THE INFLUENCE OF LOCAL DEFECTS AND NONLINEAR MATRIX BEHAVIOUR ON THE FAILURE PROCESS OF FIBER REINFORCED COMPOSITES

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Abstract
The premise for the development of improved failure criteria for fiber reinforced composites is the understanding of the failure processes on microscale. On microscale the local behaviour of the matrix polymer is crucial. It significantly differs from that found in macroscopic measurements. The present study is focused on the plastic material behaviour of a standard epoxy resin. Local plastic strains are measured in tensile test by using digital image correlation. A multilinear plastic law is implemented by the experimental data. The stress field arising in an idealised representative volume element of a unidirectional ply is analysed by finite element calculations. The change of the stress and strain field due to plastic deformation according to the multilinear plastic law is shown. The change leads to a shift of the critical locations where failure is likely to initiate. In addition, the failure type changes from stress based to strain based failure. The dominant role of the shear strains is shown.

1. Introduction

It's a matter of fact that failure of fiber reinforced composites is initiated on microscale. It is originated at microscopical imperfections in the matrix or in the interface between fiber and matrix. Since the failure process is very complex it is not possible to predict the whole process from initiation to ultimate failure. It is not even possible to observe the process because local failure may develop at several locations simultaneously and only one specific becomes critical. In addition, the ultimate failure is a high speed process taking place inside the material. There are no methods available to visualise high speed processes developing inside the material. So we are left with the final fracture surfaces. To conclude the temporal course of the failure process from the fracture surface is possible only up to a very limited extent. Due to the statistical manner of the occurrence and the size of imperfections and due to the significantly differing material behaviour of resins on macro and on microscale, which cannot directly be measured, failure criteria in principle are restricted to a certain level of simplification. However, a deeper understanding of the processes and an improved description of the material behaviour on microscale should lead to improved failure criteria which are closer to the the actual mechanisms. The aim of the study is to reveal some effects of the nonlinear behaviour of the matrix. This study, however, is still based on a high level of idealisation. The inspection of fracture surfaces of composites reveals how complex the fracture process indeed is and, in turn, how high the level of idealisation of any model necessarily must be. In addition it shows that even a standard epoxy resin can undergo large deformations at room temperature (Fig. 1). It is still an open question if it ever will be possible to predict failure on microscale with sufficient reliability.

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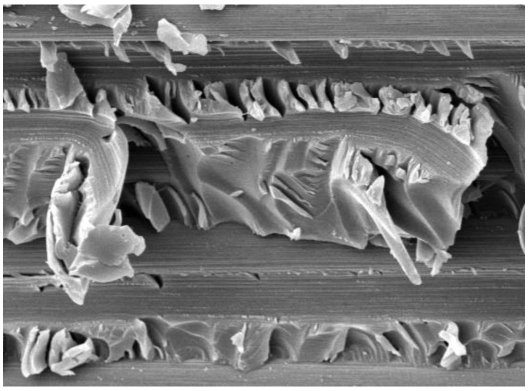

Figure 1. Fracture surface of a carbon/epoxy composite under mixed mode loading; imprints of fibers reveal locally high plastic deformations (upper fiber imprint)

2. Failure mechanisms in composites

2.1 Predefined failure planes

To predict failure of fiber reinforced composites commonly standard failure criteria are used taking into account the interaction of multiaxial stresses [1-5]. These criteria are based on models simplified concerning the material behaviour and the failure mechanisms. In contrast to homogeneous materials where the fracture plane is not determined a priori in composite materials failure surfaces are predefined by the inner structure of the material. In many homogeneous materials failure occurs under maximum normal tensile stresses. This is, the orientation of the fracture plane is defined by the stress state, the failure itself occurs under uniaxial stresses. If, on the other hand, the fracture surface is predefined by the structure of the material, failure in general occurs under a multiaxial stress state. The stress components at the location of failure initiation usually cannot be determined theoretically because the location is not known. This is a fundamental difference to homogeneous materials.

In a composite material two types of predefined failure surfaces are present. One is a plane parallel to the fiber direction where pure inter fiber failure occurs. This type of failure can be observed for example in off axis tests when the specimens are long enough to allow failure without fiber breakage. A second type of failure appears in laminates, namely inter ply failure. In this case the fracture plane again is predefined, this is, parallel to the plies. Failure takes place under a multiaxial stress state on those fracture planes. The third type of failure is fiber failure which happens if the deviation of the loading axis from the fiber direction is sufficiently small. Fiber failure in general is accompanied by inter fiber failure. Presumingly fiber failure is the initial mechanism and inter fiber failure is caused by non simultaneous fiber breakage. However, from the final fracture pattern this cannot be clearly decided.

