1 General Introduction

The determination of the strength and the failure behaviour of fiber reinforced composites is a complex task. This is mainly due to the anisotropy and the inhomogeneity of the material on the microscale. The strength or the fracture toughness of the interface between fiber and matrix is the key factor controlling the failure process on microscale, especially under loads transverse to the fiber axes. The ultimate failure of a laminate is a complex process consisting of a large number of different subprocesses. Due to its highly dynamic nature on one hand and the microscopical scale on the other it is not possible to observe the ultimate failure directly or even to analyse it theoretically. However, long before ultimate failure occurs a large number of elementary failure processes take place in a rather static manner or at least at low crack propagation speed.

2 Failure of 0°/90° cross ply laminates

2.1 General aspects of the failure of plies under transverse loading

The strain distribution in a ply loaded transverse to the fiber direction is very inhomogeneous due to the fact that - up to a certain extent - fiber and matrix are connected in series. Since the fibers undergo only small strains due to their high stiffness and, as a result, their contribution to the total deformation in loading direction is very small the matrix has to contribute the by far largest part of the deformation. Accordingly the strains and, as a consequence, the stresses in the matrix are very high. This leads to early failure of the transverse loaded plies, the so called "first ply failure" by cracks perpendicular to the loading direction. The origination of these cracks, mainly caused by debonding of individual fibers within the 90° ply, is analysed later.

2.2 Failure of 0°/90° CFRP laminates

First, the cracks and fracture surfaces of a 0°/90° cross ply laminate are studied on the example of a carbon fiber/epoxy laminate with a 45% fiber volume fraction (figs. 1-4). The micrographs are made with a confocal laser scanning microscope. During the growth of the external load a large number of cracks develop in the 90° ply, almost equally spaced (fig. 1a). A closer look into the crack paths shows that the cracks primarily propagate in the interface between fiber and matrix (fig. 1b). The matrix is crossed by the cracks merely in order to continue the propagation in the interface of the neighbouring fiber. Even though the cracks in the 90° ply are - in average - rather straight and mainly oriented perpendicular to the loading direction, they may be slightly deviate or they may jump between different "layers" of fibers depending on the local geometry and the respective strength of matrix and interface (figs. 2a-2c). In the vicinity of the interface between the 0° and the 90° ply the delamination predominantly takes place along the fibers in the 0° ply (fig. 1b). However, also cracking of the 90° ply is encountered. In this case shear failure transverse to the fibers occurs (see Sect. 2.3).

2.3 Fracture surfaces of a 90° ply

There is a great difference between fracture surfaces formed under pure or at least prevailing mode I stresses and those formed under mixed mode stresses implying a remarkable mode II part [1]. The matrix breaking under prevailing mode I conditions in general shows rather smooth fracture surfaces with moderate variations in height. In the example given here the crack jumps by about 20μm from fibers at a lower level (fig. 2a, bottom) after crossing...
a matrix zone to fibers at a higher level (fig. 2a, top). The jump exhibits a very steep slope (figs. 2b, 2c). The profile of the fracture surface of the matrix zone, indicated by the yellow arrow, is rather smooth (fig. 2c). One should keep in mind that the length scale and the height scale of the profile are different.

In case of failure under dominating mode II stresses the fracture surfaces are ragged showing significant shear cusps in zones where the distances between the fibers are large (fig. 3a, left and right zones). When the fibers are very close to each other the matrix rather breaks with a flat fracture surface (fig. 3a, center). Obviously the distance between the fibers has a significant influence on the fracture of the matrix. Due to the high stiffness of the fibers the deformation of the matrix is constrained. The profile (fig. 3b) reveals the roughness of the fracture surface. In the debonded zones almost no matrix is left on the fiber surface. This means that the adhesion to the fiber surface is lower than the cohesion within the matrix. Even though this seems to be the most likely material behaviour, it is known, especially in case of thermoplastic matrices, that an interphase may form around the fiber possessing a higher fracture toughness than the bulk matrix. In this case the fibers are covered by a thin matrix film after fracture. However, this is not typical in case of epoxy resins. As an extreme example the fracture surface of a matrix rich region is shown in fig. 4 exhibiting a large number of shear cusps. All fibers are torn out, their former positions can be identified by the canals. Only a single fragment of a fiber is left (upper left corner).

