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MAXIMISING THE FRACTURE RESISTANCE OF GLASS FIBRE COMPOSITES BY CONTROLLED LARGE SCALE FIBRE BRIDGING

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ABSTRACT

Delamination is a common failure mode of composite structures due to their often low interlaminar fracture resistance. The use of conservative design approaches to prevent crack initiation, leads to design that are not optimal. The alternative approach is to allow damage initiation and some/limited crack propagation in the structure. Such an approach requires the use of composite materials with high damage tolerance. In the present work, we aim to increase the fracture resistance of unidirectional composites by large scale fibre bridging. It is shown that the steady-state fracture resistance can be increased several times by modifying the fibre sizing and matrix material.

1 INTRODUCTION

Composite materials are increasingly being used in applications requiring stiff and lightweight structures such as wind turbine rotor blades and aircrafts. Structural details such as adhesive or bolted joints, T-connections and ply drops [1] create complex high-stress fields including out-of-plane stresses. This is a critical issue since composite materials exhibit low interlaminar fracture resistance and are susceptible to delamination. Once a crack initiates, subsequent interlaminar crack propagation may lead to a substantial decrease of the structural integrity of a composite structure [2, 3]. As a result, several techniques have been developed to increase the fracture resistance, and thereby creating more damage tolerant composite materials.

Improving the fracture resistance of composite materials has been achieved by various means e.g. a) tougher matrix material using interleaves [4] or thermoplastic fibres [5] or b) modifications of the fibre architecture e.g. z-pinning [6]. It should be noted that any modification in the composite material to increase the fracture resistance should not have a substantial effect on other critical properties such as in-plane fatigue.

In the present work, the aim is to significantly increase the fracture resistance of composite materials by using the fibre bridging mechanism (the fibres or ligaments of fibres that connect the crack faces in the wake of an advancing crack tip [7, 8]). It is envisioned that by increasing the amount of fibre bridging, the crack growth rate of a delamination crack under cyclic loading can be decreased or even that the delamination can be arrested. The dissipated energy in the fibre bridging zone or fracture process zone behind the crack tip is controlled by the fibre and matrix mechanical properties and by the adhesion between them. The focus of the present work is on unidirectional glass fibre reinforced vinyl ester composites. Both the fibre sizing, influencing the fibre/matrix adhesion, and the resin properties were modified to maximise the fracture resistance whereas the glass fibres properties were not modified.

In many applications, such as wind turbine rotor blades, the unidirectional composites are in the form of fabrics e.g. support threads (backing fibres) and stitching yarns are used to stiffen the fabric and maintain the fibre alignment during handling and manufacturing. However, these additional

features, despite their low amount in the composite, can have a strong negative effect in the interlaminar fracture resistance, as it will be shown in the present work.

2 EXPERIMENTAL DETAILS

2.1 Materials

Unidirectional fabric glass fibre reinforced vinyl ester composites were manufactured by vacuum infusion. E-glass fibres with different sizings in the form of rovings were used as reinforcement. Different vinyl ester resins (VE) with respect to chemistry and mechanical properties were used as matrix material. The reference composite material consisted of a commercially available grade vinyl ester resin (VE_{ref}) and E-glass fibres with a commercial sizing ($Sizing_{ref}$). In addition to the commercially grades, two experimental vinyl ester resins were used: a high elongation resin (VE_a) and a rubber modified resin (VE_b). All resins were formulated for Norox PBC-21 as a curing agent. Similarly, two experimental fibre sizings were used: a sizing with increased flexibility ($sizing_1$) and a sizing with an improved compatibility to vinyl ester resins ($sizing_2$). For the purpose of this study, the unidirectional composites included a minimum amount of support threads and a commercial stitching pattern with a direction transverse to the fibre direction. The direction of the fabrics (e.g. the support threads facing outwards or inwards) during laying up the fabric layers did not have an influence in fracture resistance values.

