
A REVIEW OF THE PULL-OUT MECHANISM IN THE FRACTURE OF BRITTLE-MATRIX FIBRE-REINFORCED COMPOSITES

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ABSTRACT

The current work addresses the role of damage mechanisms such as interfacial debonding, crack deflection, bridging and sliding during fracture of a brittle-matrix fibre-reinforced composite with respect to their energy dissipation capacity and their impact on the pull-out mechanism. The aim of the paper is to explain why fibre failure is preferably concentrated within the matrix environment to give rise to the pull-out mechanism and not within the crack flanks where fibre stress is maximum. Two approaches, mechanics of materials and fracture mechanics, are invoked to demonstrate that pull-out is triggered and dominated primarily by the fibres' surface flaw distribution rather than by fibre strength. The origins of pull-out are also explained in terms of statistics and the identified failure pattern of fibres in composites is discussed in view of its implications to experimental practice. The implications of the findings are summarized in a current need for a deeper investigation into the micromechanics of reinforcement in composites, the role of surface flaws and the interface as well as in the competing roles of strength and flaw size.

1. INTRODUCTION

It is currently well-established that, among all types of micro-reinforcement, continuous fibres can provide composite materials with the greatest fracture resistance and damage tolerance [1]. Moreover, composites embodying this specific reinforcement, often constitute a special class of a generic composite family. One good example is Continuous Fibre-reinforced Ceramic-matrix Composites (CFCCs), a designated class of Ceramic Matrix Composites (CMCs) that is used today in high-end applications with increased demands for reliable thermomechanical performance in hostile environments over a wide range of temperatures up to 2000 °C.

The prominent mechanical performance of CFCCs and other continuous-fibre reinforced composites, as compared to single-phase systems, is attributed mainly to two important features: their increased crack growth resistance and their notch insensitivity. These unique properties provide the materials with the ability to “ignore” the presence of volume flaws and cracks by effectively redistributing and consuming energy around these imperfections thus limiting the energy supplied to the crack tip for the fatal work of crack growth. The two most powerful energy dissipation mechanisms in composites are *fibre bridging* and *pull-out*. The existence of both mechanisms is directly related to the interactions of the interface with the fibres and the surrounding matrix.

The mechanical engineer or materials scientist often tends to focalize on the measurement, importance and implications of these phenomena in composites behaviour, design and optimization procedures. While these quests are of *defacto* importance, they can at times detract interest from comprehending the origins, role and reasons of existence of the mechanisms responsible for the recorded behaviour. For example, the composites community today is still far from being able to predict *a priori* a composite's behaviour based on the properties of the matrix, the fibre and the interface. This is because the behaviour of the composite's individual phases has been generally evaluated indirectly, through the analysis of macromechanical testing data. Such analyses have definitely verified the existence and quantified the magnitude [2,3,4] of energy dissipation mechanisms but they cannot explain their origin. The number of studies focusing on direct measurements of the micromechanics of reinforcement in composite materials is very limited today [5,6,7] and definitely insufficient for a fundamental comprehension of the physics and mechanics of micro-reinforcement.

One of the most challenging questions on composites behaviour remains currently unanswered: Why do bridging fibres fail within the matrix environment to give rise to the pull-out mechanism and not within the crack flanks where fibre stress is maximum? A.G. Evans was the first to address this question in 1994 [8], based on previous observations [9,10] and described the particular effect as “unusual”. Since

that time, no evidence of a well-documented explanation to this behaviour has appeared in bibliography.

The current work aims to provide explanations to this phenomenon in terms of fracture mechanical and statistical concepts, by presenting the physical interactions occurring between the fibres and the interface at successive stages during the fracture process. Additionally, the role of various fracture mechanisms is discussed *vis-à-vis* their energy dissipation capacity. The fracture behaviour presented in the current study is typical of a composite system with a moderately strong interface and is compatible with Large Scale Bridging conditions. The behaviour was established experimentally during a recent pilot investigation at the microscale of the mechanics of fibre failure of a ceramic matrix composite using the Laser Raman Microscopy technique combined with optical microscope monitoring of the fracture sequence [7].

