Composites Science and Technology 110 (2015) 103–110

Contents lists available at ScienceDirect

Composites Science and Technology

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

The influence of the nonwoven veil architectures on interlaminar fracture toughness of interleaved composites

Victor A. Ramirez [a], Paul J. Hogg [b], William W. Sampson [a,*]

[a] School of Materials, University of Manchester, Manchester M13 9PL, UK
[b] Royal Holloway, University of London, Surrey TW20 0EX, UK

ARTICLE INFO

Article history:
Received 22 July 2014
Received in revised form 19 January 2015
Accepted 21 January 2015
Available online 31 January 2015

Keywords:
A. Layered structures
A. Polymer matrix composites
B. Fracture toughness

ABSTRACT

Nonwoven veils with a range of weights formed from polyphenylenesulfide (PPS) fibres with different diameter were interleaved within unidirectional carbon fibre epoxy composites and their mode I and mode II interlaminar fracture toughness (IFT) measured. In modes I and II the IFT increases with the areal density of the veil up to a plateau; at a given areal density, the mode I IFT is greater for thin fibres than for thicker fibres. For the PPS veils, we observe no significant influence of nonwoven anisotropy on IFT; though some dependence is observed for a highly anisotropic PEEK veil. Interpretation of the results using theory describing nonwoven architectures reveals that for both modes the IFT depends on the mean coverage of the veil and hence on the fraction of the propagating crack front that contains no fibres. Results for composites formed from a veil of polyetheretherketone (PEEK) fibres exhibit behaviours consistent with those observed for PPS.

1. Introduction

Laminated fibre reinforced composites are used increasingly for structural applications in the automotive, energy [1] and aerospace [2–5] sectors. They are light-weight, relatively easy to manufacture and exhibit good performance under loading and at high temperature. More widespread application has been limited by the propensity for interlaminar failure of the composite under mode I and mode II loading. Insufficient resistance to interlaminar stresses through fibre–polymer interactions results in the generation of cracks [6,7]; above some critical stress, characterised by the critical interlaminar fracture toughness, these cracks will propagate resulting in delamination. Accordingly, interlaminar fracture toughness is a key parameter to assess the performance of composites under conditions such as fatigue [8,9], compression [10], impact [11–13] or compression after impact [14–17]; in each of these cases, delaminations due to mode I and II loading are a principle cause of material failure [9,17–20].

Now, cracks and imperfections resulting in delamination may arise from local variability in materials properties, voids or other imperfections occurring during manufacturing, or through damage sustained in-service. Regardless of the nature of flaws, it is inevitable that those located at the surface of the material will be more readily identified, *e.g.* by simple inspection, than those located inside the composite; here more advanced analytic techniques must be applied [21]. Interlaminar fracture toughness testing protocols and standards [22,23] offer an approach to assess the energy required for crack initiation and eventual propagation by measuring delamination length in samples under mode I or mode II loading to obtain the fracture energy release rates, G_{Ic} or G_{IIc}, respectively.

Strategies to improve the interlaminar fracture toughness in composites include modifying fabric architectures [24–29] by manufacturing, *e.g.* 2D patterns [28,29], weft-knitted fabrics [26] or 3D textiles [29,30]. In each case the volume fraction of fibres in the composite and its homogeneity is key; the absorption of stresses induced during loading being improved for a greater and more homogenous fibre fraction. However, such approaches do bring complexity to manufacturing processes, and can result in poor resin infusion, generating voids and wrinkles, which may eventually lead to micro cracks [31]. These are found also when elements are inserted out-of-plane *e.g.* Z pins or stitching tows. Thus, although such techniques may considerably improve interlaminar fracture toughness in both modes [32,33], the increased propensity for discontinuities can compromise their in-service performance [20].

Interleaved composite systems, where layers of high-toughness resin are introduced within composites, have been applied with CFRP prepregs; these increase interlaminar fracture tough-

* Corresponding author. Tel.: +44 (0)161 306 3898.
E-mail address: sampson@manchester.ac.uk (W.W. Sampson).

http://dx.doi.org/10.1016/j.compscitech.2015.01.016
0266-3538/

ness in both modes I and II [34], but increase also composite thickness and weight [9,35]. Gillespie [7,36] and Smiley and Pipes [37] demonstrated that the fracture behaviour of interleaved composites is characterised by the radius of the plastic deformation zone at the propagating crack front. This radius depends on the mechanical properties of the interlayer and the opening rate; when the radius is greater than the interlayer thickness the plastic area extends beyond its boundaries and cracks appear in the bonded interfaces between interlayer and fibres, rather than those between fibres and matrix, which are typically stronger. Although advances in resin toughening technology have improved the interlaminar fracture toughness of interleaved composites, Kuwata [16] considers that these gains are approaching the limit of what may be achieved. Following Lee and Noguchi [38–40], they considered instead the use of nonwoven textile veils for interleaving CFRP composites and found significant improvements in mode I and mode II fracture toughness when using polymer veils.