In the preceding remarks the fracture surfaces were regarded as planes. This, however, is valid only from a mesoscopical point of view. If we look at inter ply failure, for example, the general orientation of the fracture surface is defined – parallel to the plies. On a microscopical scale the fracture surface is not a plane but decomposes into a large number of small facets with different orientations in the range of some microns. The fracture surface even jumps between the fibers of the two plies and also intra ply failure may occur. Accordingly, if we talk about predefined fracture surfaces we look from a mesoscale level. On microscale, this is, on the scale of the facets, the fracture surface is not predefined but is determined by the local stress state and by microdefects in the polymer or the interface.

Since it is not possible to observe the failure process on microscale directly, this is, the temporary development of the individual microcracks, but the final fracture surface is got as the only result, the

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stress state being present in the moment of failure of every facet cannot be defined. As a result, we cannot say anything about the strength of the material on microscale subject to multiaxial stresses. Since the global failure of a composite is the sum of processes taking place on microscale it is necessary to analyse these processes, especially their temporal succession. This is a precondition for the development of real mechanisms based failure criteria.

2.2 Limits of continuum mechanics

There is another essential factor concerning failure on microscale. The polymer matrix in a fiber reinforced material usually is regarded as homogeneous. The validity of continuum mechanics is presumed. This means that no matter on what scale the material is observed the material possesses identical properties. Accordingly, if the material properties are determined by common macroscopical tests, these material parameters are assumed to be valid on any scale. However, the stresses that the material is able to sustain locally, e.g. at a crack tip, are much higher than what is measured in macroscopical tests. We know from classical fracture mechanics that materials behave different on microscale, e.g. at crack tips. Fracture mechanics avoids to analyse stresses at crack tips because this procedure necessarily had to fail even if a finite radius at the crack tip would be assumed, as proposed by Neuber [6]. The assumption of a finite radius bypasses the mathematical problem of unbounded stresses at the crack tip but not the physical problem of the different material behaviour on microscale. The use of the fracture toughness as a key parameter in fracture mechanics accommodates the fact that we are beyond the limits of continuum mechanics at a crack tip.

A similar phenomenon is encountered around microdefects. If a failure analysis at a microdefect, for example at a lack of bond between fiber and matrix, is performed failure is predicted at a much lower external load than found in reality. Since we have no chance to measure the material behaviour on microscale, not even with advanced methods like nanoindentation, we have to restrict ourselves to a coarser scale and assume continuum mechanics to be valid. However, even with these restrictions mentioned above the failure analysis of composite materials can substantially be improved by taking into account material behaviour which is measured on mesoscale. The nonlinear, mainly plastic behaviour of the polymer matrix changes the stress field on local scale significantly and, as a consequence, it changes the failure behaviour.

3. Nonlinear behaviour of polymeric matrices

3.1 Influence of nonlinear material behavior on the stress field in fiber reinforced composites

Failure of laminates starts with the failure of a plie which is under high stresses perpendicular to the fiber axes. The understanding of the process of first ply failure is one premise to understand the failure of composites. A number of studies was performed dealing with this topic. The debonding of fibers under transverse loading was studied thoroughly e. g. by Paris and the elasticity group in Seville [7-9] and, among others, by the authors [10]. Interface cracks propagating circumferentially around single fibers are investigated by calculation of the energy release rate with the help of the boundary element or the finite element method. These studies are based on linear elastic fracture mechanics. In the present paper a first step in improving the description of the stress field around the fibers is made by taking into account the matrix plasticity by measuring the material parameters on mesoscale.

The description of the mechanical behaviour of the respective epoxy resin still is highly idealised. First, the time dependence of the resin is not taken into account. Second, the material is assumed to be perfectly homogeneous and also perfectly bonded to the fibers. Third, the validity of a multilinear plastic law is presumed for the material. The claim of the present study is not to develop a new general plastic constitutive equation for epoxy resins but to estimate the influence of plastic behaviour of the matrix allowing large deformations and, in turn, a redistribution of the stresses.