2. ENERGY DISSIPATION IN COMPOSITES

Fracture, in the vast majority of fibre-reinforced composites, is associated with the formation and propagation of matrix macro-cracks followed by successive fibre failure. As will be shown in the following, all stages of the fracture sequence for such materials are dominated by the performance of the interface, the nano-scale phase responsible for transferring the externally applied energy from the matrix to the reinforcing fibres. It is obvious that the properties and nature of the interface are the key parameters in the behaviour of a composite. Pull-out is one of the last mechanisms to appear in the fracture sequence of a composite, yet one of the most powerful in terms of crack growth resistance. The various energy dissipation mechanisms that develop during the fracture of a continuous-fibre reinforced composite with a moderately strong interface are presented in the following for the case of stable crack growth from an initial notch or flaw root under the application of an external tensile field.

2.1 Interfacial debonding

The application of an external tensile deformation field to a notched and otherwise stress-free continuous-fibre reinforced composite will increase the system's energy (Fig. 1a). Once the accumulated surface energy density reaches the critical value of fracture toughness, new crack surfaces will form first in the matrix (that is usually weaker than the fibres) and will propagate along the initial notch plane with further energy input. The propagating crack front will

eventually approach the vicinity of fibres where it will deflect along the interface, causing the local rupture of the complex bonds between matrix and fibre (Fig. 1b). The origin of interfacial debonding lays upon the weaker nature of the interface compared to the matrix and the fibre while the extent of debonding, usually measured by a means of a debond length, depends upon interfacial strength. Interfacial debonding is a moderate energy dissipation mechanism that restrains the work of fracture at the crack front by consuming a part of the externally applied energy. Debonding endures until the energy consumed in the specific mechanism balances the work required for further crack growth.

2.2 Crack deflection

Once the externally applied energy has caused enough damage to the interface around the fibre for the mechanism of crack growth to become energetically more preferable, the crack will propagate further in front of the debonded fibre-matrix interface along the same or different plane with respect to the initial notch. The new crack-growth plane is defined by the position along the debond length that requires the minimum amount of energy to crack; this may not always fall within the plane of the initial notch. In this case, the crack will *deflect* (Fig. 2).

2.3 Bridging & Sliding

Behind the crack front, bridging fibres stretch freely along the separating crack faces (exposed length) but also along the debond length where stretching is restricted by friction due to sliding of the fibre's surface along the debonded interface (Fig. 1c). Stretched bridging fibres carry directly the externally applied energy/load, thus significantly decreasing the energy supplied at the crack tip. In full analogy to Hooke's law, individual bridging fibres react to the externally applied load with a force of an inverse sign. It is in this sense that bridging stresses are often referred to as "crack closure stresses". Additional energy is lost as friction at the fibre/matrix interface due to the sliding mechanism. However, friction is a less powerful energy dissipation mechanism than direct load transfer on intact bridging fibres. The mechanisms of interfacial debonding, crack deflection, bridging and sliding evolve self-similarly along the material, each time the crack front encounters new fibres (Fig. 1d).

2.4 Pull-out

With further loading, the energy stored within individual stretching bridging fibres will eventually reach a critical level, sufficient to cause fibre failure.

