Composites: Part A 69 (2015) 64–71

Contents lists available at ScienceDirect

Composites: Part A

journal homepage: www.elsevier.com/locate/compositesa

Pseudo-ductility and damage suppression in thin ply CFRP angle-ply laminates

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ARTICLE INFO

Article history:
Received 30 July 2014
Received in revised form 28 October 2014
Accepted 1 November 2014
Available online 10 November 2014

Keywords:
A. Laminates
B. Delamination
C. Laminate mechanics
D. Mechanical testing

ABSTRACT

Composite materials usage is limited by linear elasticity and the sudden, brittle failure they often exhibit. It is possible to mitigate this inherent limitation and enlarge the design space by using thin plies. This paper presents an experimental study, using a spread tow thin ply carbon–epoxy prepreg material with a cured ply thickness of 0.03 mm, which shows that highly non-linear stress–strain behaviour can be achieved with angle-ply laminates, whilst suppressing the damage mechanisms that normally cause their premature failure. Several angles between 15° and 45° are investigated in a $[\pm\theta_5]_s$ layup. It is shown that for all angles delaminations are suppressed, allowing considerable pseudo-ductile strains to develop. Significant fibre rotations take place, permitted by matrix plasticity, leading to a post-yield stiffening of the laminate, as the fibres reorient towards the direction of loading.

1. Introduction

Carbon fibre reinforced polymer (CFRP) composites are well known to possess high stiffness and strength. They are, however, limited by brittle failure, which often occurs without warning and is catastrophic. This linear-elastic to failure stress–strain behaviour reduces design allowables and precludes the realisation of the materials' full potential. Achieving non-linear stress–strain behaviour, with the ability to yield, as metallics do, is therefore highly desirable. Non-linear behaviour and high strains to failure have been demonstrated previously with ±45° angle-ply laminates, often used to determine the shear properties of materials. In these cases, however, the large fibre angle leads to relatively low values of modulus and failure stress. Reducing the fibre angle (towards the loading direction) leads to a higher initial modulus, but despite promise of high strains to failure, laminates of $\pm\theta$ (where θ is in the range 15–30°) often fail prematurely – before the development of non-linearity. Failure of angle-ply laminates is primarily due to matrix cracking and delaminations due to high free-edge interlaminar stresses [1]. These failure mechanisms have been widely studied [2–6].

O'Brien [2,3] characterised the onset and development of delaminations in $[+\theta_n/-\theta_n/90_n]_s$ laminates. Delaminations at $\theta/-\theta$ and $-\theta/90$ interfaces were seen to initiate from the edges of the specimens following matrix cracking in 90° plies. A non-linear stress–strain behaviour was associated with a 'stiffness loss' brought about by the accumulation of damage. This stiffness loss was coupled with a strain energy release rate (G) approach employed to predict the initiation of delamination. The value of G was found to depend only on the laminate stacking sequence and location of delaminations.

Crossman et al. [5] also used a strain energy concept to determine the failure mechanisms of $[\pm 25/90_n]_s$ laminates ($n = 1, 2, 3$), highlighting the importance of the ply thickness in calculating the value of G. It is shown that increasing the number of 90° plies, not only decreases the stress levels required for matrix cracking and delaminations to occur, but also alters the order in which they take place – showing a direct interaction between the damage modes. Treating the adjustment in the number of 90° layers as effectively changing ply thickness, it is postulated that reducing ply thickness could suppress microcracking and delaminations.

Investigating angle-ply laminates, Leguillon et al. [7] examined edge delamination initiation in $[\theta_n/-\theta_n]_s$ laminates ($n = 1–8$), comparing tensile test data with predictions. Of the two methods implemented, both showed decreases in delamination initiation stress with increased layer thickness. Herakovich [8] investigated these edge effects using $[(+\theta/-\theta)_2]_s$ and $[+\theta_2/-\theta_2]_s$ laminates, where $\theta = 10°, 30°, 45°$. For all angles tested, the tensile strength, tensile strain and toughness (in this case defined as the area under the stress–strain curve) were each increased for the laminates with dispersed plies. For the 30° and 45° laminates containing dispersed plies, the increased failure stress and strain allowed more non-linearity to develop, highlighting the potential this type of laminate possesses for ductility.

