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Influence of matrix ductility and toughness on strain energy release rate and failure behavior of woven-ply reinforced thermoplastic structures at high temperature

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ABSTRACT

The purpose of the present work is to investigate damage evolution in 5-harness satin weave carbon fabric reinforced PolyPhenylene Sulphide (PPS) structures with an initial edge notch. To understand how the physical properties of the constituents (e.g. matrix toughness and ductility) and the architecture of reinforcement (woven-ply) affect the fracture behavior of C/PPS laminates, it is useful to have analytical representation of the translaminar failure modes based on fracture mechanics concepts, the strain energy release rate *G* especially. Translaminar failure is determined by the combination of loading, location of defect and material heterogeneity (presence of matrix-rich regions at the crimp area in woven-ply laminates). When translaminar failure is initiated from an existing notch, a sequence of energy-absorbing events (fiber breakage, matrix cracking, fibers pull-out, fiber/matrix debonding) occurs in a region surrounding the notch tip. The knowledge of energy-absorbing processes is therefore important since they are responsible for the toughness of the composite. Depending on laminates' stacking sequence, the contribution of matrix behavior to strain energy release rate can be evaluated during damage in both brittle and ductile composite laminates subjected to high temperature conditions ($T > T_g$) when matrix ductility and toughness are enhanced. Depending on the initial notch orientation (0 or 45°), the failure mode is either a mode I or a mixed mode (I + II). The acoustic energy associated with translaminar failure was correlated with the strain energy release rate during translaminar failure. The total strain energy release rate in quasi-isotropic (QI) laminates is 6 times as low as in angle-ply (AP) laminates, suggesting that large plastic deformation (due to a matrix-driven behavior and an enhanced matrix ductility at $T > Tg$) are instrumental in dissipating a great portion of the mechanical energy brought to the specimen in AP laminates. The "material" effect is combined with a structural one (rotation of the fibers) at the crack tip, and leads to ductile failure. Both effects contribute to high fracture toughness in AP laminates.

1. Introduction

Edge cracks can initiate from notches at the surface of composite laminates due to processing (consolidation, machining) or stress concentrations due to the geometry of composite structures. They may result in the failure of composite structures subjected to service conditions (monotonic or cyclic loadings). It is therefore necessary to evaluate the influence of prominent factors (notch orientation, stacking sequence, matrix ductility, service temperature) on the overstress distribution at the notch tip in order to better understand the subsequent damage mechanisms and failure modes (opening or mixed-mode). In orthotropic or quasi-isotropic composite laminates, transverse matrix cracking and fibers breakage (also known as translaminar failure modes) are usually the primary damage mechanisms occurring in the early phase of mechanical loading. A comprehensive review of techniques for the experimental characterization of the fracture toughness associated with the translaminar failure modes of continuously reinforced laminated composites is presented in Ref. [1]. The micro-structural mechanisms of damage, including fiber breakage and matrix

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cracking, fiber/matrix debonding, transverse-ply cracking, delamination, and void growth, form a discrete but complex damage zone (see Fig. 1) [2].

1.1. Fracture toughness and damage mechanisms in fiber-reinforced plastics

Fracture toughness is utmost important for structural applications as it represents the material capability to resist to fracture. The toughness of a polymeric matrix depends on several factors: the amount of amorphous phase (in semicrystalline polymers such as PPS), the size and size distribution of crystals, the amount of rigid amorphous phase, the glass transition temperature (T_g), the strain at failure. In fiber-reinforced plastics, fibers are usually tougher than the polymer matrix [3], and the various origins of fracture toughness in composites may be characterized by considering the sequence of microscopic fracture events that lead to crack propagation macroscopically under monotonic increasing loads. The cracks in composites can propagate preferentially along the fiber-matrix and laminar interfaces (i.e. longitudinal splitting) or transversely right through the fiber and matrix (i.e. transverse cracking), depending on the properties of the interface relative to the fiber and matrix (see Fig. 2). When a crack present in the matrix approaches an isolated fiber, the following failure mechanisms may be expected to take place [14]: (1) matrix cracking - (2) fiber-matrix interface debonding - (3) post-debonding friction - (4) fiber fracture - (5) stress redistribution - (6) fiber pullout.

