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Early stage damage in off-axis plies under fatigue loading

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ABSTRACT

The very early stages of damage evolution under fatigue loadings were investigated by testing $[45/\text{-}45/0]_s$ glass/epoxy specimens. Microscope observations revealed that the first damage event was the initiation of micro-cracks in the 45° plies within the inter-fibre region. Off-axis cracks then formed by the coalescence of these micro-cracks and propagated along the fibres direction with the same mechanism. The orientation of the micro-cracks was proven to be normal to the direction of the local maximum principal stress in the matrix. The results presented here are the first published evidences of fatigue damage initiation at the inter-fiber scale and represent a further validation of the fatigue crack initiation criterion recently presented by the authors.

1. Introduction

Multidirectional laminates made of unidirectional (UD) plies are frequently adopted in many structural applications as they offer optimal specific strength and stiffness, combined with the possibility to select and design material properties according to specific needs. Very often the in-service conditions are characterised by cyclic loads, which can lead to a progressive degradation of the laminate stiffness, mainly due to the damage evolution in the form of multiple off-axis cracks [1–9]. This behaviour is reported in several works in the literature for multidirectional laminates under uniaxial [1–7] or multiaxial loads [8,9].

Within this scenario, predicting the initiation of off-axis cracks is, first, necessary to define a procedure for describing the progressive damage evolution and stiffness degradation of laminates. Since the stress state in the off-axis plies of a laminate is in general multiaxial, even in the presence of uniaxial loads [7,10,11], a criterion for predicting the crack initiation must be capable of accounting for multiaxial stress states.

In addition, in the authors' view, a reliable initiation criterion should be based on evidences and observations derived from an extensive experimental activity, aimed not only at characterising the global material response, but also the damage mechanisms at the microscopic scale responsible for the damage onset.

The matrix-dominated fatigue behaviour of UD plies was characterised by testing flat unidirectional off-axis laminates under uniaxial loadings [12–16] and tubular specimens under combined tension-torsion loadings [11]. In these situations damage evolution was confined to the very end of the life of the samples without significant stiffness reduction and off-axis crack formation during the fatigue life. This clearly indicates the absence of a progressive damage evolution at least at the macroscopic scale.

However at the microscopic scale, i.e. the length scale of the inter-fibre distance, damage evolves from the early stages of fatigue life due to irreversible processes leading to a critical condition for the initiation of an off-axis crack. In UD laminates the onset of an off-axis crack corresponds to the final separation, whereas in constrained plies the initiated off-axis cracks can propagate steadily in the fibre direction [1–4,6,7,17–19]. Accordingly, the sensitivity of off-axis plies and laminates to fatigue loadings can only be justified by assuming the presence of a damage evolution at the micro-scale [20].

Notwithstanding this, in the authors' best knowledge, no experimental evidences are available in the in the literature allowing the fatigue damage initiation and evolution at the microscopic scale in off-axis UD plies to be understood. Only post-mortem observations of fatigued specimens are reported.

The authors presented SEM analyses of the fracture surfaces of glass/epoxy tubes with three 90° layers constrained by an external and internal thin fabric ply subjected to combined tension-torsion cyclic loadings [17]. The stress state in the 90° plies was characterised by the presence of the in-plane shear stress σ_6 and transverse stress σ_2, combined in four different values of their ratio

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$\lambda_{12} = \sigma_6/\sigma_2 = 0$, 0.5, 1, 2. Fracture surfaces were rather smooth when the shear stress component was null or very low ($\lambda_{12} = 0$, 0.5), whereas a significant presence of shear cusps in the matrix was observed in the case of higher shear stress contributions ($\lambda_{12} = 1$, 2).

A similar morphology was reported by Shiino and co-authors [21] for $[45/0/\text{-}45/90]_{2s}$ carbon/epoxy laminates brought to fatigue failure under a uniaxial load. These authors took in-plane images of the external 45° ply, showing the presence of shear cusps in the matrix.

Damage evidences identified on the fracture surfaces of samples under combined transverse and shear stresses were considered by the present authors in the development of a fatigue crack initiation criterion recently presented in Ref. [22].

The main bases of the criterion are the following:

i) in the presence of combined transverse and shear stresses (with the shear component high enough, i.e $\lambda_{12} > 0.5$) damage evolves at the micro-scale by means of the initiation and subsequent coalescence of micro-cracks in the matrix (producing the above mentioned shear cusps);
ii) the initiation of these micro-cracks is controlled by the Local Maximum Principal Stress (LMPS) in the matrix, calculated at the fiber-matrix scale;
iii) in the presence of a near transverse stress state, initiation of micro-cracks is controlled by the Local Hydrostatic Stress (LHS) in the matrix, extending to cyclic loads the approach proposed by Asp and co-authors for static loadings [23].

