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CLASSIFICATION OF THE MASTERCURVE CONCEPT WITHIN THE FRAMEWORK OF SAFETY ASSESSMENTS BASED ON MECHANICS OF MATERIALS

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ABSTRACT
Fracture mechanics based assessment methods are required for various purposes within the warranty of the safe operation of nuclear power plants components. Independent of the purpose of the analysis, it is a basic requirement that the material characteristics used to define allowable loadings are suitable to quantify reliably the considered material property. For fracture mechanics analysis under quasistatic loading different, phases of the failure process can be considered, i.e. initiation of crack growth, growing of the crack and subsequent instability. At low temperatures also a brittle failure without prior macroscopic deformation can occur. The failure process is strongly influenced by the material toughness and strength properties on the one hand and by the multiaxiality of the stress state on the other hand. Depending on which phase of the failure process has to be considered, different safety factors have to be called for.
During the last decade the so called mastercurve approach has been developed and suggested for many applications. This approach considers the cleavage failure instability, i.e. the event of major concern in all failure analyses due to the consequences of this failure type. On the background of the increasing applications of this approach it is felt to be necessary to point out the material mechanics basis of the failure processes described by the mastercurve procedure.
Within this presentation the different phases of the failure process in the transition range of fracture toughness are demonstrated on the example of a typical RPV steel. Material mechanics explanations of the failure process are given, also the modelling of the process by means of micromechanical approaches is shown.
Keywords: fracture mechanics, ductile crack initiation, cleavage instability, fracture toughness, transition range.

1 INTRODUCTION
In all branches of engineering and plant construction material characteristics are used for different purposes. In the following two categories "**qualifying material characterisation"** and/or "**quantification of specific material properties"** have to be distinguished.
The characteristics of "qualifying material characterisation" means the qualification of the material for a certain application purpose or to allow a comparative assessment of different materials. Material acceptance is a typical example. In tests must be proved if a certain minimum requirement is fulfilled. The transition temperature T_{27J} is a typical example. T_{27J} is determined in Charpy tests. To use a steel for a specific purpose, it's T_{27J} must be below a defined temperature. If this requirement is fulfilled, the material will be qualified for the intended application purpose. A quantitative component analysis, however, cannot be performed using the characteristic value T_{27J}. The minimum requirements are often based on the experience. The physical importance is subordinate.
The requirements for quantifying material characterisation are different. Here it is very important to determine characteristic values, which quantify reliably a certain material property. Thus, it can be used directly to predict quantitatively the component behaviour using calculative assessment methods. This means that the material characteristic value must be transferable to the component behaviour. Quantifying material characterisation becomes more and more important, like risk orientated monitoring and maintenance strategies get more and

more introduced or like loading optimised constructions become more and more necessary in view of obtaining ecologically and economically improved balances.
Quantifying material characteristic values cannot only be used for quantitatively assessing component analyses but also for the qualitative material characterisation. The reversion of this statement is not valid. Using qualitatively characterizing values in component assessments often lead to incorrect and sometimes also to overestimating results.
Consequently, it always must be questioned if the characteristic value used in component assessments meets the criteria of the “transferability” and thus, is able to correctly describe the behaviour of the component.
This is particularly applied to fracture mechanic characteristic values, as in many cases fracture mechanics characteristic values are the key role in assessing component failure, where transferability is very important. On the other hand, a large number of recommendations, guidelines, and standards concerning the determination of fracture mechanics characteristic values are available due to various developments in the last years. Their results have different meanings because of different approaches and simplifications. This has to be recognized when they are used for component behaviour prediction.
Against this background, the basics of materials physic of the failure process of ferritic steels will be observed subsequently in dependence on the temperature. Furthermore, the different fracture mechanics characteristic values available to characterise the failure process conditional on cracks, should be assessed with respect to the materials physical importance.

2 MATERIALS PHYSICAL BASICS OF THE FAILURE PROCESS OF FERRITIC STEELS

Fracture is an occurrence which stops deformation of materials. The material fails due to local or global separation. In the following, fracture processes taking place above the crystal recreation temperature (T > 1/6 TM) as well as dynamic fracture and fracture affected by corrosion should be excluded. Furthermore, the considerations are limited to metallic materials. Usually, failure of metallic materials can be characterised by means of macroscopic plastic deformation or with the help of fracture surface.
If fracture is described by means of **macroscopic** deformation, it must be distinguished between brittle and ductile fracture:

- **Brittle fracture** - If only little or no macroscopically plastic deformations can be observed until rupture we talk about brittle fracture. Only little energy is consumed until crack initiation and during fracture. Fracture growth is instable. Brittle fracture of ferritic steels may occur at low temperatures and/or at stress conditions with high multiaxiality.
- **Ductile fracture** - In contrast to brittle fracture, visible plastic deformation can be observed before and during fracture. Before fracture, crack initiation takes often place with subsequent crack growth. Crack growth is stable. Higher energies are consumed during stable crack growth.

Observing **microscopically,** it must be distinguished between cleavage fracture and ductile dimple fracture:

- **Cleavage fracture** - At cleavage fracture, cleavage fracture surfaces can be observed at adequate enlargement of the crack surfaces, fig. 1. During material separation no or negligible locally limited plastic deformation occurs. Cleavage fracture may also occur after higher macroscopic plastic deformation. The energy consumed for crack growth is very low. Usually crack growth is instable. The crack surface is perpendicular to the main stress. The crack surfaces are shining metallically .
- **Ductile dimple fracture** – Ductile dimple fracture of technical metals and alloys is characterised by dimple structures often with inclusions on the crack surface, fig. 2. Ductile dimple fracture is marked by its locally high plastic deformation. The energy locally consumed during material fracture is considerably higher as the energy consumed during cleavage fracture. At macroscopic ductile fracture, ductile dimple fracture is observed microscopically. However, microscopic ductile dimple fracture may have macroscopic brittle appearances for an extremely high stress multiaxiality e.g. at the crack tip. Usually crack growth is stable. Only for the high stress multiaxialities already described it may be instable.

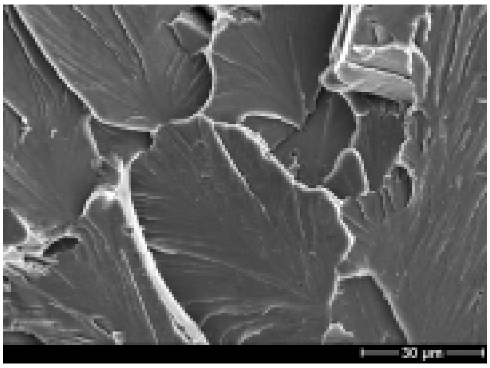


Fig. 1: Cleavage crack surface, material 17MoV8-4 mod., SEM-picture

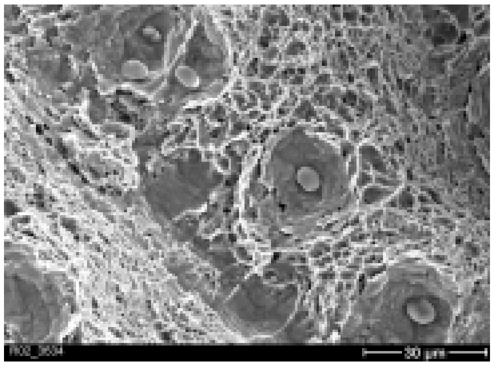


Fig. 2: Dimple crack surface, material 22NiMoCr3-7, SEM-picture

2.1 Basics of Cleavage Fracture

Transcrystalline cleavage fracture is the separation of atom bonding of crystalline solids along the defined crystalline plane. In inclusions and particles, as well as in ferrite, separation occurs primarily along the crystallographic cleavage planes, fig. 3. In the case of ferritic steels this corresponds to {100} - planes, fig.4. During failure due to cleavage fracture, the crack may change microscopically its direction when exceeding the grain boundary. This is due to the crystallographic orientation differences of the grains involved, fig. 3. Thus, the favourably orientated cleavage planes are activated. Macroscopically observed cleavage fracture grows perpendicular to the direction of the main stress.