The non-conservative deformation processes have been discussed in details in the literature [4]. In highly anisotropic composites with high fiber content, a crack normal to the fibers will often refuse to propagate in mode-I, but will be diverted into a splitting mode. In unidirectional (UD) carbon-fiber reinforced plastics, this may result in a brittle, end-to-end splitting failure which simply eliminates the crack. By contrast, in fibers-reinforced plastic laminates containing woven-roving or chopped-strand mat reinforcement, crack-tip damage may remain localized by the complex geometry of the fiber array in the Fracture Process Zone (FPZ), and the crack may proceed through this damaged zone in a fashion analogous to the propagation of a crack in a plastically deformable metal [5]. The crack-stopping ability of composites, which results from their inhomogeneity on a fine scale (the fiber/matrix interface) and on a gross scale (laminated structure) makes

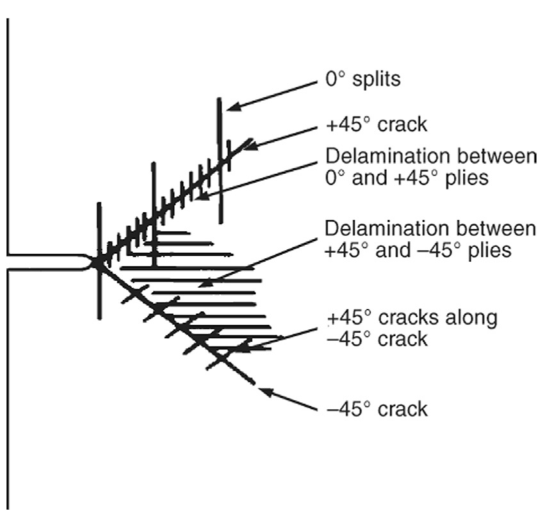

Fig. 1. Typical damage zone at a sharp crack in composite laminates subjected to mode I loading [2].

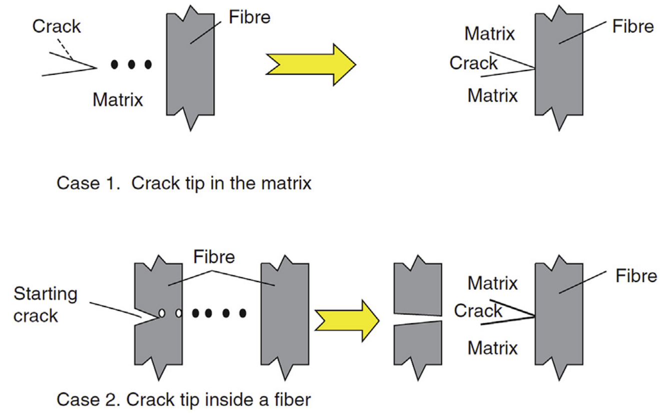

Fig. 2. Different cases of a crack in a composite laminate [3].

it difficult in many cases to generalize a fracture mechanics approach to fiber-reinforced composites with different reinforcements (UD- or woven-ply) and various stacking sequences. In UD-ply composite laminates, stress concentrating effects of notches and holes may be almost completely eliminated by large-scale splitting in the 0° and 45° plies and by delamination between the plies, the net result often being disintegration of the composite. By contrast, in woven-ply, the scale of this damage is limited by the woven structure of the composite, and cracks and notches will often propagate in a more self-similar fashion, especially in wide plates [5]. It also appears that woven-ply laminates are characterized by matrix-rich regions at the crimps (where warp fiber bundles undulate over weft fiber bundles), and the potential benefit of these matrix-rich regions is the development of plastic yield zones at the cracks tip as intra- and inter-ply cracks propagate [6]. Depending on matrix toughness and ductility, the localized matrix plasticization is instrumental in ruling damage mechanism, as these matrix-rich regions may act as cracks barriers and subsequent propagation in unnotched laminates, as well as in notched laminates [7]. It is therefore expected this mechanism to reflect on material toughness measurements either in fiber-dominated laminates (quasi-isotropic) or matrix-dominated laminates (angle-ply). Toughness is therefore an intrinsic property of a material and it is the ability of a material to dissipate deformation energy without propagation of a crack [8]. The design of tough microstructures in structural materials demands a compromise between resistance to intrinsic damage mechanisms ahead of the tip of a crack (intrinsic toughening) and the formation of crack-tip shielding mechanisms, which act behind the tip to reduce the effective “crack-tip driving force” (extrinsic toughening) [9]. The introduction of soft regions into fiber-reinforced composites to provide barriers to crack growth depends on matrix ductility, resulting in raising the intrinsic toughness of the material [7] [10]. As far the crack propagation is concerned, it is therefore potentially interesting to associate woven fabrics with highly ductile TP matrices [11], the effect of which is even more noticeable when service temperature is higher than the material glass transition temperature [12], even in Q-I laminates whose behavior is fiber-dominated. However, as underlined by Sih et al., the material in the immediate vicinity of the crack front is highly stressed and its behavior is not exactly known, therefore the fracture mechanics analysis must necessarily be restricted to regions outside of a small zone surrounding the crack tip [13].

1.2. Fiber/matrix interface-related fracture toughness

During fracture, the local response of the fiber-matrix interface