Experimental and Numerical Fracture Mechanics—An Individually Dyed History

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Abstract Almost half a century ago, fracture mechanics started in Germany with the foundation of the *DVM Working Group Fracture Mechanics* in 1969. The present authors have been partly involved in the further development of fracture and damage mechanics, one with particular interest in elastic-plastic fracture and modelling, the other in thin-walled structures, fatigue and assessment. They take the colloquium in honour of the 65th birthday of Professor Meinhard Kuna as occasion to highlight some significant achievements on the background of personal experience. In particular, they intend to show that both fracture and damage mechanics started with paradigm changes which were partly looked at with distrust in the beginning but turned out to be seminal.

1 Introduction

Today, in the early 21st century, both fracture and damage mechanics appear as established and acknowledged domains of science in the continuity of continuum mechanics. This has not at all been the case at their respective implementation. As the American physicist and philosopher of science Kuhn [46] described in his fundamental book on "The Structure of Scientific Revolutions", which first appeared in 1962, science does not progress via an accumulation of new knowledge, but undergoes periodic "paradigm shifts", in which scientific exploration within a particular field is abruptly transformed. This is most obvious for the great

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scientific revolutions like the change of a geocentric to a heliocentric system or from Newtonian to relativistic mechanics.

At a much smaller scale, fracture and damage mechanics required paradigm shifts as well and had to fight against deadlocked concepts of what "mechanics" was regarded to be. It additionally interfered with an obsolete (though still existing) understanding of physics as a science not only dealing with real objects but presenting "genuine reality", which antagonises with the idea that science can do nothing but develop "models" of reality.

Classical fracture mechanics is a direct application of classical continuum mechanics. But whereas the traditional science of strength of materials did not know anything about "defects", fracture mechanics came up with this unfamiliar term in order to explain and predict failure of structures which had been designed properly according to the state of engineering science as established in the 19th and beginning of 20th century. This first and foremost challenged engineers and companies to admit that their products were not "defect-free", a demand which particularly people from the nuclear industry refused to comply to, even after applying (more or less of necessity) fracture mechanics concepts.

Moreover, fracture mechanics introduced a so far unknown length parameter into structural assessment, namely the size of a presumed or existing defect, the crack length. How was it to be defined, particularly since it affected the load bearing capacity and lifetime of a structure significantly? Furthermore, new physical quantities and material parameters of seemingly obscure dimensions emerged with the new theory, which was used as argument to discredit its reputation.

First-hand examples from the personal experience and history of the authors will be highlighted not only as witty contributions to an anniversary but as "writing on the wall" dedicated to the present generation of scientists that similar unreasonableness can and will recur in presence and future.

By looking at the number of publications, it may be interesting to note that during the first two or three decades the majority of papers and hence of research work was done in the US. During the following decades a shift occurred towards Europe and Asia, where now substantially more publications than in the US have their origin.

2 Linear Elastic Fracture Mechanics (LEFM)

2.1 Fundamentals

2.1.1 Historic Development

Whereas the theoretical background of linear elastic fracture mechanics, that is the mathematical description of stress and strain fields at stress concentrators [37] and an assessment of the energy balance in cracked bodies [28], dates back a century ago, its significance for structural integrity was spotted not until some spectacular accidents had occurred and Irwin published his seminal paper [40] in the mid 20th century. Remarkably, Irwin did not only depict the relationship between the stress intensity and the energy approach but also introduced a two-parameter description of the stress field, long before similar concepts were discussed intensively within the fracture-mechanics community: “The influence of the test configuration, loads and crack length upon the stresses near an end of the crack may be expressed in terms of two parameters. One of these is an adjustable uniform stress parallel to the direction of a crack extension. … The other parameter, called the stress intensity factor, is proportional to the square root of the force tending to cause crack extension”. Actually, the first parameter is Rice’s T-stress [64], and the crack driving force is Griffith’s strain-energy release rate [28].

It took another ten years to launch fracture mechanics in Germany with the foundation of the DVM Working Group Fracture Mechanics in 1969, eight years after a DVM meeting, where “studies on failure mechanics and fracture research had been reported to an international audience in the presence of Dr. George Irwin” [52]. By no means, fracture mechanics had been finally established, then. Acknowledged scientists and engineers fought the new concepts as non-scientific. The “father” of notch mechanics, Heinz Neuber, attacked fracture mechanics in the 3rd edition of his seminal book [58] as follows: “The present new edition provides evidence for the various deficiencies of fracture mechanics. This is primarily about the violation of the stress distribution in the vicinity of the notch or crack tip. Furthermore the lateral dimensions of the crack and the radius of surface curvature are disregarded. Moreover, all effects related to deviations from linear elasticity are ignored.”[1] Likewise in the 1980s, the retired president of the Federal Institute of Materials Testing (BAM) in Berlin regarded fracture mechanics as pseudo-science, since it introduced a surface energy (Griffith’s energy release rate) and a material parameter with the weird dimension of $\mathrm{MPa}\sqrt{\mathrm{m}}$.

These arguments may sound quite amusing today but culminated in fierce disputes not that long ago. They were not very honest, of course, as field singularities are common in physical theories from Newtonian to relativistic mechanics. Singularities arise as limit cases of mathematical equations, but nobody expects a quantity actually to become infinite, and their existence does not devaluate the significance of a theory provided the latter describes the surrounding neighbourhood correctly. The crack tip itself, $r \to 0$, is a mathematical artefact where continuum mechanics is not applicable anyway. Likewise, boundary layer theories are known from fluid mechanics, for instance, in which the thickness of the boundary layer is negligibly small but nevertheless embodies a finite energy.

Removing the “unphysical” stress singularity remained an issue, however, and resulted in Irwin’s small-scale yielding approach [41], Dugdale’s strip yield model [21], Fig. 1a, and Barenblatt’s cohesive zone [4], Fig. 1b, ancestor of modern cohesive models (see Sect. 6.2).

[1] Translation from the preface of the German edition [57] by the present authors who made an effort to meet Neuber’s particular terminology as authentically as possible.