Unidirectional composites, without support threads and stitching yarns, were also produced by filament winding for comparison purposes. In this case, the commercial vinyl ester resin and fibre sizing were used. Fig. 1 shows the difference in the microstructure between the fabric-based and filament-winded composites. For all laminates, a 35 μ m thick and 60 mm wide perforated release film was placed in the middle of the laminates prior to infusion. The release film (slip foil) acted as crack starter.

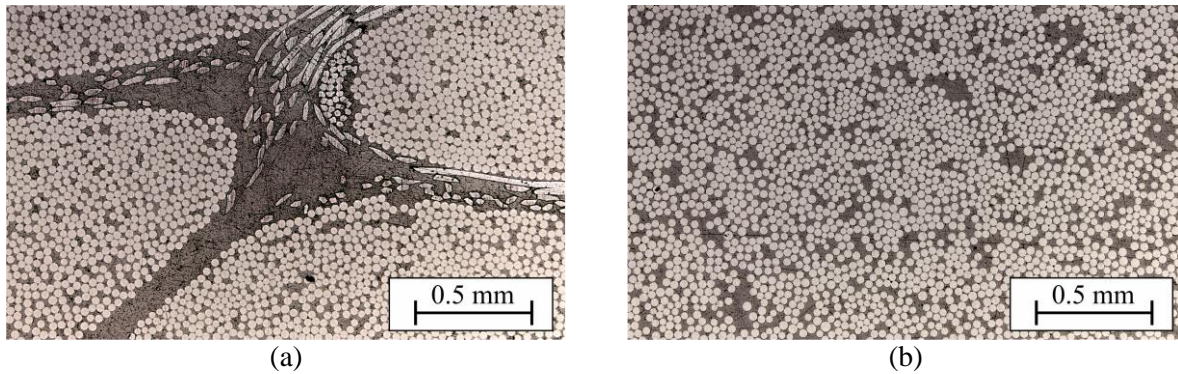


Figure 1: Cross-sectional photographs of a) fabric-based composite with rubber modified resin and sizing with compatibility to vinyl ester resins, b) filament-winded composite with commercial resin and sizing.

2.2 Specimen geometry and test method

The test specimens (beams) were subsequently cut from the laminates. The specimen length, L , was 500 mm, the width, B , was 30 mm and the height, $2h$, was approximately 2 mm. Steel beams of height, H , equal to 6.67 mm, were glued to the composite laminates to produce the sandwich Double Cantilever Beam (DCB) specimen shown schematically in Fig. 2a. The position of the slip foil is also shown with $a=60$ mm as mentioned above.

2.3 Test method

Mode I fracture mechanics tests were performed using a test fixture (see Fig. 2b), which applies pure bending moments [9]. The force, P , was measured by two load cells and the average values was

used to calculate the applied moment, M . An extensometer was mounted on the steel beams at a position corresponding to the end of slip foil ($x_1=0$) to measure the end-opening, δ , of the crack. The experiments were conducted at constant displacement rate of the lower beam (see Fig. 2b). The data from the load cells and the extensometer were recorded with a frequency of 25 Hz.

To calculate the fracture resistance, J_R , the path independent J integral [10] was used. The J integral along the external boundaries of the DCB specimen, under plane strain conditions, is given by Bao et al. [11]:

$$J_R = \frac{1 - \nu_s^2}{E_s} \frac{M^2}{B^2 H^3 \eta^3 I} \quad (1)$$

where E_s and ν_s are the Young's modulus and Poisson's ratio of the steel beams. The parameters η and I depend on the geometry of the DCB specimens and the elastic properties of the steel and composite beams and are given in [11].

Two acoustic emission sensors from Physical Acoustics Corporation were mounted on the two ends of the DCB specimens to record the acoustic emission (AE) signals from the fracture process. The first sensor was placed close to the end of the slip foil ($x_1=0$) and the second sensor at $x_1=300$ mm.