The fracture surfaces reveal that the failure of a composite ply depends on several parameters. Even though the onset of failure cannot be observed directly inside a loaded specimen it is most likely that the debonding of individual fibers is the initiator of the failure process. Accordingly, the analysis of interfacial debonding is the key to understand the failure of plies under transverse stresses.

The fiber-matrix interface was extensively investigated in regard to develop test methods for measuring the interface strength. Micromechanical tests such as fragmentation, pull-out, microdroplet and push-out tests were developed. A comprehensive overview is given by Kim and Mai [2]. However, in the respective tests shear stresses parallel to the fiber axis are the dominating loading of the interface. The failure of a composite ply though is initiated by stresses transverse to the fiber direction. The failure under transverse stresses was investigated, among others, by Varna and Paris [3,4] starting in the midst of the 1990th. Initiated by their work a strong activity took place in this field and several authors dealt with this topic. Fracture mechanical analyses of the interfacial crack propagation were performed and the mixed mode energy release rate was calculated either by means of the boundary element method or by the finite element method [5-9]. In addition the debonding processes was studied experimentally, e.g. on single fiber specimens [10-16].

3 Mode ratio of energy release rate depending on length of crack increment

3.1 Differences between cracks in uniform materials and in bimaterial interfaces

It is well known that fundamental differences exist in linear elastic fracture mechanics between cracks inside a uniform material on one hand and cracks in interfaces between dissimilar materials on the other hand. It is needless to mention that in uniform materials the ratio of the mode I and mode II energy release rates depends on the geometry of the specimen and the external load. In particular, cracks can propagate under pure mode I conditions. In contrast to that interface cracks between dissimilar materials produce a mode II energy release rate in any case, independent of geometry and external loading. While in case of a uniform material the ligament of the crack may remain straight e.g. if it is a symmetry line, the interface between two different materials necessarily has to bend, because it never can be a symmetry line. Normal stresses are acting parallel to the interface in the vicinity of the interface. These are, for example, due to the different lateral contractions of the respective materials. As a result, shear stresses occur in the zone around the interface and in the interface itself. The normal stresses acting tangential to the interface cause a contraction or elongation, respectively, parallel to the interfaces when the crack forms allowing the different materials to deform without
restraining each other. The deformation gives rise to a mode II energy release rate.

3.2 Analytical treatment of interface cracks

In uniform materials an analytical solution exist for the standard configuration, this is, a straight crack in an infinite plate (Griffith crack). The respective bimaterial case is much more complicated. In case of a straight interface crack several attempts were made to find an analytical solution. Williams was the first who solved the problem of an interface crack by applying his solution developed for the bimaterial wedge problem [17]. He found the classical $1/\sqrt{r}$ singularity also to be valid for interface cracks, however, a second term appears which is the sine of the logarithm of the distance to the crack tip [18]. This results in an oscillating singularity in the vicinity of the crack tip. This type of solution causes an interpenetration of the crack faces, indeed restricted to a very small zone next to the crack tip. However, the interpenetration violates a fundamental presumption, this is that at a specific geometrical point, only one material point can be present at a time. As a result, the solution must be rejected from a pure mathematical point of view. In order to rescue the solution for the physical problem it is argued by several authors that the zone of interpenetration is so small that it has no meaning in a real interface crack problem because a crack in a real material never can be that sharp that the zone of interpenetration could exist. From an engineering point of view this may be acceptable, from a mathematical it is not. Following Williams solution some further attempts were made, e.g. by England [19] and Rice [20,21], however, all of them could not get rid of the problem of an oscillating singularity.

3.3 Analytical contact zone models

Based on the idea of Malyshev and Salganik [22], Comninou developed a solution allowing a small contact zone near the crack tip [23]. With this assumption the problem of the oscillating singularity is overcome. The solution was later extended by Atkinson [24] and Gautoesen and Dundurs [25]. The length of the contact zone depends on the elastic parameters of the two materials. In all technical relevant material combinations the zone is so small that again the question arises whether it is of physical relevance. A critical review about the work done so far in the field of interface cracks was given by Comninou [26].

