Collection of Abstracts

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ADHESION FRACTURE RESISTANCE. STRUCTURAL EFFECTS

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An analysis of different mechanisms and models of adhesion fracture of joint materials of different natural was performed accounting for the features of microstructure. The adhesion fracture resistance is considered as a result of an interaction of the processes at the interface and within its boundary layers.

Within the framework of a model of interface bonding accounting for nonlinear laws of bond stretching and different mechanisms of deformation in the interface boundary layers the adhesion fracture energy was analyzed. The contributions of the interface surface tension and energy dissipated within the interface crack end zones to the adhesion fracture energy were evaluated.

In case of polymer-polymer joints the mechanisms of interface bond rupture and bond pull out were analyzed.

For metal-metal joints an influence of vacancies, their clasters, interstitials and dislocations on the interface tension was evaluated by a thermodynamic approach.

Some models of a functional dependence of the energy dissipated within the crack end zone on the interface surface tension were discussed. Some experimental data which confirm existence of such dependences are also considered. The characteristic type of the interrelation between the energy dissipated within the crack end zone and the interface surface tension was derived for the polymer-polymer joints when polymers A and B are joint by means of connector polymeric chains of polymer C. It was assumed that the polymeric chains of polymer C can be mixed with polymer A and, on the other hand, their chemical reaction with polymer B is possible.

Special attention was paied to modeling of adhesion fracture resistance for the systems with a thin coating under mechanical and thermomechanical loading. For this aim a beam approach was suggested. The estimates of the effective fracture resistance of the coating-substrate joints were obtained for various affecting factors, including variations of materials properties with temperature and across the coating thickness. A possible effect of the variation of the interface boundary layer on the coating thickness was discussed.

Possibilities for optimizing the adhesion fracture resistance by adjusting the bonding parameters and/or micro- and nanodefects concentrations were demonstrated. In particular it was shown that in the metal-metal joints surface tension and separation work essentially depend on the parameters which characterize the relative ability of the interface to adsorb the point lattice defects.

The paper is mainly based on the results of studies performed by the author and his colleagues in the A.Yu. Ishlinsky Institute for Problems in Mechanics RAS.

Dislocation modelling of short fatigue crack growth

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The behaviour of micro-structurally short cracks subjected to fatigue loading is difficult to predict due to the pronounced influence on the crack extension from material inherent micro-structural features as well as from local plasticity developing through the load cycles. For poly-crystal materials it turns out that the static plastic zone has a dominating role in guiding the crack growth and in limiting the dislocation movements in the cyclic plastic zone. This demonstrates the necessity of tracking the development of the plasticity by tracing the movements of the individual dislocations forming it.

The present formulation for investigation of short fatigue cracks rests solely on discrete dislocations, where the geometry is described by dislocation dipole elements and the plasticity by discrete dislocations, each of size the Burgers vector of the material, moving along preferred slip directions within the material. The formulation makes it possible to demonstrate features such as changes in growth direction, growth rate and sensitivity to grain orientation, crack geometry and grain boundaries, applying to short fatigue crack propagation but not present at studies of long cracks viewed in a global sense, employing the Paris law to describe the crack growth.

From the formulation a zigzag shaped crack path, typical for micro-structurally short cracks, emerge naturally as a consequence of the interaction between grain geometry and plastic zone. The crack path geometry makes entities such as crack growth rate unpredictable. The momentanious crack growth rate depends on if the crack is about to change growth direction or not. The crack growth rate is thus found to correlate with change in crack path direction on the micro-scale, so that the crack accelerates immediately after a change of direction, and slows down prior to such a change. Deceleration thus signals a geometry change of the crack path. As regards crack closure, the dislocations forming the plastic zone determine the extent of closure. A general conclusion is that a straight crack, growing in one direction with only one active slip plane, closed at higher load levels than straight or zigzag shaped cracks with several active slip planes.

Mechanism of Factory-Roof Formation

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The factory-roof (F-R) indisputably belongs to the most extraordinary fractographical patterns in fatigue of metallic materials. Nowadays, there is only very limited knowledge about the mechanism of F-R formation.

The presentation attempts to fill this gap and tries to answer some basic questions as:
(i) Is the torsion an exclusive kind of loading that creates the F-R?
(ii) What is the characteristic 3D picture of factory roofs?
(iii) Which physically based relationships control the initiation and growth of F-R?
(iv) Why the factory roofs are not observed in the region of a very low cycle fatigue?

In order to answer the questions, a 3D picture of a characteristic F_R morphology was constructed. Moreover, a physically justified and geometrically consistent model of the initiation of F-R was developed.

FE study of slip localization and its effect on micro-crack initiation

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Slip localization is often observed in deformed metallic polycrystals. Cyclic loadings induce the formation of persistent slip bands [1-4]. In precipitate-hardened alloys, slip bands appear after shearing of precipitates [5]. And in pre-irradiated metals and alloys, sweeping of irradiation-defects leads often to channeling [6-7]. Following the literature, slip localization could accelerate grain boundary crack initiation observed during some cyclic tests [1-2] or stress corrosion cracking experiments [8]. The damaging effects of slip localization can be modeled by computing analytically the stress field induced by individual pile-ups [2]. But, several TEM observations [6] and AFM measurements [3-4] show clearly that slip occurs often homogeneously through the slip band of finite thickness rather than on one individual slip plane as it is assumed when applying the pile-up theory.

