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Multi-scale digital image correlation for detection and quantification of matrix cracks in carbon fiber composite laminates in the absence and presence of voids controlled by the cure cycle

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ABSTRACT

Digital image correlation is applied to the images of a deforming composite to obtain strain maps at three length scales: micro-scale (ply level, hundreds of micrometers), meso-scale (laminate level, millimeters), and macro-scale (specimen level, tens of millimeters). The images are acquired *in-situ* with optical cameras and an electron microscope. The strain mapping at the macro- and meso-scales allows semi-automatic detection of matrix cracks and quantification of their density evolution in function of the applied strain. The micro-scale examination provides additional insights into the failure mechanisms. The technique is developed and then applied to characterize transverse cracking in cross-ply carbon fiber/epoxy composites in the absence and presence of manufacturing defects (including voids). Laminates with defects were produced by lowering the autoclave pressure and the cure temperature, intentionally. The strain for cracking onset and the saturation crack density are found to be different in the inner and outer transverse plies of both types of laminates. The change in processing conditions that led to the presence of voids and incomplete matrix cure resulted in a lower strain for cracking onset and up to 3.5 times increase of the crack density in comparison with the reference material without defects.

1. Introduction

Matrix cracking is one of the first damage mechanisms in composite laminates containing plies with different fiber orientations and subjected to thermo-mechanical loads. Under tension, matrix cracks form when a ply exceeds a certain level of strain, and upon further deformation, they propagate intralaminarly through the ply thickness and in-plane along the fibers. In cross-ply laminates, loaded in the direction of 0° plies, these cracks run perpendicularly to the load direction in the 90° plies, and are called “transverse cracks”. Transverse cracking leads to a decrease in ply stiffness, especially in off-fiber Young's moduli and shear moduli. In carbon fiber composites, this degradation of ply elastic properties is not always detectable on the laminate level because of a low transverse stiffness of carbon fiber plies. In glass fiber composites, on the other hand, transverse cracking causes a characteristic knee in the stress-strain diagram [1] of the laminate explained by a significant contribution of glass fibers to the ply transverse stiffness due to the fibers' high transverse stiffness. Additionally, transverse cracking can cause other forms of damage such as inter-laminar cracks, delamination, and fiber breakage in longitudinal plies [2].

The density of transverse cracks, i.e. the number of cracks per unit laminate length, increases upon further loading. The evolution of the crack density in function of the applied load depends on properties of the matrix and the fiber/matrix interface. In composites with brittle matrices and weak interfaces, the crack density may reach a plateau (referred to as “saturation crack density”). The crack density evolution, i.e. the strain for cracking onset, density growth rate, and saturation density, depend on numerous parameters including the properties of fiber, matrix, and interface as well as fiber volume fraction, ply thickness, laminate stacking sequence, and the presence of manufacturing defects such as non-uniform fiber distribution, incomplete matrix cure, and voids.

The literature about the effect of voids on transverse cracking, reviewed in Ref. [3], is consistent. As stress concentrators, voids cause crack initiation. This effect diminishes with the load increase since more and more voids become already associated with cracks, and the

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number of voids with no associated cracks decreases. This has been reported for carbon/epoxy laminates in Refs. [4,5]. Recently in Ref. [6], the initiation of cracks from voids was explained by strain concentrations around voids, measured using Digital Image Correlation (DIC), which leads to local plastic deformation of the matrix. Through numerical modelling [7], the strain concentration factor around a void with actual shape was computed and correlated with the global strain for cracking onset.

The detection of transverse cracks is challenging, especially for carbon fiber-reinforced composites, as they are not transparent. Indirect approaches for estimating the transverse crack density have been developed based on the analytical models that correlate the crack density with change in the laminate stiffness (for glass fiber composites) [8,9], Poisson's ratio [10], or Lamb wave velocity [11,12]. The accuracy of these approaches is not high due to the indirect nature of the measurement, its low sensitivity to the crack density, and numerous assumptions and approximations. As an alternative, acoustic emission registration is an easy-to-use method for monitoring damage in composites, by which the critical strain thresholds for different stages of damage development can be derived [13–15]. This approach faces important challenges since interpretation of acoustic emission signal "signatures" and identification of the damage type responsible for the signal is ambiguous [16–18].

Direct observation of cracks has been possible with (edge replication) optical microscopy [19], electron microscopy [20–22], X-ray radiography [19] or computed tomography [23], and ultrasonic C-scan [24]. Though these analysis can be performed post-mortem, tracking the evolution of transverse cracking asks for stopping the test and sometimes removing the sample for inspection, which is the main deficiency of such approaches. This is accompanied by stress relaxation, which can affect the cracking process, and closure of cracks, making them difficult to detect. Another challenge is related to the resolution: in order to detect cracks with small opening, the observation window needs to be sufficiently small, which may not be representative.