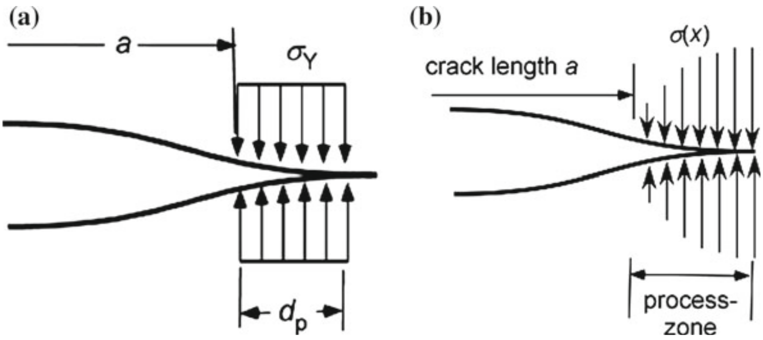


Fig. 1 Models avoiding the stress singularity at the crack tip: **a** Dugdale [21], **b** Barenblatt [4]

2.1.2 Non-singular Terms

A series expansion of the stress state near a crack tip has been presented by Williams [100] in 1957 showing that beside the terms with a $1/\sqrt{r}$ singularity there is a constant term of the normal stresses parallel to the crack face,

$$\sigma_{xx} = \frac{A_{-1}}{\sqrt{r}}\cos\frac{\theta}{2}\left(1 - \sin\frac{\theta}{2}\sin\frac{3\theta}{2}\right) - \frac{C_{-1}}{\sqrt{r}}\sin\frac{\theta}{2}\left(2 + \cos\frac{\theta}{2}\cos\frac{3\theta}{2}\right) + 2A_0, \quad (1)$$

which does not vanish for $r \to 0$. It depends on the biaxiality of the external loading. For instance, in a cracked infinite panel under biaxial loading by tensile stresses, $\sigma_{xx} = \lambda\sigma_\infty$, $\sigma_{yy} = \sigma_\infty$, this term is

$$A_0 = \sigma_\infty(1 - \lambda). \quad (2)$$

It becomes a maximum for $\lambda = 0$ and vanishes for $\lambda = 1$. Since the singularity of stresses appeared so dominant, the constant term had been neglected and forgotten for a long time until the "geometry dependence" of fracture parameters alerted the community.

Larsson and Carlsson [49] investigated the influence of non-singular stress terms and specimen geometry on small scale yielding at crack-tips in elastic-plastic materials. They found that the size and shape of the plastic zone, which is important for the definition of valid K_{Ic} values, was significantly affected by the biaxiality of loading and the specimen geometry.

On this background, Rice [64] introduced the T-stress,

$$\sigma_{ij}(r,\theta) = \frac{K_I}{\sqrt{2\pi r}} f_{ij}(\theta) + T\,\delta_{1i}\delta_{1j}, \quad (3)$$

which characterises the "inherent stress biaxiality in fracture specimens" [50] in small scale yielding and became the starting point for all considerations on a "second parameter" in fracture mechanics affecting the fracture behaviour [6, 20].

2.2 Crack Extension by Fatigue

In the 1960s the fast growing aerospace industries faced challenges towards reducing structural mass and increasing reliability, thus calling for sound methods and codes for quantifying the reliability of their products. The major item to be dealt with was—and still is—structural fatigue. It was in that decade when pioneers such as Paul C. Paris, Richard Hertzberg, Art J. McEvily, Jaap Schijve and many others laid the ground for quantifying life expectancy of structural components containing already some fatigue damage in the form of cracks.

A major break-through was achieved by Paris and Erdogan [62] who correlated the rate of crack extension, *da/dN* with the cyclic stress intensity factor, ΔK. It turned out that there is a power law relationship between these two parameters:

$$da/dN = C\Delta K^m, \tag{4}$$

where the exponent, m, for many metallic materials is typically between 2 and 4. However, when experiments were conducted at very low and very high values of ΔK, researchers found that an *S*-shaped curve resulted, where Eq. (4) is valid only in the intermediate section of the curve, Fig. 2a. It is worth noting that the fractographical appearance of a fatigue fracture surface changes with the rate of crack propagation, thus demonstrating that different mechanisms are at work, Fig. 2b.

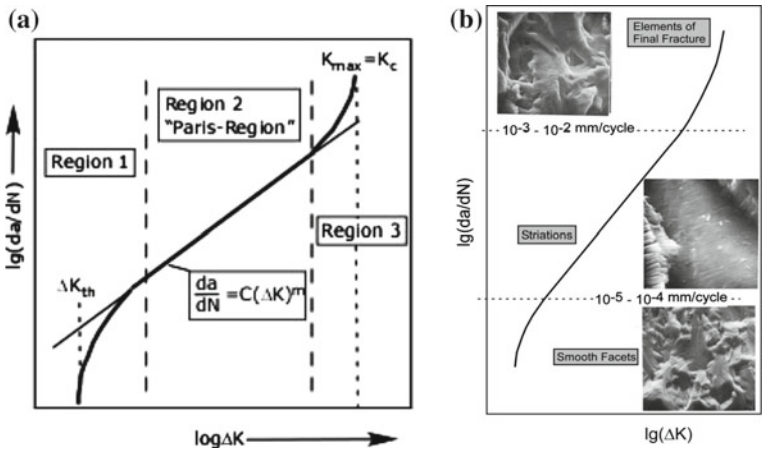


Fig. 2 Fatigue crack extension diagrams: **a** Schematic **b** Fractographic features of the aluminium alloy AlZnMgCu0.5 F46 in correlation with the crack propagation curve, after Schwalbe [73]

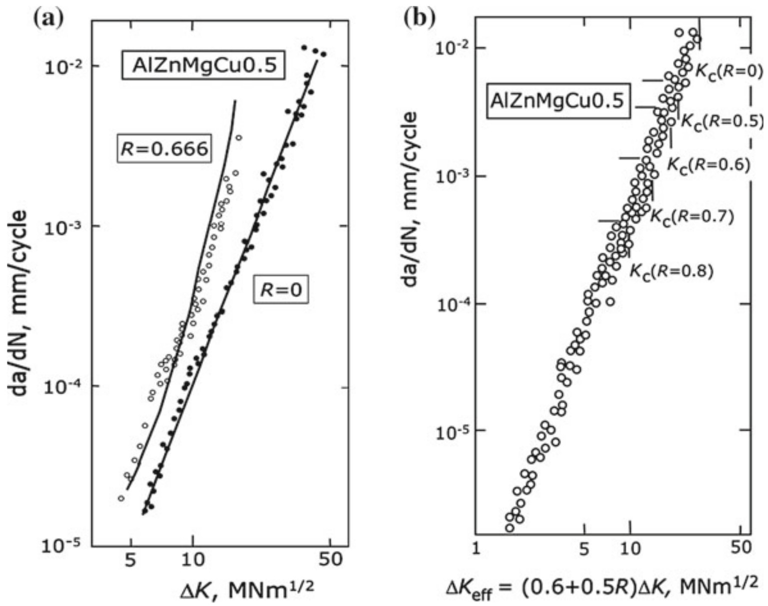


Fig. 3 Fatigue crack propagation in AlZnMgCu0.5 for two stress ratios, Schwalbe [73]: **a** plotted versus ΔK, **b** plotted versus ΔK_{eff}

