Properties and failure mechanisms of z-pinned laminates in monotonic and cyclic tension

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Abstract

The effects of through-thickness reinforcement of carbon/epoxy laminates with thin pins on the in-plane tensile properties, tensile fatigue life and failure mechanisms are investigated. Tensile studies in the 0°/C176 fibre direction are performed on unidirectional and quasi-isotropic laminates reinforced with different volume contents and sizes of fibrous composite z-pins. Microstructural analysis reveals that z-pinning causes several types of damage, including out-of-plane fibre crimping, in-plane fibre distortion, mild dilution of the in-plane fibre volume fraction due to laminate swelling, and clusters of broken fibres. In unidirectional composites, resin pockets form around pins and coalesce into continuous resin channels at higher z-pin contents. Young’s modulus falls only a few percent at most, due partly to in-plane fibre dilution and partly to fibre waviness. Monotonic tensile strength is degraded more significantly, falling linearly with both pin content and pin diameter. Comparison with prior data shows that the rate of degradation is evidently a strong function of the particular pin insertion method used. Failure mechanisms include fibre rupture, presumably affected by broken fibres, and, in unidirectional laminates, longitudinal splitting cracks emanating from resin pockets. Whereas non-pinned laminates show very modest fatigue effects, the pinned laminates exhibit strong fatigue effects, with strength falling by as much as 33% at 10^6 cycles. The slope of the fatigue life (S–N) curve tends to increase in magnitude with pin content and density. Limited evidence and prior literature suggest that the dominant fatigue mechanism may be progressive softening and fibre damage in misaligned segments of in-plane fibres.

Keywords: A. Polymer–matrix composites; B. Fatigue; B. Mechanical properties; Z-pins

1. Introduction

A long-standing problem with fibre-reinforced polymer laminates is their low delamination resistance and poor impact damage tolerance due to the lack of through-thickness reinforcing fibres. This problem has restricted the use of laminated composites in structures prone to impact, in-plane shear or through-thickness tensile loads. New materials and techniques have been developed to increase the delamination toughness and impact resistance of laminates, including toughened resins, fibre treatments, stitching and through-thickness weaving. Another technique that has great potential is z-pinning, which involves inserting thin metal or fibrous composite pins through a laminate to provide through-thickness reinforcement.

Various z-pinning methods have been proposed or used since the 1970s for inserting pins oriented through-the-thickness of a laminate [1–12]. The early insertion methods were rudimentary, and basically involved the manual pushing of each pin into the composite, which is slow and potentially imprecise. The particular method used in the present study, and perhaps the most widespread for high-performance polymer composite laminates over the last decade, is the process of pushing pins from a foam bed by pressure and acoustic vibration into prepreg [8,12]. An advantage of this process is that the insertion process can be semi-automated for wide-area reinforcement of a composite component.

Z-pins can increase the in-plane shear strength, interlaminar fracture toughness and impact damage resistance
of laminates as well as improve the ultimate failure strength of composite joints by a crack bridging mechanism [3,8,9,13–17]. For these reasons, z-pinned laminates are used in military aircraft such as the F/A-18 E/F Hornet and in Formula 1 racing cars [18]. However, all methods of pinning, including insertion by vibration and pressure from a foam bed, are associated with deformation of the in-plane fibres, which may lead to degradation of the in-plane properties of the laminate. An important step towards certifying the safety of structures that contain z-pins must therefore be to develop methods of accounting with confidence for in-plane property knockdown and its dependence on defects induced by the pinning process.

In prior studies of in-plane property knockdown, Steeves and Fleck [19] and Stringer and Hiley [15] report that the compressive strength of carbon/epoxy laminate is degraded by local distortion and crimping of the load-bearing fibres around the z-pins, which lowers the compressive strain required to create kink bands that cause failure. Freitas et al. [8,9,20], Steeves [21] and Stringer and Hiley [15] found the tensile strength of carbon/epoxy laminate is also reduced with z-pinning. The mechanisms responsible for the loss in tensile strength are not well understood, although it is attributed to a reduction in the fibre volume content caused by fibre spreading together with fibre distortion and resin-rich regions around the z-pins. Published data on the tensile properties of z-pinned laminates are scarce [8,9,20,21], and the influence of the microstructural damage caused by z-pinning on the strengthening processes and failure mechanisms have not been determined. Furthermore, the tensile fatigue properties and cyclic damage mechanics of z-pinned laminates have not been evaluated, despite the potential application of these materials in structural components subject to fatigue loading, such as wing panels and joints on aircraft.

The aim of this research is to investigate the effect of z-pinning on the in-plane tensile properties, tensile fatigue performance and failure mechanisms of carbon/epoxy laminates. The influence of the volume content and size of the z-pins on the Young’s modulus, tensile strength and fatigue life in the 0° fibre direction of unidirectional and quasi-isotropic laminates is determined, and changes to these properties are related to changes in the microstructure and failure mechanisms caused by z-pinning.