While for a crack in a uniform material the energy release rate can be decoupled into a mode I and mode II part, or equivalently, two independent stress intensity factors can be identified this is not possible in case of interface cracks. Here the stress intensity factors are coupled. Concerning the energy release rate this leads to the somewhat paradox consequence that - in a numerical solution - the ratio of the mode I and mode II energy release rate is depending on the length of the crack increment. This was studied by Sun and Jih [27] by a finite element analysis of an interface crack in a large bimaterial plate. They show that the mode I vs mode II ratio decreases with decreasing ratio of the length of the crack increment vs the total length of the crack. They conclude that if the crack increment tends to 0 the mode ratio tends to 1. However, finite element analyses performed by the authors show that the mode ratio can even become smaller than 1 and there is no indication that the decrease of the mode I part ever will stop (to be published). From the results obtained so far it is to be expected that in the limit of a vanishing crack increment length the mode I energy release rate even completely vanishes. This would be in accordance with the contact zone solution by Comninou. The problem of the influence of the crack increment on the mode ratio and on the stress intensity factors was also studied by Beckert and Lauke [28] as well as by Mantic and Paris [29].

The question arises if these mathematical phenomena are relevant for a physical problem, e.g. for a circumferential crack around a single fiber under transverse loading. On one hand there is a physical limitation of the crack increment length given by the structure of the material (at least the distance between neighbouring atoms) and by the limitations of continuum mechanics. On the other hand a stable crack not necessarily propagates fully continuous but, looked on a microscale, rather goes step by step with phases of standstill in between. So the question if there exists a material specific crack increment, which is proposed by some authors, is still unanswered. However, we have to realise that the partitioning of the energy release rate in case of interface cracks is of limited significance.
4 Interfacial crack propagation in a unidirectional ply under transverse tension

4.1 Finite element model

The debonding of a fiber inside a lamina is investigated by analysing the energy release rate of an interface crack propagating circumferentially around the fiber. The unidirectional composite ply is idealised by a representative multi fiber volume element comprised of a 12-fiber regular hexagonal array [10,11]. Linear elastic behaviour of both constituents is presumed. Plane strain conditions are applied because they are prevailing inside a composite material. Except of the free surfaces, where plane stress conditions are valid and the adjacent zone plane strain conditions are in good agreement with the conditions encountered in a 3D model. For the simulation of the crack propagation fixed loads conditions are applied.

4.2 Modified virtual crack closure method

The method used for the calculation of the energy release rate is a slightly modified virtual crack closure method (VCC). In the classical VCC method the work necessary to (virtually) close the crack the displacements of the preceding increment are taken while the nodal forces are taken at the nodes which will open in the next step. That is, forces and displacements belong to different time steps. In general, if a sufficiently fine increment is used the discrepancies of the classical procedure are tolerable. If, however, the kinking of a crack is to be analysed the classical method fails in the moment of kinking. Because the crack abruptly changes its direction, the forces and the displacements do not fit. In order to avoid this discrepancy the method was modified in that forces and displacements of the same nodes were taken, this is, the forces before the crack extends by the next increment and the displacements developed during the respective crack increment. Obviously, forces and displacements belong to the same time step.

4.3 Influence of the fiber volume fraction on the energy release rate

A central question in analysing the debonding of a single fiber is how it is influenced by the fiber volume fraction. This is studied by varying the fiber volume fraction between 5% and 85%. In case of a low fiber content the interaction between the fibers is very small. Accordingly, the results are very similar to the case of a single fiber.

4.4 Mechanical properties of fiber and matrix

As very common in the analysis of fiber reinforced materials linear elastic material behaviour is presumed. In addition, a geometrical linear theory is applied, this is, small deformations are presumed. While the presumption of linear elasticity is in good agreement with the behaviour of the fibers, it is a rather strong simplification of the real behaviour of epoxy resins. At strains higher than 2% epoxy resins show a significant nonlinearity which is partly due to a nonlinear elastic behaviour and even more caused by a viscoelastic behaviour. As a result all the analyses based on linear elasticity should be regarded as first approaches in understanding the failure of fiber reinforced composites on the microscale. Still they give some valuable insight into the dominating elementary failure processes.

4.5 Initiation of the interface crack

Before the interfacial crack propagation can be analysed the moment of crack initiation has to be identified. This can be done in two different ways. The first way is to presume a small starting crack, e.g. a microdefect in the interface, and the external load is increased until the critical energy release rate of the fiber/matrix interface is reached. The second way, following Leguillon [30], is to predict the onset of debonding by a stress criterion. This procedure is chosen in the present paper by using a simple maximum stress criterion for the radial stresses. Since the resistance of the fiber-matrix interface against radial stresses is significantly lower than against shear stresses the failure is governed by the radial stresses. The highest radial stresses - presuming a regular hexagonal distribution of the fibers - develop in the central zone of the fiber surface, this is, where the fiber surface is perpendicular to the loading direction.