Furthermore, the unique correlation with ΔK was soon questioned when it was found that crack propagation was also dependent on the R-ratio of the applied stress. A large number of equations have been developed to describe the S-shape and the stress ratio effect. With the following two modifications the S-shape can be modelled:

$$\frac{da}{dN} = \frac{C(\Delta K - \Delta K_{\mathrm{th}})^m}{(1 - R)(K_{\mathrm{c}} - \Delta K)}. \tag{5}$$

However, this dilemma was soon solved by Elber [22] who found that the R-ratio effect was due to partial crack closure during unloading the specimen. If this effect is quantitatively included in the horizontal axis—by using only that cyclic stress intensity factor, $\Delta K_{eff,}$ describing the load range during which the crack is open, and thus effective for stresses and strains at the crack tip—then the stress ratio effect disappears in the graph. This effect is demonstrated for an aluminium alloy in Fig. 3.

With a simple analytical model, the crack extension rate can be estimated, Schwalbe [72]. It is based on the assumption that under each load cycle a crack extends by that amount which corresponds to the distance from the crack tip where the true fracture strain is reached

$$da/dN = \frac{(1 - 2\nu)^2}{4\pi\sigma_Y^2(1 + n)}\left[\frac{2\sigma_Y}{E\varepsilon_f}\right]^{1+n}\Delta K^2 \tag{6}$$

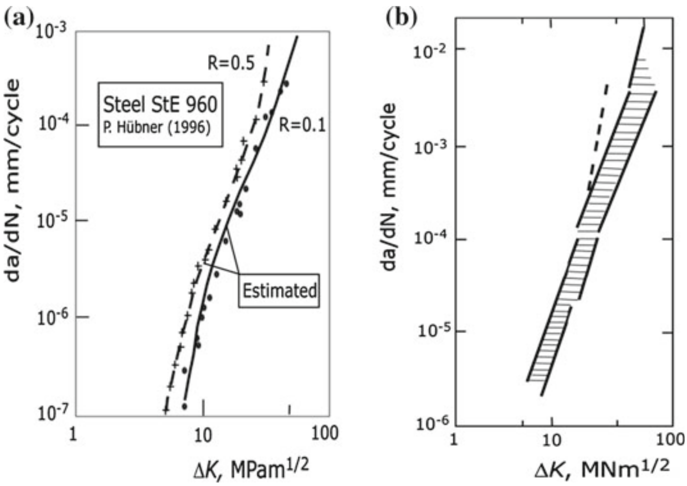


Fig. 4 Fatigue crack propagation diagrams: **a** Experimentally determined crack propagation rates of a high strength steel at two *R*-ratios with Eq. (6), Hübner [35]; **b** Variety of structural steels, with yield strengths between 250 and 1660 MPa, Schwalbe [73]

where σ_Y is yield strength, n strain hardening exponent, E modulus of elasticity, ν Poisson's number and ε_f the true fracture strain. Several comparisons with experiments on various metallic materials yielded very good results. An example is shown in Fig. 4a where Hübner [35] compares his experiments on a high strength steel tested at two *R*-ratios.

Interestingly, as opposed to popular opinion, the material's strength has no substantial effect on the propagation rate. Figure 4b shows a compilation of data obtained on 17 steels with a wide range of yield strengths. The only outlier is a very brittle steel (broken line) with low fracture toughness.

It may be worth noting that there was substantial resistance in the service load fatigue community against using fracture mechanics. The reason was that there were two communities—fracture mechanics and fatigue—with little contact with each other. Fracture mechanics was still looked at with reservation as something esoterical.

However, the activities in the application of fracture mechanics methods to the behaviour of cracks under cyclic loading—primarily in the US—gave rise to the quantitative predictability of the fatigue life of structural components by simply integrating the crack extension equation. Prerequisite is either the existence of a crack found by inspection or of an assumed crack. Due to the pioneering developments at NASA, such predictions have even become possible for variable amplitude loading, Newman [59]. This has been the break-through for general application of fracture mechanics to fatigue problems: It is now an unquestioned standard tool for qualifying structures with crack-like flaws, being used world-wide.

3 Elastic-Plastic Fracture Mechanics (EPFM)

LEFM based structural assessment finally was accepted and required for the life assessment of "high-risk" structures, particularly in the aerospace industry and in nuclear engineering.[2] It became increasingly obsolete, however, with the increasing ductility of structural steels, for example in the nuclear industry which became a major driving force in the development of EPFM. "Valid" data for "plane-strain fracture toughness" [1] at room temperature require specimen dimensions at the meter scale, which are often beyond structural dimensions and raise the question of the significance of the data. Structural engineers were encountered by a new demand: plasticity.

The advantage of LEFM is that due to the linearity of the constitutive equations and backed by the assumption of small strains, closed form solutions for stress and strain fields at a crack tip could be obtained. This is generally impossible in incremental plasticity, as the constitutive equations are not only non-linear but the current stress-strain state depends on the loading history. People tend to keep to the well known, and Griffith's concept of an (elastic) strain-energy release rate had become familiar by now. Its mathematical equivalent was a "path-independent integral", the J-integral, which independently Cherepanov [17] and Rice [63] introduced. Its significance as an intensity parameter of the crack-tip fields was demonstrated by Hutchinson [36] and Rice and Rosengren [67], hence named HRR singularity, and formulas for its experimental determination as energy release rate were provided by Rice et al. [66].

Thus, a perfect analogy to LEFM had been finally established, and those disliking integrals could recall its physical property as a plastic energy release rate. Aside from this, it was and still is accepted in some sectors of industry to convert J-values to K-values by $K_J = \sqrt{JE'}$ as in elasticity even under large scale plastic conditions. Specimen size conditions for measuring valid J_{Ic} data [2] are much less restrictive than for K_{Ic} [1].

One basic limitation of this theory should be kept in mind, namely the underlying assumption of "deformation theory of plasticity", which is a theory of hyper-elasticity, rather, requiring the existence of a strain-energy density as potential of stresses,

$$\sigma_{ij} = \frac{\partial w}{\partial \varepsilon_{ij}} \quad \text{with} \quad w = \int_0^t \sigma_{ij} \dot{\varepsilon}_{ij} \mathrm{d}t. \tag{7}$$

This assumption forbids not only global unloading of a structure but also any local re-arrangement of stresses due to yielding. Numerous numerical analyses based on incremental plasticity and large strains, e.g. [12, 54], have shown that

[2]Where it has later been replaced by the J-integral, see below, or by K_J-values calculated from J according to LEFM.

under plane-strain conditions and in sufficiently thick 3D structures the stress states are approximately J-dominated as long as the crack does not extend, and criteria for J-dominance [55, 86] were established.