2. Materials and experimental techniques

2.1. Fabricating z-pinned laminates

The laminates were made using a carbon/epoxy (CYCOM 970) prepreg tape supplied by Cytec. The epoxy matrix is a high cure temperature, non-toughened resin. The specimens contained 20 plies of tape stacked in a unidirectional or quasi-isotropic [0/±45/90/±45/0], pattern. Prior to curing, the laminates were debulked by vacuum bagging and then z-pinned using pultruded carbon/bismaleimide Z-Fibers™ supplied by Aztex Inc. The z-pins were 8 mm long, and their tips were chamfered to an angle of 45° to ease their insertion into the prepreg stack.

The z-pins were inserted in the orthogonal direction through the uncured prepreg using a hand-held ultrasonically actuated horn in a process that is shown schematically in Fig. 1, and is described in detail by Freitas et al. [9,10]. The z-pins as supplied are arranged in a square pattern within a low-density foam preform. The foam is used to ensure an even spacing between the z-pins and to provide lateral support to the pins during insertion. The z-pins were inserted progressively by moving the ultrasonic horn over...
the foam bed several times until all the pins had fully penetrated the prepreg, which was determined to be when their leading tip protruded slightly from the underside of the prepreg stack. The z-pins protruded by less than 0.5 mm, and the excess length was carefully abraded away after curing using fine-grade polishing paper without scratching the surface ply. The z-pins also protruded from the entry side because they were considerably longer than the thickness of the prepreg stack, which was nominally 4 mm thick. The excess pin length on this side was about 3.5 mm, and this was removed by first shearing the pins along the laminate surface using a sharp blade and then carefully polishing away any excess length. A considerable shear force needs to be applied to cut the z-pins, and this caused some lateral displacement and deformation of in-plane fibres, which are described later. This process of cutting the z-pins is commonly practised in the manufacture of z-pinned laminates [8–10]. After cutting, two final plies were placed on the stack to stop dimpling of the surface caused by protruding z-pins. This ensured a smooth surface finish. The addition of these two plies was included in the final thickness of the laminate.

The laminates were reinforced with thin (0.28 mm) diameter z-pins to volume contents of 0.5%, 2.0% and 4.0%. These specimens were used to investigate the effect of z-pin content on tensile properties and fatigue performance. In addition, the unidirectional laminate was reinforced to a volume content of 2.0% using thin (0.28 mm) or thick (0.51 mm) diameter z-pins to study the effect of pin size on the mechanical performance.

The entire gauge region of the tensile coupon specimens was z-pinned, with the z-pins aligned in parallel rows along the specimen (see Fig. 2). The spacing between the axial rows of small diameter z-pins was 3.5, 1.75 and 1.2 mm for the volume contents of 0.5%, 2.0% and 4.0%, respectively. The row spacing for the large diameter z-pins was 3.2 mm, thus maintaining the volume content of 2.0%.

After z-pinning, the laminates were consolidated and cured in an autoclave at an overpressure of 500 kPa and temperature of 115 °C for 1 h and then 750 kPa and 180 °C for 2 h. The average fibre volume fraction of the laminates was nominally 62%. For reference, unpinned unidirectional and quasi-isotropic laminates were manufactured using the same curing process.

2.2. Fiber volume fraction and laminate thickness

The laminates were constrained within a picture frame during z-pinning to suppress lateral spreading of the fibres due to swelling that can occur due to the displacement of in-plane fibres. Table 1 gives the thickness of the laminates with different volume contents and sizes of z-pins. The nominal thickness of the laminates is 4.0 mm, but the thickness varies by a few percent from specimen to specimen.

In the unidirectional laminates, the variation is correlated with the volume displaced by the pins: the thickness rises by a fraction that is approximately the same as the pin volume fraction. Thus the volume fraction of the in-plane fibres in the total laminate volume less the volume of the pins is approximately conserved. However, as detailed in Section 3, inserting pins into the unidirectional laminates also creates significant resin pockets next to each pin along the fibre direction. Since the pockets represents further volume from which fibres have been displaced, of magnitude roughly twice that occupied by the pins themselves (see below), it follows that fibre compaction must also occur in some zones within the laminate. Indeed, fibre displacement is accommodated approximately one-third by thickness increase and approximately two-thirds by fibre compaction. The spatial distribution of the fibre compaction is not easily determined.