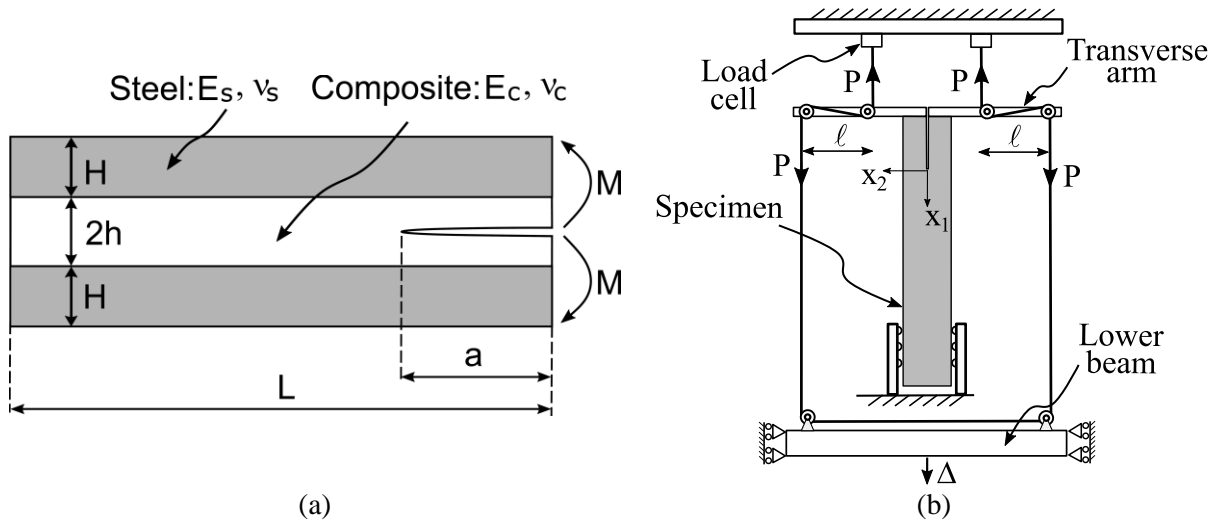


Figure 2: Schematic illustration of the a) test fixture using a wire and rollers to apply pure uneven bending moments, and b) of the DCB specimen.

3 RESULTS

Fig. 3 shows the fracture resistance curve for three unidirectional, fabric-based, composites (UD fabric) made by the commercial VE resin and commercial sizing. The steady-state fracture resistance is approximately 1300 N/m. In the same figure, the fracture resistance curves for three pure unidirectional composites (UD winding) are plotted. The average steady-state fracture resistance is approximately two times higher (2600 N/m) than the unidirectional, fabric-based, composites. The steady-state fracture resistance is attained at an end-opening approximately equal to 2 mm for the fabric-based composites and at a much higher opening, $\delta \approx 6$ mm, for the filament-winded composites. The results of Fig. 3 indicate that the support threads and stitching yarns have a strong negative effect on the fracture resistance.

Next, in Fig. 4, the fracture resistance curves for the unidirectional, fabric-based, composites with experimental resins and sizings are shown. When a rubber modified VE resin (VE_b) is used in combination with a sizing with improved compatibility ($sizing_2$) to VE resins, the steady-state fracture resistance is significantly improved and approaches the steady-state fracture resistance of the filament-winded composites. The average steady-state fracture resistance is approximately 2400 N/m.

By using a high elongation VE resin (VE_a) together with a flexible sizing ($sizing_1$), the steady-state

fracture resistance is even higher (almost two times) than the unidirectional filament-winded composites with the commercial resin and sizing. The average steady-state fracture resistance is approximately 4500 N/m. When the steady-state fracture resistance of these composites is compared with the steady-state fracture resistance of fabric-based composites using commercial grades, the increase is approximately by a factor 0.35.

