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Heterogeneity of discontinuous carbon fibre composites: Damage initiation captured by Digital Image Correlation

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ABSTRACT

This paper aims to identify architectural features which lead to damage initiation and failure in discontinuous carbon fibre composites formed from randomly orientated bundles. A novel multi-camera Digital Image Correlation system was used to simultaneously view strain fields from opposing surfaces of coupons, in order to map progression of failure.

The highest strain concentrations were found to occur when the ends of fibre bundles aligned in the direction of loading coincided with underlying transverse bundles. The failure plane was observed to grow between a number of strain concentrations at critical features, coalescing sites of damage to create the final fracture surface. Although potential failure sites can be detected at low global strains in the form of strain concentrations, the strain field observed at low applied loads cannot be extrapolated to reliably predict final failure.

1. Introduction

Composites utilising discontinuous, long fibre reinforcement offer versatility in mechanical properties with relatively low manufacturing costs [1]. A number of discontinuous deposition processes have been developed, offering the ability to tailor fibre architectures with varying fibre length, bundle size, alignment and fibre volume fraction [2,3]. The high level of automation offered by these processes reduces manufacturing costs further. However, as with most discontinuous fibre architectures, these materials are heterogeneous at the mesoscale (bundle level), resulting in both inter and intra-component variation. This variability is the largest single factor preventing wider uptake for structural applications within the composites sector.

A high level of material variability also complicates the prediction of mechanical properties compared with textiles. Modified Rule of Mixtures techniques and classical laminate theory provide sufficiently accurate global stiffness predictions [4–8] for discontinuous fibre architectures, but these 'smeared' approaches fail to account for the complex failure mechanisms which influence the onset and propagation of damage.

Research into the failure of discontinuous fibre composites tends to focus on the microscale, studying the failure of individual filaments, fibre debonding, cracking of the matrix and subsequent stress redistributions [9]. The stress state at filament ends is commonly determined using analytical shear-lag approaches [10–12] or Finite Element (FE) studies [13,14], but it is difficult to extend this work to the bundle level because of complex interactions between contacting fibres. FE models have been developed to establish how the micro mechanical behaviour contributes to failures at the meso and macro scale [15–17], indicating that composite failure is influenced by the packing arrangement in the fibre bundle [18].

The failure mode for mesoscale discontinuous fibre composites is therefore complex and is dependent on additional architectural features that result from kinking and interweaving of short fibres within bundles, local alignment of filament ends and inter-bundle resin rich regions [19]. The tensile failure mode is strongly influenced by the bundle aspect ratio, where the damage mechanism can be interpreted in terms of damage zone size. According to [20], damage tends to initiate at large bundles orientated transversely to the loading direction, providing a natural pathway for crack growth. The crack stops or deviates at points where bundles cross, spreading across large volumes of material and therefore dissipating high levels of energy. As the bundle aspect ratio reduces (fewer filaments or shorter bundles), the absence of large bundles provides less resistance to crack growth. Crack propagation is unhindered as fibres fracture, producing a relatively smooth failure surface and a much smaller damage zone.

Other studies have shown that large stress concentrations exist at bundle ends [21], as the chopped filaments remain coplanar. This end synchronisation is another significant failure initiator and yields lower failure strengths compared with evenly dispersed filamentised materials [22]. Choi and Takahashi [23] investigated

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how the failure mode changes as the depth of these stress concentrations increases from the surface of the sample. Microcracks occur at surface-based tensile stress concentrations due to plane stress conditions. These cracks then propagate along the tip of the fibre into the bulk matrix material. The failure mechanism changes to a shear mode however, as the sub-surface depth of the stress concentration increases, leading to interfacial cracks along the fibre/matrix boundary. Work using acoustic emissions investigating the failure initiation in discontinuous carbon fibre moulding compounds has shown that cracks tend to initiate at the specimen surface [24], but that these early cracks do not necessarily feature in the final fracture plane.

Detailed studies of the surface strains can potentially provide an understanding of how the fibre architecture influences damage initiation and propagation [25,26]. Digital Image Correlation (DIC) has been developed as a practical and widely accepted method of full field strain measurement [27–31]. Local displacements, and therefore induced strains, can be calculated for a specimen by comparing a succession of images taken over time during mechanical loading. As a 3D method, it has the added benefit of eliminating image-plane displacement gradients and associated errors, commonly experienced with 2D approaches [32].

Whilst the use of this technique is well established, strain fields are typically determined from just one side of the specimen. This is satisfactory for characterising the behaviour of homogeneous, isotropic materials, but clearly differences can exist between the strain fields from the outer surfaces for composites with random fibre architectures. Feraboli et al. [33,34] used DIC to study local strain effects in discontinuous fibre composites and reported that there was no correlation between strain concentrations associated with voids or resin rich areas and the site of final failure. Critical features in the fibre architecture may have been overlooked, since strain data was only collected from one side of the sample. It is therefore rational to try and use a multi-camera DIC system to observe the strain fields on both sides of the specimen in stereo.