In the quasi-isotropic laminates, the thickness rises by a fraction that is approximately the same as the pin volume fraction only for the laminate with the highest content of pins. The other laminates have unchanged thickness, within experimental uncertainty. Part of the difference with

<table>
<thead>
<tr>
<th>Laminate</th>
<th>Z-Pin volume content (%)</th>
<th>Z-Pin diameter (mm)</th>
<th>Thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Unidirectional 0.0</td>
<td>0</td>
<td>3.94 ± 0.13</td>
<td></td>
</tr>
<tr>
<td>Unidirectional 0.5</td>
<td>0.28</td>
<td>4.00 ± 0.07</td>
<td></td>
</tr>
<tr>
<td>Unidirectional 2.0</td>
<td>0.28</td>
<td>4.02 ± 0.07</td>
<td></td>
</tr>
<tr>
<td>Unidirectional 4.0</td>
<td>0.28</td>
<td>4.07 ± 0.07</td>
<td></td>
</tr>
<tr>
<td>Unidirectional 2.0</td>
<td>0.51</td>
<td>4.00 ± 0.07</td>
<td></td>
</tr>
<tr>
<td>Quasi-isotropic 0.0</td>
<td>0</td>
<td>4.00 ± 0.04</td>
<td></td>
</tr>
<tr>
<td>Quasi-isotropic 0.5</td>
<td>0.28</td>
<td>4.01 ± 0.03</td>
<td></td>
</tr>
<tr>
<td>Quasi-isotropic 2.0</td>
<td>0.28</td>
<td>3.99 ± 0.06</td>
<td></td>
</tr>
<tr>
<td>Quasi-isotropic 4.0</td>
<td>0.28</td>
<td>4.10 ± 0.03</td>
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Fig. 2. Dimensions of the tensile specimens.
unidirectional laminates could be that resin pockets are much smaller in the quasi-isotropic laminates (see below), so that fibres are displaced from a lesser volume, not much greater than the volume of the pins themselves. For the lower pin densities in the quasi-isotropic laminates, most of the fibre displacement is evidently accommodated by fibre compaction alone.

2.3. Monotonic and fatigue tensile testing

Monotonic and fatigue tensile tests were performed in the 0° fibre direction on the laminates using rectangular coupons with the dimensions given in Fig. 2. The monotonic tests were performed using a 250 kN MTS machine at a loading rate of 1 mm/min. The fatigue tests were performed using the same MTS machine operated under a cyclic load sinusoidal waveform with a stress (R) ratio of 0.6 and loading frequency of 5 Hz. The materials were tested to a range of peak fatigue stress levels between 70% and 95% of the monotonic failure strength to generate fatigue life (S–log N) curves. The minimum peak fatigue stress of 70% was used because this is about the fatigue limit of the laminate, and testing at lower stress levels failed to cause fatigue-induced fracture. The number of load cycles to failure (N) was taken to be when the laminate could no longer carry the peak fatigue stress, and this coincided with complete fracture of the specimen. Fatigue tests performed on specimens that did not fail were stopped at one million load cycles.

3. Results and discussion

3.1. Microstructural characterisation

Microstructural changes to the fibre architecture of composites caused by the insertion of z-pins are known to alter the mechanical properties [8–10,19,21]. Therefore, the microstructure of the unidirectional laminate was examined to aid in understanding the effect of z-pinning on the mechanical properties and failure mechanisms. Fig. 3 shows a cross-section view of a z-pin in the laminate, and the pin is seen to be inclined at an angle (ϕ) from the orthogonal (through-thickness) direction. Microstructural analysis revealed that ϕ was statistically distributed, as shown by the histograms and fitted Gaussian density functions of Fig. 4. The median angles for the small and large diameter pins were 14° and 23°, respectively.

The cause of the offset was investigated by measuring the distribution of ϕ at different stages in the manufacture of the z-pinned unidirectional laminates, viz. (i) in the as-received condition in the foam preform, (ii) after z-pins had been inserted about half-way through the laminate, (iii) after z-pins had been inserted completely through the laminate, (iv) after removal of the excess pin lengths from both surfaces, and (v) in the laminate after consolidation and curing in the autoclave. The median offset angles for the two pin sizes after these manufacturing stages are given in Fig. 5. The small Z-pins are not offset significantly from the original (as-received) condition during insertion. The offset angle of the small Z-pins increased suddenly when their excess length was removed by shear cutting and abrading. The direction of the rotation is that in which the shear force was applied. In the case of large diameter

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**Fig. 3.** Cross-section view of a z-pin in the unidirectional laminate. Note that the pin is inclined at an angle (ϕ) from the orthogonal direction.

**Fig. 4.** Median offset angles for the (a) small and (b) large z-pins during the different manufacturing stages of the z-pinned laminates.
pins, shear cutting caused a relatively small increase in $\phi$, with the largest increase caused when the laminates are consolidated under pressure in the autoclave.