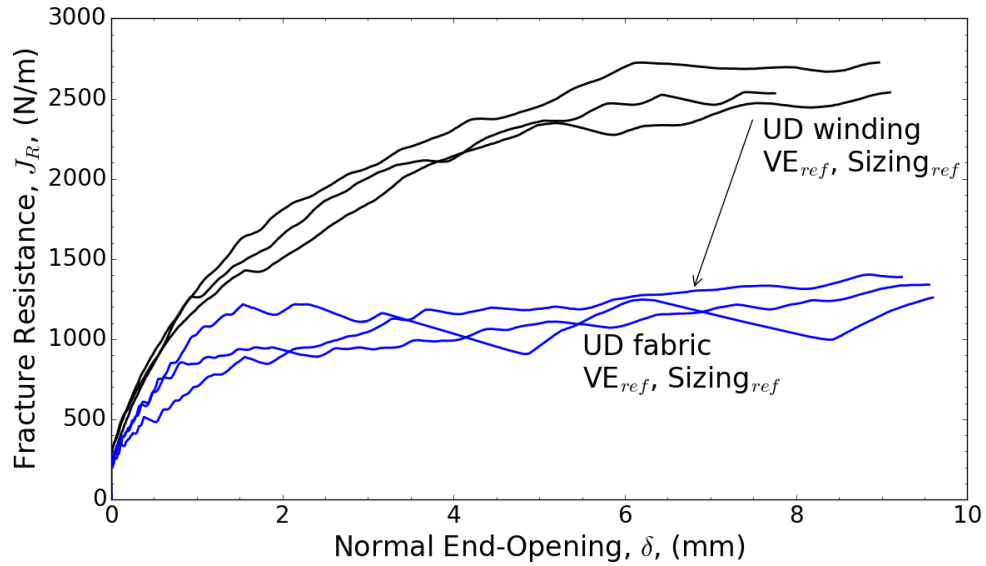


Figure 3: Measured fracture resistance curves, J_R , as a function of the (normal) end-opening, δ , for unidirectional, fabric-based, composite and a pure unidirectional composite manufactured by filament winding. A commercial VE resin (VE_{ref}) and a commercial sizing ($sizing_{ref}$) were used.

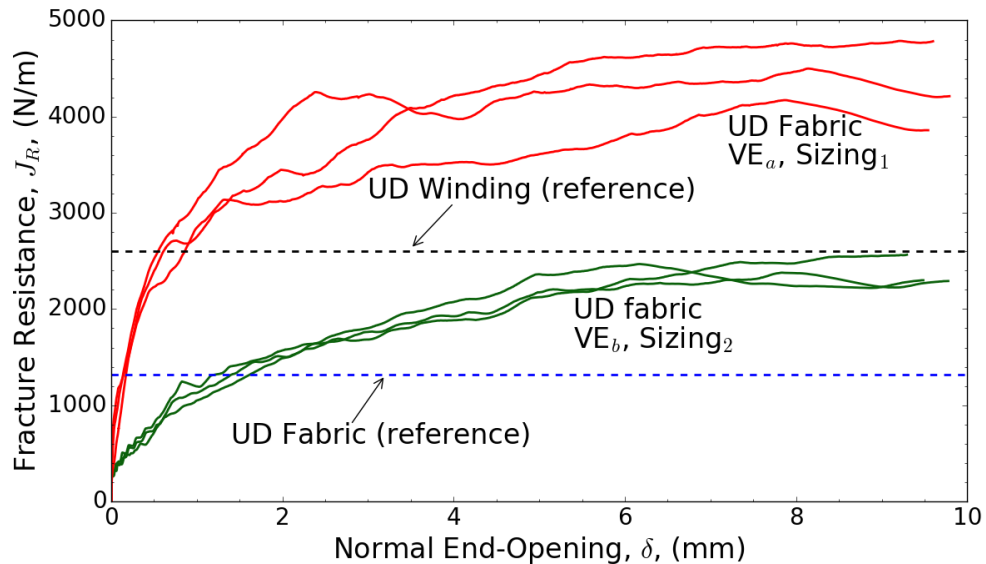


Figure 4: Measured fracture resistance curves, J_R , as a function of the (normal) end-opening, δ , for unidirectional, fabric-based, composites with difference VE resins and sizings. The dash lines indicate the average steady-state fracture resistance values of the unidirectional, fabric-based and filament-winded composites from Fig. 3.

