FE MODELLING OF WEAR MECHANISMS OF CF/PEEK COMPOSITES

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ABSTRACT

The sliding wear behaviour of polymer composite materials has been the subject of many studies in order to improve their tribological performance. To characterise the wear mechanisms mostly experimental studies were performed. According to these findings the wear of these materials is a result of simultaneous wear of different components of the composites. The mechanisms to dominate the process of material removal are matrix wear, fibre sliding wear, fibre cracking and wear by fibre/matrix debonding at the interface. If fibre/matrix debonding is appeared, the local separation initiates additional fibre cracking, wear debris formation and more intensive wear process.

FE micro-models have been developed in order to determine contact, stress and strain conditions produced by a steel asperity sliding on the surface of a fibre-reinforced polymer composite. Three cases were studied, i.e. a normal, a parallel and an anti-parallel fibre orientation relative to the sliding direction. In order to get more realistic simulation results about the failure conditions in the composite structure, FE contact macro/micro-models were used, contrary to the so far widely applied anisotropic analytical or numerical macro-models. To model a ”micro-environment” as part of a ”macro-environment”, the displacement coupling technique was introduced. On the basis of the stress results, conclusions were drawn on the possible wear mechanisms of the fibre-reinforced polymer composites. For each fibre orientation, surface failure of the matrix material occurs due to high shear stresses. The other characteristic source of failure is fibre/matrix debonding, eventually followed by fibre cracking events.

According to the results, the high shear strain as well as the high tensile strains can produce local debonding of the fibre/matrix interface in a limited vicinity of the actual contact area. The experimental verification based on a scratch test using a smaller diamond indentor with the radius of R=100 μm, in order to represent the behaviour of the asperities in sliding contact. The scratch surface was observed by SEM.

1. INTRODUCTION

The reinforced polymers are widely used in different tribological applications such as rollers, sliding bearings, gears etc. Design or selection of these structural elements are usually based on experiences gained by specific wear tests. A theoretical approach for a better understanding of the wear mechanisms is not so common, due to the complexity of the wear process, the specific material behavior of the polymers and the inhomogenity of the composite materials.

The aim of the present study is to observe and to model the wear mechanisms of carbon fibre-reinforced PEEK composites by using scratch tests, scanning electon microscopy, and FE contact techniques applied to macro/micro-models.

2. CHARACTERIZATION OF THE COMPOSITE MATERIAL STUDIED

The material studied is a unidirectional carbon fibre-reinforced, thermoplastic polyether-etherketone (PEEK) composite.

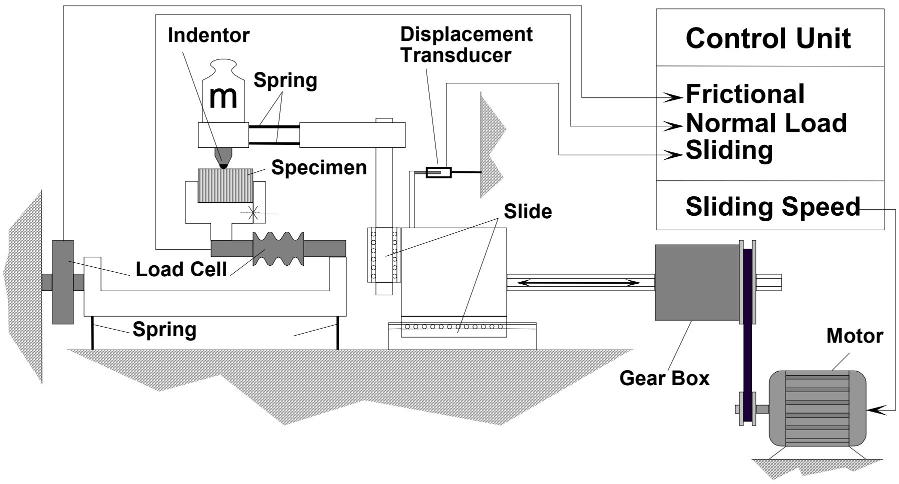
The strength of the carbon fibers is very high in the fibre direction, due to the high covalent bonds between the carbon atoms within the graphite lamellae oriented in the fibre length axis. In the radial direction, on the other hand, the strength of the carbon fibers is rather low, as a result of the weak van der Waals bonds between the graphite lamellae.

The thermoplastic PEEK, used as the matrix material, has a crystallinity of about 20-35%. The polymer is characterized by a high strength and toughness, a good chemical resistance, and an excellent tribological behavior [1]. In addition the semicrystalline PEEK polymer has a high resistance against repeated cycling loading and against heat and radiation.