Low temperatures, unfavourable material states due to manufacturing or service, multiaxial stress states or high loading velocity are parameters promoting cleavage fracture in materials with a body-centred cubic or hexagonal lattice structure.

Failure of specimens or components made of ferritic materials due to cleavage fracture is a multistage process where several barriers must be overcome. This process can be classified into three phases. Micro crack onset, micro crack growth through the grain and exceeding of the grain boundary. A micro crack occurs during loading e.g. due to an accumulation of displacements along the grain boundary or due to fracture of a two-phases-particle. These are mostly brittle particles such as carbides or oxides.

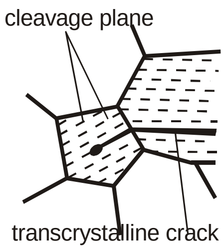


Fig. 3: Transcrystalline cleavage crack

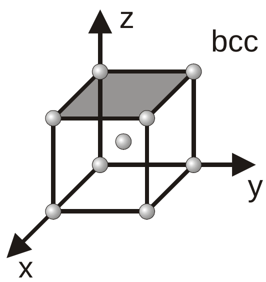


Fig. 4: Crystallographic crack plane in body-centered cubic lattice

Dynamic particle fracture may stop at the particle-ferrite grain boundary if the energetic conditions of continuing growth such as cleavage fracture are not given or if crack blunting takes place. The energetic condition of continuing growth depends on the principle stress, on the particle magnitude and also on the grain magnitude due to the additional impact of the shear stress on the crack initiating gliding plane.

Not only the particle-ferrite grain boundary is a barrier but also the following ferrite-ferrite grain boundary where a dynamically running micro crack may stop and then blunt. It seems that only the energetic condition is relevant to further growth. If the micro crack overcomes this barrier too, it usually will continue to grow unresisted. This microcrack will only stop again if there is a macroscopic inhomogeneous distribution of stress, temperature or structure.

As a growing micro crack causes increasing energy consuming plastic deformation which decelerates crack growth with increasing temperature it becomes more and more difficult to meet the requirements of instable or cleavage fracture-shaped crack growth during and after exceeding the barrier. The critical occasion in cleavage failure is the exceeding of the highest barrier. This barrier depends on the structure, temperature and stress multiaxiality.

2.2 Basics of Dimple Fracture

Micromechanics procedures at ductile dimple crack initiation are usually affected by the micro structure. Therefore, precipitations and inclusions are of particular importance. Basically the magnitude, quantity, chemical composition, distribution and shape of this second phases manage the micro structural procedures which cause crack initiation.

In nearly all technical metals and alloys such a second phase exists as inclusions or precipitations. Magnitude, distribution and the shape of a second phase depend strongly on the production process. Investigations show that with increasing volume fraction of such second phases fracture toughness decreases.

The micromechanical procedures causing crack initiation in alloys having such a second phase can be divided into three phases, fig. 5.

- **Formation of voids**. The first phase of crack initiation, the formation of voids takes place
  - at particles of a second phase
  - at grain boundaries and grain boundary gussets and
  - in the perlit at steels with high C-concentration

  However numerous investigations have shown that void formation is mostly initiated at particles of the second phase. It can be observed that particles with a spicular or ramified shape mostly fail due to fracture, fig. 6, whereas in the presence of spherical inclusions decohesion between particle and matrix form voids, fig. 7.

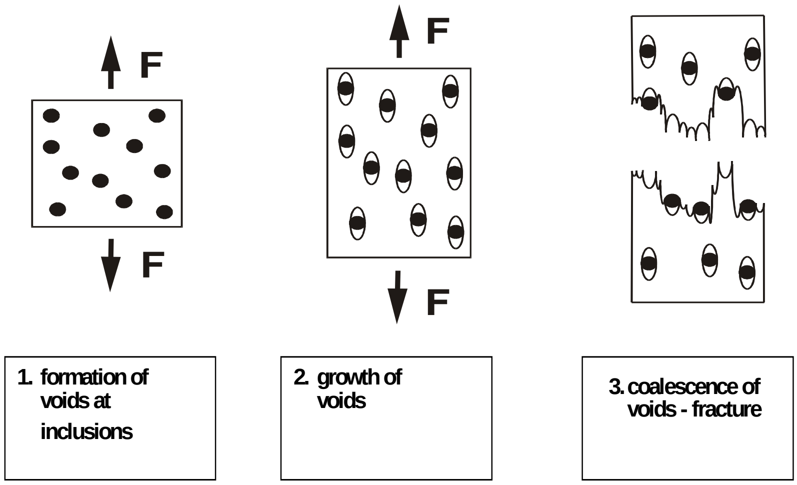


Fig. 5: Micro mechanic stages of ductile dimple fracture

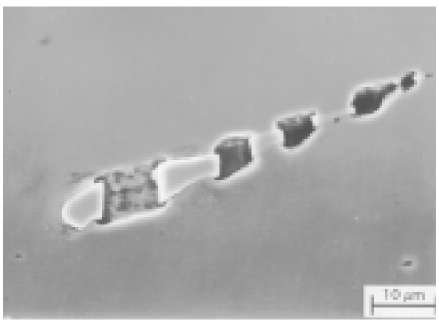


Fig. 6: Formation of voids at a Niobium-carbonitride inclusion, material 1.4550, SEM-picture

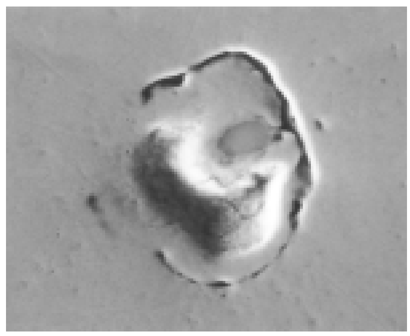


Fig. 7: Formation of a void at a MnS- segregation, material 20MnMoNi5-5, SEM-picture

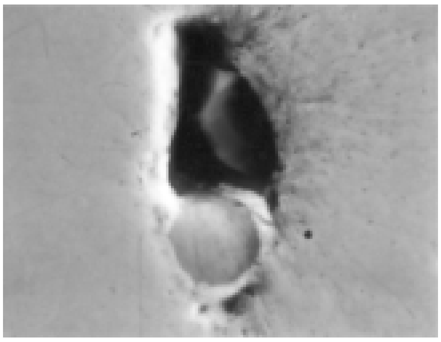


Fig 8: Enlarged void at a MnS segregation, material 20MnMoNi5-5, SEM-picture

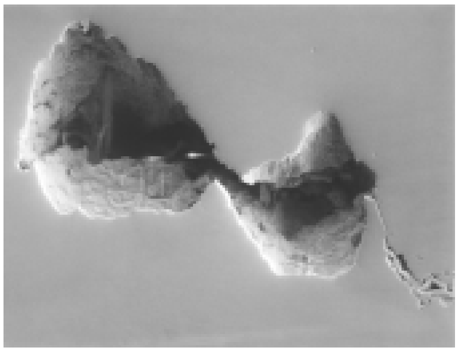


Fig. 9: Coalescence of voids, material 20MnMoNi5-5, SEM-picture

- **Voids growth**. With increasing plastic deformation the voids developed at the inclusions start growing, fig. 8. The growth rate and the shape of the voids are affected by plastic deformation and the multiaxiality of the loading condition. It can be observed that the growth rate of the voids developed at large inclusions is considerably higher than at small particles.
- **Coalescence of voids**. Voids do not grow arbitrarily. From a certain magnitude of the voids a coalesce is detected, fig. 9. Three mechanisms can be observed primarily (shear band formation, formation of small secondary voids, constriction of the material bond bridges between voids) where the individual mechanisms may superimpose. This coalescence of voids causes a small crack.

2.3 Rupture Process at Crack Tips

During loading of a cracked component a strong stress superelevation can be observed at the crack tip. This superelevation may lead, already at low rated nominal stress to the fact that the above-mentioned failure mechanisms are able to act . In order to figure out which one of the failure mechanisms can be expected, the material temperature has an important role.
The fact that it is impossible to measure directly the acting macroscopic failure mechanisms is an additional difficulty in experimental fracture mechanics material characterisation. These mechanisms have to be determined after testing using fractographic and metallographic features. This is a very time-consuming process. Furthermore, only the macroscopic deformation magnitudes (crack opening, load line displacement) can be determined during testing.