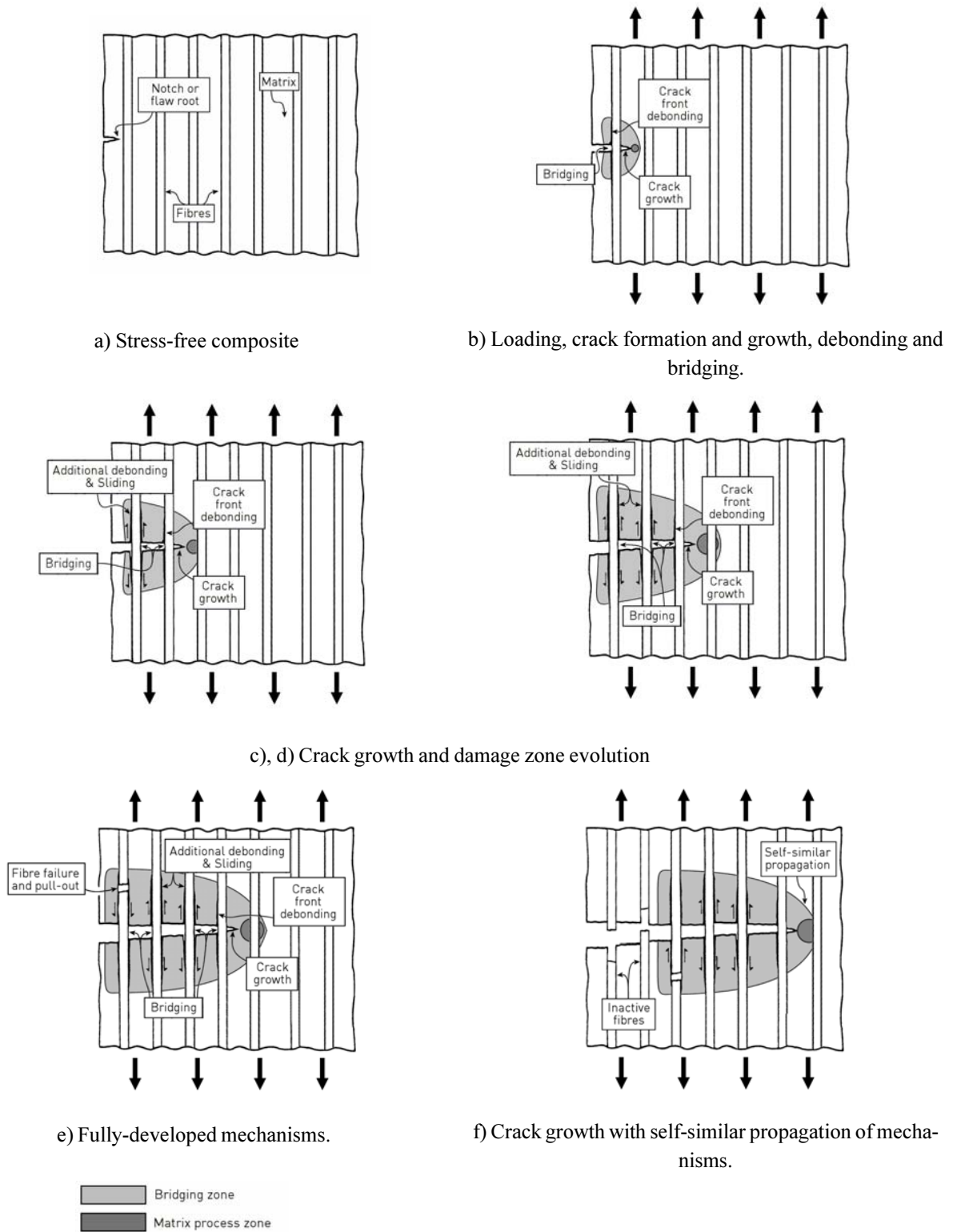


Fig. 1: Consecutive phases in the fracture sequence of a continuous-fibre reinforced composite (qualitative, not-to-scale diagram).

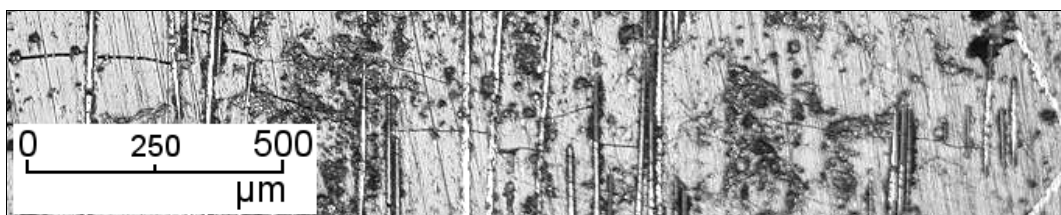


Fig. 2: Crack deflection at the interface during the initial fracture stages of a Glass-ceramic/ SiC_f composite.

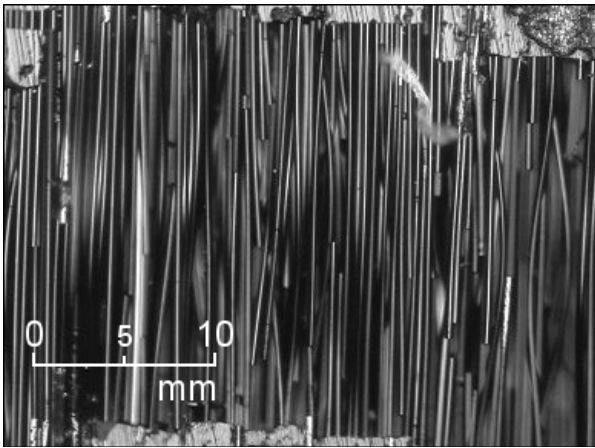


Fig. 3: Fibre bridging and pull-out of a Glass-ceramic/ SiC_f composite.