3. EXPERIMENTS

To study the typical wear mechanisms, a diamond indentor with a radius of R=100 μm was compressed under a normal load of F_N=1 N onto the polished composite surface. Subsequently, the indentor was slowly moved, in order to create a scratch on the specimen surface. The scratch tests were carried out under normal (N) as well as parallel (P) and anti-parallel (AP) fibre orientation. With repetition of the scratch tests on the same groove, effects of cycling loading can also be studied. However, in this case, it is very important to ensure the same contact path for the diamond indentor during the repetitions. The test apparatus is described in Fig. 1. To study the typical wear mechanisms, the scratched surfaces of the composite specimens were examined by scanning electron microscopy. Typical SEM micrographs can be seen in Figs. 2-7.

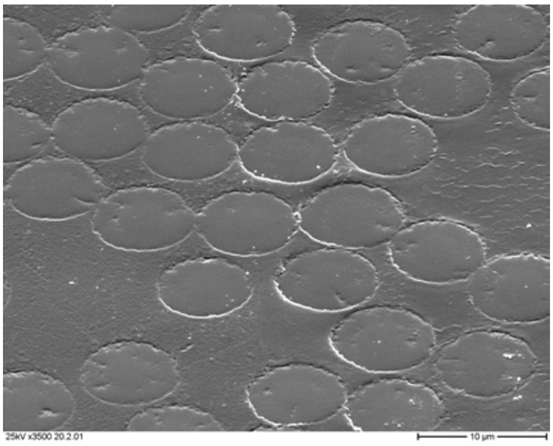
Fig. 2 shows a higher magnification SEM-micrograph of a transition between a polished, unscratched composite surface region (left side of figure, with the matrix slightly below the fibre level) and a diamond tip scratched region (right side, with matrix push up and shear features).

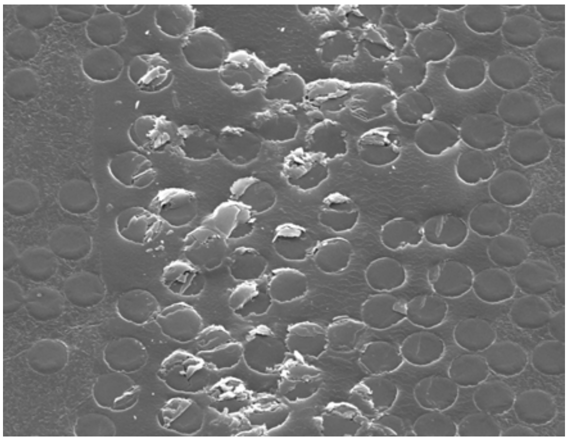


"**Fig. 1.** Scheme of the Scratch-Tribometer."

Fig. 3 represents the conditions after five repetitions. The corresponding wear features are much more intensive than after one scratch. Most probably, already during the first scratch some sub-surface deformations have been induced (but were not yet visible), which allowed further damage mechanisms to occur during the subsequent scratches. Based on the SEM photos (Figs. 2-3) the following wear mechanisms can be observed: (a) small matrix shear lips or cracks, transversely oriented to the scratch direction, (b) first indication of fibre edge cracking occurred at the rear edge of the fibers, and (c) the push-up of matrix material at the lateral fibre edges (relative to the slightly lower level of the matrix compared to the fibre ends on the polished, but unscratched composite surfaces).

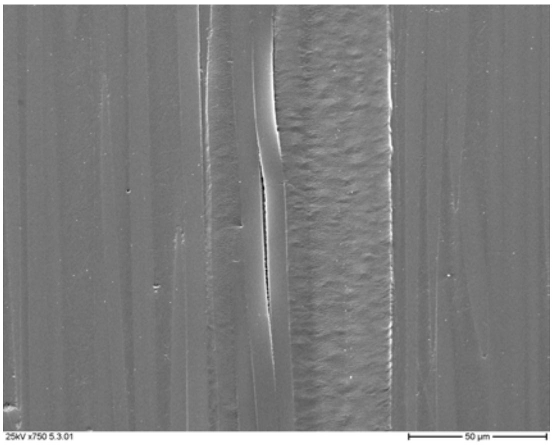
In the case of parallel fibre orientation, Fig. 4 shows the typical wear mechanisms, which are as follows: fibre/matrix debonding, fibre bending, fibre cracking, and matrix shear features. In Fig. 4, right and left from the groove, the polished surface remained in its undamaged state. At higher magnification (see Fig. 5), a fiber crack transferring from one fibre to the next is noticeable. In addition, the formation of the shear features can also be found in the SEM-photos.

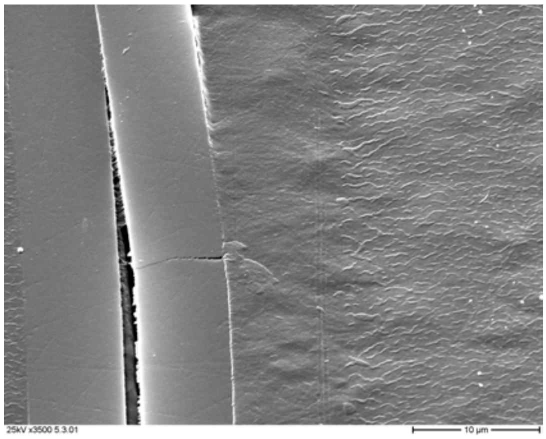




"**Fig. 2.** SEM-micrograph of a normally oriented composite surface scratched by a diamond tip (the tip was slid from top to bottom)."