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3.2 Plastic behaviour of epoxy resins

Epoxy resins exhibit a distinct plastic behaviour. In addition they are highly time dependent. This means that the determination of the elastic and plastic material parameters, even though they are independent of time, requires time consuming creep and relaxation tests [11]. The typical values of strain to failure found in literature range from four to eight percent. These values presumingly are not taking into account the local nature of plasticising. Tensile tests show that at a certain load a plastic deformation develops in a small zone. These deformations can become very large leading to strains up 50%. The contribution of these deformations to the total change in length of the specimen, however, is small due to the short length of the zone. It must be doubted that all strain to failure values given in literature are indeed true local strains but rather averages over the specimen length.

The resin used for this study is a L135/H137 epoxy system manufactured by Hexion. Common dogbone tensile specimens of 10mm in width and 4mm in thickness were used. The specimens were cut out of plates by a milling process. The tests are performed with a loading speed of 1mm/min. In order to determine the plastic behaviour of the material the local strain, i. e. the strain in the respective plastic zone has to be measured. To this end, a strain field measurement system is used (Aramis). The strains measured are logarithmic strains (Hencky), the stresses are the true stresses (Cauchy). Two typical stress strain curves are given in Fig. 2a. The blue curve represents the strain in the zone of high plastic deformations while the red curve is taken outside this zone where the material behaves predominantly viscoelastic. For the finite element analysis a multilinear plastic law is used [12]. The material parameters are determined from the strains in the plastic zone (blue curve). The multilinear plastic stress/strain curve used in the analysis is given in Fig. 2b. One should keep in mind that in the present study the instantaneous stresses and strains are taken for the determination of the material parameters. This is, the viscoelastic part of the strains is not separated. The separation of the viscoelastic strains is left to a further study.

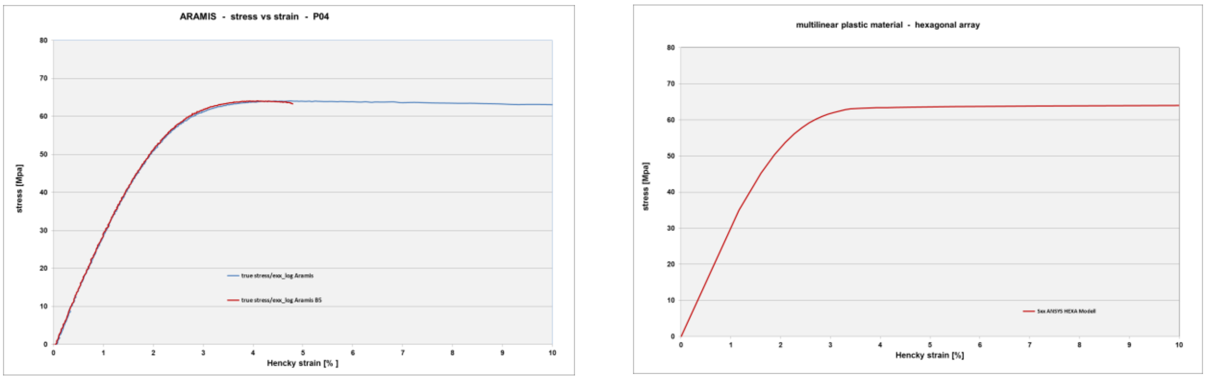

Figure 2. **a)** Stress/strain curves; blue: plastic zone, red: elastic zone **b)** stress/ strain curve of multilinear plastic law

4. Analysis of the stress and strain field in a unidirectional ply under transverse loading

4. 1 Finite element model of the composite ply

The stress field developing in the matrix between the fibers in a unidirectional ply is analysed by using a representative volume element. The model consists of a regular hexagonal 12-fiber array representing a fiber volume fraction of 50%. On the edges the respective symmetry boundary conditions are prescribed. The load is applied by prescribing displacements in x-direction on the right edge of the model while the left edge is fixed in x-direction. The interface between fiber and matrix is assumed to be perfect. The composite is comprised of glass fibers embedded in an epoxy matrix. Two

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types of matrix material behaviour are analysed. Linear elastic behaviour is compared with multiaxial plastic behaviour. In case of plasticity large deformations are allowed. The fibers are regarded as linear elastic materials. Plane strain conditions are applied because they represent the appropriate approximation of the stress/strain state within the composite away from the edges. The elastic properties of fiber and matrix are given in Table 1.