To evaluate the influence of slip localization on crack initiation, crystalline finite element (FE) computations are carried out using microstructural inputs (slip band aspect ratio/spacing, physically-based critical resolved shear stress (CRSS) values…). Slip bands (low CRSS) are embedded at the free surface of a matrix or small aggregates (high CRSS).

The following results are obtained concerning slip and grain boundary axial stress concentration:

- strong influence of aspect ratio, matrix CRSS and neighboring grain orientations
- weak effect of slip band CRSS, spacing and grain boundary orientation
- analytical formulae have been proposed to predict surface/bulk slips [9]. The load range of accuracy of those formulae is determined by comparing FE and analytical predictions of surface slips depending on the applied load
- the computed surface slips are generally in reasonable agreement with experimental AFM/TEM measures whatever the localization mechanism and material (austenitic stainless steels [7], nickel [4], α-brass [3]…) even if only microstructural inputs and applied load are used but no adjustable parameter.

Finally, grain boundary stress concentrations are studied. They are induced by the strong shearing of the grain boundaries. The computed surface strain and GB stress concentrations lead to strong accelerations of damage mechanism (surface oxide / grain boundary brittle fracture because of slip band impingement, vacancy cavity nucleation due to stress concentration [10]). Finally the computed stress fields are compared with the ones calculated assuming slip localized on on individual lip plane and considering the induced pile-up.

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Simulation of stage I crack growth using a 3D-model

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In the high-cycle fatigue regime the propagation of short fatigue cracks can determine up to 90% of a component's lifetime. After initiation the crack often grows on individual slip bands in stage I and exhibits strong interactions with microstructural features such as grain boundaries. Experimental investigations on a duplex steel have shown that the crack propagation rate decreases significantly or the growth even stops when the crack tip approaches a grain boundary. In order to simulate the propagation of such a stage I crack a three-dimensional model has been developed, which considers the influences of the free surface and the real orientation of the slip plane. The model is extended to elastic-plastic material behaviour by allowing a plastic deformation due to slip on the active plane. The spread of the plastic zone is blocked by grain boundaries. To solve the boundary value problem a numerical procedure based on the dislocation loop technique is applied.

MODELING FATIGUE CRACK GROWTH IN A BIMATERIAL SPECIMEN WITH THE CONFIGURATIONAL FORCES CONCEPT

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The concept of configurational forces is a powerful computational tool for the quantitative description of the behavior of cracks propagation in materials and structural components.
A two-dimensional finite element analysis is performed to model the behavior of fatigue cracks in bimaterial specimens made of diffusion-bonded ARMCO-iron and SAE 4340 steel with an interface perpendicular and inclined to the crack plane. The concept of configurational forces is used to evaluate by post-processing the crack-tip shielding and anti-shielding effects that occur due to the differences in yield stress, hardening and coefficients of thermal expansion of ARMCO-iron and SAE 4340 steel. For given values of applied stress intensity range ΔK_{app} and applied load ratio R_{app}, the near-tip J-integral J_{tip} is calculated at the maximum and minimum load. To allow for crack closure, the effective near-tip stress intensity range $\Delta K_{\mathrm{eff,tip}}$ is evaluated from the near-tip stress intensity range ΔK_{tip} and the near-tip load ratio R_{tip}. Calibration curves for homogeneous ARMCO-iron and SAE 4340 steel are used to calculate the crack growth rate da/dN from the values of $\Delta K_{\mathrm{eff,tip}}$. The distance between interface and crack tip L is varied to simulate the behavior of a growing crack. The computed curves da/dN versus L curves are in good agreement with the experimental results.

Micro mechanisms in Piezoceramics near morphotropic phase boundary

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Piezoelectrics are materials able to transform an electrical field (as stimulus) into mechanical strain (as response) and vice versa. Many crystalline materials exhibit piezoelectricity. But, a few materials exhibit these phenomena strongly enough to be used in technical applications. Thereof, polycrystalline piezoceramics are the strongest because highly nonlinear ferroic effects such as ferroelectricity and ferroelasticity dominate the piezoelectric effect. Up to now Perovskite-type Lead Zirconate Titanate (PZT) based ceramics near morphotropic phase boundary are commonly used in piezoelectric devices.

The PZT-crystallite itself shows cubic centro-symmetry above Curie temperature. When cooling down below the Curie temperature a phase transition from the paraelectric into the piezoelectric symmetry takes place. The lattice structure becomes then deformed and less symmetric (no inversion-centre). As a result, the piezoelectric phase exhibits spontaneous polarisation. Due to the misfit of the distinct variants of the unit cells, the microstructure becomes stressed and charged and hence twinned in order to reduce mechanical constraints and electric depolarisation fields. The material becomes piezoelectrically active at a macroscopic level due to texturising or "poling" (i.e. the alignment of "domains" similar to ferromagnetics).