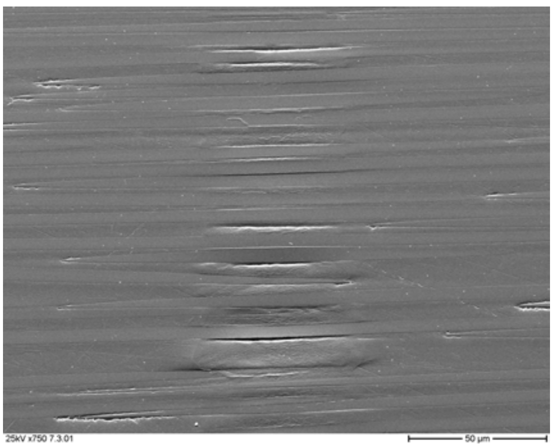
"**Fig. 3.** SEM-micrograph of a normally oriented composite surface after five scratch repetitions (the diamond tip was slid from top to bottom)."

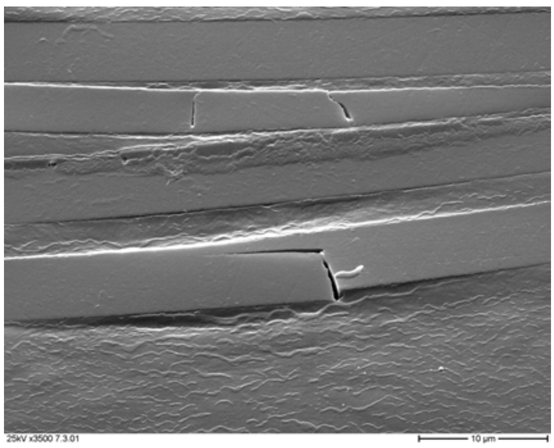




"**Fig. 4.** SEM-micrograph of a parallel oriented composite surface scratched by a diamond tip (the tip was slid from top to bottom)."

"**Fig. 5.** An enlarged segment of the contact area in the case of P-fibre orientation (the diamond tip was slid from top to bottom)."





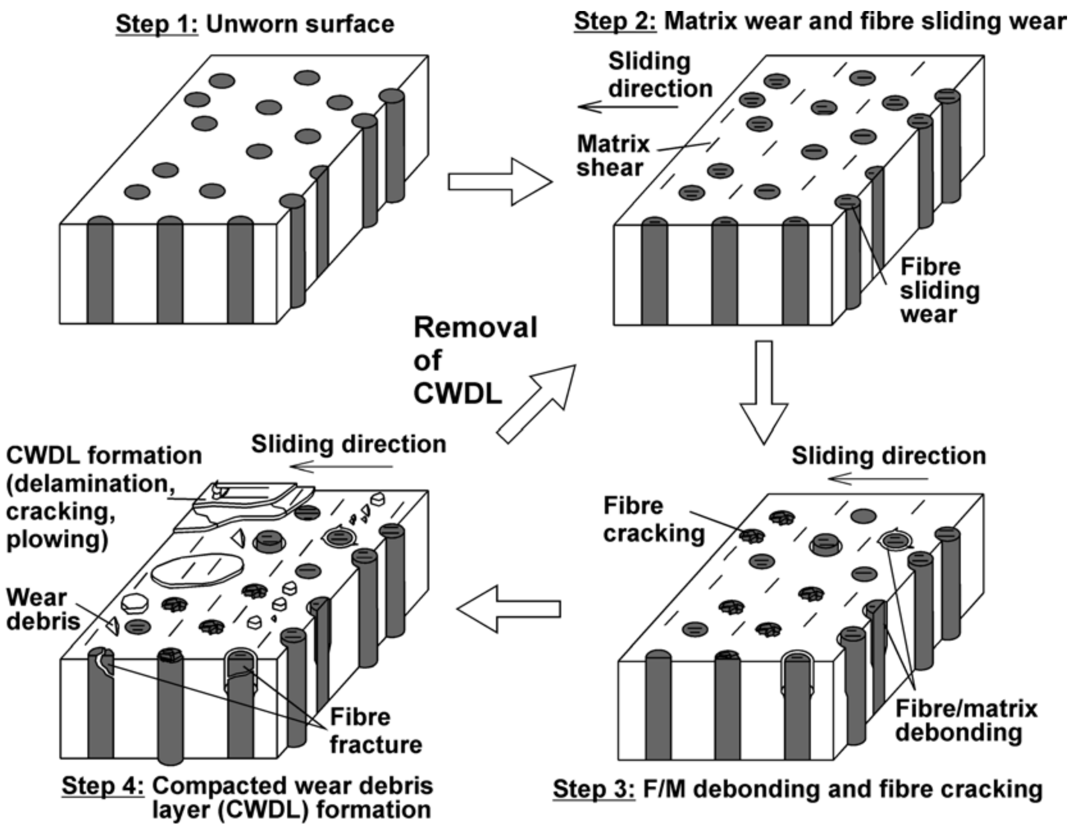
"**Fig. 6.** SEM-micrograph of transversely oriented composite surface scratched by a diamond tip (the tip was slid from top to bottom)."