DIC systems with multiple cameras have been successfully used in the literature [35,36], but not to view opposing faces of a tensile specimen. Up to 4 cameras were used to determine the properties of unidirectional laminates in [35]. Two cameras were placed either side of specimens to view the thickness of the laminate during a four point bend test, demonstrating the ability to use DIC to identify through-specimen differences in the strain fields of multilayer laminates.

Combinations of different optical techniques, including DIC, infrared thermography and X-ray tomography have also been recently used to investigate damage initiation and propagation in carbon–epoxy woven laminates [26]. It was demonstrated that each technique was better suited to detecting different phenomena, for instance local fibre fracture was captured more reliably by infrared thermography, therefore a combination of techniques may prove more reliable for detecting damage events.

The aim of the current paper is to determine if DIC techniques can be used to identify potential failure initiation mechanisms for discontinuous fibre composites, by relating localised strain variations on both sides of the sample to features in the mesoscale fibre architecture, such as bundle ends and resin rich regions. Studies using low-load surface strain maps will be used to determine if the final fracture location can be predicted from such features.

2. Methodology

2.1. Preparation of composite specimens

Two composite materials were chosen for the experimental work; a discontinuous fibre architecture and a biaxial non-crimp fabric (NCF). The NCF was chosen as an orthotropic benchmark for comparison against the quasi-isotropic discontinuous material, to examine how the ply layup sequence influenced the surface strains.

Preforms with a discontinuous fibre architecture were manufactured using the automated Directed Carbon Fibre Preform (DCFP) process [37]. A revolving chopper head randomly deposited carbon tows, cut to 15 mm lengths from a continuous bobbin of Toray T700 60E 12 k. The head moved across a perforated metal grid, through which a vacuum was drawn, depositing fibres over a 600 mm $\times$ 400 mm region. Fibre was initially deposited following a series of North–South linear paths across the entire area, and then completed by depositing in an East–West direction. Binder was applied along with the fibres and the process was repeated until the desired fibre areal mass was achieved (see Table 1).

For the NCF samples, 400 mm $\times$ 300 mm preforms were created by hand laminating individual plies of 200gsm/PW-BUD/T700SC 12 K 50C/0600 mm UD CF NCF (supplied by Sigmatex) into a tool. 12 plies were placed at alternating 0° and 90° orientations to achieve a target volume fraction of 40% in a 3 mm cavity (NCF_1 – $[0/90]_6$). A second unbalanced architecture (NCF_2) was created with the following ply orientations, [0/90,0/90,0/90,0/90,90/0,90/0], achieving a preform where both external plies are in the 0° orientation. 3%wt Reichhold Pretex 110 binder was added between plies and the preform was compacted and cured at 120 °C for 5 min to stabilise.

All preforms were consolidated in a press before being punch cut to fit the 400 mm $\times$ 300 mm Resin Transfer Mould (RTM) tool. Both NCF and DCFP preforms were injected with Huntsman XU3508 resin in a vacuum assisted closed mould tool. The resin was preheated to 80 °C and the tool temperature maintained at 90 °C during the 15 min injection cycle, at pressures of up to 5 Bar. Plaques were then cured for 4 h at 110 °C before being removed from the press and air cooled for 2 h. Five plaques were created in total, with the target thicknesses and volume fractions presented in Table 1.

Dog bone samples were milled from the completed composite plaques (Fig. 1), using a gauge width of 38 mm to avoid problems with material size effects [38]. A stochastic speckle pattern was applied on the front and rear faces using water based paint. Care was taken at this stage to optimise the size and distribution of the pattern to ensure image data was being collected across the whole region of interest for the duration of the test. A pattern too fine or too coarse can result in data being lost across the sample. Each specimen had a horizontal line drawn perpendicular to the sample edge, across the width of each surface. This was used during the post processing phase to align the coordinate system within the software to that of the actual specimen and was designated the *X*-axis. (The direction of tensile elongation was defined as the *Y*-axis).

2.2. Multi-camera cluster DIC

2.2.1. System development

Full field strains were measured using a Dantec Dynamics Digital Image Correlation System (Fig. 2). The system consisted of four 5.0 Megapixel CCD cameras with 28 mm lenses. In a conventional 2 camera Digital Image Correlation (DIC) system, one camera is typically used as a reference. Points from images taken by the reference camera are defined on a grid basis and discretised into facet subsets. These facets can then be identified within subsequent images, from the same or a second camera, and their new position identified relative to the reference. A limitation to this approach is that it is only possible to measure the displacement on surfaces which are within view of the reference camera.

A cluster based approach has therefore been developed for conducting a multi camera analysis on two opposing sides of a tensile