Fig. 3 also shows crimping of carbon fibres around a z-pin near the mid-plane of the laminate. The crimping was severe, with the fibres being misaligned by as much as 60–70° from their original direction. However, damage was confined to a small region close to the z-pin (within 0.3 mm). It is seen in Fig. 3 that the crimping was not symmetric around the pin, but was more severe on the side inclined to an angle under 90°. This indicates that crimping is caused by a combination of the force on the pins during insertion and the rotation of the pin when they become misaligned. Thin elongated voids were present inside some of the z-pins, including the pin shown in Fig. 3. These voids were created during the manufacture of the z-pins, presumably because the matrix resin could not infiltrate regions of high fibre content.

It is possible that some of the carbon fibres were broken during the z-pinning process. The pin tip was chamfered to assist their insertion through the laminate by spreading the fibres. However, it is possible that some fibres could not be spread sufficiently to avoid being crushed under the pin tip. The proportion of the fibres that may have been fractured by the z-pins could not be accurately measured, although it is expected that the broken fibres occur in clusters close to the pins. There was clearly far less damage to fibres than depicted in prior work by Steeves and Fleck [19], which is attributed to the superior insertion technique used in this study. However, anecdotal evidence has made clear that results in this method of pin insertion, viz. by pressure and vibration from a foam base, depend on the tool used and operator skill.

Fig. 6 shows an in-plane view of a z-pin in the unidirectional laminate, and it is seen that the carbon fibres have been spread to accommodate the pin. This causes distortion of the fibres and the formation of resin-rich pockets where the fibres have been displaced. The fibre distortion angle ($\theta$) was measured from photomicrographs to be 4.0 ± 0.5° for the small diameter pins and 5.4 ± 0.8° for the large pins. The fibres within a small volume of material surrounding each z-pin were displaced in the axial ($x$-) and transverse ($y$-) directions, as shown schematically in Fig. 7. The dimensions of the affected area were measured from photomicrographs of specimens containing a single z-pin, and it is seen the fibres were distorted in both directions over a distance of ~1.0 mm for the small pin size and 1.5 mm for the large pin.

While the fibres surrounding each z-pin in the unidirectional laminate were only distorted over a small region, when the pins were spaced closely together the resin-rich regions join together to form continuous resin channels along the axial rows of z-pins, as shown in Fig. 8. It is seen in Fig. 6 that the length of the resin-rich region along the fibre direction in the laminate reinforced with the small pins is ~2 mm. When the spacing between the z-pins is less than this length, neighbouring resin-rich regions join together to form a continuous resin channel. The resin channels did not occur in the laminate with a z-pin content of 0.5% because the pins were spaced 3.5 mm apart. However, resin channels occurred in the unidirectional laminates with z-pin contents of 2.0% and 4.0% because the pins were spaced less than 2 mm apart. In some cases, small voids formed in the resin channels because the epoxy resin gelled during curing before it had filled the cavity created by the displacement of the fibres. The 0° plies in the
quasi-isotropic laminates also developed continuous resin channels when the z-pin content was 2% and 4%. The ±45°/C176 plies also develop resin rich channels in these directions and resin channels formed along the transverse direction in the 90°/C176 plies.

3.2. Monotonic tension

3.2.1. Summary of trends

The modulus and strength under tension were determined by dividing the measured load in an experiment by the product of the measured width of the specimen and either (1) the actual specimen thickness or (2) a fixed, nominal thickness (4.0 mm). Since, during pinning, the ply fabrics were constrained from lateral (in-plane) spreading by fences, the density of fibres per unit width is expected to remain very close in the pinned laminates to its value in unpinned laminates. On the other hand, small variations in thickness arise from specimen to specimen. Since the load in the tests is carried almost entirely by the fibres (which are two orders of magnitude stiffer and stronger than the matrix), variations in specimen thickness for fixed density of fibres per unit width will lead to changes in modulus or strength simply by fibre dilution. Comparison of stiffness or strength calculated assuming fixed thickness with values found using actual thickness will reflect changes due to other factors.

The effect of increasing either the z-pin content or the pin size on the longitudinal Young’s modulus of the unidirectional and quasi-isotropic laminates is shown in Fig. 9. The closed and open data points represent the modulus values calculated using the actual specimen thickness and a nominal thickness of 4 mm, respectively. The solid and

Fig. 7. Schematic illustrations showing the dimensions of the disturbed region surrounding the (a) small and (b) large diameter z-pins in the unidirectional laminate.

Fig. 8. Photograph of a resin-rich channel along a row of z-pins in the unidirectional laminate.