"**Fig. 7.** An enlarged segment of the contact area in the case of AP-fibre orientation (the diamond tip was slid from top to bottom)."

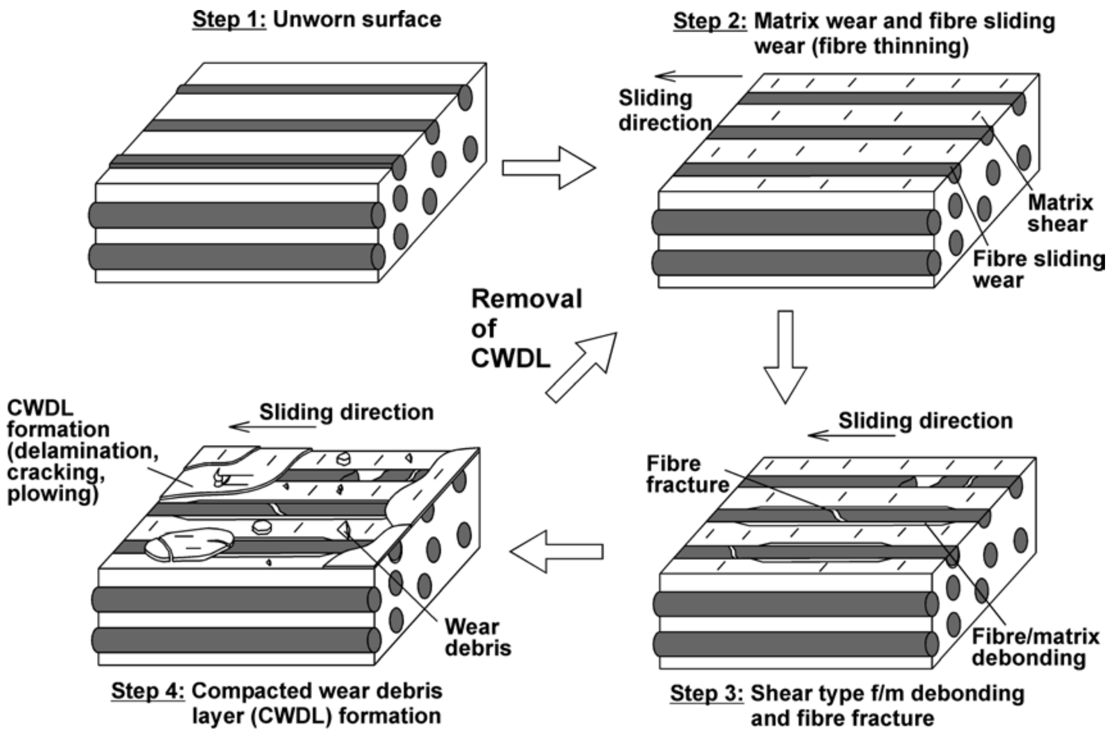
Figs. 6-7 show the typical wear mechanisms in the case of AP-fibre orientation. Due to the compressive and frictional loads, the fibers bend in the depth and in the sliding direction respectively. Both were associated with fibre fracture events in the center region of the core as well as in the transition between the groove and the undamaged area. Some of the broken fibre segments were removed out of the actual groove area, which can be a result of the additional torsion acting in the scratched material surface. Repeated sliding of the counterpart may lead to cutting up of the fibre fragments at a later stage of the wear process.
Fig. 7 shows an enlarged segment of the scratched area. Based on the above figures the following typical wear mechanisms can be observed: fibre/matrix debonding, shear features of the matrix material between the broken fibers (as already described for the P-fibre orientation), fibre cracking and their removal from the fibre beds.

