GPa</td>
<td>1.21 GPa</td>
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<tr>
<td>( v_{12} )</td>
<td>0.2</td>
<td>0.38</td>
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Fig. 1. X-ray CT images for (a) [±30]s, and (b) [±45]s, specimens. Each image is taken from the laminate mid-plane. It may be noted that the small triangular shape and line across (a) and criss-cross of grey lines on (b) are artefacts from CT scan of the laminates’ surface layers.

Fig. 2. Micrograph showing the failure surface of a [±45]s specimen. The brush-like fibres demonstrate that failure was similar to the mechanisms seen in woven composites.

Table 3

| Plasticity terms for micromechanical and laminate level modelling of Skyflex USN020A. (All values quoted require stress in Pa.) |
|---|---|
| $\beta$ | $4.0 \times 10^{-33}$ |
| $n$ | 3.86 |
| $A$ | $2.1 \times 10^{-31}$ |
| $r$ | 3.52 |
| $a_{66}$ | 2.1 |

Fig. 3. Diagram of unit cell representation of fibre–matrix regions for micromechanical model.

Fig. 4. Experimental data for [90]16 laminates, overlaid with micromechanical model output.
**Fig. 5.** Off-axis stress – strain curves for Skyflex USN020A, as produced by the micromechanical model.

**Fig. 6.** Effective stress against effective plastic strain for off-axis angles. A single value of $a66$ has collapsed the curves on to one that can be represented by a power law.

**Fig. 7.** Off-axis stress – strain curves for Skyflex USN020A, as produced by the micromechanical model.

**Fig. 8.** Applied stress – axial strain plotted against model output for $[\pm 15]_{5s}$ and $[\pm 20]_{5s}$.

**Fig. 9.** Applied stress – axial strain plotted against model output for $[\pm 25]_{5s}$, $[\pm 26]_{5s}$, $[\pm 27]_{5s}$ and $[\pm 30]_{5s}$.

**Fig. 10.** Applied stress – axial strain plotted against model output for $[\pm 25]_{5s}$.

**Fig. 11.** Model and experimental fibre rotation trend for $[\pm 27]_{5s}$ laminate.
Fig. 7. Schematic of fibre reorientation concept. Fibres are shown to rotate towards the longitudinal axis of the specimen with the application of a tensile load.

Table 4
Mean strength, strain to failure and fibre rotation for model and experimental results. Coefficient of variation (CV) appears below the relevant experimental result.

<table>
<thead>
<tr>
<th></th>
<th>[±15]_s</th>
<th>[±20]_s</th>
<th>[±25]_s</th>
<th>[±26]_s</th>
<th>[±27]_s</th>
<th>[±30]_s</th>
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<td>σ_s (MPa)</td>
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<td>Test</td>
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<td>5.8</td>
<td>2.6</td>
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<td>1316.5</td>
<td>1050.5</td>
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<td>ε_x (%)</td>
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<td>4.29</td>
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<td>8.1</td>
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<td>3.43</td>
<td>3.57</td>
<td>3.76</td>
<td>4.52</td>
<td>16.18</td>
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<tr>
<td>θ' (°)</td>
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<td>16.8</td>
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<tr>
<td>CV (%)</td>
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<td>1.8</td>
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<td>20.3</td>
<td>22.9</td>
<td>33.6</td>
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References

1. X. Xu and M. Wisnom. "An experimental and numerical investigation of the interaction between splits and edge delaminations in [+20\textdegree, 0\textdegree, -20\textdegree]_m carbon/epoxy laminates". Proceedings of ECCM15, Venice, 2012.