Material	E-Modulus [MPa]	Poisson ratio
fiber	72000	0.21
matrix	3000	0.35

Table 1. Elastic constants of glass fiber and epoxy matrix

4.1 Normal strains in elastic and plastic material model

First, the strains in x-direction, ε_{xx}, are studied (Hencky strains). In case of elastic material behaviour the strain maximum arises in the center between the fibers (Fig. 3a), this is, where the fibers have the smallest distance in x-direction from each other. The strains slightly decrease towards the interface. In case of the plastic behaviour again high strains (total strains) occur in the center (Fig. 3b). They are several times higher than the strains in the elastic case. The zone of high total strains is very limited and they rapidly decrease in x- and y-direction. However, extremely high total strains arise in the interface and in the adjacent in zones centered at about 45° (grey colour). Due to the large plastic deformations also a redistribution of the elastic part of the strains occurs, resulting in a significant change of the stress field. The stresses in the interface will be studied in detail in a separate section.

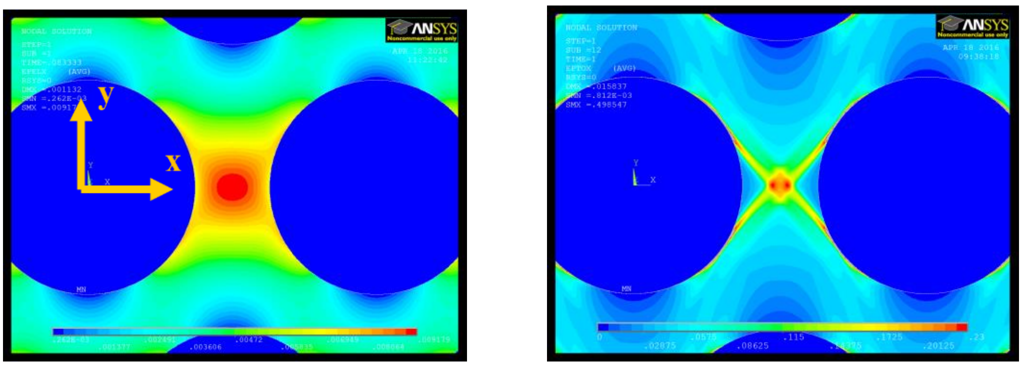

Figure 3. a) Strain ε_{xx}, elastic material model **b)** Total strain ε_{xx_total}, plastic material model

4.3 Normal stresses in elastic and plastic material model

The stress transfer between fiber and matrix concentrates in the center caused by the higher stiffness of this region due to the small distance between the fibers. In the elastic model the stresses σ_{xx} are dominated by the corresponding strains ε_{xx} (Fig. 4a). The stress field in case of plastic material behaviour is significantly different. The transverse stresses σ_{xx} exhibit four distinct maxima in the central region (Fig. 4b) instead of a coherent zone of almost constant high stresses (Fig. 4a). Due to yielding in the center region the adjacent zones carry more load. The distribution of the normal stress in the matrix accordingly is somewhat homogenised by the plastic deformations. In contrast, the shear stress distribution becomes very inhomogeneous and exhibits high stress concentrations in the vicinity of the interface, caused by pronounced shear stress concentrations (section 4.4).

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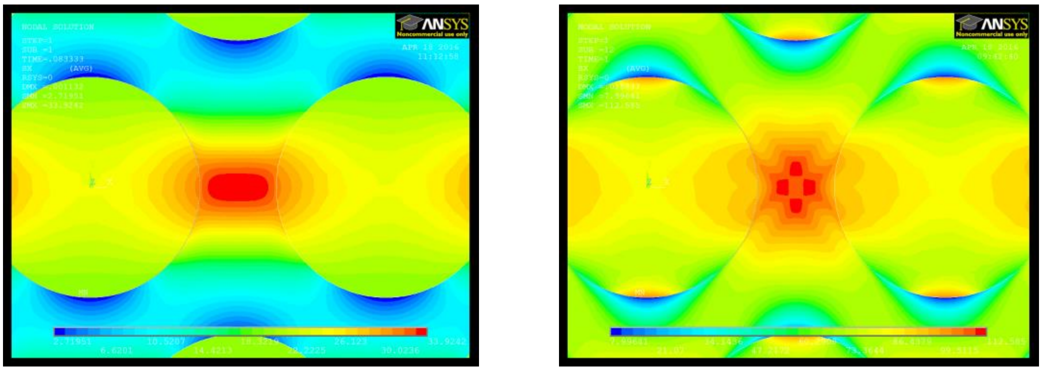

Figure 4. a) Stress σ_{xx}, elastic material model **b)** Stress σ_{xx}, plastic material model

4.4 Shear strains in elastic and plastic material model

The shear strains have two maxima in the interface in case of elasticity (Fig. 5a). The dominating one appears at 25°, a smaller one at 65°. They rapidly decrease in radial direction, that is, the shear stresses concentrate at the interface. Due to plastic deformation high shear strains develop between the fibers along diagonal lines touching the fibers (Fig. 5b). At an angle of 45° they become extremely high at the interface (grey zones). The plasticity of the material leads to a large increase of the shear strains and, as a result, shear stresses. The shear deformations are not restricted to small zones at the interface but large areas of the matrix deform. Still, the maxima of the shear stresses appear at the interface.