3.1 Crack Extension: JR-Curves

Since the J-concept worked satisfactorily for predicting crack initiation, at least in thick-walled components, its limitation that any crack growth causes local stress redistributions which violate Eq. (7) sank into oblivion. Regardless of this, the fracture mechanics community started extending the concept to growing cracks and developed the concept of resistance curves [2] in terms of $J(\Delta a)$.

What followed was, adopting Kuhn's conception [46], a period of "normal science", when scientists attempt to enlarge the central paradigm by "puzzle-solving", which is extremely productive. Journals and conferences were flooded with R-curves. Anomalies were found: R-curves depending on specimen size and shape [26], though by definition, a material parameter must be geometry independent. These findings further increased the number of measured R-curves. It was not even clear whether all of them were measured and evaluated correctly, which opened up the chance of intensive discussions whether or not the measured effects were "real". Evaluation formulas for J were actually still controversial beyond the turn of the millennium [8].

During a period of normal science, the failure of a result to conform to the paradigm is not seen as refuting the paradigm, but as the mistake of the researcher. When the present author published a numerical study [15] on the path dependence of J for large crack extension in 1989, showing that J became zero at the crack tip, $r \to 0$, thus questioning the significance of J as a parameter governing crack growth, he was blamed for false FE analyses. However, Rice et al. [65] had shown in 1980 already, that the singularity of the strain energy density at a moving crack is $\ln(r^{-1})$ in incremental plasticity, whereas in order that J remains finite for $r \to 0$ it has to be an r^{-1} singularity as in elasticity and "deformation theory" of plasticity. The same authors also proposed a "far-field" value, J_{f}, which is different from the deformation theory value of J.

Likewise, geometry dependent R-curves were attributed to faulty testing. The necessity of introducing a "second parameter" characterising the "constraint" of a structure was fiercely fought by the US apologists of a "one parameter characterisation" [86] until finally one of them came up with his own "two-parameter approach" [60, 61]

$$\sigma_{ij}(r,\theta) = \sigma_{ij}^{\mathrm{HRR}}(\theta,n) + Q\sigma_0\delta_{ij} \quad \text{for} \quad |\theta| < \frac{\pi}{2}, \tag{8}$$

mimicking Eq. (3) and unobtrusively ignoring other people's prior suggestions of triaxiality parameters based on the hydrostatic stress [11, 13]. Q is no constant second term of an analytical series expansion like Rice's T-stress, however, but a phenomenological approximation of several higher order terms [85].

Two ASTM conferences on "Constraint Effects" were held in the United States in 1993 and 1995, and in Germany, a Priority Programme (*Schwerpunktprogramm*) on "Ductile Fracture Mechanics" (*Fließbruchmechanik*) was funded by the German Research Foundation (DFG) from 1989–1996 with a budget of 12.8 million DM. The existence of "constraint effects" had been accepted, finally, and the "conventional" along with the "anomalous" results, i.e. the geometry dependent R-curves, were subsumed into one framework. It remained a phenomenological patchwork after all, as a global quantity, J, was combined with a local field parameter, the crack-tip triaxiality, which was not even uniquely defined, and any evident and physical background was lacking.

From today's point of view, the fracture-mechanics community missed two essential points:

- The cumulative quantity J, which increases with crack length, ceases to be an energy-release rate, as soon as the crack starts extending, and an incremental quantity would be required instead, as Turner [94] pointed out. What has been understood to be an extension of Griffith's theory was not, in the end.
- Although physically meaningful models of the failure processes occurring at the crack tip were available [68, 93], the purely phenomenological J concept did not consider them. Respective local models came up as the "local approach" [69], "micromechanical models" [92] and "damage mechanics" [10, 56] but were eyed with distrust by the J-community, in the beginning.

3.2 Energy Dissipation Rate

In the middle of the discussions on the geometry dependence of J_R curves, Turner [94] introduced the energy dissipation rate as an alternative measure of tearing resistance,

$$R = \frac{\partial U_{\mathrm{dis}}}{B\partial a} = \frac{\partial W_{\mathrm{ext}}}{B\partial a} - \frac{\partial U_{\mathrm{el}}}{B\,\partial a}, \tag{9}$$

where W_{ext} is the external work and U_{el} the (recoverable) elastic strain energy. This definition is a straight transfer of Griffith's elastic energy release rate [28] to plastic processes which is consistent with the incremental theory of plasticity. The dissipation rate has the same dimension as J and characterises the increment of irreversible work per incremental crack extension, da. It falls with increasing crack length in gross plasticity and consists of two contributions, namely work of remote plastic deformation and local work of separation,

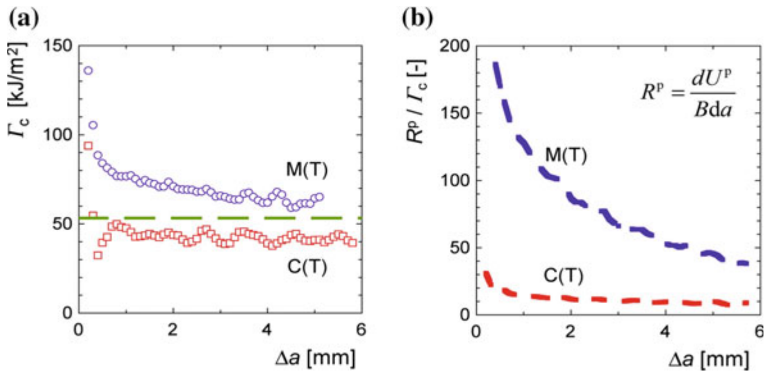


Fig. 5 Energy dissipation rates in M(T) and C(T) specimens calculated from FE analyses with the GTN model [89]: **a** local work of separation, **b** (normalized) remote plastic work

$$R = \frac{\partial U_{\text{pl}}}{B\partial a} + \frac{\partial U_{\text{sep}}}{B\,\partial a} = R_{\text{pl}} + \Gamma_{\text{c}}. \tag{10}$$

A simple energy balance put the dilemma of R-curves straight: What people measure as "fracture resistance" results to a great deal from remote plasticity and not from local material separation [95], and the problem of geometry dependence is hence inherent and unsolvable. Turner's approach was enlightening but showed no way out. There was no possibility based on continuum mechanics to split the two contributions in Eq. (10), and attempts to establish "an alternative view of R-curve testing" [90] based on the dissipation rate did not become accepted. Only understanding of the energy dissipation mechanisms in the process zone at the crack tip is precisely what is necessary to identify "fracture toughness" as a material property.

