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Micromechanical modelling of the transverse damage behaviour in fibre reinforced composites

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ABSTRACT

A micromechanics damage model is presented which examines the effect of fibre–matrix debonding and thermal residual stress on the transverse damage behaviour of a unidirectional carbon fibre reinforced epoxy composite. It is found that for a weak fibre–matrix interface, the presence of thermal residual stress can induce damage prior to mechanical loading. However, for a strong fibre–matrix interface the presence of thermal residual stress is effective in suppressing fibre–matrix debonding and improving overall transverse strength by approximately 7%. The micromechanical model is subjected to a multiple loading cycle (i.e. tension–compression–tension), where it is shown to provide novel insight into the microscopic damage accumulation that forms prior to ultimate failure, clearly highlighting the different roles that fibre–matrix debonding and matrix plasticity play in forming the macroscopic response of the composite. Such information is vital to the development of accurate continuum damage models, which often smear these effects using non-physical material parameters.

1. Introduction

The importance of transverse plies in composite laminates is significant as they provide essential stiffness and strength to components undergoing multi-axial loading, helping to maintain structural integrity. The failure behaviour of these transverse plies has a major influence on the overall failure of composite laminates. Transverse fracture often occurs early in the loading history and as a result is one of the limiting design criteria in composite structures. Transverse fracture is a result of fibre–matrix debonding and/or matrix micro-cracking which, through the process of damage accumulation, can induce further damage in laminates such as inter-ply delamination or fibre fracture in neighbouring plies [1]. The interaction of these microscopic damage mechanisms is dependant on a number of factors, the most important being the adhesion between the constituent phases.

Experimental studies have shown that the dominant damage mechanism involved in transverse ply cracking is debonding occurring at the fibre matrix interface [1,2]. The process is shown in Fig. 1a where the initial failure of a carbon fibre/epoxy system was examined by Hobbiebrunken et al. [2], using in situ SEM experiments. It was found that damage initiated through widespread fibre–matrix debonding. As these interfacial debonds grew, small resin bridges formed between them. Under sustained or increased loading these resin bridges underwent significant plastic deformation until they ruptured, causing final fracture of the ply as shown in Fig. 1a. It is well established that the local fibre distribution significantly affects the interfacial stress state [3] and, consequently, the onset and evolution of fibre–matrix debonding. The transverse fracture behaviour of composite plies is further complicated by the presence of thermal residual stress, caused by the cooling of the material after the curing process. Due to the mismatch in thermal expansion coefficients of the constituent phases, thermal residual stresses develop at the fibre–matrix interface throughout the cooling stage. The presence of thermal residual stress is greatly affected by the local fibre distribution [3] and it has been shown to significantly alter the microscopic stress state upon subsequent mechanical loading [4]. Thermally induced stresses can even be significant enough to cause damage to initiate in composite laminates prior to any mechanical load being applied. Fig. 1b shows damage in a graphite fibre composite due to thermal residual stress in post cure specimens observed by Gentz et al. [5].

It is therefore clear that the transverse fracture behaviour of composite materials is dependant upon numerous contributing factors, such as constituent properties, interfacial properties, local fibre distribution and the presence of thermal residual stresses. Due to the complex nature of damage progression, many micromechanical studies to date have focused on transverse fracture behaviour from viewpoint of damage initiation [3,6]. While, a number of previous studies have made the assumption of a periodic fibre arrangement in the microstructure [2,4] which, following numerous studies, may not be representative of actual material behaviour

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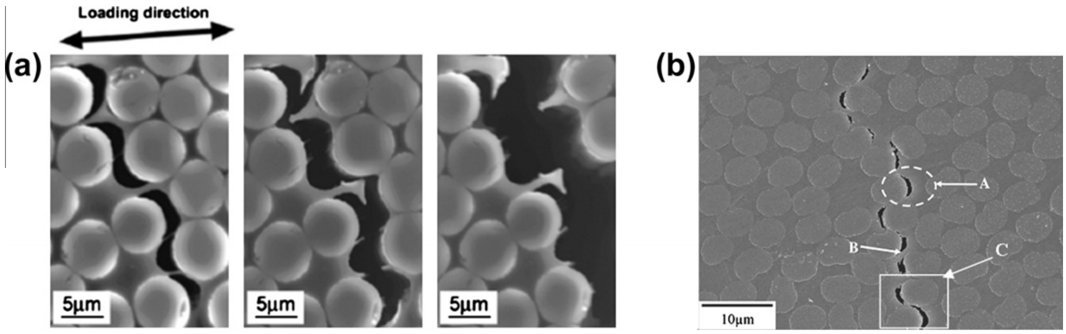

Fig. 1. (a) In situ experimental failure observation of a carbon fibre/epoxy composite (from [2]) and (b) interfacial damage as a result of thermal residual stress (from [5]).