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Fig. 9. Effect of (a) volume content and (b) size of z-pins on the Young’s modulus of the laminates. The closed and open data points represent the modulus values calculated using the actual laminate thickness and a nominal thickness of 4 mm, respectively.
dashed lines show linear regressions for these two data types, fitted under the constraint that they have the same value at zero pin content. The modulus values for both unidirectional and quasi-isotropic laminates show scatter among different pin diameters and volume fractions that is comparable to overall trends. Therefore, fitted lines have modest correlation factors. The trends in data for the unidirectional laminates are that (1) the modulus decreases approximately 1.5% for each 1% of z-pin volume content for pins of diameter 0.28 mm when computed using a constant specimen thickness, but remains almost independent of pin content when computed using the actual specimen thickness; and (2) the modulus decreases approximately 5.0% for each 1% of pin volume content for pins of diameter 0.51 mm, with only weak dependence on which specimen thickness is used. For the quasi-isotropic laminates, there is no statistically significant change in the modulus with increasing content, whichever specimen thickness is used.

The fitted trend lines suggest that the greater part of changes in the modulus with pin content for both laminate types reinforced by 0.28 mm pins are due to laminate swelling; the modulus decrease is not significantly different from zero when a constant specimen thickness value is used in calculations. For the larger diameter pins, the decline in modulus is much greater and significantly exceeds the decline attributable to swelling. For the larger pins, the greater part of the modulus decline is attributed to fibre misalignment. However, it is not known what contribution the offset of the pins had on the loss in modulus. Figs. 3, 6 and 7 suggest that, at most pin densities, fibre misalignment, both in-plane and out-of-plane, is localized around each pin and extends over a volume that is a small multiple of the pin volume, approximately 3–4. Assuming that the misalignment (which is quite severe in the out-of-plane direction—Fig. 3) reduces the effective modulus of the affected material by a factor of 2 or so, one comes, to order of magnitude, to reductions of 1–2% per 1% of pin content. This is in the range that could explain the measured modulus changes for the larger diameter pins. That this effect should be significant for the larger pins and not the smaller pins could be attributed to a greater proportion of misalignment damage.

The effect of z-pinning on tensile strength is shown in Fig. 10. The closed and open data points present the strength values calculated using actual and constant thickness values. The strength of the unidirectional laminate decreased rapidly with both increasing pin volume content and pin diameter. The reduction ranges, for the cases tested, up to approximately 25%. The trends are approximately linear and may be summarised by the following empirical expression for the ratio of the tensile strength, \( \sigma_t \), to that of unpinned laminate, \( \sigma_{t(0)} \):

\[
\frac{\sigma_t}{\sigma_{t(0)}} = [1 - \alpha D c_i]
\]

where \( c_i \) is the area fraction of the pins and \( D \) is their diameter. For the unidirectional laminates, \( \alpha = 14 \text{ mm}^{-1} \). This fit is shown in Fig. 10 for the unidirectional laminate data. Since thickness variations are much smaller than strength variations, no significant change in \( z \) results from using fixed (4 mm) rather than actual specimen thickness in computing strengths. The model does not consider the effect of the continuous resin channels on the failure strength despite the channels causing splitting cracks in the laminates pinned at the intermediate and high densities.

For the quasi-isotropic laminate, the reduction with pin volume fraction is much smaller, not exceeding approximately 7% for the range of pin densities studied. But a linear trend with even this mild slope has relatively low correlation factor, because the trend is not far from the variance in the data. Freitas et al. [8,9,20], Steeves [21] and Stringer and Hiley [15] have reported higher reductions of tensile strength for multiaxial carbon/epoxy laminates due to z-pinning. At least in the work of Steeves, this could
be attributed to inferior methods of pin insertion leading to a greater degree of in-plane fibre damage. The other papers contain insufficient reporting of in-plane fibre deformation to address the question.

If one assumes that the strength of the quasi-isotropic laminates would also decrease linearly with pin diameter (this was not tested here), then the strength summary of Eq. (1) would again hold approximately. The data of Fig. 10 imply \( z = 4 \text{ mm}^{-1} \). The curve for the quasi-isotropic laminate in Fig. 10 shows Eq. (1) for this value of \( z \).

The only other tensile strength data available to check the trend of Eq. (1) are those of Frietas et al. [8], which are for a cross-ply carbon/epoxy laminate reinforced with z-pins up to a volume content of 10%. The change in laminate thickness with increasing z-pin content was not reported, although substantial swelling or in-plane fibre dilution must be expected at 10% pin content. The authors leave unclear whether the swelling effect was considered when calculating the failure stress. Fig. 11 shows the data of Freitas et al. [8]. Once again, tensile strength falls linearly with pin content. The closed data points show the strength values as reported by Freitas et al. If it is assumed that these data were calculated using a constant specimen thickness and that the actual thickness increased in proportion to the pin content (e.g., 5% z-pins caused the laminate to swell by 5%), then the strength values that would be found using the actual specimen thickness can be estimated. These are shown by the open data points in Fig. 11. The lines in Fig. 11 were calculated using Eq. (1) fitted with the constraint that for constant and adjusted specimen thickness they should have the same value at zero pin content. The linear fits are a reasonable representation of the data, but the slopes are much higher than for the quasi-isotropic laminates tested in the present work. For constant thickness values, the slope \( z = 25 \text{ mm}^{-1} \); while for adjusted thickness values, \( z = 29 \text{ mm}^{-1} \) (compare with \( z = 4 \text{ mm}^{-1} \) in the present work). For reasons that are not well understood, the pinning process used by Freitas et al. has apparently caused far more severe damage to the laminate.