Regions of identically orientated unit cells and spontaneous polarisation-vectors are called domains, like in magnetic materials. In polycrystalline piezoceramics the domains are oriented randomly at the as-sintered state. Viz. no macroscopic piezoelectric behaviour is observable. Because of the ferroic nature of the material, it is possible to force permanent alignment of the different domains using a strong electrical field, the so-called poling process. The poled state of the material can be modified by exceeding the electrical and mechanical limits of the material. The ceramic now behaves macroscopically as piezoelectric, and the material is remanently polarised and remanently strained.

In this work the micro mechanisms (domain-switching) have been measured using complementary methods such as hysteresis measurements of electric displacement v. electrical field and mechanical stress v. strain on the one hand and texture analyses via XRD-measurements on the other hand. These methods can help to develop an better understanding of domain-movements (change of the texture due to poling and de-poling mechanisms during electro-mechanical loading), which play a critical role in modern devices such as multilayer piezoelectric actuators.

Fracture behaviour of layered ceramics under different loading scenarios

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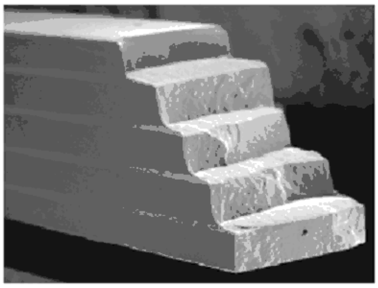
Abstract: Layered ceramics are a promising structural design that combines layers with different elastic-thermal properties which will influence the mechanical response of the laminate. From this viewpoint, much effort has been put in the fabrication of laminates with a tailored residual stress profile arising from mismatch of thermal expansion coefficients between layers, selective phase transformation and/or chemical reactions.

In the present investigation, zirconia-containing laminar ceramics have been designed to develop **compressive stresses** in the internal layers by means of the tetragonal to monoclinic phase transformation that takes place in the zirconia phase when cooling down during sintering. The corresponding volume increase associated with such transformation determines the residual stress field within the multilayer. Under certain conditions, these compressive stresses may act as a **barrier to crack propagation**. In other cases, crack deflection at the interface and/or crack bifurcation due to the high compressive stresses in the internal layers of the composite result in an **increase of fracture toughness** and energy absorption capability, with respect to conventional monolithic ceramics.



Keywords: Layered Ceramics, Alumina, Zirconia, Residual Stresses, Fracture Behaviour, Thermal Shock Resistance, Fatigue Response.

Numerical modelling of crack propagation and R-curve behaviour in layered Al_2O_3/ZrO_2 composites

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Numerical study of crack growth and apparent R-curve behaviour in layered Al_2O_3/ZrO_2 composites prepared by electrophoretic deposition [1] is attempted. This composite exhibits strong interfaces. The mismatch of thermal expansion coefficients between different layers inevitably generates thermal residual stresses during subsequent cooling of layered ceramics with strong interfaces. The relative thickness of different layers determines the relative magnitudes of compressive and tensile stress, while the strain mismatch between the layers dictates the absolute values of the residual stresses. The toughening effect of the residual stress state is often predicted by means of the weight function method, e.g.[1], [3]. The classical weight function concept is used to calculate the stress intensity factor, considering an inhomogeneous distribution of the residual stresses in a homogeneous body (mostly with the elastic modulus of the first layer). According to Fett et al. [4] an approximate weight function method can also be applied to heterogeneous, graded or laminated materials with a variable Young's modulus. However, the elastic mismatch of the layers induces an additional crack driving force term which is not taken into account by the weight function method. The propagation of a crack in a direction orthogonal to the laminate planes can be promoted (anti-shielding) or retarded (shielding) by the different elastic properties [5]. In this paper, the apparent R-curve obtained by means of the weight function method is compared with the prediction obtained using a detailed full field analysis by FE.
FE modelling of a cracked three point bending specimen was carried out assuming that the specimen contains maximum 100 layers (50 alumina layers and 50 zirconia layers). For comparison, also specimens with fewer layers were considered. The 2D FEM mesh was generated in the finite element system ANSYS. Due to symmetry of the problem, only one half of the specimen was modelled. A combination of quadratic isoparametric and quarter-point elements were used to model the specimen. The apparent R-curve of a laminate with a given residual stress profile was calculated considering the equilibrium condition at the crack tip, i.e. crack propagation is possible if the stress intensity at the crack tip, K_{Itip} for the crack length a equals or exceeds the intrinsic material toughness K_{IC},

$$K_{Itip}\left(a\right)=K_{Ic},\; K_{Itip}\left(a\right)=K_{Iappl}\left(a\right)+K_{IR}\left(a\right) \qquad (1)$$

and solving for K_{Iappl}, where K_{Iappl} is the applied stress intensity and K_{IR} the stress intensity contribution from the residual stresses. The displacement-matching approach was used to the calculation of the stress intensity factor. The results were obtained with quadrilateral elements collapsed to triangular quarter-point elements.
All the layers made of the same material (A or Z, respectively) have the same thickness, so the laminate is well defined by the thicknesses t_A and t_Z, or the total thickness W and the ratio $\lambda = t_A/t_Z$.. In this paper, the total thickness W was 2.55mm and the laminate thicknesses $t_A = t_Z$ = 0.00255 mm. The A layers are under compression since their thermal expansion coefficient is minor than that of the Z layers which are under tension. The calculated values of the apparent fracture toughness as a function of the crack length a are shown in Fig.1. Horizontal lines in individual layers denote the intrinsic material toughness K_{IC}.
The apparent toughness increases in the layers with compressive stress with increasing crack length, and it decreases in the layers with tensile stress as the crack continues to grow. As to