[3,7]. Recently however, a number of advanced micromechanics damage models have been developed, such as those by Llorca and co-workers [8,9], which enable the prediction of microscopic damage progression and the ultimate failure of carbon fibre/epoxy composites. These have been achieved through the use of cohesive zone models at the fibre–matrix interface coupled with non-linear constitutive material models to describe the behaviour of the constituent phases. Computational micromechanics can therefore provide a platform to analyse microstructural damage evolution through heterogeneous materials for a range of thermal/mechanical loading scenarios.

Hence, the aim of this paper is to collectively investigate the effects of thermal residual stress and fibre–matrix debonding on the transverse fracture behaviour of a fibre reinforced composite. The computational framework used to model deformation in the carbon fibre/epoxy composite is similar to that pioneered by Llorca and co-workers [8,9] where, a cohesive zone model is used to predict the onset of fibre–matrix debonding while the non-linear behaviour in the matrix phase is modelled using the Mohr–Coulomb plasticity theory. The recently developed Nearest Neighbour Algorithm (NNA) [10], that can accurately reproduce a statistically equivalent fibre distribution for the high volume fraction composites, is used to generate the finite element models for the analysis. The effect of thermal residual stress on the transverse fracture process is thoroughly examined. Finally, the effect of damage accumulation due to cyclic loading is assessed in order to try and better understand the role that fibre–matrix debonding and matrix plasticity play in the overall macroscopic response of the composite.

2. Finite element modelling

The material under study is HTA/6376, a high strength carbon fibre reinforced plastic (CFRP) manufactured by Hexcel and used extensively in the aerospace industry. The material has a fibre volume fraction of 59.2% and an average fibre diameter of 6.6 μm. In this paper, three studies have been carried out to examine the transverse fracture behaviour of the composite material. Firstly, a linear elastic analysis is carried to examine the microscopic stress state following the thermal cool-down from cure temperature. In particular, the Interfacial Normal Stress (INS) and Interfacial Shear Stress (ISS) which develop during the thermal loading phase are analysed. Secondly, the transverse fracture behaviour of the composite as a result of fibre–matrix debonding is examined. A number of loading cases are considered which examine the effect of thermal residual stress on the transverse fracture behaviour of the material. Parameter studies are carried out examining the influence of interfacial strength and the interfacial toughness on the material behaviour. Finally, the effect of cyclic loading on the transverse fracture behaviour is considered. The material is subject to a tensile/compressive loading regime and the effect of microscopic damage accumulation on the macroscopic response is analysed.

2.1. Generation of micromechanical models

The fibre distribution for the HTA/6376 material has been recently characterised and the Nearest Neighbour Algorithm (NNA) has been developed [10] which enables statistically equivalent fibre distributions to be generated. A typical fibre distribution generated by the NNA is shown in Fig. 2a. The NNA uses experimentally measured nearest neighbour distribution functions to define the inter-fibre distances and an experimentally measured diameter distribution function to assign fibre diameters in the Representative Volume Element (RVE). This allows the short range interaction of fibres in the microstructure to be reproduced enabling an accurate representation of the local microscopic stress state. Five separate RVEs, each statistically identical but topologically different to that shown in Fig. 2a, were generated and each measured 66 × 66 μm and contained approximately 80 fibres. Previous studies have shown that this size is more than sufficient in order to produce the overall macroscopic response of a composite [8]. Python scripts were used to generate finite element models of the RVEs in the ABAQUS [11] finite element code. The fibre and matrix regions were discretised using a quad-dominated mesh which consisted of predominantly 4-noded (CPEG4) and a relatively small number of 3-noded (CPEG3) full integration generalised plane strain elements. To ensure accurate prediction of damage in the matrix, sufficiently dense meshes consisting of approximately 85,000 elements were used. A discrete layer of cohesive elements (COH2D4) was introduced between the fibre and the matrix to predict the onset of fibre–matrix debonding.

Periodic boundary conditions, similar to those used by [12], were applied to the micromechanical models to ensure a macroscopically uniform stress/displacement field existed across the boundaries of each RVE. These consist of a series of tie constraints which require corresponding nodes on each opposing face of the RVE to undergo identical displacements. The periodic boundary conditions can be expressed in terms of the nodal displacement vector, $\mathbf{u}$, such that,

$$\mathbf{u}_{North} - \mathbf{u}_{n_4} = \mathbf{u}_{South} \quad (1)$$

$$\mathbf{u}_{East} - \mathbf{u}_{n_2} = \mathbf{u}_{West} \quad (2)$$

where the subscripts *North*, *South*, *East* and *West* correspond to nodes situated on each edge of the RVE and subscripts, n_2 and n_4 correspond to the control nodes which are located at each of corner of the RVE, as shown in Fig. 2a. Two load cases are considered here, the first being a mechanical transverse tensile load imposed on the RVE by applying a horizontal displacement boundary condition to the active control node, n_2, as shown in Fig. 2a. This is referred to