The failure strain for the unidirectional and quasi-isotropic laminates decreased at a linear rate with increasing pin content. The stress–strain curves for the unpinned and pinned laminates were linear, and therefore any loss in elastic modulus and strength caused by pinning resulted in a corresponding reduction in the failure strain. The reduction in failure strain due to pinning was more significant with the unidirectional laminate because its modulus and, in particular, strength were degraded more by the pins than the quasi-isotropic laminate.

### 3.2.2. Mechanisms of failure

Photographs of failed unidirectional tensile test specimens with and without z-pins are presented in Fig. 12. The unidirectional laminate without z-pins failed by ligament rupture whereas the failure mode changed with z-pinning to a combination of ligament rupture and multiple longitudinal splitting in the axial direction along the rows of z-pins. Since failure of the z-pinned specimens was catastrophic, it is not known whether fibre rupture or splitting occurred first; the two mechanisms may well interact dynamically during the failure process. Microstructural examination of the failed specimens revealed the presence of small cracks within the resin-rich zones adjoining the z-pins, as shown in Fig. 13. These cracks were not detected in the specimens prior to tensile testing, which indicates that they were not caused by thermal effects during curing. Very few of these cracks were detected in the failed specimens, although those that were found were aligned in the load direction. It is believed the cracks were initiated under the action of a local transverse tensile strain generated as the misaligned fibres around a z-pin experience a small degree of straightening under axial tensile loading, as depicted schematically in Fig. 14. Close to the failure stress, these cracks may propagate unstably along the continuous resin-rich channels and trigger the long splitting cracks observed in Fig. 12.
There is some similarity between the failure observations around pins and the failure sequence at ply-drop locations in tape laminates. Similar mechanics are likely to be pertinent, as described, for example, in Varughese and Mukherjee [22], Vidyashankar and Krishna Murty [23] and Xia and Hutchinson [24]. Under far-field tension, the driving forces for damage initiation around a pin (or ply end) are: (1) tension across the resin pocket, in the vertical direction in Figs. 6 and 7, caused by the tendency of deflected in-plane fibres to straighten; and (2) tension at the pin/resin boundary, in the horizontal direction in Figs. 6 and 7. Since an accurately calibrated failure criterion for local cracking under these stresses is unavailable, no attempt is made here to analyze these phenomena quantitatively.

While the z-pins altered the failure mechanism of the unidirectional laminate, no change to the failure process was found for the quasi-isotropic laminate. Post-mortem examination of unpinned and pinned quasi-isotropic laminate specimens revealed the same failure mode, as shown in Fig. 15. Longitudinal splitting cracks were much more limited in extent than in the unidirectional laminates. Further, they were not correlated with the locations of the pins. This can presumably be attributed to (1) the minimal extent of continuous resin-rich channels were between pins in the quasi-isotropic laminates, due to constraint of fibre deformation during pinning by the off-axis plies; and (2) load transfer across the potential splitting plane by off-axis fibres, which reduces the driving force for splitting.

The volume fraction of the fibres that were crimped out-of-plane or deflected in-plane by the z-pins in the unidirectional laminate increased with the pin density, as did the line-density of resin-rich channels across the specimen width. These worsening defects are the only apparent source of the increasing deterioration in the tensile strength with increasing z-pin content. However, strength, much more than modulus, can also be strongly affected by clusters of fibre breaks, which need not involve more than a handful of fibres and are therefore not easily detected or measured. Some fibre breakage undoubtedly accompanies the insertion of each z-pin, creating an effective defect at each pin. Since the strength of the defect at each pin is likely to be statically distributed, the extreme defect (weakest link for failure initiation) will present a decreasing

![Fig. 13. Photograph showing a small crack in the resin-rich region next to a z-pin. The load direction is indicated by the arrows.](image)

![Fig. 14. Schematic of the process of longitudinal cracking in the resin-rich region in a z-pinned unidirectional laminate.](image)

![Fig. 15. Failed tensile specimens of the quasi-isotropic laminate (a) without z-pins and (b) with z-pins tested under monotonic loading.](image)
strength as the pin density (and therefore the sample size) rises.