A numerical analysis of crack extension in a C(T) and an M(T) specimen with the model of Gurson et al. [29, 56] brought additional quantitative evidence [89]. There is a minor geometry effect on the local work of separation, Γ_c, Fig. 5a, but a major effect on the global plastic work, R_{pl}, Fig. 5b. Note that R_{pl} is normalized by Γ_c.

Regardless of Turner's arguments, J_R-curve testing continued, and damage mechanics developed in parallel, suspiciously eyeballed by the mainstream fracture community.

3.3 The CTOD Concept

Historically the first elastic-plastic fracture mechanics concept was developed in the 1960s at the Welding Institute, Cambridge, U.K. Wells [99] started with the idea that the spot where fracture initiates—the crack tip—should be looked at, and that the deformation there at the moment of fracture should be taken as a property characteristic of the material tested. In the experimental method developed at The

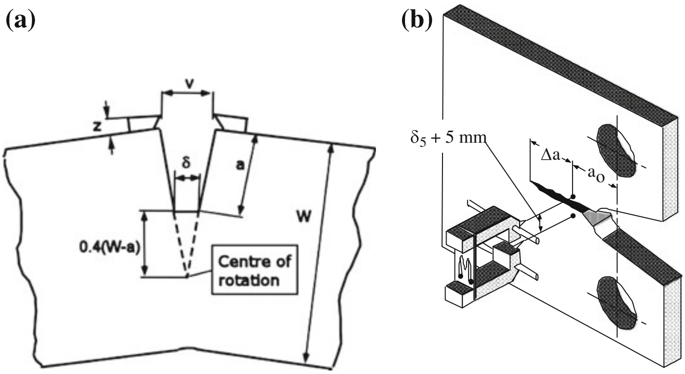


Fig. 6 Crack tip opening displacement (CTOD) **a** Determination according to BS 5762 [16]; **b** δ_5 according to Schwalbe [74]

Welding Institute (TWI) the crack tip opening displacement (CTOD), δ, is determined indirectly from measurements of the displacement, v, at the specimen's front face, Fig. 6a,

$$\delta = \frac{K^2(1-\nu)}{2\sigma_Y E} + \frac{0.4(W-a)}{0.4W + 0.6a + z} v_{pl}, \tag{11}$$

where v_{pl} is the plastic part of v.

The CTOD test method became the British Standard BS 5762 in 1979 and was later also integrated in ISO 12135 [38] and ESIS [23] methods.

Another method for determining the crack tip opening displacement was developed in the authors' group, the δ_5 method. It is particularly suited for determining the crack extension properties of a material and is measured at the specimen's side face, Fig. 6b, Schwalbe [75] and Schwalbe et al. [82]. This experimental technique found also its way into standards: ISO 22889 [39] and ASTM E 2472 [3].

A further application of the δ_5 technique has been demonstrated by Hellmann and Schwalbe [33], where it was shown that both definitions of the CTOD yield practically identical results. The main difference between both method consists in the applicability to test piece geometry: Whereas the BS 5762 technique can only be used with C(T) and SE(B) specimens, the δ_5 technique can be applied to any geometry with a surface breaking crack, including structural components.

It turned out that the δ_5 method is particularly suited for testing and analysing thin sections, and it is for these cases that criteria for the validity of CTOD crack extension resistance curves have been established, Heerens and Schödel [32]. Such a curve is independent of the specimen's width, W, if

- $\Delta a \leq 0.25(W - a_0)$ for a C(T) specimen,
- $\Delta a \leq W - a_0 - 4B$ for an M(T) specimen, $W - a_0 > 4B$,

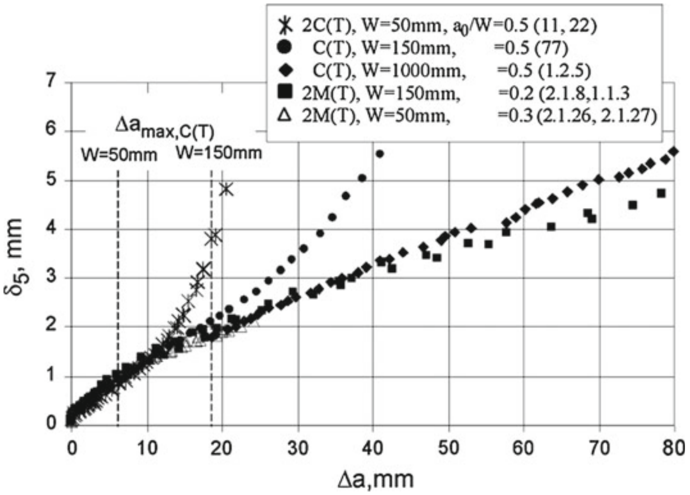


Fig. 7 δ_5 R-curves obtained on M(T) and C(T) test pieces with different width dimensions, Heerens and Schödel [32]

where B is the thickness of the specimen, ISO 22889 [39]. An example is shown in Fig. 7. It is clearly seen that at an amount of 25 % of the original net section width, $W - a_0$, the R-curves become width dependent, hence, are no longer valid. In contrast to this finding, an M(T) specimen exhibits a much longer curve. The δ_5 R-curve method is particularly suited for thin-walled materials used in light-weight structures.

At very high loads, leading to full plasticity of the remaining net section, the cyclic CTOD, $\Delta\delta_{5,}$ can successfully be used to correlate the fatigue crack propagation rate, da/dN, Hellman and Schwalbe [34].

The two elastic-plastic fracture mechanics concepts are not independent of each other, they are compatible. Otherwise, at least one of these concepts would be wrong. A very simple correlation is

- $J \approx \delta\sigma_Y$ for plane stress,
- $J \approx 2\delta\sigma_Y$ for plane strain,

A more rigorous correlation is given in Sect. 5 on assessment procedures.

Interestingly, a further fracture parameter, the crack tip opening angle, CTOA, whose theoretical background can be found in the analysis of the near-tip field at a growing crack by Rice et al. [65], is practically identical with the stabilised slope of the δ_5 R-curve, Heerens and Schödel [32]. Its direct measurement on a specimen surface causes difficulties insofar as the tip of the crack is often not detectable due to the amount of plastic deformation causing a blackish area which obscures the crack

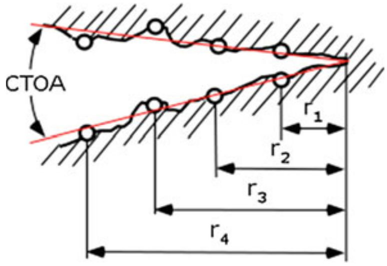


Fig. 8 Determination of the crack tip opening angle using four pairs of reference points, Schödel [71]

tip. A solution was found by Schödel in his PhD Thesis [71], defining a few pairs of reference points along the crack contour which allow easy determination of the CTOA, Fig. 8. The CTOA test method is included in the test standard ISO 22889 [39].