2.3.1 Upper Shelf of Toughness

In the case of ferritic steels in the upper shelf, loading of a cracked body causes a plastic zone in the highly stressed area of the crack tip. This plastic zone grows super-proportionally and covers more and more volume of the material. In these plasticised areas the above-mentioned processes such as the formation, growth and coalescence of voids takes place. The extend or progress of these processes increases with the local extend of macroscopically visible plastic deformation.
This firstly leads to a measurable blunting of the originally sharp crack tip. Further loading what e.g. means further formation and coalescence of voids, tears the surface of the blunted crack tip, see fig. 10. Continuing the process of growth and coalescence of voids due to further movement of the highly plasticised areas into the ligament finally leads to distinct crack growth. The respective fracture surface areas are clearly marked by dimple fracture. However, the stretched-zone (extremely stretched dimples) of the original blunting area at the crack tip differs from the areas of stable crack growth (regular dimple structure), fig. 11. This kind of crack growth is called stable crack growth as long as the input of further energy (increase of load and/or deformation) is necessary to propagate the crack and without this, crack growth will stop. The crack propagation is called instable if it grows without further input of energy until it penetrates the remaining wall thickness.
This mechanism shows macroscopically an elastic-plastic load-crack opening-behaviour (Load-COD) fig. 10. It has to be said, that the material plasticise in the crack tip area already at low loading where the global load-crack opening behaviour appears still linear elastic and doesn't show any signs of plastic deformation. Inverse it can be assumed, that when load-crack opening behaviour shows an over-elastic behaviour at least in the area near the crack tip, distinct plastic deformation (blunting) has occured and in some cases crack growth already took place.

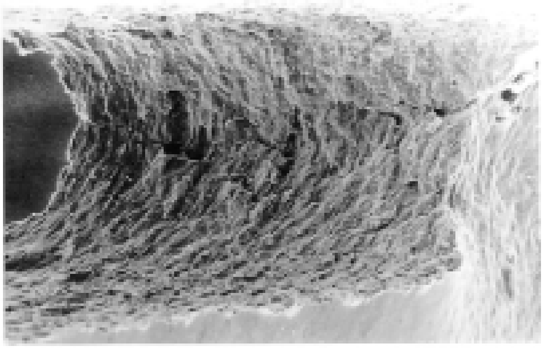


Fig. 10: Blunting before onset of ductile tearing of the crack (acc. to Blumenauer)

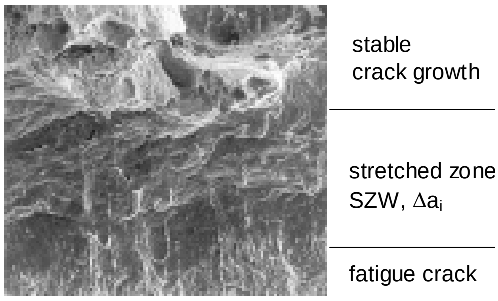
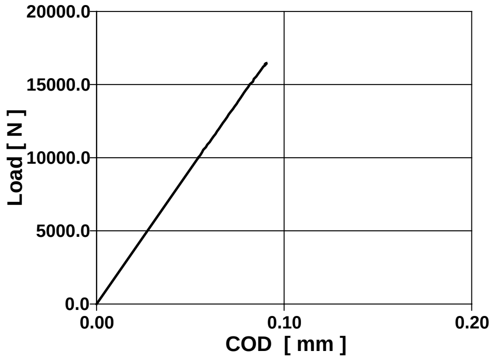


Fig. 11: Crack surface after ductile crack initiation and stable crack growth (SZW and dimple crack surface)

2.3.2 Lower Shelf of Toughness

Also in the case of ferritic materials in the lower shelf, the highest loaded area occurs at the crack tip when loading a cracked body. As mentioned in section 2.1 failure due to cleavage fracture is a multi-stage process where several barriers must be overcome. The criteria are affected by the fact that a minimum stress level must exist over a specific (small) area at the crack tip in order to trigger cleavage fracture. According to section 2.1, reaching the critical barrier depends on the structure, stress/loading multiaxiality and temperature. In the case of fracture mechanics tests using a certain material with a certain structure and standardised specimens with

identical stress multiaxiality reaching the critical barrier to trigger cleavage fracture only depends on the temperature.
This means that performing fracture mechanics tests on specimens with a constant geometry but with variable test temperature, cleavage fracture becomes more and more unlikely at increasing temperature. This is due to the fact that a minimum stress level required for cleavage fracture cannot be reached any more because stress peaks at the crack tip are reduced due to plastic deformation before reaching the necessary magnitude and distribution. Stress hardening may cause a stress state also after limited plastic deformation which promotes cleavage fracture due to the combination of a critical stress level and weak spots in the structure.



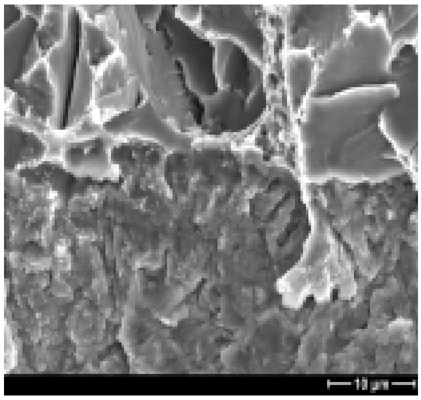


Fig. 12: F-COD and fracture surface of the specimen C(T)25 CMBA1 at lower shelf (-150°C), conditions for KIc-validity are met

In many caseseven at very low temperatures small deformed areas can be determined between the fatigue crack front and the cleavage fracture area, sometimes only locally restricted, fig. 12. This deformed sections are to be regarded as the local beginning of a stretched zone. The corresponding global load-crack opening characteristic determined in fracture mechanics tests is nearly linear and the small plastic deformations close to the crack tip area are usually neglectable in the F-COD-diagram.

2.3.3 Transition Area of Toughness

In view of fracture mechanics analysis, the temperature range in which usual fracture mechanics specimen fail after more or less macroscopic plastic deformation due to instable crack growth (instability) is called transition area.
For a clear definition for the transition region two terms are of importance:

- “Macroscopic plastic deformation”, to distinguish the transition region from lower bound and cleavage fracture and
- “instable crack growth” to distinguish between the transition region and the upper shelf.

1. Macroscopically measurable plastic deformation must take place. According to the above-mentioned statements, this means that similarly distinct plastic deformation must exist in the area of the crack tip which can be measured macroscopically, e.g. as a plastic part of COD. Plastic COD means also, that a stretched zone of a certain magnitude was formed.
2. Instable crack growth means that the specimen fails without any further energy input. This crack may also occur after similarly distinct stable crack growth. The question how distinct plastification and crack growth are at the given temperature, depends strongly on the specimen’s geometry and magnitude. That means on the respective stress state in the specimen.

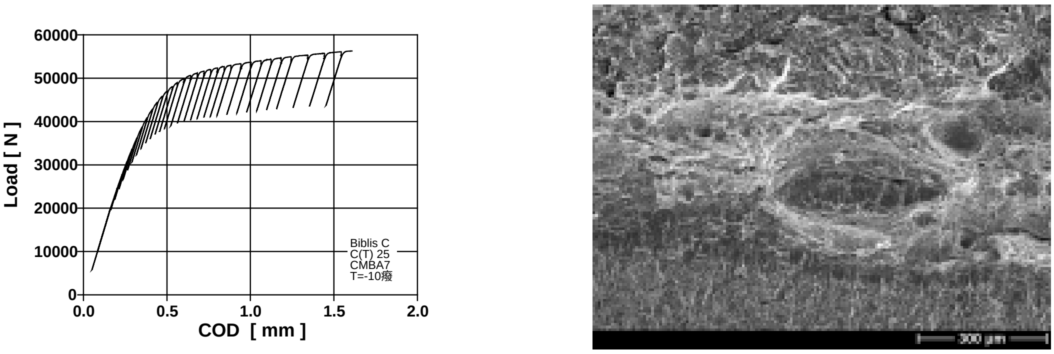


Fig. 13: F-COD and fracture surface in the upper transition region, C(T)25 CMBA7 (-10°C)

From the two conditions it results inevitable, that specimens which are assigned by this definition to the transition area, show temperature dependent more or less distinct stretched zones, followed by ductile crack growth (dimple area) with subsequent instability (mainly cleavage fracture), fig. 13.