Since in-plane fibre deformation is similar in the angles of deflection for different z-pin sizes (Fig. 7), perhaps the reduction in strength with increasing z-pin diameter at fixed pin density is related to increased fibre breakage at each pin insertion. A second possibility is a size effect in the strength of the resin pockets formed at each pin: while the fibre deformation is geometrically similar for different pin diameters (Fig. 7), the length of the resin pocket scales with the pin diameter. Crack initiation in resin pockets is known to exhibit gauge effects for pockets that are ~1 mm in size or less, which is within the size range in the z-pinned specimens [16,22,23].

The lower fractional loss of strength with rising z-pin density in the quasi-isotropic laminates compared to the unidirectional laminates ($\alpha = 4$ vs. $\alpha = 14$) is presumably related to fewer defects being created in the quasi-isotropic laminates. Off-axis plies constrain in-plane fibre deformation, and the extent and width of resin-rich channels; and may possibly, by reducing fibre deflections, also reduce fibre breakage.

Some of the z-pins themselves were fractured during tensile testing in both the unidirectional and quasi-isotropic laminates. Fig. 16 shows a cross-section of a broken z-pin with a crack aligned normal to the tensile load direction. Cracks such as these are believed to initiate at microscopic voids within the z-pins. However, the cracks do not link with the strength controlling failure events (splitting or rupture) and are therefore probably unimportant.

3.2.3. Tests using specimens with a single pin

The effect of z-pin diameter on the tensile strength was investigated further by inserting single rods of different sizes into unidirectional laminate specimens. The rods were embedded by hand into the center of the gauge area, rather than using the ultrasonic horn. The rods were off-set from the orthogonal direction by an angle less than $20^\circ$, which is similar to the range of offset angles when pins were inserted using the ultrasonic horn. Fibrous z-pins with diameters 0.28 and 0.51 mm were supplemented by larger steel pins, with diameters between 0.96 and 5.72 mm. The amount of swelling in the region surrounding the steel rods increased with their diameter. The swelling was greatest at the site of the rod and extended several millimeters away. Since it was not uniform, thickness variation was not considered in calculating the tensile strength. The laminate strength was found to drop rapidly with increasing pin diameter (Fig. 17), but not linearly. The linear summary relation of Eq. (1) should therefore not be used beyond the maximum pin diameter, 0.5 mm, of the pins for which the data it represents were obtained.

3.3. Fatigue properties

Tensile fatigue life ($S$–log$N$) curves are presented in Fig. 18 for the unidirectional and quasi-isotropic laminates reinforced with different densities of z-pins. Like most polymer composites, the unpinned unidirectional and quasi-isotropic laminates show very modest fatigue effects: only 8% and 13% loss of strength over $10^6$ cycles, respectively. Despite the limited amount of data, it is apparent that much stronger fatigue effects arise in the pinned laminates. While the fatigue resistance of the unidirectional laminate was affected only slightly by the lowest volume content of z-pins (0.5%), a large reduction occurred at the higher z-pin contents (33% loss of strength over $10^6$ cycles). The fatigue resistance of the quasi-isotropic laminate also deteriorated with increasing z-pin content, although the reduction was less severe than for the unidirectional laminate (25% loss of strength over $10^6$ cycles for 4% pin density). The effect of increasing z-pin size on the $S$–$N$ curve for the unidirectional laminate is shown in Fig. 19. A similar fractional loss of strength is found for both pin sizes, with the larger pins leading to lower absolute strengths at all cycle counts.

Fig. 20 shows the gradient of the $S$–log$N$ curves, $m$, plotted against the volume content of the z-pins, for fixed pin diameter (0.28 mm). While the data show significant scatter, especially for the quasi-isotropic laminates, there

![Fig. 16. Photograph showing transverse fracture of a z-pin under tensile loading. The load direction is indicated by the arrows.](image-url)
is a trend for fatigue effects to increase with pin density up to approximately 2%, beyond which levelling in $m$ suggests that they have saturated. The increase of $m$ with density is less for the quasi-isotropic laminates, implying weaker fatigue effects due to pinning than in the unidirectional laminates when measured in terms of absolute strength; but the difference is less if one considers instead the fractional loss of strength with cycles.