4 Test Procedures

4.1 Some Test Techniques

The new concept of fracture mechanics, of course, required adequate test techniques to take account of the effect of a crack on material properties. The explosion of a Polaris rocket in the US due to brittle fracture gave rise to the foundation of the first committee dealing with fracture mechanics, the ASTM Committee E 24. Now it is Committee E 8. This committee created in the 1960s the first test standard, E 399 [1], which has become a world-wide used method. This method deals with materials and test conditions for linear elastic behaviour and plane strain conditions. The resulting material parameter was dubbed K_{Ic}. This test method has been in use with only minor modifications until now. Key researchers involved in this area were John Srawley, Bill Brown, Gil Kaufman and John Shannon.

In the beginning, all kinds of fracture mechanics tests were very unpopular in industry because of their high cost. Much time and money go into the pre-cracking procedure, making fracture mechanics testing expensive.

Fracture mechanics tests owe a remarkable attribute: the values determined are dependent on the size and geometry of the test piece; however, in the regime of linear elastic behaviour, beyond a certain size plain strain conditions prevail and the value of the fracture property remains fairly constant. And beyond that size, the values are regarded as "valid". As a consequence, the "validity", in other words, size independence, of such values can only be checked after the test has been done. This is unique to the world of fracture mechanics and can make the determination of useful parameters quite cumbersome.

The minimum size requirement for achieving a size independent value is given by

$$B, (W - a) \geq 2.5 \left(\frac{K_{\mathrm{Ic}}}{\sigma_{\mathrm{Y}}} \right)^2, \tag{12}$$

where W designates the width of the specimen, B the thickness and a the crack length. Usually, materials with relatively moderate strength levels have high toughness and therefore require enormous specimen sizes and hence very big and expensive test equipment. This is e.g. the area of pressure vessels where materials with high toughness are preferred. Consequently, German institutions, e.g. the *Bundesanstalt für Materialprüfung*, *University of Stuttgart* and also the *University of Aachen* installed test equipment for performing tests on large-scale specimens [48].

This dilemma was solved when the emerging nuclear industry in the U.S. was in need of information about the fracture properties of their high toughness materials. This led to the developments of elastic-plastic methods as described in Sect. 3.

The test alone is only a part of the problem when fracture mechanics is to be applied to an actual structural component. It had to be demonstrated that the component would fail under the same conditions as were present in the test. Therefore, in the earlier years of fracture mechanics, huge components such as thick-walled pressure vessels were tested, requiring appropriate budgets.

The development of elastic-plastic fracture mechanics and the utmost exploitation of the structural mass and the increasing necessity to define safety margins brought along the need to know quantitatively the material resistance beyond the initiation of crack extension. To his end, appropriate experimental techniques for the determination of the most important quantity—the increasing crack size during the test—had to be developed.

After having played around unsuccessfully with standard ultrasonic equipment, the very first usable technique was developed by the Westinghouse research group in Monroeville near Pittsburgh, PA. It was dubbed the "unloading compliance method" since it consists of periodic unloading during the loading path of the specimen. The unloading traces are elastic, and their slope is a measure of the actual crack size. Garth Clarke from the Westinghouse group worked on the computerisation of this method [18]. It is the first method that was ever standardised.

However, the first technical realisation of this method had a problem: In the initial part of the loading the crack seemed to decrease its length because the unloading slopes got steeper than the initial loading slope of the specimen. It turned out that the friction between the loading bolts and the loading holes of the C(T) specimen prevented proper unloading. The solution was the introduction of flat-bottomed holes in the loading clevises. Still the same artefact of "negative crack growth" can be found in rather recent publications.

The second technique—the electrical potential drop method—exploits the effect of crack size on the electrical resistance of specimens made of metallic materials. AC and DC methods were developed. Its standardised form which makes use of direct current was developed in the authors' research group, Schwalbe and Hellmann [79]. It is based on the disturbance of a DC electrical field by the presence of a crack. Johnson [42] developed a closed form relationship between the crack size, a, and the potential drop measured across two well defined points. Solving that equation for the crack length yields [79].

$$a = \frac{2W}{\pi}\cos^{-1}\left(\frac{\cosh\left(\frac{\pi y}{2W}\right)}{\cosh\left[\left(\frac{\varphi}{\varphi_0}\right)\cosh^{-1}\left\{\frac{\cosh\left(\frac{\pi y}{2W}\right)}{\cos\left(\frac{\pi a_0}{2W}\right)}\right\}\right]}\right), \tag{13}$$

Here

$2y$ signifies the distance between the potential pick-up points,

φ_0 is the potential related to the starting crack length, a_0,

φ is the current potential drop related to the increasing crack length.

Although this equation was derived for an M(T) specimen geometry it was demonstrated that it is applicable to C(T) and SE(B) geometries as well, whereas compliance based techniques have to be calibrated for each specimen geometry. On the other hand, it can be used for metallic materials only.

Initially, arguments were raised against this method such that it should depend on the material and on the test temperature due to the variation of resistivity. However, it is the normalisation of the current potential by the starting value: φ/φ_0 ruling out these arguments. This technique is being used in a number of test methods.

4.2 *Harmonisation of Test Procedures*

Whereas in the beginning of standardisation of fracture mechanics tests, a method for each fracture parameter, K, J, CTOD and for plane strain and plane stress was developed, over the time a harmonisation was attempted. The idea was that a single method should be sufficient for all fracture parameters and that the specimen response should tell whether the evaluation is to be done according to either linear elastic or elastic-plastic procedures. Furthermore, the test interpretation either in terms of J or CTOD should be given in the test method to be developed.

A first attempt was undertaken by the European Group on Fracture (EGF) by creating its first test method, EGF P1-87D, Schwalbe et al. [81]. This procedure describes the determination of crack extension resistance curves in terms of J and δ where δ had to be determined according to BS 5762 [16]. It allows also the determination of J and δ at initiation of stable crack extension. Later, this procedure was superseded by ESIS P2-92 [23], which includes the stress intensity factor as a fracture parameter and an additional method for determining initiation of stable crack extension using measurement of the critical stretch zone width as outlined by Heerens et el. [31]. A further initiative by ISO led to the standard ISO 12135 [38] very similar to ESIS P2-92 [23].