This paper focuses on the damage onset in the off-axis plies where the crack initiation is usually controlled by the LMPS. In this situation the initiation criterion was already validated, at least at a macroscopic level, by re-analysing multiaxial fatigue data at crack initiation coming from different sources in terms of this local stress parameter. The use of the LMPS produced a single, narrow scatter band capable to contain all the data analysed.

This result represents indeed an implicit validation of the approach proposed, since it was directly applied on the last stage of the damage evolution at the micro-scale, i.e. the onset of an off-axis crack.

Something that is still missing, instead, is the direct observation of the damage mechanisms taking place at the microscopic scale (assumed to be micro-cracks between the fibres) before the formation of an off-axis crack. This would provide physical, experimental evidences to support the criterion. According to the authors' best knowledge the only evidence in the literature showing such micro-cracks in a specimen before its final separation was presented by Cox and co-authors [24] for a [±45] laminate under quasi-static loadings. To completely validate the initiation criterion it remains also to show that the plane of nucleation of these micro-cracks is normal to the orientation of the LMPS in the matrix.

Within this context, the aim of the present work is to identify by direct observations the damage mechanisms responsible for the damage onset at the micro-scale. A dedicated specimen with lay-up $[45/\text{-}45/0]_s$ was designed and tested under a uniaxial fatigue load. The surface of the external 45° ply was accurately polished to make the fibres and the matrix clearly observable. The specimen was removed from the testing machine at regular cycles intervals and observed under an optical microscope, revealing the presence of micro-cracks between the fibres, before the initiation of off-axis cracks. In addition, the orientation of the plane where micro-cracks nucleate was proved to be normal to the direction of the Local Maximum Principal Stress.

2. The damage-based criterion for crack initiation in composite laminates

To better motivate the aim of the work, in this section the crack initiation criterion recently proposed by the authors [22] is briefly presented. As already mentioned, it is based on the assumption that when the in-plane shear stress is high enough, damage at the microscopic scale occurs in the form of micro-cracks in the matrix and their initiation is driven by the Local Maximum Principal Stress. In particular it was assumed that these micro-cracks start to grow in a plane, named the *local nucleation plane*, which is normal to the direction of the Local Maximum Principal Stress. The LMPS is therefore representative of the driving force for this damage evolution, leading to the initiation of an off-axis crack, as schematically shown in Fig. 1a.

As discussed in Ref. [22], this holds valid for stress states involving a reasonably high amount of shear. If the shear stress component is null, or very low, the damage initiation is instead driven by the Local Hydrostatic Stress (LHS) in the matrix, as already proved by Asp and co-authors [23] in the case of static loadings.

The criterion proposed in Ref. [22] is applicable when the fibre-matrix adhesion is strong enough to prevent the initiation of damage in the form of fibre-matrix debonding. An extensive discussion on this subject is presented in Ref. [25].

Provided that this condition is verified, it was shown that crack initiation data can be collected into two scatter bands relating the number of cycles for the off-axis crack initiation to the LHS or LMPS, depending on the amount of the shear stress [22]. In the present work the attention is focused only on the LMPS-driven mechanism, which indeed characterises most of loading conditions of practical interest.

To calculate the LMPS resulting from a general in-plane stress state (σ_1, σ_2 and σ_6) the authors proposed to carry out Finite Element (FE) analyses on a fibre-matrix unit cell subjected to periodic boundary conditions as shown in Fig. 1b [22]. For a square or hexagonal fibre array the highest value of the LMPS is located at point A, where the LMPS value and the orientation of the local nucleation plane, β_c, can be evaluated as

$$\mathrm{LMPS} = \frac{1}{2}\left[\sigma_{rr} + \sigma_{zz} + \sqrt{\sigma_{rr}^2 + 4\sigma_{rz}^2 - 2\sigma_{rr}\sigma_{zz} + \sigma_{zz}^2}\right] \tag{1}$$

$$\beta_c = \pi - \frac{1}{2} ArcTan\left[\frac{2\sigma_{rz}}{\sigma_{rr} - \sigma_{zz}}\right] \tag{2}$$

The orientation of the *local nucleation plane* is a parameter of fundamental importance as it quantifies the degree of local multiaxiality in the matrix. The capability of the LMPS-based criterion of collecting data for different multiaxial conditions lies, indeed, in the correct identification of the plane in which micro-cracks initiate, i.e. the *local nucleation plane*, as extensively discussed in the appendix of Ref. [22].

The assumption of a regular fibre array for computing the *local nucleation plane* orientation may be seen as too simplicistic and a potential limitation of the range of validity of the proposed approach.

The authors, however, already proved that the orientation β_c of the *local nucleation plane* does not depend on the unit cell type, orientation and volume fraction [22]. In addition, in Appendix A it is proved that a random distribution of fibres, which is more likely to occur in actual laminates, results in the same β_c as evaluated by a regular fibre array, thus justifying the use of the LMPS and β_c calculated from unit cells even in the case of real laminates with