Photographs of the fracture process zone, at the wake of the crack tip, are shown in Fig. 5. Extensive fibre bridging can be observed for the filament-winded composites (based on commercial resin and sizing). The fracture is not confined to a thin damage zone; rather a damage band of nearly the entire composite develops. There is no clear fracture plane at some distance away

from the crack tip. On the other hand, for a fabric-based composite, also based on commercial resin and a sizing, fibre bridging from the reinforcing fibres is nearly suppressed by the presence of the stitching yarns as can be seen in Fig. 5b. The bridging mechanism in this case is mainly due to the stitching yarns being separated and pulled apart as indicated by the regularly-spaced pattern connecting the crack faces behind the crack tip. This can explain the significantly lower fracture resistance of these composites in comparison with the filament-winded composites (see Fig. 3) and the small end-opening at which the steady-state fracture resistance is attained. Fig. 5c shows the fracture process zone for the fabric-based composite with the highest steady-state fracture resistance (VE_a and $Sizing_1$, see Fig. 4). The fracture process zone consists of both reinforcing fibres and stitching yarns bridging the crack faces. This observation can explain that the end-opening at which the steady-state is attained is larger than the fabric composite with commercial grades (Fig. 5b) and smaller than the filament-winded composite (Fig. 5a). From the photographs of Fig. 5, it is not clear why the fabric-based composite of Fig. 5c has a much higher steady-state fracture resistance than the filament-winded composite.