Contrary to one's expectation, it will be shown in a following section that this energy level is less than the fibre strength. If fibre failure occurs along the length exposed within the crack flanks, the fibre will remain inactive in terms of load contribution (Fig. 3). If, on the other hand, fibre failure occurs along the debond length, the fibre will *pull-out* (Fig. 1e). In the latter case, the fibre has an active contribution to the composite due to the development of shear frictional forces during sliding of the failed fibre's surface along the debonded interface. The energy consumed as friction at the interface is the contribution of the pull-out mechanism to the energy dissipation capacity of the composite. Considering the extremely high number of fibres in a composite and the corresponding vast surface area available for sliding and thus for consumption of energy as friction, it is understood that the pull-out mechanism is a powerful energy dissipation mechanism.

2.5 Damage and matrix process zones

Stress concentration at the crack tip is responsible for the development of a local *matrix process zone* where a number of micro- and nano-scale phenomena occur. Such phenomena include microcracking and crazing, rupture of atomic bonds, phase transformation, matrix shattering and debris effects. The exact dimensions and the geometry of the matrix process zone depend on the matrix properties, on the specimen geometry and are, in general, extremely hard to measure for such complex materials. The matrix process zone is qualitatively represented in Fig. 1 with dark grey colour. The area occupied by the mechanisms of interfacial debonding, crack deflection, bridging, sliding and pull-out constitutes the *bridging zone* of the material. The dimensions and shape of this zone also depend on the properties of the

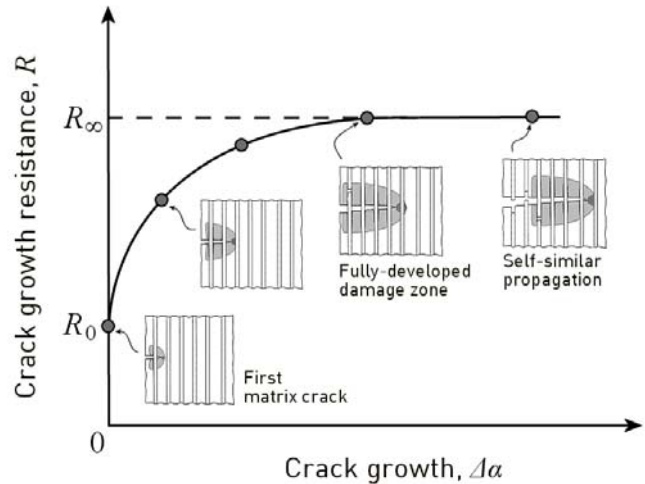


Fig. 4: Typical rising R-curve behaviour as a result of energy dissipation in composites.

composite's constituents and loading configuration. Together, the matrix process and bridging zones constitute the composite's *damage zone*, shown qualitatively in Fig. 1 with light grey colour. Depending on specimen configuration, the damage zone may span the whole width of the material, as for example in a Double-Edge Notched (DEN) sample [7], or attain steady-state shape and dimensions that propagate in a self-similar manner through the material, as for example in the Compact Tension (CT) specimen [11].

2.6 Rising behaviour of crack growth resistance

The energy dissipation mechanisms presented above are responsible for the increased crack growth resistance of fibre-reinforced composites compared to single-phase materials. It is important to note that the magnitude of these mechanisms is not constant throughout the fracture sequence but increases during the initial fracture stages from the stress-free case to the steady-state value corresponding to the self-similar propagation of the whole damage zone through the material (Fig. 1f). For example, the number of bridging fibres-and hence also of the energy dissipated through the bridging mechanism-increases during early fracture. This behaviour is known as *rising crack growth resistance* and is schematically represented by the increasing first stage of the composite's energy density versus crack-growth curve, i.e. the R-curve, Fig. 4.