The handbook EFAM GTP02 [78] developed at GKSS Research Centre is probably the most comprehensive procedure. It is based on ESIS P2-92 and includes additional features such as M(T) specimens, the δ_5 technique, determination of K-based fracture parameters, crack tip opening angle, rate of dissipated energy, testing of weldments, statistical treatment of scatter and special validity criteria for tests on specimen with low constraint. In addition to material parameters related to stable crack extension, the determination of parameters for unstable fracture is also described.

5 Assessment Procedures

A number of engineering assessment methods have been developed in different institutions from various countries. Due to the number of procedures and the complexity of the subject, only a very coarse overview can be given here.

Methods for assessing the integrity of structures with crack-like defects have to be standardised in order to make assessments independent of individual methods and persons using them.

Industry specific methods such as for pipelines, pressure vessels and aircrafts have been developed, however, basically they all have to satisfy the inequality

$$\text{Crack Driving Force} < \text{Material Resistance}.$$

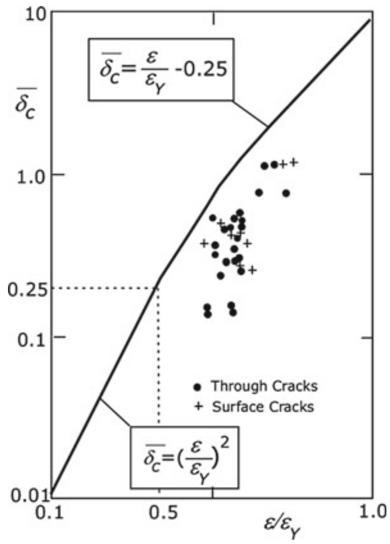
The difference lies in the details used for expressing either side of this inequality. The left hand side poses most of the problems, e.g. using finite element analysis or one of several analytical expressions, whereas on the right hand side it has to be decided which material properties have to be used and how this has to be done.

The most important information from such an assessment are maximum load a component is able to carry, critical crack size and residual life time. From this information inspection intervals can be quantified in order to enable the operator to find a crack before it becomes critical.

As long as the structural behaviour can be characterised by the framework of LEFM, the problem of the left hand side, as given by $K = \sigma(\pi a)^{0.5} Y(a/W)$, is relatively easy to solve. The stress, σ, acting in the structural cross section containing the crack can be determined by linear-elastic stress analysis. The dimensionless function, $Y(a/W)$, depending on geometrical and loading conditions, can be found in handbooks for numerous cases. Otherwise, modern computational methods such as finite element analyses make it possible to generate solutions for unusual problems.

When it comes, however, to elastic-plastic conditions in the cross section under consideration, then J or δ have to be determined using elastic-plastic analyses. In order to facilitate the task of determining the elastic-plastic crack driving force parameter, several “engineering” assessment schemes have been developed, the first one again at TWI, Harrison et al. [30]. They called it the *Design Curve*. In this method, the CTOD is expressed in normalised form such that for two degrees of yielding the normalised critical CTOD, $\bar{\delta}_c$, is given by

Fig. 9 Experimental data compared with the Design Curve, data from [30]



$$\begin{aligned} \bar{\delta}_c &= \frac{\delta_c}{2\pi\varepsilon_Y a} = \left(\frac{\varepsilon}{\varepsilon_Y}\right)^2 \qquad &&\text{for } \frac{\varepsilon}{\varepsilon_Y} < 0.5 \\ \bar{\delta}_c &= \frac{\delta_c}{2\pi\varepsilon_Y a} = \left(\frac{\varepsilon}{\varepsilon_Y}\right) - 0.25 \qquad &&\text{for } \frac{\varepsilon}{\varepsilon_Y} > 0.5 \end{aligned} . \tag{14}$$

These equations are supposed to be used for short cracks where the CTOD is proportional to the local strain. The advantage is that no solution for the stress intensity factor is needed. Similar equations were derived for J as a driving force parameter. Design curve approaches do not attempt to provide more or less accurate failure conditions; they rather give the user conditions which are supposed to be "safe". The example shown in Fig. 9 demonstrates that all experimental values are below the Design Curve and can hence be regarded as safe.

A further method, the "*Failure Assessment Diagram*" (FAD) was developed in the U.K. by the then nationalised Central Electricity Generating Board (CEGB).

In its early version, a simple curve was proposed which—similarly to the COD Design Curve—divides an area of safety and one of unsafe conditions; that is to say, this curve is not a curve for predicting failure. Its vertical axis is given by the applied stress intensity factor, K, normalised by the critical value, K_c: $K_r = K/K_c$. The horizontal axis is again a normalised quantity, a quantity representing the applied load related to a plastic limit load: $S_r = F/F_L$.

The FAD line was formulated as

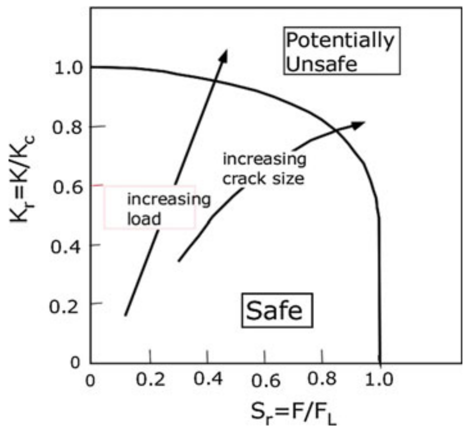


Fig. 10 Failure Assessment Diagram, as cited by [102]

$$K_{\mathrm{r}} = S_{\mathrm{r}}\left[\left(\frac{8}{\pi^2}\right)\ln\sec\left(\frac{\pi S_r}{2}\right)\right]^{-1/2}, \tag{15}$$

see Fig. 10. The square bracket is almost identical to the expression for the crack tip opening displacement based on the Dugdale model [21] shown in Fig. 1a. This expression—although frequently related to Dugdale—in fact was derived by Goodier and Field [27].

Several additional formulations followed to adjust the FAD to the development in fracture mechanics, see e.g. [103].

In the US, the following developments can be observed: In the 1970s/1980s, driven by the needs of the developing nuclear industry and by the contemporary development of the J-integral theory, an attempt was initiated to develop a handbook for J as a driving force parameter, similar to the stress intensity factor handbooks. This way the *EPRI*[3] *Handbook* emerged [47, 87].