3. FRACTURE MECHANICS MODELS TO DESCRIBE FAILURE PROCESS

The use of fracture mechanics models to describe failure and to assess safety margins of cracked components respectively, is based on the fundamentals of strength analysis:

acting load < allowable load

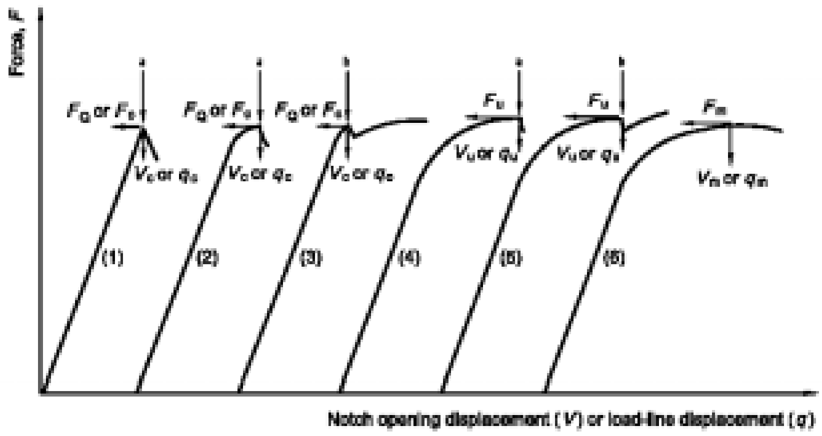
In the classic strength analysis, the acting load is quantified as accurate as possible using a suitable assessment method. Allowable load is defined as quotient of tolerable load (defined using measured material characteristic values) and a safety margin. Load is described by a suitable parameter e.g. stress.

In fracture mechanics analyses, stress is not suitable as a parameter due to the singularity at the crack tip. Therefore, different parameters were created due to the different deformation behaviours of steels. The parameters predominantly used are the stress intensity factor K_I and the J-integral. Further parameters to characterise the procedures at the crack tips such as crack tip opening displacement CTOD or different modifications (COD, δ) can also be used instead of the parameters mentioned above. But these parameters are less common despite their advantages.

To permit using the most common parameters K_I and J, it is necessary to define validity margins. These validity margins result from the fact, in the strictest sense, the stress intensity is only valid for linear-elastic stress states. The J-integral is valid for over-elastic loading, assuming nonlinear-elastic material behaviour, that means assuming the validity of the deformation theory.

Because of these limitations concerning the validity of parameters a definition of the validity criteria regarding the determination of corresponding material characteristics must be found. Thus, numerous recommendations, guidelines and standards were made.

The validity criteria of different characteristic values defined for this aim, are firstly orientated on the macroscopic deformation characteristic of the specimens, measured as force – crack opening displacement (F-COD). Brittle behaviour, that means non-deforming or low-deforming behaviour until fracture (brittle fracture) allows the use of stress intensity K_I as a parameter. Over-elastic deformation behaviour requires the J-integral as the parameter to be used.



NOTE 1 F_Q is the maximum force used in the determination of a provisional K_{Ic} (see Figure 16).

NOTE 2 F_c, F_u and F_m correspond to either δ_c, δ_u and δ_m respectively, or J_c, J_u and J_m respectively.

NOTE 3 Pop-in behaviour is a function of the testing machine/specimen compliance and the recorder response rate.

a Fracture.

b Pop-in.

Fig. 14: Illustration of possible load-displacement behaviours /1/

In some cases when macroscopic over-elastic deformation takes place before fracture (ductile fracture), cleavage fracture may occur after more or less ductile crack growth (dimple crack growth). This behaviour is mainly found in the transition region of toughness. Thus, the upper shelf of ductility is characterised by the fact that an existing crack does not fail directly during loading, but blunts due to plastic deformation, tears and growths stable due to plastic deformation. This means that the crack stops and does not fail without further energy input. In this case the value at the beginning of stable crack growth is used as fracture mechanics characteristic value and not the value at cleavage as used in the lower shelf and in the transition region. The fracture surface of stable crack growth is predominantly characterised by dimple fracture.

A summary of the different possible force-deformation behaviours described in /1/ is shown in fig. 14. Curve 1 describes the force-crack opening characteristic of brittle fracture. Curves 2 to 5 are corresponding for the transition region. Curve 6 is typical for upper shelf behaviour.

In the past, an individual test standard regarding the determination of the corresponding characteristic values was developed for every failure type, e.g. the ASTM-standards. Here, standard ASTM E 399 /2/ was only designed for brittle fracture, standard ASTM E 813 /3/ for ductile fracture and standard ASTM E 1921 /4/ for the transition region. Recent standards, such as DIN ISO 12135 /1/ and ASTM E 1820 /5/ are designed for all fracture mechanics parameters.

Not only the above-mentioned main characteristics or validity characteristics of the force-deformation behaviour and crack surface morphology must be met, but several other validity criteria, may vary from one standard to the other. An overview and resulting requirements are given in the following.

3.1 Brittle Fracture / Fracture Mechanics Characteristic Values at Linear-Elastic Behaviour

In the past, there was a consensus on the interpretation of the test results in the case of brittle fracture without relevant previous plastic deformation, compare curve type 1 in fig. 14. The load at the onset of cleavage can be used to determine the K_{Ic}-values, if the requirements from the standards (e.g. ASTM E 399 /2/, ESIS P2 /6/, ISO 12135 /1/) are fulfilled. These requirements are requirements concerning e.g. the specimen geometry and size, the regularity of the crack front and further details of the test. The most important magnitude is the specimen thickness B.

Fulfilling these criteria, it was assumed that the characteristic values determined, are extensively independent of the specimen geometry and size. These characteristic values are the basis for brittle fracture curves in several codes, e.g. the KTA /7/ or the ASME Code /8/.

In recent work, e.g. /9/, where the weakest-link theory is considered, it is assumed that K_{Ic}-values depend on the length of the crack front of each specimen and must therefore be size corrected, accordingly. However, corresponding recommendations for a change in standards are still discussed and not yet implemented in valid standards.

3.2 Fracture Mechanics Characteristic Values during distinct Elastic-Plastic Behaviour

In the case of pure ductile fracture, a crack resistance curve is usually determined with the help of the F-COD-diagram of the specimen observed, compare fig. 15. The crack initiation values can be deviated from this curve, according to the given criteria.