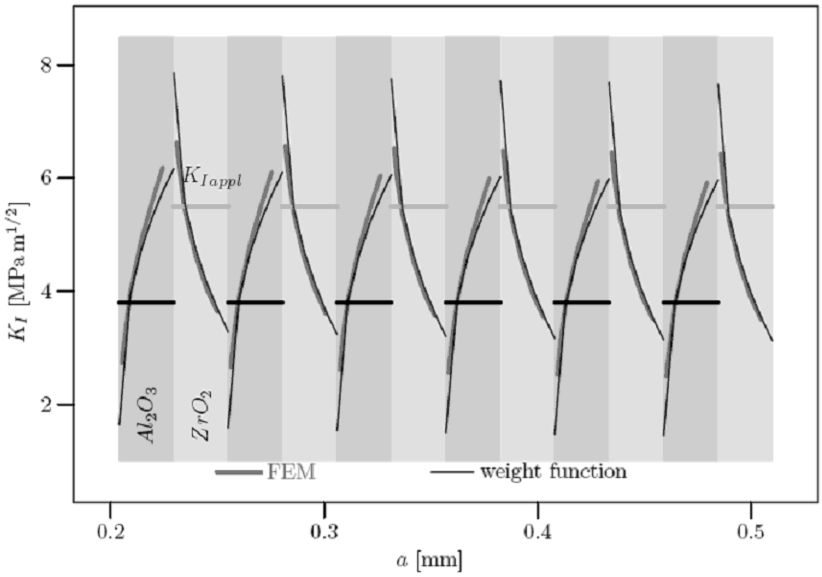

Fig. 1

the weight function approach, note that the stress distribution $\sigma(x)$ acting normal to the fracture plane was also obtained by FEM.

In the case of a matrix crack impinging on the interface, a differential energy analysis is unsuitable due to the discontinuity in the elastic properties which entails that the exponent δ_1 of the displacement singularity is different from the classical value 0.5. *Finite* crack extensions l are to be considered (instead of infinitesimal one) and the competition between deflection and penetration at the interface is evaluated using the condition that the crack will follow the path which maximizes the additional energy ΔW released by the fracture. If crack penetration occurs preferentially to deflection at the interface, the following condition [6] must be satisfied: $\Delta W_p = \delta W_p - G_c l_p > \Delta W_d = \delta W_d - G_{ci} l_d$, where G_{ci} is the interface toughness, G_c is the toughness of the lamina material and δW is a change of the potential energy between the original and new crack position. Since the interfaces of the investigated layered composite were found to be very strong [1], the previous condition is presumed to be fulfilled. As the crack grows through layered composite, it impinges on the interfaces between laminae and two situations are encountered: a strong singularity (δ_1 <0.5) if the crack propagates from stiffer to softer lamina and a weak singularity (δ_1 >0.5) if the crack propagates from softer to stiffer lamina. In the case of strong singularity, the total excess of energy may cause that the crack propagation across the interface is unstable. In the case of weak singularity, the energy terms ΔW_d and ΔW_p are negative and correspond to amounts of the energy which must be provided to the system. The crack instability cannot be invoked since the applied loading must be continuously increased to allow the crack tip to reach the interface. A suggestion is made how to link together the results concerning the crack touching the interface with those pertaining to the situation when the crack tip is very close to the interface.

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The influence of defects on the development of deformation and fracture of Ni_3Al intermetallide nanofiber

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The peculiarities of structure-energetical reconstruction of Ni_3Al nanofiber taking place at the influence of deformation of uniaxial tension were studied by the method of molecular dynamics. The calculated block of nanofiber crystal was presented in a form of segment of rectangular parallelepiped. The sides of parallelepiped containing 24x24x48 atoms corresponded to planes [100] of FCC lattice. Free boundary conditions were applied to the boundaries of the calculated block in the directions of axes X, Y. Rigid conditions were applied in the places of deformation influence. Tensile stresses were applied along the axis of tension – Z with the velocity 20 m/s. The interactions between different pairs of atoms were given by Morse semiemperic potential functions. The graph of the dependence of accumulated energy of deformation on time was made for the segment of segment of nanofiber which was ideal in structure and composition. Four stages of reconstruction of atomic structure and superstructures were clearly seen in the graph.