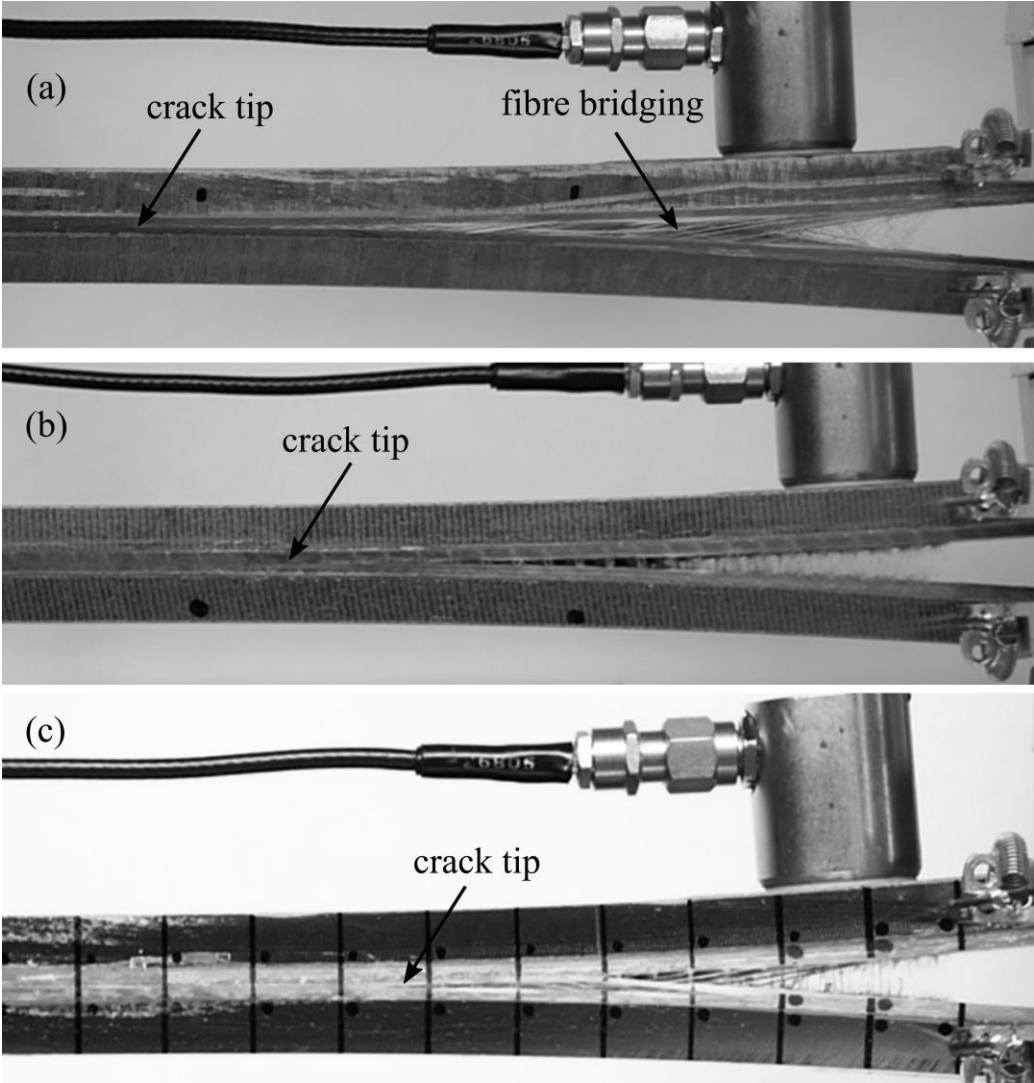


Figure 5: Photographs of the DCB specimens when the crack has grown several millimeters showing the fracture process zone at the wake of the crack tip for three unidirectional composites: a) filament-winded using commercial resin and sizing (V_{ref} , $sizing_{ref}$), b) fabric-based with commercial resin and sizing, and c) fabric-based with a high elongation resin (VE_a) and flexible sizing ($sizing_1$),

The AE signals recorded during the tests were analysed for all the composites tested. The various

temporal features (e.g. energy, amplitude, counts etc) of the AE signals were used to identify possible differences between the various composites. Fig. 6 shows plots of the duration vs rise time of the AE signals at steady-state for three composites. No significant differences can be observed for the composites shown. For the filament-winded composite (Fig. 6a) and the fabric-based composite with commercial resin and sizing (Fig. 6b), AE signals with large duration time recorded. For the fabric-based composite with the highest fracture resistance (Fig. 6c), the number of AE signals with high duration time is lower. The fracture of this composite gives a larger number of AE signals with low duration and rise times.

The cumulative AE energy was evaluated for all composites for the entire duration of the tests. The filament-winded composite had the larger cumulative energy and the shortest testing duration. The fabric-based composite with the highest fracture resistance had the lowest cumulative energy (the difference in cumulative energy between these two composites was approximately 20%) even if the duration of the tests was longer.