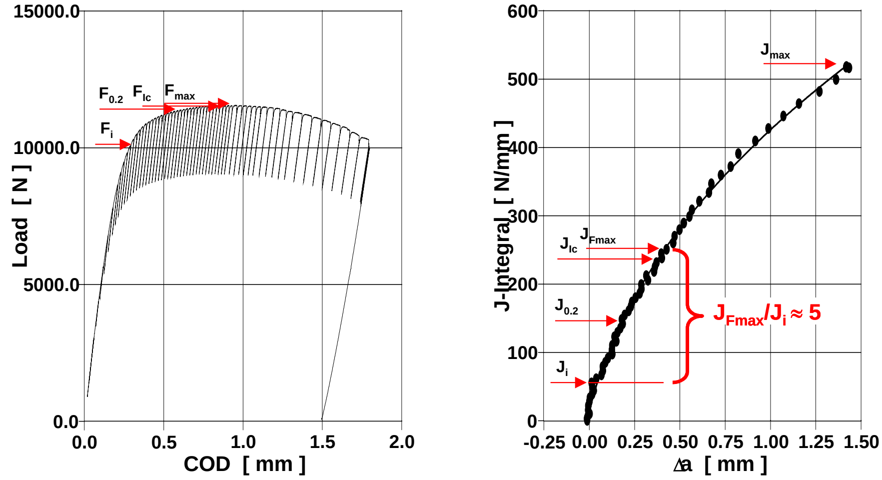


Fig. 15: Course of F-COD and the resultant JR-curve with different initiation values

The J-integral is mainly used as a parameter to characterise loading in the crack tip area. In the past, the procedure to determine crack initiation values of ductile initiation was revised and re-approved several times. In doing so, improved data acquisition and processing possibilities were taken into account. Today, numerous standards, codes and recommendations to determine fracture mechanics characteristic values exist due to these national and international standardisations e.g. /1,5,6/. All codes describe nominally a crack initiation value, that means the value where an existing crack starts to grow at further increase of load. The difficulties result from are different opinions of the interpretation of further growth of a crack and the necessary resolution of the measurement technique used. Finally, the different approaches in the different standards lead to the determination of characteristic values, according to these standards, which differ extremely from one standard to the other, c.f. fig. 15.

Against this background, it must be distinguished between qualifying characteristic values (e.g. J_{Ic} according to ASTM E 1820 /5/, $J_{0,2}$ and $J_{0,2/bl}$ according to ESIS /6,10/) and quantitative, transferable material characteristic values, used in fracture mechanics safety analyses. Such a quantitative transferable material characteristic value is the value J_i according to /1/ or /11/ (determined with the help of the potential drop technique or by measuring the stretched zone width, /1/).

If the extend of crack growth is higher than the extent of the stretched zone (Δa > SZW or $J > J_i$), the crack resistance curve (J_R-curve) represents no characteristic value or material characteristic. Besides of the influence of the material toughness, the course of the J_R-curve is strongly characterised by the specimen geometry, specimen size and the type/kind of loading. This finally leads to a dependence on the multiaxiality of the stress state. Therefore, the crack resistance curve (and also characteristic values of the area of crack growth higher than the SZW extension) is not directly transferable on another specimen or component geometry. This requires basically an analysis of the multiaxiality of the specimen's or component's stress state /12/.

In case of such distinct elastic plastic behaviour, also the determination of a **characteristic value** for failure (fracture) is not possible, since fracture depends on the type of testing, the specimen geometry and the size chosen, /12/. To determine a valid crack resistance curve according to the standard, the specimen should not crack instable but a distinct amount of stable ductile crack growth must take place before unloading. At the same time, the maximum value of J-Integral should not exceed J_{limit} (characterised by the specimen geometry and the material strength) as otherwise the J-integral is not valid any more.

3.3 Fracture Mechanics Characteristic Values at limited Elastic-Plastic Specimen Behaviour (Transition Region)

In the transition region of ductility, specimens of ferritic steels fail due to cleavage fracture after more or less plastic deformation and limited stable crack growth depending on the temperature, the specimen size and the multiaxiality of the stress state. Thus, two characteristic events to describe failure in this area can be used.

First, an existing crack starts growing after loading (ductile initiation). In doing so, the criteria correspond to these of the upper shelf of toughness. Secondly it is the cleavage event (instability).

3.3.1 Crack Initiation Characteristic Values

The lower the testing temperature, the less is deformability of the material in comparison to the upper shelf, c.f. fig. 16. This causes a sudden change from stable crack extension into instable crack extension during the fracture mechanics test in the transition temperature range. In the upper shelf and in the upper transition region, fracture may occur after exceeding the maximum load point. The extent of stable crack growth attained is considerably higher than the extent of the stretched zone. A distinct crack resistance curve can be determined where the curve is approximated by all J-Δa-points. It may be the case that the calculated crack extension must be adjusted to the crack extension measured on the crack surface.

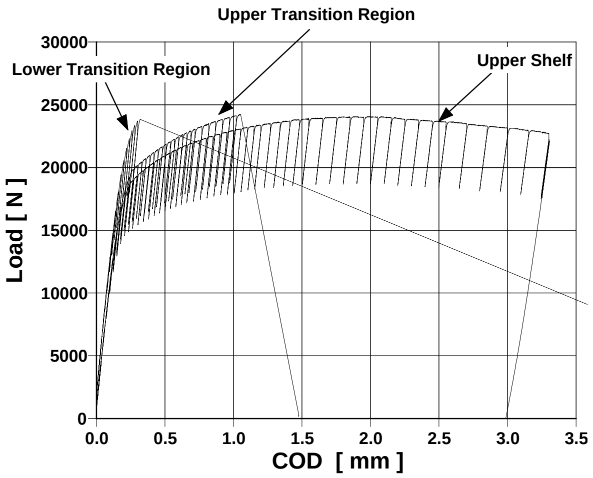


Fig. 16: Load-deformation behaviour of fracture mechanics tests in the upper shelf and in the transition region of the material toughness

A further decrease in temperature causes sooner instable cleavage fracture. Plastic deformation energy and therefore the J-integral value at fracture become lower. Also crack tip blunting until the beginning of ductile crack growth becomes lower. Therefore also the J-integral value at crack initiation becomes lower. Furthermore, ductile crack extension between initiation and instability decreases. That means that the J-Δa-points of the curve approximation reduce. Thus, the course of the J_R-curve becomes less clear due to additional procedure dependant inaccuracies and the already existing materials scatter. These inaccuracies directly affect the initiation value and may be partly compensated using precise measurement methods.

3.3.2 Instability Values

In general the onset of cleavage is preferably used to determine characteristic values, as there are many advantages: instability values are easy to determine experimentally as there are relatively few requirements for testing. To determine instability values in the transition region of ductility different standards and codes are available, e.g. ESIS P2-92 /6/, ASTM E 1921 /4/, ISO 12135 /1/.

The "characteristic value" J_C at fracture (instability due to cleavage fracture) is determined according to ASTM E 1921 /4/. Besides numerous other validity criteria, e.g. regarding crack geometry and loading during fatigue, a further criteria must be fulfilled. The fracture surface must predominantly show cleavage fracture. A small extend of dimple fracture occurring before cleavage fracture (that means ductile crack growth before changing into cleavage fracture) is tolerable. This failure behaviour may occur on fracture mechanics specimens at extremely low macroscopic deformation (almost brittle fracture), but can also be observed at higher plastic deformation (ductile fracture). This is the reason why in these test standards the validity criteria are not based on the macroscopic deformation criteria but on micro-fractographic features. The ASTM standard E 1921-03 for example, allows 0.05(W-a) or maximal 1 mm of ductile crack growth (that means ductile dimple fracture) before changing into cleavage fracture. Other standards, such as ESIS /6/ apply the same criteria but other values are mentioned (ESIS allows for J_c only ductile crack growth of maximal 0.2 mm before changing into cleavage fracture).

It is known and in the meantime largely accepted, that these fracture values J_C are affected by different variables. The most important one is the impact of the stress state which strongly affects the failure behaviour after exceeding the physical crack initiation value. Thus, all parameters which are affected by the multiaxiality of the stress state also affect the J_c-values. These are for example the specimen geometry and size, crack geometry, type of loading and the materials strain- and hardening behaviour. The plastic specimen volume plays also a dominant role in actuating cleavage fracture after ductile crack initiation assuming, that the Beremin-model is valid /13/ (weakest link model assuming a Weibull-distribution): the larger the highly-stressed material volume, the more likely is the presence of a cleavage fracture causing initiation spot in the volume.

Generally, these facts are not taken into account in standards concerning the determination of characteristic values (ESIS, ISO). The parameters affecting the fracture values mentioned are not taken into account. This is only the case in standard ASTM E 1921 where the impact of the crack front length (equal to the specimen thickness) is considered under a statistic point of view (probability to cause cleavage fracture). Approaches in order to consider the impact of multiaxiality of the stress state on fracture in the transition region is still investigated. But they are not included in standards yet.