The first stage – quisielastic deformation; the second – plastic deformation; the third – flow, the fourth – failure. The first stage when accumulated energy of deformation increased by parabolic law was finished by the appearance and accumulation of Frenckel pairs – vacancies and interstitials. Point defects united in dislocation loops at the definite concentration. The shear of the pairs of deformed nanocrystal took place along plane [111]. In this connection, the level of accumulated energy of deformation went down sharply.

The changes of distribution of deformation stages, the obtained levels of accumulated energy of deformation of nanostructural fiber were noticed at the presence of vacancies and interstitial atoms. The first stage of deformation nearly disappeared at the definite concentration of Frenckel pairs. Antiphase boundaries were introduced into the nanofiber. The differences in structure-energetical stages of deformed nanofiber were seen in the dependence on their types.

Fracture Behaviour of Tungsten and Tungsten Alloys

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Abstract

Future nuclear fusion reactors will bring about demanding environment for particular materials. High power loads, disruption events and erosion of first wall materials in general and of divertor materials in particular will lead to the usage of tungsten and tungsten alloys. The divertor region is the lower part of the vacuum vessel, where plasma particles interact with the wall due to their open trajectories. Thus, the material properties asked for are high thermal conductivity, high melting point, erosion resistance and low sputtering yield.

A problem one is confronted to when working with tungsten, is its low ductile-to-brittle transition temperature (DBTT), complicating handling at low temperatures. We investigate pure tungsten as well as different tungsten alloys, such as tungsten – rhenium alloys (W 13Re, W 26Re), potassium doped tungsten and lanthanum oxide dispersion strengthened (1 wt% La_2O_3) tungsten, in consideration of the influence of microstructure, dislocation density, stages of processing and temperature on the fracture behaviour.

Rhenium is known to have a positive effect on the ductility of tungsten, i.e. lowering the transition temperature, whereas the mechanisms governing this change in material property have remained an open question. Double cantilever beams specimens and compact tension specimens were manufactured and tested at temperatures ranging from room temperature up to 900°C in-situ as well as ex-situ in a specially designed furnace- and vacuum chamber equipped tensile testing machine.

For analyzing the extent of plastic deformation and the interaction of dislocations with the fracture process, we use electron backscatter diffraction, revealing a certain amount of plastic deformation close to the crack path; this amount is especially dependant on testing temperature and chosen tungsten material.

Plasticity and Damage Evolution in a Stamping Tool

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High-strength steels are increasingly used for lightweight structures, such as car bodies. Stamping, cutting and blanking of such high-strength steels cause extremely high loads, especially near the cutting edges of the tools, which might lead to early failure. There is a high demand to understand and to predict the local loading conditions and the damage processes at and near cutting edges in order to search for possibilities to increase the lifetime of such tools by optimising the cutting process, the tool geometry and the tool material.
A special edge-loading experiment has been developed in order to investigate the material response under cyclic loading and high compression. The local deformation behaviour and the damage evolution is analysed after different load cycles.
Two different high-strength tool steels have been investigated, a cold-work steel produced by the electro slag remelting method and a powder-metallurgically produced high-speed steel.

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Anisotropy of the Fracture Behaviour of Severly Deformed Iron and a Pearlitic Rail Steel

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Keywords: Severe plastic deformation, Armco iron, Pearlitic steel, fracture toughness.

Abstract

Among the various types of Severe Plastic Deformation (SPD) High Pressure Torsion (HPT) is an outstanding and simultaneously simple technique of producing submicro- and nanocrystalline materials. Due to the limited dimensions of HPT-deformed metals little is known about their fracture properties. Therefore in this study the fracture behaviour of pure iron and a steel with a fully pearlitic structure was investigated. For this, CT-specimens were cut out of the penny shaped HPT specimens, such that the load could be applied along different axes with respect to the material microstructure. Afterwards fracture toughness tests were performed. The study shows that there is a strong mechanical anisotropy with respect to different loading directions. This was found in both the one phase iron and the pearlitic steel, consisting of two phases. This anisotropy can be related to the resultant microstructure obtained by the HPT shear deformation process.

Micromechanical analysis of ductile fracture initiation in steam pipeline steel

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Keywords: Ductile fracture, Steam pipeline steel, Complete Gurson model, Rice-Tracey model

The objective of this paper is micromechanical analysis of damage level of steam pipeline steel 14MoV 6 3, in the conditions of the crack growth initiation by ductile fracture micromechanism. The state of the material is analysed after 100,000 service hours of the steam pipeline in the power plant. The level of damage is determined by experimental and numerical analysis of the material taken from the pipeline (old material) and new (virgin) material of the same grade, for comparison. Experimental investigation is conducted on round tensile (RT), notched tensile (NT) and pre-cracked SENB specimens made of both materials. From previously published results, based on the investigation on NT specimens, it is concluded that the level of damage in old material can be determined using micromechanical approach. However, previous work did not include the analysis of the pre-cracked specimens using this approach. The subject of this paper is determining the value of J integral at the onset of ductile crack growth – J_{Ic}, using micromechanical models.