4. THE REPRESENTATIVE WEAR MECHANISMS

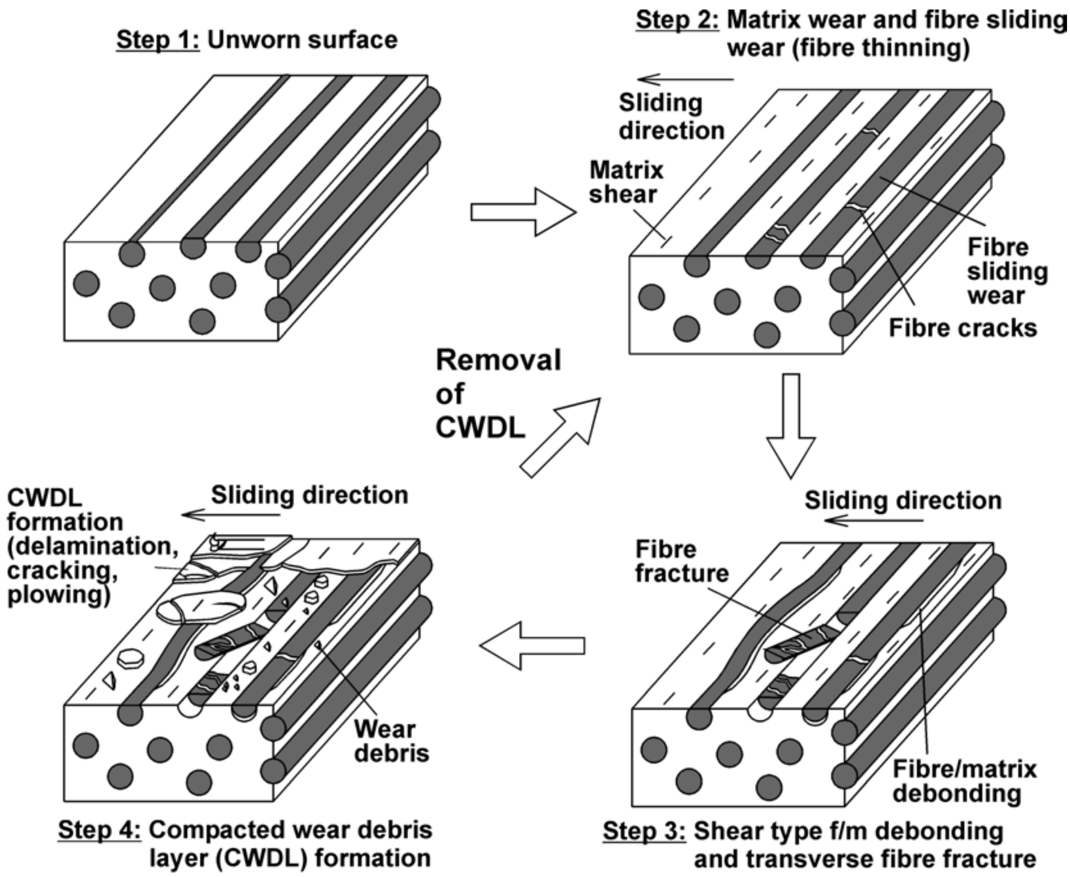
Figs. 8-10 illustrate schematically the typical wear mechanisms of the unidirectional fibre-reinforced polymer composites in the form of wear cycles. The wear sequence may occur as follows: matrix wear, fibre sliding wear, fibre cracking, wear by fibre/matrix debonding at the interface, fibre edge cracking. As it is illustrated, in the initial stage of the wear process, matrix wear and fibre sliding wear occur. Following these mechanisms, fibre cracking appears due to the matrix removal and the asymmetric contact pressure distribution produced by the sliding steel asperity.
In addition, as a typical failure mechanism, fibre/matrix debonding occurs due to the shear and tension type loading. If fibre/matrix debonding has taken place, the local separation initiates additional fibre cracking, wear debris formation and a more intensive wear process. In the steady state wear process, a so-called compacted wear debris layer (CWDL) covers the surface, composed of pulverized wear debris and matrix material. During the wear process this layer is continuously formed and removed by the surfaces sliding over each other.



“**Fig. 8.** Typical wear mechanisms of a normally oriented unidirectional fibre-reinforced polymer composite.”



"**Fig. 9.** Typical wear mechanisms of a parallel oriented unidirectional fibre-reinforced polymer composite."



"**Fig. 10.** Typical wear mechanisms of a transversely oriented unidirectional fibre-reinforced polymer composite."

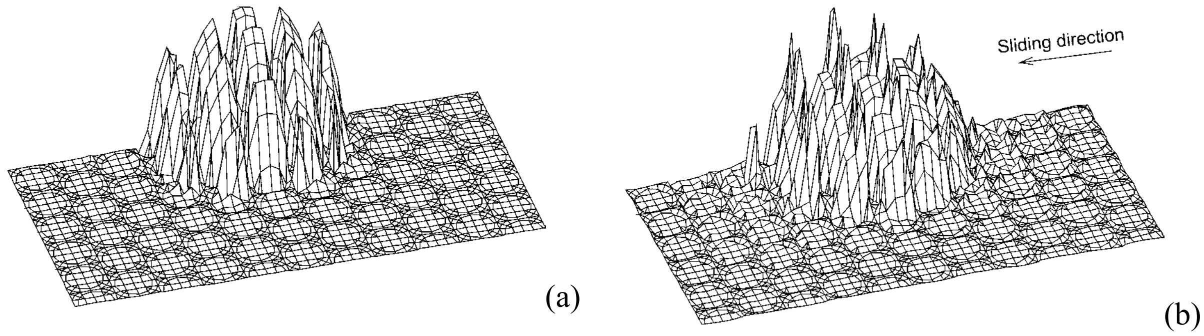
5. WEAR AND FAILURE CONSIDERATIONS BY FE ANALYSIS