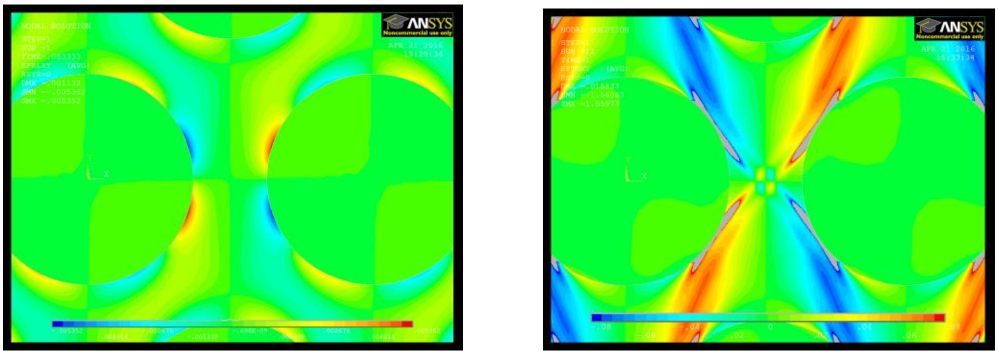

Figure 5. a) Shear strains ε_{xx}, elastic material model **b)** Total shear strains ε_{xy_total}, plastic material model

4.5 Stresses at the interface and in the adjacent matrix

Since inter fiber failure frequently is initiated by interfacial debonding the stresses and strains acting in the interfaces and in the adjacent matrix are examined. A cylindrical coordinate system with the origin in the center of the respective fiber is used. Since an interface has no thickness the stresses acting in the matrix adjacent to the fiber surface are analysed. For brevity the word interface still is used when normal stresses parallel to the interface $\sigma_{\varphi\varphi}$ or out of plane stresses σ_{zz} are addressed. The radial and tangential stresses in the interface are given in Fig. 6a for the upper half of the left fiber. The angle counts counterclockwise from the center line ($y = 0$). The level of the prescribed displacements is chosen such that the same external force at the edges in the elastic and in the plastic model are met. In the elastic model (Fig. 6a, red lines) the radial stresses reach their maximum where the neighbouring

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fiber is closest, that is in the center ($\varphi = 0$). The radial stresses σ_{rr} decrease rapidly in circumferential direction. Between 60° and 120°, this is at the top of the fiber, small radial tensile stresses of about 10% of the maximum are left. The tangential stresses $\sigma_{r\varphi}$ reach a maximum around 35° which is about half of the radial stress maximum. Then subsequently the shear stresses decay almost linear.

In case of plastic material behaviour the radial stresses form a small plateau at the center (Fig. 6a). The maximum is about 20% lower compared to the elastic case for the given loading. At the top of the fiber the stresses even become slightly compressive. Within the first 15° the shear stresses have a similar course in both cases. Due to the yielding process an extended plateau forms. From the plateau the shear stresses decrease rapidly. In addition the Mises equivalent stress in the interface is shown. We have to keep in mind that the model is based on a plane strain conditions. That is, in addition to the in plane stresses also out of plane stresses are present (σ_{zz}). The in plane stresses as well as the out of plane component are shown in Fig. 6b for the plastic material model. For small angles where the shear stresses are low the other normal stress components $\sigma_{rr}, \sigma_{\varphi\varphi}, \sigma_{zz}$ remarkably contribute to the equivalent stress. At about 15° the equivalent stress reaches the yield stress and plastic deformations takes place. The yielding zone covers most of the interface and the adjacent region.