It is well known that the micromechanism of ductile fracture includes three stages: void nucleation, their growth and coalescence. Two micromechanical models are used in this paper for determining the crack growth initiation by ductile fracture micromechanism; one uncoupled (Rice – Tracey, modified by Beremin) and one coupled (Complete Gurson Model – CGM, proposed by Zhang et al). With Rice – Tracey model, it is possible to simulate the void growth and to determine the critical damage parameter for prediction of the crack initiation. This parameter is determined on NT specimens with three different notch radii, to take the effect of triaxiality into account. CGM is a micromechanical model that incorporates the effects of void coalescence, through plastic limit load model proposed by Thomason. Very important advantage of this model is direct determination of the damage parameter (critical void volume fraction), without transferring this parameter from another geometry (e.g. NT specimen).

Volume fraction of MnS inclusions and volume fraction of secondary void nucleating particles are determined from chemical composition, using Franklin's formula and lever rule, respectively. Initial void volume fraction and size of the finite element (FE) near the crack tip are varied, and calculations are performed with and without taking the nucleation of secondary voids into account, in order to analyse the influence of these parameters. Values of J_{Ic} for both materials are determined using the micromechanical models mentioned in the previous paragraph. Experimentally determined ratio J_{Ic}(new)/J_{Ic}(old) is compared to the values obtained numerically, and it is determined that FE size has significant effect on this ratio. Based on these observations, optimal correlation between the FE size and microstructural parameters is established. It is possible to assess the level of damage (ratio J_{Ic}(new)/J_{Ic}(old)) using either of the presented models, but the uncoupled model exhibits larger deviations of both J_{Ic} values in comparison with experimental results.

MICROMECHANICAL CONSIDERATIONS REGARDING CRITICAL PLANE METHODS FOR FATIGUE ASSESSMENT

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ABSTRACT

The underlying micro-/mesomechanical setting and the possible physical mechanisms of fatigue failure are discussed. Within this setting, a general modular continuum mechanics framework able to represent these geometrical and physical features is proposed.

The basic idea of the concept is that a statistical treatment of the microstructure (grain and flaw ensemble) at a macroscopic material point leads directly to its fatigue assessment. For macroscopically isotropic materials, all orientations are distributed homogeneously across the hemisphere. In this case, the weakest link hypothesis leads to the classical critical plane concepts; the critical distance hypothesis leads to integral methods (such as the effective shear stress hypothesis and the shear stress intensity hypothesis, cf. [1]), where an average over a range of microscopic orientations is used to assess macroscopic failure.

Based on the aforementioned considerations, a critical assessment of some more prominent criteria [1] [2] is performed. It is most interesting to note that, from a micromechanical point of view, the integral shear stress hypotheses [1] [2] may be interpreted, depending on their treatment of superimposed normal stresses, either as elastic shakedown criteria (without accounting for normal stress components) or as Mode II crack arrest criteria (accounting for crack closure due to normal stresses). An extended version of the integral shear stress hypothesis [1] fits well into the framework presented above and has improved predictive capabilities, notably for anisotropic materials.

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Damage in the Wheel – Rail – Switch System

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Rolling contact fatigue and plastic flow are significant examples of surface damage occuring on rails, crossings and wheels, limiting the life time of these components. Wheel - rail contact forces are one of the principal factors giving rise to surface damage of wheels and rails. For most problems with surface damage the tangential components of these forces are more important than those normal to the wheel/rail contact.

Some types of damages arising in wheels, rails and crossing components are shown. Since our aim is to simulate the development of surface damage a hierarchical model for describing the rolling/sliding contact of a wheel on a rail is introduced. The wheel – rail forces and consequently the near-surface deformation of the contacting partners are depending on the vehicle system dynamics. Therefore, at the start of simulations a proper description of this dynamic process is necessary. Either vehicle dynamics simulations or dynamic finite element calculations can be used to deliver the input for a static finite element model dealing then with the local elastic-plastic rolling/sliding contact.

To include also a more realistic picture of rough surfaces of wheel and rail in the calculations, which is absolutely necessary to predict the highly deformed surface layer and microstructural changes as observed in wheels and rails, a further refinement of the model (i.e. finite element representation) is necessary.

The results of our simulations are compared to observed damage in wheel – rail and crossing components. It will be discussed, whether the simulations can describe the reasons for microstructural changes and damage in wheel – rail components at the present state of research.

Fatigue of Copper at Very High Numbers of Cycles

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Abstract

In an attempt to study the deformation features of fcc metals in the very high cycle fatigue (VHCF) regime, experiments have been performed on polycrystalline copper at cyclic stresses slightly below the conventional PSB threshold for more than 10^{10} cycles. The study was performed at an ultrasonic frequency of approximately 19 kHz. Surface appearance and dislocation structures were studied with SEM and TEM. First, the distribution and extent of slip activity was determined for several defined strain and stress values below the conventional PSB threshold. The amount of irreversible slip, expressed in terms of the number of grains containing slip features, increases with both, stress amplitude and number of cycles.

Secondly, the identification of PSBs with the method of polishing and reloading was applied and it was found that PSBs are formed below the threshold, if the number of cycles is above the conventional $2x10^6$.

Finally, measurements of fatigue life were performed until 10^{11} cycles which shows that there is no fatigue limit below at least 10^{10} cycles.