To discuss the wear mechanisms, let as consider the contact and stress states developed between a counterpart asperity (e.g. represented by a diamond tip or a small steel ball) and fibre-reinforced polymer composites under different fibre orientations relative to the sliding direction. The numerical models and the detailed results can be found in [2] and [3]. The further statements, referring to the wear mechanisms are based on these numerical results.

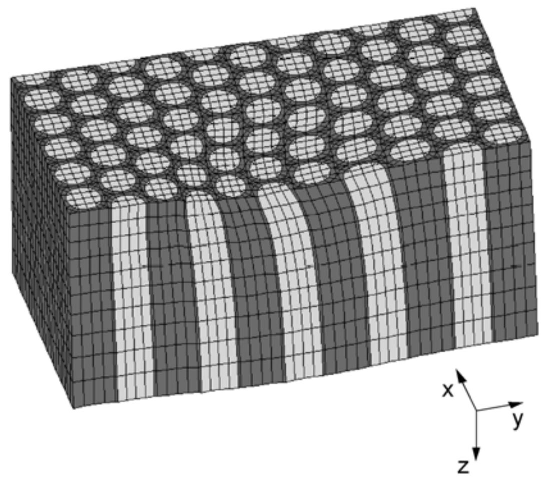
Normal (N) Fibre Orientation

Fig. 11 shows the contact pressure distribution in the cases of normal (Fig. 11a) and frictional (Fig. 11b) contacts. In both cases, mostly fibers transfer the load. The frictional force induces an asymmetric contact pressure distribution on the top of the loaded fibers. The local pressure peaks appear at the rear edges of the loaded fibers located within the contact area.

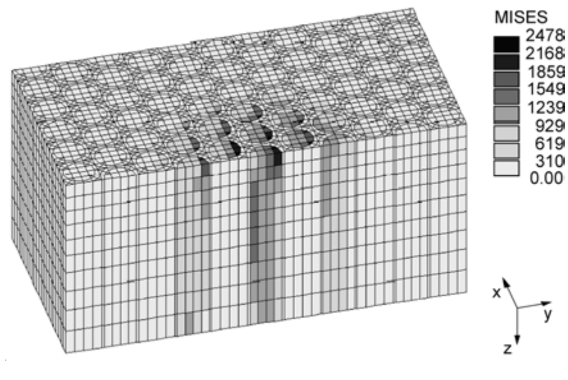
These results were obtained by FE macro/micro-models fitted by the displacement coupling technique [2]. In the case of frictional contact problem, the deformed shape of the micro-model can be seen in Fig. 12. Since in the present case and in [2] the radius of the steel ball is different from that of the diamond indentor, these results can, in a first approximation, only be regarded as illustrative results to understand the wear processes.



"**Fig. 11.** Contact pressure distribution without friction (a) and with friction (b) in the case of N-fibre orientation."



"**Fig. 12.** Deformed shape of the contact region in the case of N-fibre orientation (deformation scale 5:1)."



"**Fig. 13.** The Von Mises equivalent stress in the composite for N-fibre orientation."

The vertical stress component σ_z, that is the dominant component of the Von Mises Equivalent stress (Fig. 13), can cause failure at the rear edges on the top of the fibers. The compression limit strength of CF/PEEK composite is 1200MPa [4]. As a rough assumption, the compression strength of a single carbon fibre is about 1900-2000MPa, obtained by using a rule of mixture type relationship. This value and the maximum σ_z stress component at the rear edges of the mostly loaded fibers are about in the same range, predicting fibre cracking at these local region. This phenomenon can be explained by the asymmetric contact pressure distribution appearing on the top of the loaded fibers.

Due to the high normal and shear stresses, equivalent stresses, exceeding the yield limit of the matrix material, are produced near to the surface and along the fibre/matrix interfacial regions [2]. As a result, the matrix material becomes deformed and eventually can be sheared off in the form of thin wear debris layers.

The horizontal σ_y tensile stress component reaches its maximums at the rear edge of the contact area on the surface. During the sliding motion these stresses are producing a repeated compression-tension loading along the fibre/matrix interfaces at the surface, resulting in fibre/matrix debonding events on both sides of these fibers (in the sliding direction). The tension type debonding appears on the surface of the composite material, from where it propagates in the depth direction. In addition, the normal loads induce high shear stresses along the fibre/matrix interface. Therefore shear type debonding can also be initiated, propagating in the depth direction, similar to the tension one.