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http://dx.doi.org/10.1016/j.compositesa.2014.11.004
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For both the theoretical approaches taken, Leguillon et al. [7] present sharp increases in delamination initiation stress for ply thicknesses less than 0.125 mm. Expense and damage to fibres during the manufacturing process, however, have limited work on reducing the ply thickness below the standard. Recent advances in tow-spreading technology have allowed so called thin prepregs to be produced. Sasayama et al. [9] and the Industrial Technology Centre of Fukui Prefecture in Japan 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] performed both static and fatigue tension tests of un-notched and notched quasi-isotropic (QI) specimens, impact and compression-after-impact tests on thin ply laminates (ply thickness, $t_p = 0.04$ mm). In all cases these laminates showed less cracking, delaminations and splitting than specimens with thicker plies ($t_p = 0.2$ mm) of the same material. Results from fatigue testing of the un-notched specimens show the potential of thin ply material. It was demonstrated that, after 50,000 cycles at 60% of the strength, the thin ply laminates maintained stiffness and strength. The thick ply laminates lost in the region of 17% from both the original stiffness and strength. X-ray images taken prior to failure show very little development of damage within the thin ply laminates, indicating their superior damage suppression capabilities. Yokozeki et al. [11,14] conducted investigations covering the compressive strength and damage resistance of thin ply QI laminates under both in-plane [11]and out-of-plane [14] loadings. In all cases, the thin ply laminates were shown to be more resistant to damage accumulation. This is particularly noticeable in transverse indentation tests. Thick ply laminates ($t_p = 0.14$ mm) exhibited considerable delaminations on the back face, whereas the thin ply specimens ($t_p = 0.07$ mm) showed only internal delaminations at the same applied load. As presented by Sihn et al. [10], this suppression of damage led to sudden brittle failure. Ogihara and Nakatani [13] presented work on carbon/epoxy angle-ply laminates, also concentrating on the effect of ply thickness. Specimens of ±45° and ±67.5° both showed increases in tensile strength with ply thicknesses of 0.05 mm ($[\pm\theta_{12}]_s$) rather than 0.15 mm ($[\theta_4/-\theta_4]_s$). A mesoscale continuum damage mechanics model, devised by Ladeveze and LeDantec [15], was employed to show also that the thin-ply laminates were significantly more damage resistant. Highly non-linear strains, in excess of 15%, were recorded for the $[\pm 45_{12}]_s$ laminates tested under quasi-static tension. At these large strains, the effect of the geometric rearrangement of fibres towards the loading direction (known as fibre scissoring) becomes important, as stated by Wisnom [16] and Herakovich et al. [17]. Wisnom [16] showed how taking account of fibre rotations for in-plane shear testing leads to a more accurate representation of both the shear stresses and strains in a $[\pm 45]_s$ laminate. Herakovich et al. [17] coupled fibre rotations with the Ladeveze and LeDantec model to emphasise the importance of their inclusion when predicting the stress–strain response of $[\pm 45_3]_s$ laminates.

In this paper, experimental studies of thin ply angle-ply CFRP laminates loaded under quasi-static tension are presented. The effect of fibre rotation on the laminate stress–strain behaviour and the possibilities for a pseudo-ductile response are investigated. Analyses of the fractured laminates, including X-ray computed tomography (XCT) scans are presented to examine the damage resistance of spread tow thin ply prepreg material.

2. Experimental methods

All testing has been performed using Skyflex USN020A, a commercially available spread tow carbon fibre/epoxy prepreg produced by SK Chemicals. This material consists of Mitsubishi Rayon TR30 carbon fibres ($E_{11} = 234$ GPa, strain to failure, $\epsilon^*_{11} = 1.9\%$) and SK Chemicals K50 resin, a semi-toughened epoxy. Prior to testing of angle-ply laminates, full characterisation of the material was necessary. Quasi-static tensile tests were performed on $[0_{16}]$, $[90_{16}]$ and $[\pm 45_5]_s$ laminates. To ensure sufficient data was collected, batches of 10 specimens were fabricated for each layup. Unidirectional (UD) samples had a gauge length of 100 mm and width of 10 mm, with glass fibre/epoxy prepreg end tabs of length 40 mm. In all tests end tabs were a $[(90/0)_2]_s$ cross-ply laminate of 2 mm thickness. The $[\pm 45_5]_s$ samples had a gauge length of 150 mm, width of 15 mm and end tabs of 40 mm. Three-point micrometer measurements of laminate thickness, performed prior to testing, yielded a cured ply thickness (CPT) of 0.03 mm (CV = 0.34%). All tests were conducted, using an Instron hydraulic-actuated test machine, under displacement control, using cross-head rates of 1 mm/min for UD samples and 2 mm/min for the $[\pm 45_5]_s$. The results are shown in Table 1. Determination of E_{11} allowed a fibre volume fraction (V_f) of 42% to be calculated using the rules of mixtures.

The angles chosen for further tensile testing were ±15°, ±20°, ±25°, ±30°. All layups were of the same stacking sequence: $[\pm\theta_5]_s$, as used for the ±45° laminates. The dimensions and rate of displacement for these samples were also the same as for the ±45°. Batches of five specimens were prepared for each layup.

All strain data was captured using an Imetrum Video Extensometer and associated software. A rectangular grid of video gauge targets was set up, in order to record both longitudinal, ϵ_x and transverse, ϵ_y, strains. Calculation of fibre rotations and shear stress and strain requires knowledge of both of these. In all cases, the true stress and strain have been computed from the captured engineering strains to account for the change in cross-sectional area at high strains.

2.1. Calculation of fibre rotations

Fibre rotations have been considered in a similar fashion to the approaches taken by other studies [17,16,18]. The fibres are taken as inextensible and idealised to act in a scissoring motion, realigning towards the direction of applied stress. This gives rise to the concept of 'excess length', whereby the reorientation of the fibres allows further strain to be taken by the laminate. The updated fibre angle, θ', is related to the strains, ϵ_x and ϵ_y, in Eq. (1), where θ is the original fibre angle of the laminate.

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

2.2. Definition of yield and pseudo-ductility

'Pseudo-ductility', in this case, refers to the geometric effect of fibre reorientation as well as yielding of the matrix. For clarity, yield stress, σ_Y, and pseudo-ductile strain, ϵ_d are shown graphically in Fig. 1. The yield stress is defined as the point of intersection between the laminate stress–strain curve and a straight line of the initial modulus offset by 0.1% strain (shown as position 'A' on Fig. 1). The pseudo-ductile strain is the failure strain minus the strain at the same stress level on a straight line of the initial modulus.

2.3. Determination of shear stress and strain

As a change in the orientation of the fibres is accounted for in this study, it is therefore important to apply this to the calculation

Table 1
Elastic properties of Skyflex USN020A.

E_{11}	101.7 GPa	E_{22}	6.0 GPa
G_{12}	2.4 GPa	ν_{12}	0.3