Nonwoven textiles consist of a web of stochastically distributed fibres bonded to each other. The fibres used and the mass per unit area, or 'areal density' of the web depend on end-use application; these include gas filters, fluid absorbers, surfaces corrosion protectors, etc. [41,42]. Preliminary studies of hybrid composites using nonwovens show some improvement in delamination resistance accompanied by a reduction in tensile strength and shear modulus [38]. Subsequently, the improved mode II interlaminar fracture toughness was found to depend on the location of crack tip propagation such that the mode II fracture toughness, G_{IIc}, was greatest when crack propagation is close to the mid-plane [40]; improvements in mode I fracture toughness were limited however [39]. Subsequently, Kuwata and Hogg [31,43] examined the properties of interleaved composites formed from different veils. They showed that for mode I, veils with low areal density and with thermoplastic fibres seemed to be the most effective for absorbing fracture energy. For mode II the fracture energy was seemingly dependent on the interleaved nonwoven architecture.

Nonwoven textiles used in composites are usually wet-laid, *i.e.* they are formed by continuous filtration of a dilute fibrous suspension in water. The resultant nonwoven mat typically exhibits a preferential orientation of fibres in the direction of manufacture (MD) and accordingly exhibits a corresponding mechanical anisotropy such that tensile strength and modulus in the MD are greater than those in the perpendicular, or cross-direction (CD) [42].

Nonwoven textiles are characterised by their heterogeneous, stochastic structures. Frequently these are modelled as random fibre networks where the location of each fibre is independent of those of all other fibres and the distribution of fibre orientation to any given direction is uniform. There is a considerable literature describing the structure of these random fibre networks, see, *e.g.* [44], and statistical models of these structures have been widely applied to the characterisation of paper, fibrous filter media [45,46] and, more recently, electrospun polymer fibre networks [47]. Typically, such models seek to give structural properties of the network in terms of network properties such as porosity and areal density, and fibre properties such as length, width and linear density; note that the linear density of a fibre is given by the product of the cross-sectional area and density, irrespective of the cross sectional shape, so for fibres with circular cross-section, with diameter, ω (m), formed from a polymer of density, ρ (kg m^{-3}), the linear density, δ (kg m^{-1}), is given by

$$\delta = \frac{\pi\omega^2\rho}{4}. \tag{1}$$

The specific surface area, S_f (m^2 kg^{-1}) of fibres with circular cross-section is

$$S_f = \frac{\pi\omega}{\delta} = \frac{4}{\rho\omega}. \tag{2}$$

The number of fibres covering a point in the plane of support of the network is a discrete random variable called coverage, c, and classical statistical models for network structure assume that the distribution of coverage at points is given by the Poisson distribution such that in a network with mean coverage, $\bar{c}$, the probability that a point in the network has coverage, c is given by

$$P(c) = \frac{\bar{c}^c e^{-\bar{c}}}{c!} \quad \text{for } c = 0, 1, 2, 3 \ldots \tag{3}$$

The mean coverage is given by the ratio of the mean areal density of the network, $\bar{\beta}$ (kg m^{-2}) to that of the constituent fibres, β_f (kg m^{-2}). Intuitively, the mass per unit area of fibres is given by the ratio of the linear density to the fibre width, so the mean coverage is given by

$$\bar{c} = \frac{\bar{\beta}}{\beta_f} = \frac{\bar{\beta}\omega}{\delta}. \tag{4}$$

For extensive discussion of the use of Poissonian statistics to yield structural statistics of fibrous architectures, see [44]. For now, we note two results arising from such considerations: the fraction of the network that is not covered by fibres, where resin can infuse directly from one side of the veil to the other is given by the Poisson probability of coverage zero, *i.e.*

$$P(c = 0) = e^{-\bar{c}}. \tag{5}$$

In a network with porosity, ε, the expected dimension of inter-fibre voids differs in the plane (x–y) and plane-perpendicular (x–z) directions of the network such that [48].

$$\bar{d}_{xy} \approx 2\bar{d}_{xz} = \frac{2\omega}{\log(1/\varepsilon)}. \tag{6}$$

Eqs. (2)–(6) provide the dependencies that give rise to this study. From Eq. (4) we observe that changing fibre dimensions allows the mean coverage to be varied independently of areal density; coupled with Eq. (5) it follows that the fraction of pin-holes in a network with given areal density will depend on fibre dimensions also. Similarly, Eqs. (2) and (6) reveal that at a given porosity the mean pore size is directly proportional to the diameter of the constituent fibres, whereas the specific surface area of fibres is inversely proportional to their diameter. So, the structural characteristics of nonwoven veils that we expect to influence composites performance can be influenced by choice of fibres.

Here, we present an experimental study designed to investigate the interlaminar fracture toughness of interleaved composites manufactured from nonwoven veils with systematically varying structural characteristics. Interpretation of the resultant data is guided by the statistical theory of fibre and network properties as given by Eqs. (1)–(6).

2. Materials

Composites with a symmetrical construction of layers of unidirectional carbon fibre fabrics around a central layer of nonwoven polymer veil were prepared for testing. For all specimens, the resin used was Epoxy Araldite® LY 564 with Hardener XB 3486, from Huntsman Advanced Materials. Unidirectional carbon T300 fibre fabric, UD1 was supplied by FORMAX; it had areal density 340 g m^{-2} and thickness 260 µm. Unidirectional carbon T300 fibre fabric, UD2 was supplied by Sygmatex; it had areal density 300 g m^{-2} and thickness 300 µm.

Nonwoven veils were supplied by Technical Fibre Products Ltd., Cumbria, UK. One veil sample had nominal areal density 11 g m^{-2} and was formed from polyetheretherketone (PEEK) fibres with