The maximum tensile stresses do not appear exactly at the interface but in a small distance of some percent of the fiber diameter apart from the interface. With respect to the fact that the interface between fiber and matrix usually is weaker than the matrix itself it is most likely that the failure starts in the interface. This is confirmed by experimental
observations of fracture surfaces where no matrix is found on the fibers in the respective region.

The stress concentration in the center zone near the interface is caused by the high stiffness of the fibers compared with the matrix. With decreasing distance of the fibers, the stress concentration strongly grows. As a result, the debonding of individual fibers starts at rather low external loads in zones with small distances between the fibers. Since the fiber distribution in a real composite is not regular, even in case of moderate fiber volume fraction zones exist where the distances between the neighbouring fibers are very small. For the study of the influence of the fiber volume fraction on the energy release rate the crack initiation is assumed to occur at the same load level, this is at the same magnitude of the radial stresses in the interface (maximum stress criterion) in all cases under consideration.

4.6 Energy release rate for interface cracks at various fiber volume fractions

First we focus on a low fiber volume fraction, this is, 5% (fig. 6a, green line). This is very close to a single fiber embedded in a pure matrix. The interface crack is initiated by pure radial stresses in the center zone, this is, at 0°. The crack is assumed to propagate simultaneously clockwise and counterclockwise. The abscissa of following figures represents the total crack angle. This is, a crack angle of 120° corresponds to a crack whose tips are positioned at +60° and -60°, respectively. Directly after crack initiation the energy release rate grows very fast reaching a maximum at about 120°. Then it decreases in a similar manner vanishing at 300°. The fast increase of the energy release rate in the first phase indicates that the crack propagation is very unstable.

Experiments on single fibers loaded transverse to their axis show that the crack indeed propagates unstable. However at an angle of about 120° the crack stops. At first glance this would perfectly fit with the Griffith crack criterion, because after the maximum the energy release rate decreases, indicating stable crack propagation. However, during the unstable phase of crack propagation a large portion of surplus of energy was released, mainly in form of mechanical waves which is not covered by the Griffith criterion. These waves on the other hand can lead to crack propagation even though the energy release rate is decreasing. These findings show that the phenomenon of interfacial crack propagation cannot fully covered by linear elastic fracture mechanics under static conditions.

The total energy release rate for different fiber volume fractions, normalised with respect to the maximum value, is shown in fig. 6a. With increasing fiber volume fraction the slope of the energy release rate becomes steeper and steeper and the maximum shifts to lower crack angles. In case of high fiber volume fractions (70%, 85%) the slope of the energy release rate becomes very high indicating that the debonding process becomes highly dynamic.

The picture looks completely different if the absolute values of the energy release rate are observed (fig. 6b). The absolute values are strongly decreasing with increasing fiber volume fraction. This is one hand because the debonding takes place at lower external loads as a result of the higher stress concentrations. On the other hand it is due to the fact that the matrix volume which releases the main part of the energy becomes very small in case of high fiber volume fractions. Both effects in sum lead to the dramatic decrease of the energy release rate at high fiber volume fractions.

5 Interface cracks kinking into the matrix

Experimental observations of cracks in laminates show that, at a certain crack length, the interface crack kinks into the matrix and jumps to the next neighbouring fiber. The specific position of the crack tips in the moment of kinking depends on the geometrical configuration, primarily on the distance to the neighbouring fiber.

5.1 Interface crack kinking at 60°

As first example an interface crack in a 30% fiber volume fraction composite is studied. The crack is assumed to propagate to the next neighbouring fiber at the shortest distance between the fibers, this is, at an circumferential angle of 60°. In the model chosen the crack propagates circumferentially in the interface in both directions in the first phase as analysed in the preceding example. When the crack has extended by 60° in clockwise as well as in counterclockwise direction the interface crack stops.
At the upper crack tip positioned at 60° the crack kinks into the matrix perpendicular to the interface (Fig. 7). The crack then propagates on a straight line towards the neighbouring fiber. The lower crack tip is fixed during matrix cracking.