Post-mortem examination revealed the same failure modes as in the static tensile samples (see Figs. 12 and 15). That is, the unidirectional composite without z-pins failed in fatigue by ligament rupture, whereas the z-pinned unidirectional laminates experienced ligament rupture and longitudinal splitting along the axial rows of z-pins; while the quasi-isotropic laminates always failed by fibre rupture across an inclined band. Nevertheless, the significant slopes of the $S$–$N$ curves imply that substantial cyclic damage must in fact occur that is not present under monotonic loading. The most likely mechanism is damage in regions around pins where fibres have been misaligned. Prior work on tape and textile polymer composites has put forward two possible mechanisms: the attrition of fibres where microfracture of the fibre–matrix interface has permitted relative sliding between the fibres and the matrix [25]; and softening and damage of the resin itself under local cyclic shear, for which evidence has been most compelling in studies of compression fatigue [26,27]. Under nominally aligned loading, both of these mechanisms act where fibres are misaligned, i.e., where significant local shear stresses arise in coordinates that are aligned with the local fibre direction. The result of progression of fibre damage would be a progressive loss of fibre strength. Progressive matrix damage can initiate delamination or splitting cracks; and, since it allows straightening of the misaligned fibres, it can also promote the initiation of cracks in resin pockets (the mechanism illustrated in Fig. 14).

No certain conclusions are possible about how variations in these mechanisms might account for the trends in fatigue effects (i.e., in the slope $m$ of the $S$ – $\log N$ curves) with pin density. Nevertheless, the fact that fatigue effects are weak in the absence of pins and strong in their presence

![Fig. 18. Effect of volume content of z-pins on the $S$–$N$ curves for the (a) unidirectional and (b) quasi-isotropic laminates.](image)

![Fig. 19. Effect of z-pin size on the $S$–$N$ curve for the unidirectional laminate.](image)

![Fig. 20. Plot of the slope of the $S$–$N$ curve against the z-pin content.](image)
points very strongly to the presence of fatigue mechanisms in the misaligned fibres, as described. Further, one might speculate that the apparent saturation at a density of approximately 2% in the unidirectional laminates is associated with the extension of resin channels from one pin to the next, which also occurs for all pin densities above 2%. This would suggest that the strength determining mechanism for the unidirectional laminates is cracking of the resin pockets (Fig. 13); and that the principal fatigue effect is propitiation of this by the progressive softening of resin allowing fibre straightening. But the evidence for this sequence is very incomplete.

4. Conclusions

The process of pin insertion caused out-of-plane crimping, in-plane distortion and fracture of fibres near the z-pins. Crimping and distortion of fibres were exacerbated by pin rotation when excess pin length was sheared off and during laminate consolidation and cure. In unidirectional laminates, resin-rich regions formed adjacent to the pins and extending in the fibre direction; and the laminate swelled in the thickness direction, diluting the in-plane fibre volume fraction. Resin pocket formation and swelling were minimal in the quasi-isotropic laminates.

The Young’s modulus of laminates reinforced with 0.28 mm pins fell with increasing pin density to a degree that was consistent with the degree of swelling, i.e., slightly in the unidirectional laminates and not significantly in the quasi-isotropic laminates. The modulus fell at a greater rate for 0.51 mm pins (unidirectional laminates only), suggesting that, for these pins, it was affected significantly by fibre distortion, which is assumed to rise with pin diameter.

The tensile strength was degraded more than Young’s modulus by z-pinning. The strength decreased approximately linearly with increasing pin density and size and at a far greater rate in the unidirectional laminates than in the quasi-isotropic laminates. Key mechanisms are believed to be clusters of broken fibres near the z-pins, augmented in the unidirectional laminates by cracking in the resin pockets, where fibre straightening initiates tensile rupture and longitudinal splitting. Both the strength of flaws associated with fibre breaks and the propensity for resin pocket cracking could reasonably be expected to increase with pin density and diameter, accounting for the observed trends. The more modest strength loss in the quasi-isotropic laminates is attributed tentatively to the constraint of off-axis fibres, which can suppress longitudinal splitting cracks and perhaps minimizes breakage damage to fibres.

The tensile fatigue lives of the unidirectional laminates and, to a lesser extent, the quasi-isotropic laminates were also reduced substantially by z-pinning. Softening and damage of misaligned fibre regions may be the primary fatigue mechanism, but this has not been clearly established.

Comparison with other data in the literature reveals order-of-magnitude variability in the rate of damage with increasing pin density, which must presumably be attributed to differences in the details of the pin insertion process, via unknown mechanisms. Such variability may pose a significant barrier to more widespread use of pinning technology.

Trends in the rate of degradation of modulus and strength with pin diameter, at fixed pin area fraction, suggest that using finer pins than those used here might avoid significant degradation of in-plane properties under monotonic loading altogether. A goal of pin diameters less than approximately 0.1 mm would seem reasonable for this, based on the present data. However, it is not as clear that loss of strength in cyclic loading would be as easily avoided, because the fractional loss of fatigue strength to $10^6$ cycles is similar for both diameter pins studied.

Microstructural and geometric details such as specimen thickness should be regarded as critical elements of data reporting, since variations in fibre density can vary with processing and can easily be comparable to the knock-downs in modulus and even strength that can occur with z-pinning.

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References

on bolted/bonded joints in polymeric composites. Florence Italy, Paper 17, 2–3 September 1996.