One can conclude that debonding seems to be one of the first phases of the wear process. If an asperity is sliding over the same region after debonding, the stresses produced (especially in the matrix) are substantially higher due to the separate deformation of the fiber and matrix components. At the same time, stronger bonding can reduce the debonding depth and in this way, can increase the wear resistance of the composite material.

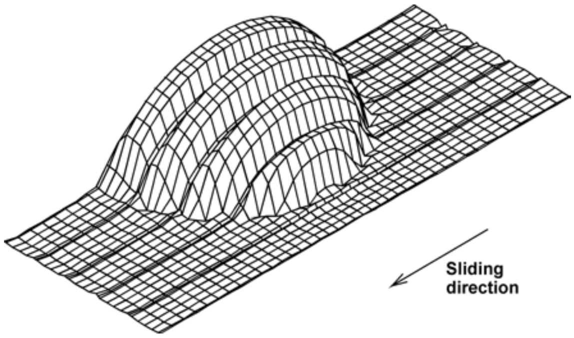
Comparing the failure mechanisms predicted, matrix shearing and fibre/matrix debonding seem to be the most dominant ones at the beginning of the wear process. They are followed by fibre cracking, as a further wear phenomenon, because the maximum compression stress at the top of the fibers is near to the compression limit strength.

Parallel (P) Fibre Orientation

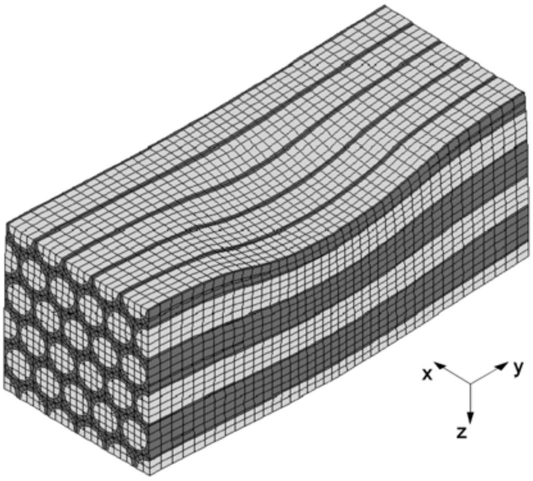
Fig. 14 and 15 show the contact pressure distribution and the deformed shape of the micro-model, respectively. The combined loading of compression, traction and bending can cause high σ_y stresses, that may lead to fibre cracking as it can be verified experimentally. If no friction is considered, bending produces a symmetric stress distribution (without traction) relative to the *x-z* plane. An additional friction force, however, modifies the stress pattern (Fig. 16). Behind the contact area, tension appears in the mostly loaded fibers, and the magnitude of the maximum compression stress in the middle of the contact area is increased. The location of the maximum compression region is moved further to the sliding direction due to the effect of the friction force.

In respect of the wear process the most important wear mechanisms are the follows: fibre thinning, fibre/matrix debonding, fibre cracking, shear events of the matrix material. Since the reinforcing fibers are oriented parallel to the contact plane they are, first of all, subjected to bending.

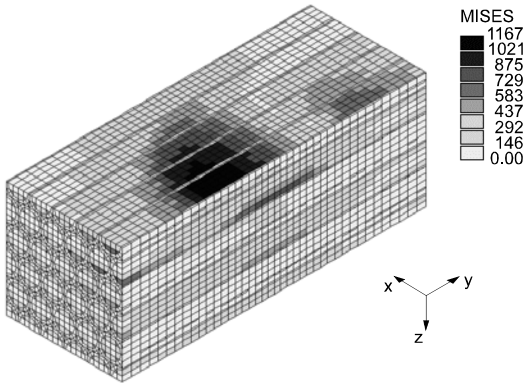
The matrix material, in a small vicinity of the contact area and especially in depth direction, is also subjected to high stresses. As a results, plastic deformation of the matrix under shear loading conditions should occur. The highest shear strain takes place around the mostly loaded surface fibers, at the front side of the contact area, and it is produced by the deformation due to normal and friction forces. According to these results, shear deformation and compression can finally yield to a shear type failure of the matrix material. The high shear strains can also produce shear type fibre/matrix debonding. During the repeated sliding motion this debonding can propagate to the surface, producing a complete debonding of the surface fibers. It can, therefore, be concluded, that shear type debonding is a dominant failure mechanism in the case of P-fibre orientation. This debonding can produce a different, partly separated micro-structure, that has less wear resistance than the original one, due to higher, more critical stresses.



"**Fig. 14.** Contact pressure distribution in the case of P-fibre orientation."