Fatigue Damage Evolution in a Duplex Stainless Steel under Low Amplitude Loading Conditions

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The fatigue behaviour of a stainless duplex steel was characterized with respect to the damage evolution. The microstructure consists of 50% ferrite and 50% austenite, thus leading to a beneficial combination of both phases. Since ferrite is the stronger phase local plastic deformation in terms of slip markings is only observed in the austenitic phase, at least in case of low amplitude loading conditions. Specimens either survived 10^8 cycles or failed before reaching 2×10^6 cycles. Only few specimens failed in the range of 10^7 cycles from internal inclusions. Although slip markings increased in intensity and occurrence during the tests an endurance limit for this material seems to exist, defined by the phase boundaries as they exhibit a pronounced barrier effect against dislocation movement. Preferred damage initiation regions in the austenite are recrystallization twins or their boundaries (CSL Σ3 boundaries), depending on the local grain orientation with respect to the stress axis. Thus, twin boundaries do not exhibit a high resistance to plastic deformation. Since the austenitic phase has the character of an inclusion in the ferritic matrix, damage is not spread through the whole material.

THE TWISTING MECHANISM OF THE FATIGUE CRACK SUBSURFACE ORIGINATION AND PROPAGATION FOR Ti-6Al-2Sn-4Zr-2Mo-0.1Si ALLOY.

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The phenomena of fatigue crack origination subsurface related to the metals behavior in Very- or Ultra-High-Cycle-Fatigue (UHCF). The main idea introduced to explain the metals possibility to originate the fatigue crack subsurface based on the influence of inclusions stress state (constrain) because cracks origination take place from inclusions.

For Ti-based alloys, crack origination subsurface can be seen without influence inclusions. The discussed situation takes place for bi-phase ($\alpha+\beta$) Ti-based alloys with lamellar or globular (granular) structures. Two situations of metals cracking were discovered in area of subsurface crack origination: (1) the point of origin places at the boundary of two grains or plates; (2) the origin area creates because of quasi cleavage one grain or one plate.

The first stage of subsurface crack propagation has discussed as the short crack propagation. Nevertheless, it is not the same situation for developed cracks at- and subsurface. The principal difference in these two situations for crack propagation takes place because of (1) environment effect, (2) stress-state (constrain factor), and (3) Bauschinger effect.

Tests were performed on titanium alloy VT3-1 (Ti-6Al-2Sn-4Zr-2Mo-0.1Si) which is widely used in manufacturing procedure for compressors disks of aircraft engines.

Primary specimens were subjected to symmetric and asymmetric tension-compression with frequency of 35 Hz at temperature 20°C on the hydraulic test machine. The maximum stress level was in the range of (140-920) MPa with stress ratio in the range of (0.3-0.67) for tension and at -1.0 for tension-compression. Tests with various R-ratios were performed at the constant mean stress level 600 MPa.

During tests, there were performed Acoustic Emission (AE) monitoring for the moment of the fatigue crack origination and short crack propagation registration. The main idea of the crack detection in fatigue tests is based on the introduced earlier "α-criterion". The summarized AE-signals in versus number of cycles have drastically increases at the time of the crack initiation at the surface. If registered, the test continued during several hundred cycles for clear evidence of the "α-criterion".

It was discovered that "α-criterion" reflects the drastically increase of AE-signals versus number of cycles so times of evidence as number of subsurface origins was created.

The fracture surface analysis of the fatigued specimens has shown that the subsurface crack origination primary performs because of material cracking by one of the global or lamellar planes of bi-phase ($\alpha+\beta$) Ti-based alloy structure. The next step of the material cracking directed to the fracture origination from the border of the first smooth facet. However, the short crack propagation performs because of quasi-cleavage of material under mixed mode I and II as a result of its local volumes rotation under external cyclic tension.

The numerical calculations of material volume deformation in the case of three-axial stress state were used to explain mechanism of material destruction.

The twisting mechanism of material cracking was introduced to explain, primary, smooth facet creation in the origin of subsurface initiated crack. The twisting process of material cracking was also evident for short crack propagation from the first smooth facet.

DEFORMATION AND FRACTURE OF WOOD

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ABSTRACT

The mechanical properties of wood are strongly determined by its structure as an intrinsic feature on one hand and by extrinsic influences, like loading condition or environment, on the other hand. The structure of wood is most complex and hierarchical, with this complexity existing on all levels of magnification. Therefore it is necessary to correlate the deformation behaviour to the structural features not only macroscopically but especially in the micro- and even nanometer range.

Different modern techniques help to better investigate and understand the mechanisms of deformation. The structure-function relationships of wood have been studied mainly at the cell and cell-wall level in this study. Investigations with an in-situ deformation unit in an ESEM (environmental scanning microscope) made recording of mechanical data and synchronous observation of the underlying changes of the microstructure possible. In addition, observation of the influence of humidity was possible. Recording the load deformation response and observation of deformation features during the test thus helped to provide information on the underlying mechanisms.