An alternative approach to interrupted tests is observation of cracks *in-situ*. This is rather straightforward in transparent materials like glass fiber-reinforced plastics (for example in Refs. [25,26]), but in carbon fiber-reinforced composites, such observation of cracks can be done only on the surface of the specimen, either on the in-plane surface or the edge [27–29]. Manual counting of cracks as test progresses is laborious and prone to errors due to poor visibility of unopened cracks or cracks with small opening, in particular for typical aerospace ply thicknesses, i.e. ~0.2 mm. The *in-situ* crack observation and quantification can be further enhanced and (semi-) automated by use of full-field displacement (or strain) measurement, the most common type of it being DIC. Discontinuities in the displacement field or large apparent local strains clearly indicate cracks, which can be recognized by image analysis algorithms.

DIC is extensively used for enhancement of crack observation in experimental damage mechanics. At the time of writing, the search for "digital image correlation" and "crack*" in *Science Citation Index* returns 1518 records and when narrowed to "digital image correlation" and "crack*" and "composite*" and "fibre* or fiber*" the record count is 194. However, only in few recent works such as [30–32], the technique is used to measure the ply-level cracking in laminates. The methodological details of the strain fields are not discussed in these works. As a downside, specific sample preparation and establishment of imaging setup for DIC increases the test workload. DIC was also proven to be a reliable tool for micro-scale analysis of deformation in fiber-reinforced composites [33,34].

In the present study, we employ DIC at three scales to investigate transverse cracking, the most common case for the matrix cracks, in an aerospace-grade carbon/epoxy composite. The methodology includes semi-automatic detection and counting of the cracks. The different scales provide information on initiation and propagation of the cracks. To the authors' knowledge, such a detailed multi-scale application of DIC to characterization of the cracking process in fiber-reinforced polymers has not been reported. The study is conducted on two laminated composites produced with different cure cycles leading to a low-defect material and an imperfect material with voids and incomplete matrix cure. The micro-scale DIC analysis, performed inside a Scanning Electron Microscope (SEM) chamber, clarifies the relation between the voids and cracks initiation.

2. Materials

An aerospace-grade carbon/epoxy composite made from unidirectional prepreg tapes is studied. The prepreg is produced by *Cytec* from high-strength standard-modulus carbon fibers, *Tenax® - E HTS40 F13* 12 K, impregnated with a toughened epoxy resin, *CYCOM® 977-2*. Cross-ply $[90/0]_{2s}$ and $[90/0]_{4s}$ laminates are produced using automated tape laying and cured in an autoclave at *SABCA Limburg NV, Belgium*. Debulking is applied to the lay-ups before cure, which depending on the laminate's thickness, takes place once or several times. Prior to debulking, a release film is placed over the product and a vacuum film is applied on top of it. Debulking is performed by application of a vacuum level of 0.5–0.7 bar for minimum 10 min at room temperature. Afterwards, the vacuum and release films are removed and the final bagging commences. For each stacking sequence, two different types of material are produced. A reference material is made with the manufacturer's recommended (high-pressure-temperature) cure schedule, which includes curing at 180 °C and 6 bar, producing a low-defect material. An imperfect material is produced with intentional change of the cure cycle to introduce voids for research purpose. This low-pressure-temperature cure schedule, similarly to [35], includes curing at 150 °C and 0.5 bar, resulting in an intra-laminar voidage, adequate for this study (Fig. 1). The vacuum is maintained during both high- and low-pressure-temperature cycles.

Two reference and two void-containing plates each with dimensions of 200 mm × 200 mm are produced. The thickness of the reference and imperfect plates is 1.65 ± 0.03 and 1.46 ± 0.02 mm, respectively, for the $[90/0]_{2s}$ lay-up, and 2.97 ± 0.05 and 2.87 ± 0.04 mm, respectively, for the $[90/0]_{4s}$ lay-up. The slightly higher thickness of the reference plates may be due to the fact that they were designed to have one rough surface, while both surfaces of the imperfect plates were made smooth. The different surfaces of the two plates would not affect

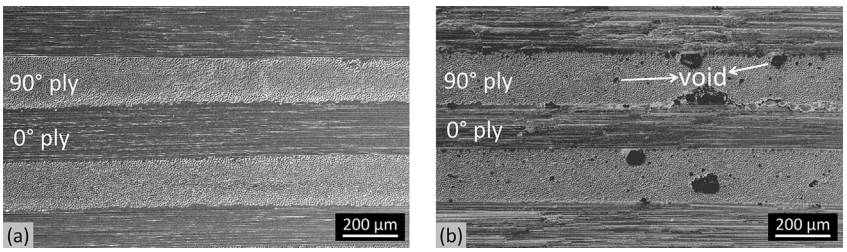

Fig. 1. SEM micrographs of the composite cross-sections for (a) the reference laminate produced with a manufacturer's recommended cure cycle and (b) the imperfect laminate produced with a different cure schedule.