In the following the procedure of standard ASTM E 1921 is examined in detail, because of the importance of these material values and the assessment methods based on them.
Standard ASTM E 1921-03 primarily instructs the determination of the transition temperature T_0. It bases on the description of instability values of ferritic steels in the transition region by the Master Curve according to Wallin. In doing so it is assumed that fracture probability can be described by a three parametric Weibull distribution. Two parameters are fixed, where one is the temperature independent lower bound value K_{min} = 20 MPam ^{½}. The course of the Median curve $K_{Jc(med)}$ (T) (Master Curve) with the temperature and the corresponding fractiles are defined in the standard. Only the reference temperature T_0 has to be determined in material testing. This is the temperature at which the J_c-value of the Master Curve is 100 MPam^{½}.
Using standard ASTM E 1921 two topics must be taken into account. The first relates to the testing method and the observation of the validity criteria of each individual specimen to obtain valid instability values according to this standard. The second relates to the mathematical use of instability values to fit the Master Curve and to determine the reference temperature T_0.

- **Criteria for the individual specimens:**

SE(B) and C(T) of different size but proportional dimensions are the most common used specimen geometries in fracture mechanics. Standard ASTM E 1921 /4/ recommends these specimens as equivalent. Testing small scale specimens with a cross section of 10x10 is allowed as long as certain further criteria are fulfilled. In fatigue pre-cracking different criteria must also be fulfilled. In doing so, maximum fatigue load is very important because in contrast to the standards of upper shelf ductility it is considerably reduced and requires numerous load cycles. This leads to time consuming pre-cracking. Further criteria concern the regularity of the shape of the crack front. If they are not fulfilled the test results are not valid.
It is allowed to test the specimen until cleavage fracture using monotonic increasing loading or the method of partial unloading. The loading rate factor according to the standard may differ by a factor of 100.
Further validity criteria concern the fracture surface. Cleavage fracture must cover nearly the entire fracture surface. Ductile dimple fracture (ductile fracture) as a sign of stable crack growth before cleavage is tolerated but only up to an extension of 5% of the ligament or 1 mm.
Furthermore, it must be assured that the J-value at fracture does not exceed a certain maximum. Thus, not tolerable plastification of the specimen should be excluded. Length of the ligament and the temperature dependant material values Youngs modulus and yield stress also affect this validity criterion.
If the J-value at cleavage fracture becomes higher than the defined “limit-value”, the test result can still be used to determine the temperature T_0. In doing so, the experimentally determined J-value at cleavage fracture is replaced by the calculated “limit-value”. This procedure is called “censoring”.
J_c-values are converted into K_{Jc}-values using the relation for plane strain condition. Here is also the Youngs modulus at the corresponding test temperature required.
Finally the values are adjusted. This is often called crack length correction what is misleading. This is to take into account the increasing probability for specimens with a longer crack front (thicker specimens) to meet a cleavage causing weak spot according to the weakest link theory. Therefore, thicker specimens reach lower cleavage values than smaller specimens with shorter crack fronts. A specimens thickness of 25 mm is the standard thickness. This means, that to determine the T_0 results of specimens with a shorter crack front are corrected to lower values, results from specimens with a longer crack fronts are corrected to higher values.

- **Criteria for the data set:**

To determine the reference temperature T_0 of the Master Curve according to ASTM 1921 /4/ using the (crack front corrected) instability values K_{Jc}, a minimum number of valid test results within a defined temperature range relative to T_0 must exist.
The K_{Jc} data can be determined at one temperature or at different temperatures, fig. 17 and 18. According to the maximum likelihood-method, a curve with a fixed form (Master Curve) is adjusted by approximating the parameter T_0 to the measured data.

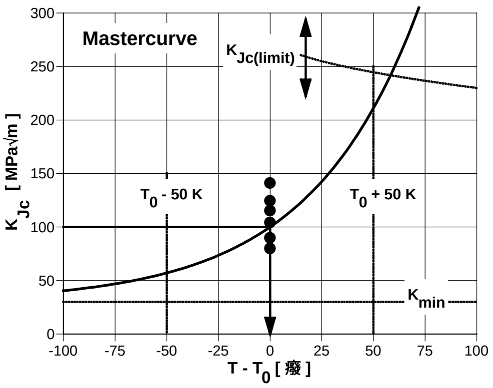


Fig. 17: MC, 6 specimen, T=const.

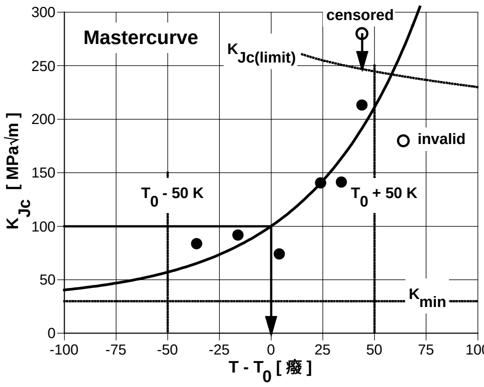


Fig. 18: MC, tested at different temperatures

It is assumed that each material basically behaves according to the Master Curve (and its given scatter. Therefore, the experimental results are only needed to adjust the position of the Master Curve on the temperature axis. The form of fracture toughness versus temperature and its scattering behaviour are fixed for all considered materials. At least 6 valid specimens are necessary to determine the temperature T_0 and therefore to adjust the curve. At unfavourable temperatures in relation to T_0 even more specimens may be necessary.

The specimens which are actually not valid (censored, because higher than the maximum value J_{limit}) are also taken into account by a special procedure. The basic idea of the procedure is that the censoring (reducing the values to the J_{limit}-value) should not adulterate the position of the Master Curve versus T_0.

Only values within the range $T_0 \pm 50$K to the determined temperature T_0 are valid for the determination of T_0. This criterion causes an iterative process for certain data sets. In the second iteration the results beyond the validity range of the first iteration are not considered. Then it may be the case that comparatively large data sets do not contain sufficient valid specimens any more even when all other validity criteria are fulfilled.

From the present state of the master curve it can be concluded, that the Master Curve approach is very attractive for application because:

- The reference temperature T_0 is determined from quasi-static fracture mechanics testing.
- Results of small scale specimens can be used. These specimens must only fulfil the requirements of the standard for K_{Jc} and not the severer requirements for K_{Ic}.
- The results from fracture mechanics testings are regarded directly, the use of results from other test types (e.g. Charpy test) is not necessary.
- Experimental data (at least 6 valid specimens) may be used, tested at a constant temperature near T_0 or distributed within the temperature range $-50K \leq T_0 \leq 50K$.
- Irradiation conditioned shifts of the reference temperature T_0, and therefore of fracture toughness curves, can be determined with irradiated material by fracture mechanics tests (e.g. pre-fatigued Charpy-specimens)

Despite of these numerous attractive points of view, several points have to be discussed and clarified before a common use of the concept is possible. On the one hand, this concerns the testing procedure used in the standard, e.g. the iteration procedure to fulfil the ± 50 K criterion or questions about the consideration of non valid test results by applying the censoring procedure. Furthermore, it should be clarified that the results are only valid for the observed specimen geometries and crack front length. They cannot be transferred to specimens and components with another multiaxiality of the stress state. Today this is known and generally accepted. Proposals were made to take into account the effect of multiaxiality. Some proposals are even approved by several experimental test series, but from the authors view a generalisation is not yet available.

4. EXPERIMENTAL RESULTS OBTAINED IN THE TRANSITION REGION

Within the scope of the research project of BMWA /14/ a sufficient number of specimens were tested in the transition region in order to perform statistic analyses of the ductile crack initiation in the transition region. Therefore fracture mechanics specimens with different geometry and size manufactured from a RPV steel were tested. This RPV steel type 22NiMoCr 3-7 was taken from a pressure vessel of a PWR, produced with original manufacturing technology.

The specimens were tested according to the method of partial unloading. Then ductile crack initiation, ductile crack growth and cleavage instability were analysed.