"**Fig. 15.** Deformed shape of the contact region in the case of P-fibre orientation (deformation scale 5:1)."

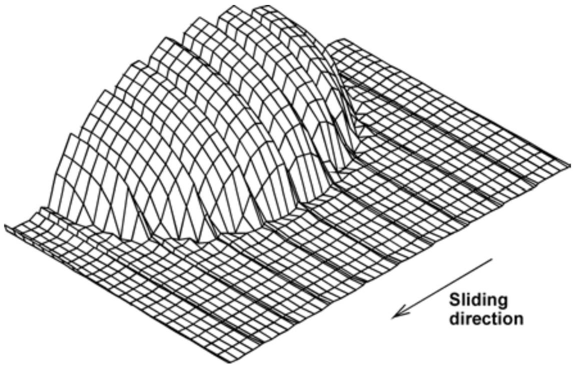


"**Fig. 16.** The Von Mises equivalent stress in the composite for P-fibre orientation."

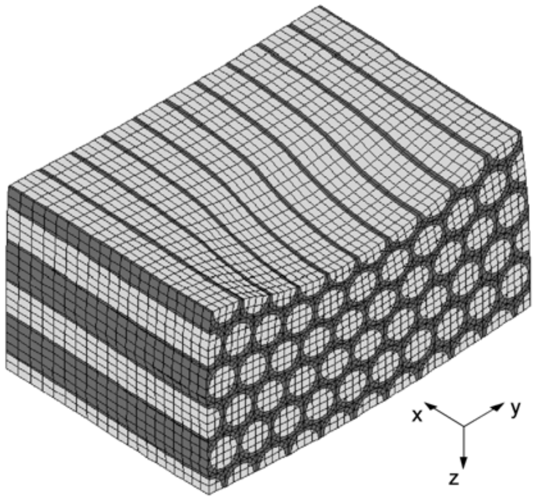
Anti-Parallel (AP) Fibre Orientation

The contact pressure distribution and the deformed shape of the micro-environment can be seen in Figs. 17-18. Considering the mostly loaded surface fibers, it can be concluded that the combined loading of compression/tension, bending and torsion produces high fibre stresses (Fig. 19). As a result, fibre cracking and fibre fragment removal can occur. Similar to the P-fibre orientation, it can also be concluded in this case that in a small region of the contact area, especially in the depth direction, plastic deformation appears in the matrix material.

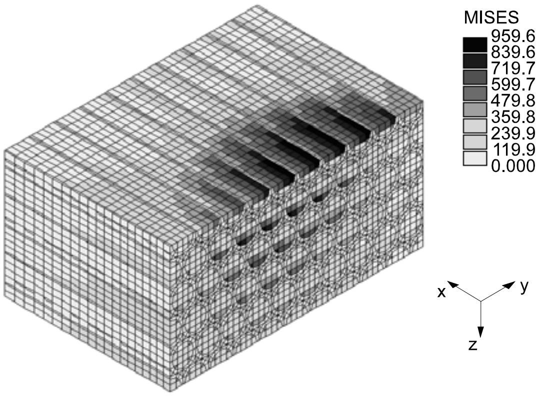
Based on the results of [3], the highest shear strains appear below the mostly loaded surface fibers, where they can produce shear type debonding. Furthermore the high horizontal □y stresses can cause tension/compression type debonding. During the repeated sliding motion this debonding can propagate to the surface, producing a complete debonding of the surface fibers.



"**Fig. 17.** Contact pressure distribution in the case of AP-fibre orientation."



"**Fig. 18.** Deformed shape of the contact region in the case of AP-fibre orientation (deformation scale 5:1)."



"**Fig. 19.** The Von Mises equivalent stress in the composite for AP-fibre orientation."

One can conclude, that the shear and tension/compression type debonding is also one of the dominant failure mechanisms in the case of AP-fibre orientation. According to the results of [3], the AP-fibre orientation produces lower fibre and matrix stresses than the P-fibre orientation. This would imply a lower probability for wear failure under AP in comparison to P-fibre orientation. This is opposite to the impression one has from the scratch experiments. Probably, in terms of the complete wear process and the complex wear mechanisms, the AP-fibre orientation is less advantageous, the wear debris has a more dominant effect on the wear process than in the case P-fibre orientation.

6. CONCLUSIONS

Typical wear mechanisms of unidirectional fibre-reinforced composites (CF-PEEK) were studied by scratch tests and SEM observations. The dominant wear mechanisms are as follows: matrix wear, fibre sliding wear, fibre cracking, wear by fibre/matrix debonding at the interface, fibre edge cracking. Based on these findings, representative wear cycles were identified for each fibre orientation.

The numerical contact results obtained by FE macro/micro-models explain the fibre cracking events, the plastic deformation of the matrix material and the fibre/matrix debonding. Finally, we can conclude that debonding seems to be one of the first phases of the wear process.

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