The analysis of the radial and tangential stresses in the vicinity of the crack tip, this is, outside the singularity dominated zone, shows that at an angle of 60° the radial stresses (normal to the interface) are very low compared with the stresses tangential to the interface. As a result the crack does not open in case of further propagation along the interface. Even though it is not clear up to now if the splitting of the energy release rate into mode I and mode II is of any physical relevance in case of interface cracks, the two parts are given here as dotted lines. The crack increment is identical with the one used for the kinked crack in the matrix.

Due to the kinking into the matrix the crack tip comes under strong stresses perpendicular to the crack line, that is, the crack opens and a large mode I energy release rate develops (fig. 8). The mode II energy release rate almost vanishes in the first phase of the kinked crack. The total energy release rate of the kinking crack is much higher than for the interface crack.

During further propagation the mode I part significantly decreases while the mode II part slightly increases. The total energy release rate declines which is due to a stress redistribution to the neighbouring zones as a result of the crack. That is, the zone ahead of the crack becomes more and more unloaded.

5.2 Interface cracks kinking into the matrix at various positions

The crack paths of a 90° ply inside a cross ply laminate observed in experiments show a variety of different angles where the crack kinks into the matrix and propagates to the neighbouring fibers. The angles where kinking occurs range from about 30° up to almost 90° in cases where the fibers are very close to each other (fig. 1b). Angles smaller than 30° are not found in the transverse loaded ply.

The total energy release rate for cracks kinking at angles between 10° and 90° is shown in fig. 9 for a fiber volume fraction of 30%. In case of a crack kinking at 10° into the matrix the energy release rate exhibits a small peak directly after kinking and decreases with further extension of the crack. Except of the small peak the energy release rate is lower compared with a crack propagating in the interface. The question is if the peak is of physical relevance. In this respect we have to keep in mind that the kinking is a process which cannot be directly observed. Presumably the kinking will not be a stable process but rather be highly dynamic as seen in the first phase of the interface crack. To answer this question a detailed study of the moment of kinking would be necessary.

When the crack kinks at 20° the energy release rate is higher not only in the moment of kinking but also during following the phase. The energy release rate is of the same magnitude as in case of a crack further propagating along the interface. For cracks kinking at larger angels (30° to 90°) the increase of energy release rate becomes very high and remains on a high level during further matrix cracking.

The question is at which position the crack would kink into the matrix in a real composite. Looking solely at the energy release rate the decision is clear, the crack would kink into the matrix if the energy release rate developing here is higher than in the interface. However, the fracture toughness of the interface usually is lower than that of the pure matrix. Accordingly, the position of kinking depends on the ratio of the fracture toughness of the fiber/matrix interface and the matrix itself. In addition, the dynamics of the crack propagation as well as nonlinear material behaviour influences the process.

References


Table 1. Elastic constants of glass fiber and epoxy matrix

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<tr>
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Fig. 1b. Cross-ply laminate under uniaxial tension - crack formation in a 90° ply, detail

Fig. 2a. Cross-ply laminate under transverse tension - fracture surface in a 90° ply

Fig. 2b. Cross-ply laminate under transverse tension - fracture surface in a 90° ply, topographical view

Fig. 2c. Cross-ply laminate under transverse tension - fracture surface in a 90° ply, height profile

Fig. 3a Cross-ply laminate under mixed mode stresses under tension perpendicular to plane and shear transverse to fiber axes

Fig. 3b Cross-ply laminate under mixed mode stresses - tension perpendicular to plane, shear transverse to fiber axes, profile of transverse cut (red line in fig. 3a)
Fig. 4. Cross-ply laminate under mixed mode stresses - tension perpendicular to plane, shear transverse to fiber axes, fracture surface in a 90° ply

Fig. 5. Finite element mesh of a representative volume element of a hexagonal fiber array

Fig. 6a. Total energy release rate during interfacial crack propagation, variation of fiber volume fraction, normalised to maximum total energy release rate

Fig. 6b. Total energy release rate during interfacial crack propagation, variation of fiber volume fraction, normalised to maximum stress

Fig. 7. Crack kinking from interface into matrix at 60°, transverse normal stress

Fig. 8. Mixed mode energy release rate for crack kinking from interface crack into matrix at 60°
Fig. 9. Total energy release rate for crack kinking from interface into matrix at various circumferential positions