Fig. 19 shows specimens tested in the transition temperature region and analysed according to ASTM E 1921-03. Temperature T_0= -72°C is obtained, using all results and applying the censoring procedure according to ASTM E 1921-03. The limitation on specimens tested within $T_0 \pm$ 50K was not taken into account. With the specimen geometry and size used in this project, cleavage fracture could not be obtained at a temperature above –10°C. Thus, no conclusion can be made for the temperature range above –10°C using this procedure.

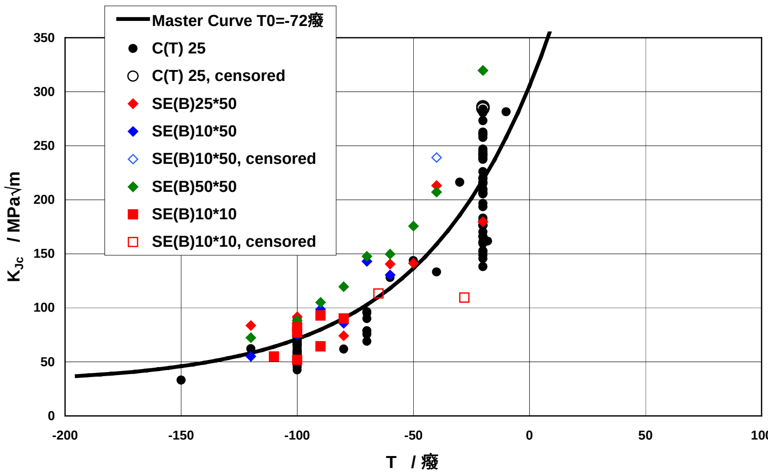


Fig. 19: Application of Master Curve procedure on instability values KJc, 22NiMoCr3-7, /14/

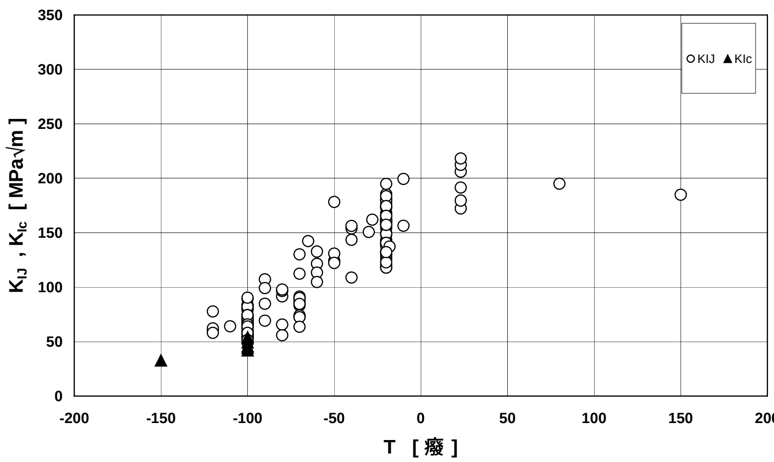


Fig. 20: KIJ (Ji) and KIc for material 22NiMoCr3-7 , /14/

The results of analysing these tests in respect of ductile crack initiation, according to the above mentioned assessment method, are shown in fig. 20. A temperature curve develops which is similar to the Charpy-temperature curve. In the entire temperature range from the lower shelf, the transition region to the upper shelf a common characteristic value exists which is of identical physical meaning. Furthermore it can be observed that K_{IJ}-values of the transition region and valid K_{Ic}-values complement one another at low temperatures. A similar curve is obtained showing SZW-values versus temperature, c.f. fig. 21. Sometimes the occurrence of ductile crack initiation and the associated processes, such as blunting, SZW-development and subsequent ductile crack growth is questioned particularly in the transition range of ductility. Sometimes even the influence of the temperature of this process is questioned. Thus in order to prove the failure behaviour in the transition range, several specimens were not loaded until cleavage fracture. They were unloaded when ductile crack initiation was assumed and before cleavage instability was obtained. These specimens were examined in view of ductile crack initiation and growth before instability with the help of metallographic and fractographic procedures. Therefore, these specimens were cut lengthwise into two halves. One half of each specimen was polished metallographically to show crack tip blunting and crack growth. The other half was bared by cyclic fatigue loading to search for dimple fracture areas (stretched zone and stable crack growth). Fig. 22 shows the K_J-values obtained before unloading, the scatter bands of K_{Jc} (cleavage fracture) and K_{IJ} (ductile crack initiation).

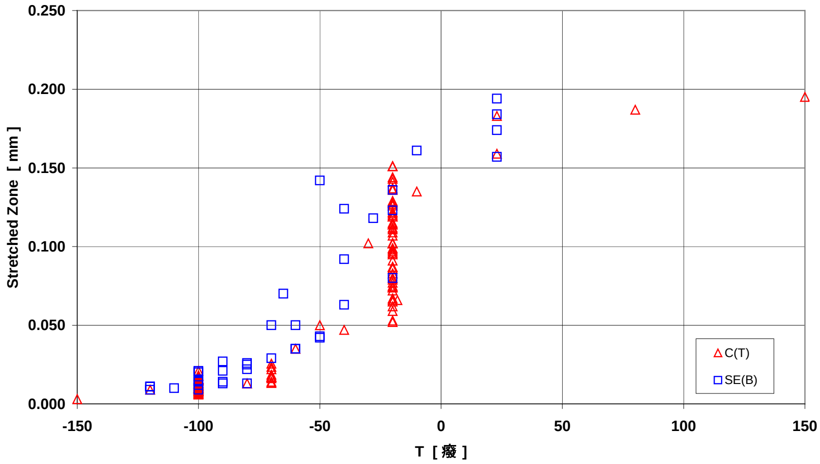


Fig. 21: SZW for material 22NiMoCr3-7, /14/

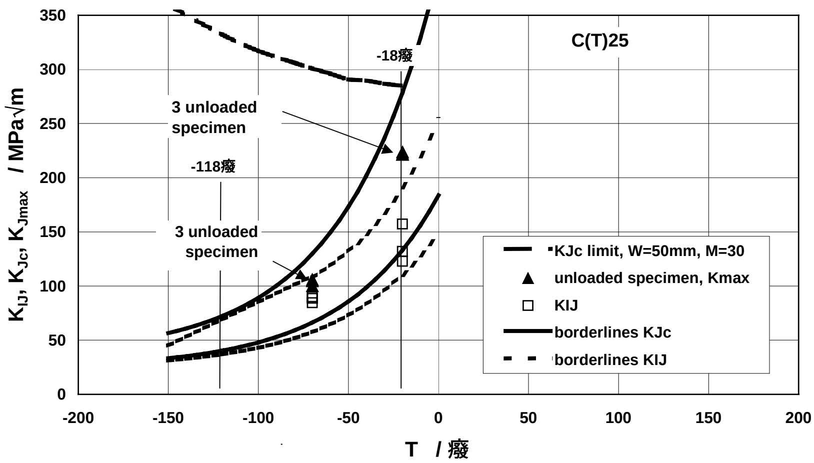


Fig. 22: Delineation of unloaded specimen in the scatterbands for crack initiation , initiation KIJ and cleavage instability KJc , /14/

The specimens tested at T = -20°C were just unloaded above the scatter band of crack initiation K_{IJ}. The specimens tested at T = -70°C were unloaded in the upper region of the scatter band of crack initiation K_{IJ}.

Fig. 23 to 26 show the test results of an unloaded specimen tested in the transition range at -20°C. The force-crack opening diagram and the J-Δa-curve show the typical behaviour of the transition region of ductility. Both the SEM-picture of the fracture surface and the polished cross-section show that the above-mentioned failure mechanism (blunting, forming of SZW, ductile initiation, ductile crack growth) occurs before cleavage fracture instability. The number and distribution of the J-Δa-points do not show a clear curvature of the J_R-curve yet. Therefore the J_R-curve is determined by a straight line through origin and $J_{max}/\Delta a_{max}$ (measured on the fracture surface). The J_i-value results from the intersection of the straight line with the stretched zone.