2.7 Importance of the interface

As discussed above, the appearance of the pull-out mechanism is triggered by fibre failure along the debond length. Given the fact that debonding involves rupture of the interfacial bonds, it is understood that the magnitude of the pull-out mechanism depends on

the nature and properties of these bonds, in other words on the interfacial strength, that defines the extent of debonding and hence also the pull-out length: Lower interfacial strengths provide larger debond lengths and thus a higher probability for a fibre failure deeper along this direction, hence producing higher frictional forces during pull-out. The same concept applies for the sliding of intact/bridging fibres along larger debond lengths. It is unambiguously entailed that the strength of the interface is the crucial parameter controlling the energy dissipation capacity and overall performance of the composite.

The above remarks do not imply that the weaker the interface is, the greater the energy dissipation capacity of the composite will be. At the limit of a very weak interface, the amount of energy required for complete interfacial rupture will be low and the loaded composite will exhibit extensive bridging and pull-out without any significant resistance stemming from sliding of intact or failed fibres, respectively. Moreover, the steady-state value of crack growth resistance in such systems would be attained rapidly enough to accelerate fatal failure. Hence, a very weak interface limits the damage tolerance and fracture resistance of a composite material. On the other hand, a very strong interface may require such large amounts of energy for rupture of interfacial bonds for the process of crack growth to be energetically more preferable. Hence, a strong interface may lead to limited debonding and decrease the magnitude of energy dissipation mechanisms that are triggered by it. Composites with very strong interfacial bonds usually fail in a catastrophic brittle manner.

Likewise, one of the biggest challenges in the composite community today remains the development of interfacial layers with moderate or, ideally, custom strength and properties according to the desirable mechanical performance of the final material and the target application.

3. APPROACHES TO FIBRE FAILURE

Failure of bridging fibres in a composite material of a moderate interface can occur within the crack flanks, leading to a mechanically inactive fibre, or along the debond length to give rise to fibre pull-out. In practice, the two options appear not to be equally preferable: fibres tend to fail almost exclusively within the matrix environment. This behaviour has been experimentally demonstrated as extensive pull-out in various composite systems [7,12,13] and has been explicitly addressed in the past [8]. In a recent study, the experimentally recorded pull-out behaviour of a glass-

ceramic/SiC_f composite was successfully approached by a model that neglected fibre failure within the crack flanks [14]. In addition, the existence of the statistical property of *mean pull-out length*, is a standalone proof that failure is not only concentrated within the debond length but, moreover, it occurs around a fixed plane (for the same composite material). On the other hand, basic mechanics of materials evaluations dictate that fibre failure should occur along the crack flanks where fibre stress is maximum.

3.1 Mechanics of materials approach

The mechanics of materials approach to fibre failure is presented in the following example of fibre stress distribution in a composite material of a moderate interface. Fig. 5 offers a qualitative demonstration of the stresses acting upon three neighbouring fibres under a macroscopical stress of σ_0 . Two of the fibres are considered failed (denoted by numbers "1" and "3") and are undergoing pull-out, while another fibre (numbered "2") is still intact and stretching. The extent of interfacial debonding is assumed constant among all fibres with the relevant debond length being equal to l . The pull-out lengths for fibres 1 and 3, defined as the distance between the failure location and the neighbouring crack face, are given as h_1 and h_3 respectively.

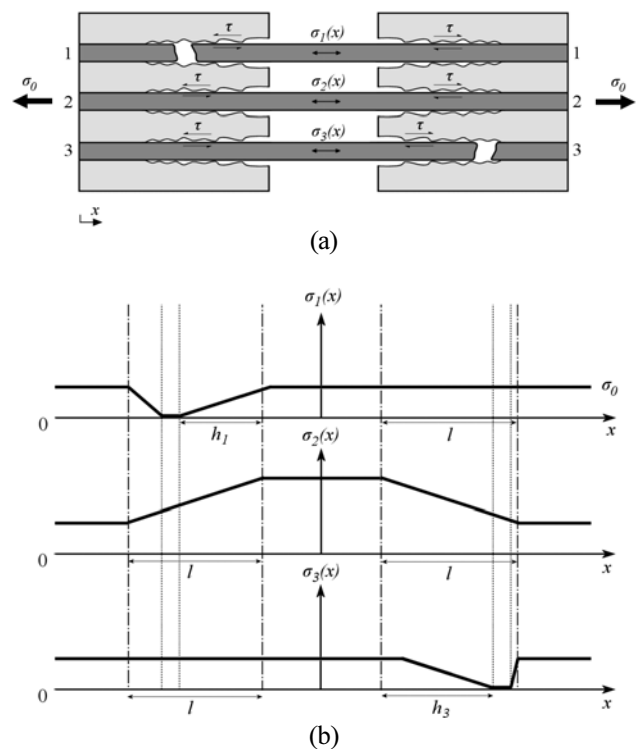


Fig. 5: a) Schematic representation of bridging by intact fibre(s) and pull-out due to failed fibres
 b) stress distribution profiles along intact and failed fibres during fracture and the effect of interfacial friction