Consequences of different microstructural features on the fracture mechanical properties will be shown and discussed, and variations of species and orientation will be considered. On the cell wall level, the influence of the microfibril angle on the fracture properties of wood will be reported.

Martensitic Transformation in the Vicinity of Fatigue Microcracks in Metastable Austenitic Steel: SEM Analysis and Modelling Concept

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The application of cyclic loading to metastable austenitic steels causes a gradual transformation from the fcc γ austenite phase into the metastable hcp ε martensite and the cubic α´ martensite phase. Martensite nucleation sites are slip bands being generated due to local stress concentration according to the elastic anisotropy of the microstructure. By means of correlating EBSD measurements with FE calculations applied to the real microstructure the stress concentration was attributed to the effect of elastic anisotropy. According to former studies, martensite formation occurs when the local plastic strain amplitude exceeds a certain threshold value. In the vicinity of short fatigue cracks the threshold condition is always fulfilled, and hence martensite formation expands over a wide area in the respective grains, depending on the local crystallographic orientation. For a closer evaluation of the transformation process, which causes a volume increase and an increase in the local strength, a piezo-driven testing machine was designed for in-situ-observation of the fatigue crack propagation. The microstructural phenomena that accompanies microcrack propagation in metastable austenitic steels have been implemented into a mechanism-based boundaryelement approach.

Crack initiation and growth in Chromium Nitride coated tool-steel

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In our study we investigate the damage evolution in a coated tool steel under monotonic and cyclic loading, as a function of the residual stress state of the coating. Chromium Nitride (CrN) coatings were deposited on a heat-resistant tool steel by physical vapour deposition. The substrate hardness was varied, as well as the residual stresses in the coatings by changing the bias voltage. The properties of the coating are characterized by investigations in the Scanning Electron Microscope (SEM), X-Ray Diffraction measurements and Nano-Indentation.
In addition to conventional 4-point bending fatigue experiments, in-situ 3-point bending experiments were conducted in the SEM to examine the initiation and growth of cracks in the CrN coating. The fracture surfaces of the Wöhler-specimens are analysed in the SEM to reveal the origins of the fatigue cracks. To identify elastic and plastic deformation at the point where the cracks are initiated, we use a special software package that enables us to determine strain maps of the in-plane strains at different loading stages of the in-situ experiments.

3D and 2D finite element simulations of the bending test are performed in parallel to resolve the local stress and strain conditions at the moment where the cracks appear. The concept of configurational forces is used to evaluate by post-processing the crack-tip shielding and anti-shielding effects that occur on the one hand due to the differences in Young's modulus and hardening coefficient of steel and CrN and, on the other hand, by the residual stress state. In order to simulate the behaviour of a growing crack, the crack length is varied between 1µm and 5µm for a given load stage.

The goal of our study is to clarify, how the fatigue resistance is influenced by hard coatings under residual compression, to analyse which factors affect the initiation and the growth of fatigue cracks, and how the coatings can be improved to increase the fatigue-life of engineering parts.

Analyses of Fatigue and Fracture on Micro-samples

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Abstract

The focused ion beam technique in combination with new micro- and nano-testing facilities enables the fabrication and the testing of samples with dimensions from 100 nm up to 100 microns. The combination of these two techniques opens a wide field for the improvement of the understanding of the mechanical properties of materials. Most of the activities in the last few years were devoted to size effects in plasticity. The main aim of the presentation is to show, what one can learn with these new technique in the field of fatigue and fracture. Different examples of micro-tension, micro-bending and micro-fatigue experiments will be presented.

Micro-mechanisms of Deformation and Toughening in Biological Materials

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Biological materials such as bone, antler or glass sponge skeletons require a high fracture resistance for their mechanical function. The primary reason is the hierarchical nature of the structure, which may be optimized for toughness from the nanometer to the macroscopic scale (1). The lecture reviews some of the recent work on bone and glass sponges, as well as some new approaches to describe crack propagation in layered systems.

Bone is a composite of collagen and calcium phosphate nanoparticles and there are toughening mechanisms operating at many different length scales (2), from the nanometer size of the mineral particles (3), to the interaction between mineralized collagen fibrils in the 100 nm size range (4-6), the lamellar organisation in the micron range (7) and the plexiform packing at the 100 micron scale (8). Most interestingly, the toughest bony tissue is deer antler, which has recently been shown to profit from nanoscale inhomogeneities in the fibrillar structure (9). Turtle shell, another type of bony tissue, uses an arrangement reminding the principle of interlocking composites to provide small stiffness at low loads, but high stiffness when the loads increase (10).

Glass sponge skeletons are even a more striking example where an inherently brittle materials, amorphous silica, becomes fracture resistant due to a tiny fraction of added protein matrix (11-14). The structural key is a multilayer structure of the silica glass, where concentric, jus a few microns thick silica layers in the glass fibres are separated by nanometer thick protein layers (11,12). Cracks are typically arrested at the interfaces between the layers (14). The structure may be described as a material where the elastic modulus varies periodically in the direction perpendicular to the layers (12). A recent theoretical treatment shows that such structures provide an effective barrier for crack propagation when the amplitude of the periodic Young's modulus variation is large enough (15).

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