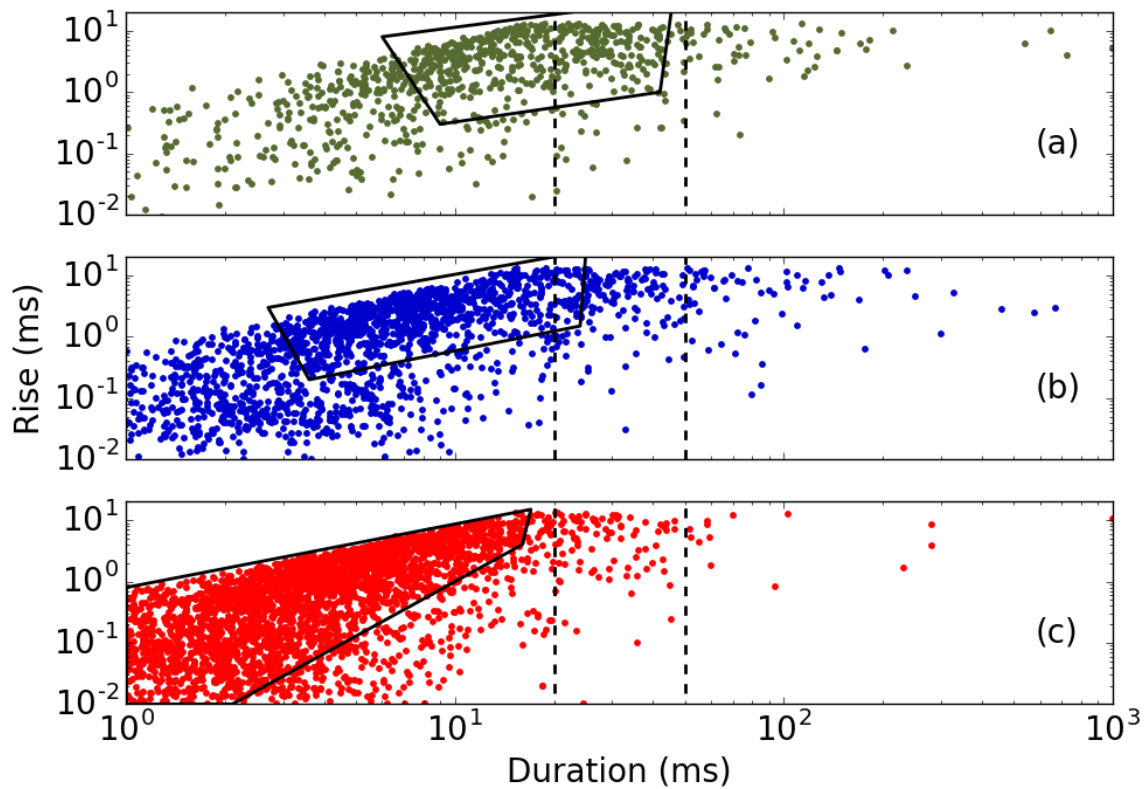


Figure 6: Duration versus time rise of the AE signals at steady-state for three unidirectional composites: a) filament-winded using commercial resin and sizing (V_{ref} , $sizing_{ref}$), b) fabric-based with commercial resin and sizing, and c) fabric-based with a high elongation resin (VE_a) and flexible sizing ($sizing_1$).

4 DISCUSSION

Unidirectional fabric-composites due to their microstructure (see Fig. 1), have a lower steady-state fracture resistance than the unidirectional filament-winded composites as it is shown in Fig. 3. By modifying the resin formulation and altering the fibre/matrix adhesion, using different fibre sizings, it is possible to obtain fabric-based composites with the fracture resistance even much higher than filament-winded composites. Based on the optical observations of the fracture process zone and the analysis of the AE signals, it is not yet possible to identify with confidence the reasons for this very large increase in fracture resistance. The amount of bridging fibres is not, qualitatively, larger than in the filament-winded composites. In addition, the differences between the AE signals characteristics are

not significant. In addition, the cumulative energy from the AE signals is higher for the filament-winded composites. Based on these observations, it may be argued that there might be a large contribution from the matrix plastic deformation both at the matrix crack tip and at the debond crack tips of the debonding fibres. The VE resin used for this composite was a high elongation resin and thus plasticity is possible mechanism contributing to the high fracture resistance. Matrix plasticity can be captured by AE. In addition, the use of the flexible sizing probably decreased the fibre/matrix adhesion (based on the visual appearance of the laminates), leading to more reinforcing fibers bridging the crack faces than the fabric-based composites based on the commercial sizing. However, probably the effect of this second mechanism is weaker than the plastic deformation of the resin.

5 CONCLUSIONS

Based on the same resin and fibre sizing system, filament wound composites can have more than two times higher steady-state fracture toughness than fabric-based composites. Experimental vinyl ester resins and sizings were developed in order to increase the fracture resistance of the unidirectional, industrial relevant (fabric-based), composites. It was shown that it is possible to significantly increase the fracture resistance of these composites; the fracture resistance can be even almost two times higher than the filament-winded composites based on commercial grades. Based on the current results, no clear conclusions can be drawn for the main mechanism contributing to this large increase in fracture resistance. It is suggested that matrix plastic deformation can have a strong positive effect on the fracture resistance.

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