The far-field stress on fibre "2" is equal to σ_0 , while sliding of the intact fibre's surface along the debond length gives rise to frictional shear stresses of magnitude:

$$\Delta\sigma = \frac{2l}{r}\tau \quad (1)$$

where r is the fibre radius and τ is the interfacial shear stress, assumed to be constant. A fundamental formulation, known as the "ACK model", for the calculation of interfacial shear stress in fibre-reinforced composites has been offered by J. Avenston et al in 1971 [15]. Thus, the bridging stress on fibre "2" within the crack flanks will be given by the expression:

$$\sigma_B = \sigma_0 + \frac{2\tau l}{r} \quad (2)$$

Equation (2) is a quantitative statement of the dual contribution of bridging/intact fibres to the crack growth resistance capacity of the composite; primarily by carrying a part of the applied axial load along the crack flanks (closure stress) but also by dissipating energy as friction at the interface.

The stress distribution scenario changes dramatically for failed fibres ("1" and "3"). While the far-field stress of these fibres remains equal to σ_0 , fibre failure leads to a local relaxation of the axial bridging stress, while it triggers the development of pure frictional forces along the interface and the appearance of the pull-out mechanism. The only stresses acting along the failed fibres are the axial equivalents of these shear stresses due to sliding, which increase within the pull-out length, h , with a slope proportional to the interfacial shear stress, τ , as given by Eq. 1. The effect of shear disappears at the point where the shear profile intersects the far-field profile (Fig. 5, profiles $\sigma_1(x)$ and $\sigma_3(x)$).

In view of the above arguments as well as of Fig. 5, it is evident that pull-out stresses can never exceed the magnitude of bridging stresses. This remark demonstrates further that the crack growth resistance capacity of the pull-out mechanism cannot exceed that of the bridging mechanism and rationalizes the previously stated observation that friction is a less powerful energy dissipation mechanism than direct load transfer on intact bridging fibres.

Nevertheless, the above mechanics of materials approach to failure leaves the following questions open:

1. why can fibre failure occur within the debond length, h , and not within the crack flanks where axial stress on the bridging/intact fibres is maximum?
2. at which location along the debond length, h , does failure occur?

It is the fracture mechanics approach to failure that furnishes the answers to these questions.

3.2 Fracture mechanics approach

Fracture mechanics approach failure from the point of view of critical flaw size, rather than strength. In this context, a fibre loaded to a uniform stress along its length, will fail at the location where the most critical flaw exists. Moreover, it has been shown and is invoked herein as a fundamental assumption that *fibres tend to fail at locations predetermined by their surface flaw distribution* [16]. To extend this concept, any factor affecting the fibre's surface flaw distribution can be held responsible for premature fibre failure. Mechanical friction of the fibre surface along the rough debonded interface is one such factor. The interaction of the rough microstructure of the debonded interfacial area with micro-flaws on the fibre surface during the relative sliding motion of these two surfaces affects the critical size and hence also the fatality of the flaws. One such mechanism is the tearing of interlocking blocks of fibre and debonded interface/matrix during the relative movement of the two (Fig. 6c); a process that magnifies the existing flaws at the fibre surface, therefore rendering them more critical in terms of failure. In accordance with the principles of fracture mechanics, this tampering of the fibre surface flaw distribution dominates as a fibre failure mechanism.

Based on the above concepts, it is expected that reinforcements never actually assume their ultimate strength during composite loading. This effect was demonstrated in a recent study [7] where the stress on a large number of individual bridging SiC fibres was measured for the first time in situ, during testing of glass matrix composite using Laser Raman Microscopy. The maximum value of bridging fibre stress was determined as only 70% of the nominal fibre strength. The latter property was representative of the fibres in the composite as it was determined on fibres extracted from the material. It is necessary to clarify that the above remarks do not imply that the strength of a fibre in a composite material is less than the fibre's strength in air, but that fibre failure is due to different patterns in each case.