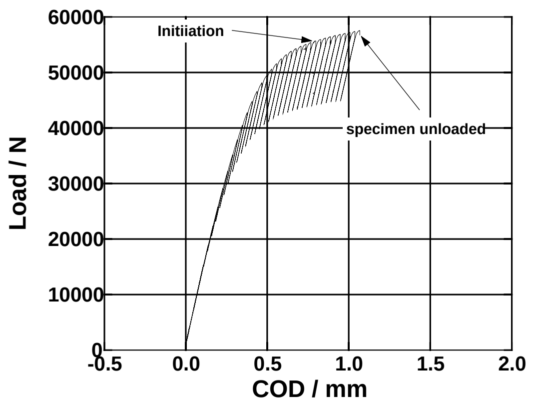


Fig. 23: F-COD-diagram (C(T)25 specimen CMACE13)

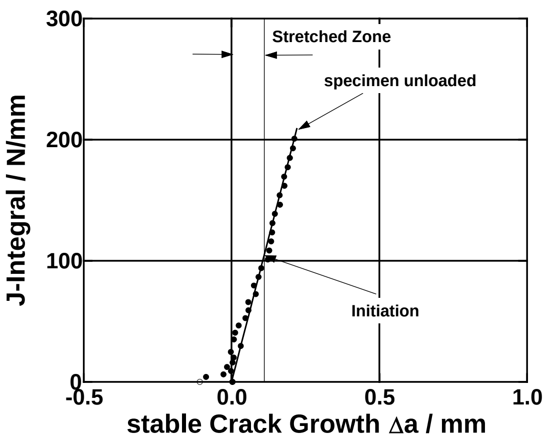


Fig. 24: JR-curve in the upper transition region (C(T)25 specimen, CMACE13)

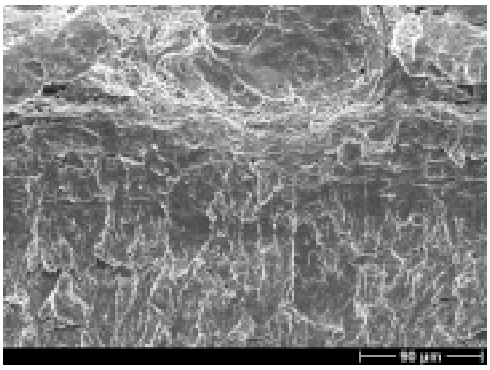


Fig. 25: SEM-exposure of the fracture surface (C(T)25 specimen, CMACE13)

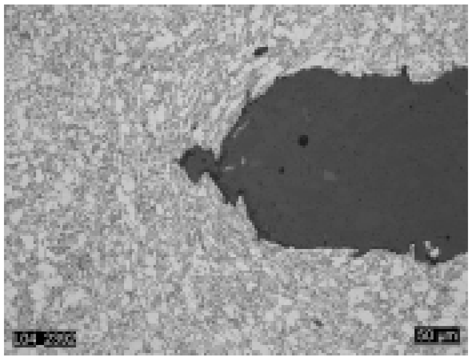


Fig 26: Polished micrograph, section of C(T)25 specimen CMACE13

Comparable tests at lower temperatures confirm that the failure mechanism described still works, fig. 27 to 30. However, the stretched zone and ductile crack growth occur only to a small extend, that means in the range of a few µm. The distance between ductile crack initiation and cleavage fracture reduces. The J-Δa- point distribution does not allow a reasonable curve approximation. In those cases the J_R-curve is determined as a line through origin with J_{max} (J_c) and Δa_{max} (measured on the crack surface). The J_i-value results from the intersection of this straight line with a parallel to the J-axis with the distance of the stretched zone width. Approximation of the J_R-curve by a straight line through the origin is a conservative assessment of the course of the J_R-curve. This is an important condition in view of the assessment of components with the use of the J_i-values determined by this method.

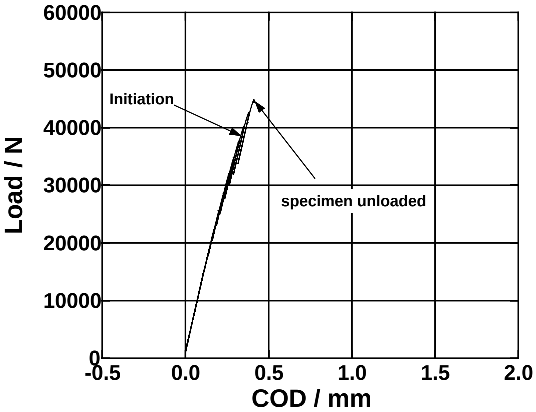


Fig. 27: F-COD (C(T)25 specimen CMACC7)

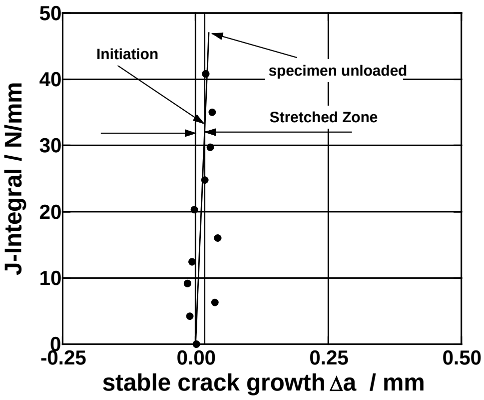


Fig. 28: JR-curve in the lower transition region (C(T)25 specimen CMACC7)

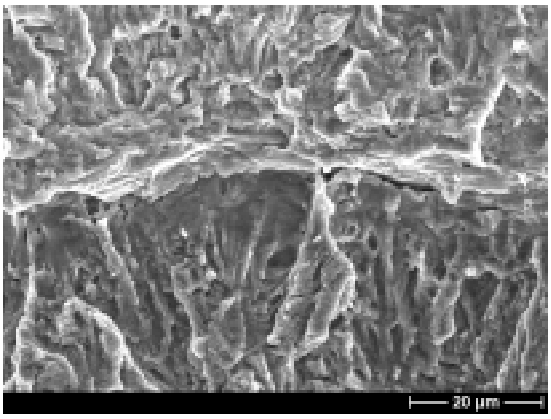


Fig 29: SEM-exposure of the fracture surface, C(T)25 specimen CMACC7

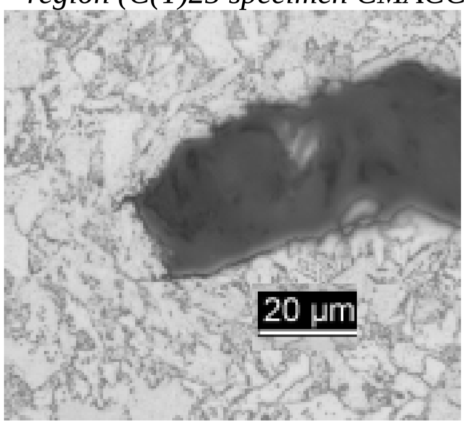


Fig. 30: polished micrograph section of the unloaded C(T)25 specimen CMACC7, tested in the lower transition region

5 CONCLUSIONS

Reliable fracture mechanics material characteristic values are of fundamental importance for quantitative component assessment taking into account the impact of cracks. To use material characteristic values of small scale specimens for component analyses it is necessary, that they fulfil the criterion of transferability.

In the upper shelf of material toughness the ductile crack initiation value J_i fulfils this condition. In the past, it was cast doubt on the ability to determine these ductile crack initiation characteristic values even in the transition region and that J_i mirror correctly the processes of material physics.

The experimental results mentioned above prove that the failure mechanisms, which act in the upper shelf region (plastification of the crack tip, formation of a stretched zone, ductile crack initiation) act also in the transition region of ductility. In the transition region they may also lead to failure due to the change from ductile crack growth to cleavage fracture. Since these procedures are affected by the material plastification, they are less extended at decreasing temperature. They even may be limited to such an extent that crack initiation directly causes instable failure. This is usually the case at very low temperatures or very high multiaxiality of the stress state.

In contradiction to the fracture toughness values K_{Jc} based on cleavage instability, as used for the Master Curve procedure, the crack initiation values K_{IJ} have the advantage of direct transferability and direct application to the component. Adjustments for the account of the crack front length and multiaxiality of the stress state as in the use of the Master Curve concept are not necessary.

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