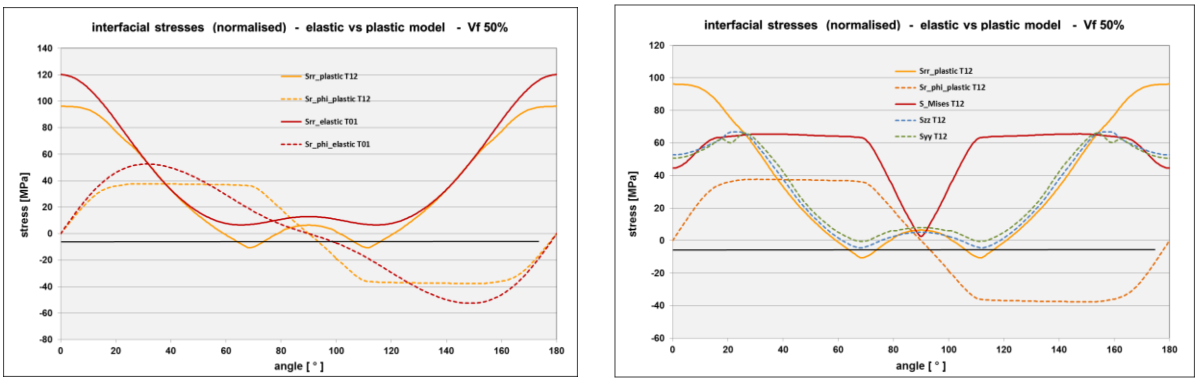

Figure 6. a) Interfacial stresses, elastic and plastic material model **b)** Interfacial stresses, plastic material model

In case of a material which undergoes large deformations the stresses may no longer be the physical quantities governing failure. Since the limit of deformability is now the key factor determining the initiation of failure the strains developing in the plastic material are investigated. The elastic strains arising in the plastic model are plotted on a different scale which is enlarged by a factor of 15 (Fig. 7a, right scale). The total shear strains (dotted yellow line) possess a pronounced maximum at 30° (Fig.7a). Here the total radial strains also show a maximum which is compressive. The elastic part of the radial strains decreases from the center ($\varphi = 0$). In the center almost no radial plastic strain is present (Fig. 7b). The same is true for the shear strains. The plastic deformation rapidly increases in circumferential direction resulting in a steep increase of the total shear strains. While the elastic part of the radial strains decreases in the beginning the plastic part increases leading to almost constant total radial strains.

These results of the stress and strain field calculated on a representative volume element of a composite show that the failure mechanisms completely change in case of matrix plasticity. While in an elastic matrix failure is caused by high radial stresses at the center line (y=0) the failure in case of a plastic matrix with the ability to undergo large deformations is caused by shear strains off the center. As a result, interfacial debonding is initiated at a different location and by a different mechanism compared to the elastic case.

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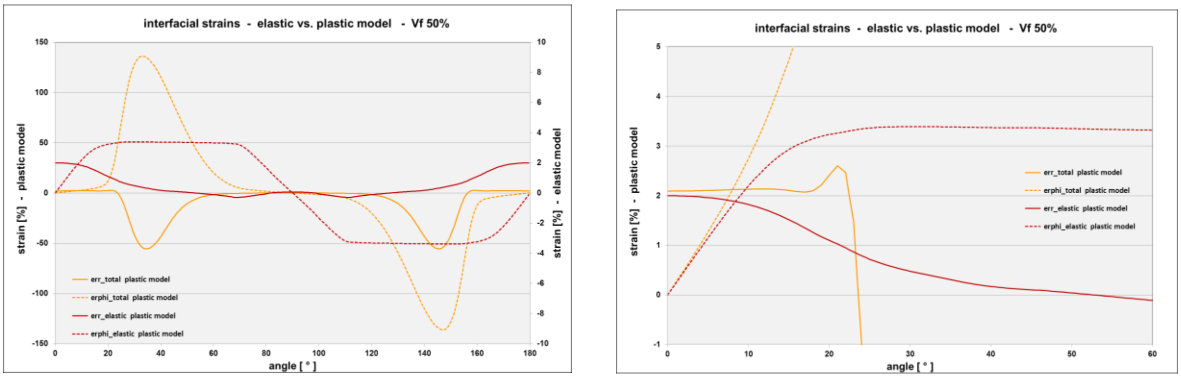

Figure 7. a) Radial and shear strains in the interface of the plastic model **b)** dto., enlarged scale

Conclusions

The plastic deformability of the matrix material significantly changes the stress field in the composite under transverse loading. This results in a change of the failure type as well as of the locations critical for failure initiation. There is a strong need for further improving the material description of the resin by further decreasing the scale from meso to micro. In addition, the effect of microdefects has to be investigated. Furthermore the time dependence of the matrix has to be taken into account even though – or just because – the ultimate failure is a high speed process.

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