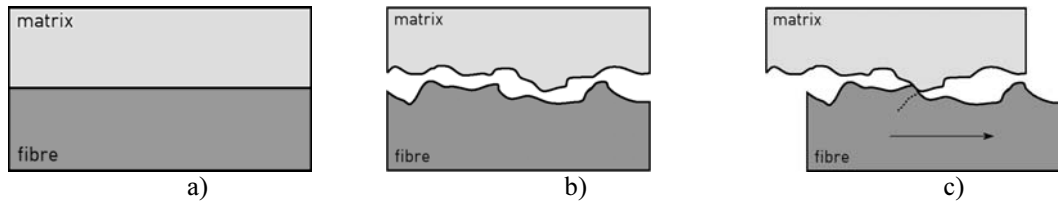


Fig. 6 : a) idealized smooth interface as predicted by the “perfect body” assumption of mechanics of materials, b) real, rough interface as assumed by the fracture mechanics approach and c) premature fibre failure due to tearing of interlocking material blocks during sliding

Another explanation to the observed behaviour is offered by Statistics. Under the assumption of a negligible effect of friction to flaw size, the relevant gauge length of a stretching bridging fibre is equal to the sum of the debond length and crack opening displacement. Due to the usually weaker nature of a moderate interface as compared to the stronger and more brittle matrix, the debonding mechanism requires less energy input than the actual cracking of the matrix. Hence, the debond length is usually greater than the instantaneous crack advance and crack opening displacement. Then, the failure location of an intact bridging fibre is much more possible to fall within the debond length than within the crack opening. The above behaviour was recently verified experimentally on a Glass-ceramic/ SiC_f composite [17] where the mean pull-out length was determined post mortem at 0.690 mm while crack opening was one order of magnitude less, approximately 0.05 mm, on the onset of fibre failure.

4. IMPLICATIONS

If the fibre failure pattern in composites is dominated by flaw size rather than strength, a need for refining the currently available fibre failure assessment methodologies appears necessary. Today, composite materials' design and performance prediction is based on more or less sophisticated “rules of mixture” of the constituents' properties. Such properties are usually the result of mechanical testing characterization experiments. For reinforcements, the widely accepted single-fibre test in air and the emerging bundle test are almost the exclusive methods for mechanical properties characterization. Both techniques enjoy advantages and disadvantages, however both respect the target material as a macroscopic medium, neglecting surface or volume micro-irregularities that appear to play a more important role in failure than was believed at the time of development of these methodologies. Moreover, the specific effect of micro-interactions of fibre surface flaws with the matrix and interface remains to be investigated separately.

Furthermore, it should not be overlooked that fibres in a composite material are expected to have undergone extensive handling during composite fabrication and to have received severe thermal treatments during composite processing/curing. Such processes affect the integrity, crystallinity and load bearing capacity of the reinforcements and hence of the overall structure. In order to accurately simulate a fibre's behaviour in a composite, the established mechanical properties should correspond to the identical physical and microstructural state as the target fibre. It appears then reasonable that mechanical testing for characterization of properties should be performed on fibres having already received the exact thermal, or any other treatment as the fibres in the composite.

5. CONCLUSIONS

An argumentation based primarily on fracture mechanics and secondarily on Statistics and on Physics has been presented to provide the fundamentals of the pull-out mechanism in fibre-reinforced composites. The argumentation explains the phenomenon of fibres failing within the matrix environment to give rise to the pull-out mechanism and not on the plane of the matrix crack, despite the fact that fibre stress is maximum at that site.

In particular, it has been demonstrated that:

1. The fibre failure pattern in composites is a fracture mechanics- rather than a mechanics of materials- dominated mechanism wherein strength plays a less critical role with respect to failure than flaw size.
2. Interactions of the fibre surface with the debonded interface -due to relative sliding of the two surfaces during stretching of intact fibres- affect the fibre surface flaw distribution and are responsible for premature fibre failure within the matrix environment.
3. In composites of moderate interfaces, the usually larger debond length as compared to matrix crack opening displacement, increases the probability of fibre failure occurring within the matrix environment

rather than within the crack flanks.

4. The energy dissipation capacity of the pull-out mechanism cannot exceed that of the bridging mechanism.
5. For the better understanding of the composite failure process it is essential that the role of fibre surface flaws is examined and quantified.

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