It is based on partitioning the J integral into an elastic and a plastic component:

$$J = J_{\mathrm{el}} + J_{\mathrm{pl}}, \tag{16}$$

with

$$J_{\mathrm{el}} = \frac{K_{\mathrm{eff}}^2}{E}, \tag{17}$$

$J_{\mathrm{p1}} = 0$ for contained yielding conditions ($F \leq F_{\mathrm{Y}}$), where F_{Y} is the applied force at the attainment of net section yielding,
and

[3]EPRI is the acronym for Electric Power Research Institute which is financed by American power generating companies.

$$J_{\mathrm{pl}} = \alpha g_1(a/W, n)\left(\frac{F}{F_{\mathrm{Y}}}\right)^{(1+n)/n} \tag{18}$$

for fully plastic conditions ($F \geq F_Y$), where α is the coefficient and n the strain hardening exponent, respectively in the Ramberg-Osgood strain hardening law, and $g_1(a/W, n)$ is a function providing the effects of the geometry of the component and of strain hardening. Functions $g_1(a/W, n)$ have been determined for a number of configurations from finite element analyses.

Whereas the stress intensity factor is only a function of geometry variables, elastic-plastic parameters such as J are also dependent on the deformation properties of the material considered. Although the originators of the handbook simplified the procedure by the above mentioned method, the enormous efforts needed for a suitable handbook led then to a stop, simply because of the parameter explosion following from the interaction of geometrical parameters with the deformation properties of the material. Nevertheless, the handbook has set a landmark, and its way to develop J-expressions has influenced other authors.

It should be kept in mind that the EPRI Handbook provides only the left hand side of the assessment equation, whereas the COD Design Curve and FAD methods deal with both sides of the equation.

The *Engineering Failure Assessment Method* (EFAM) [76] is the only comprehensive assessment scheme which includes both the experimental determination of the relevant material properties including corrosive environments, strength mismatched welded joints and high temperature behaviour, and driving force estimating schemes, including creep conditions and mismatched welded joints. All elements are written in procedural form. The formal document for this assessment scheme can be found under [74].

The *Engineering Treatment Model* (ETM) [83] is one out of six elements of the EFAM. It describes the determination of driving force parameters in analytical form. The detailed procedure is given in [77].

Below the yield load, when $F < F_{\mathrm{Y}}$, the crack driving force expression for CTOD is

$$\delta_5 = \frac{\beta_1}{E}K + \frac{1}{mE\sigma_{\mathrm{Y}}}\frac{F}{F_{\mathrm{Y}}}K_{\mathrm{eff}}^2. \tag{19}$$

For fully plastic conditions ($F \geq F_{\mathrm{Y}}$), the ETM driving force formulations is obtained by transferring the stress—strain curve of the material under consideration to the yielding net section. For this purpose, the stress strain curve is represented by a power law for stresses beyond the yield strength, σ_{Y}

$$\frac{\sigma}{\sigma_{\mathrm{Y}}} = \left(\frac{\varepsilon}{\varepsilon_{\mathrm{Y}}}\right)^N, \tag{20}$$

where $0 < N < 1$.

The resulting ETM equation reads

$$\frac{\delta_5}{\delta_{5\mathrm{Y}}} = \left(\frac{F}{F_\mathrm{Y}}\right)^{\frac{1}{N}} = \left(\frac{J}{J_\mathrm{Y}}\right)^{\frac{1}{1+N}}. \tag{21}$$

This expression shows a clear correlation between δ_5 and J. It can be easily used for sensitivity analyses, because the analytical expression shows clearly how the various parameters affect the result. The relationship in Eq. (21) has been verified in numerous experimental and numerical investigations.

For very short cracks, i.e. for $a \ll W$,

$$\frac{\delta_5}{\delta_{5\mathrm{Y}}} = \frac{\varepsilon_\mathrm{a}}{\varepsilon_\mathrm{Y}}. \tag{22}$$

δ_5 is directly proportional to the applied strain, ε_a

An application of the ETM to mixed mode cracks was reported by Dalle Donne and Döker [19], Fig. 11. For their tests on cruciform specimens with mixed-mode loading a modified δ_5 technique was developed, namely a vector defined by the Mode I and Mode II components. More validations see e.g. in [84].

The ETM was extended for analysing yield-strength mismatched welded joints [80, 102]. To this end, a welded joint is characterised by the base metal, BM, and a strip of weld metal, WM, between two pieces of base metal. Thus, a bi-materiel model emerged where both components are given piece-wise power law

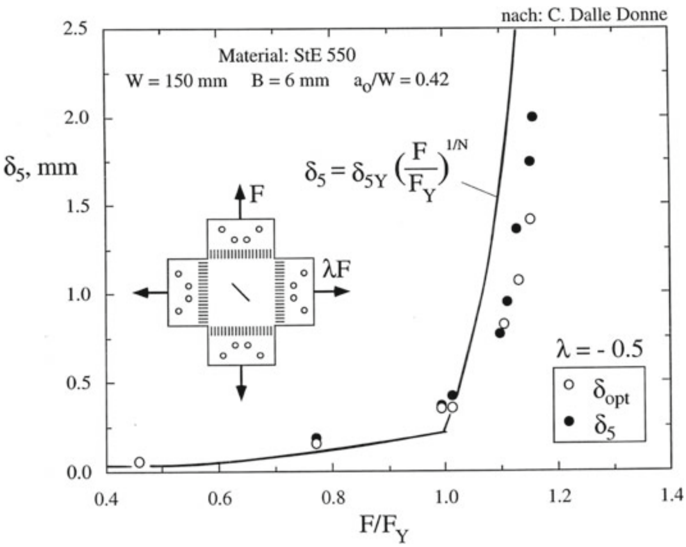


Fig. 11 Experimental data obtained for a crack under mixed-mode loading in comparison with the ETM prediction [19]

deformation behaviour. The model is for the δ_5 route only. The δ_5 driving force is given by the strain applied to the base metal. The procedure is quite complex, nevertheless, all conditions can be described with analytical expressions. For the details, the document EFAM ETM-MM 96 [80] should be consulted which includes also yield load solutions for some standard cases.

In the framework of a European Brite-Euram project, the "Structural Integrity Assessment Procedure for European Industry (SINTAP)" was developed. Seventeen institutions from nine European countries contributed to this project [103] SINTAP offers the FAD routine as well as the crack driving force (CDF) routines which deliver identical results; it is a matter of personal preference which method is going to be used. And also, the user has the option to use the J-integral or the crack tip opening displacement.

The principle of the FAD routine is shown in Fig. 10, however, the ordinate is based on J or δ instead of K. Several levels of analysis are offered [103]:

- The *Basic Option* requires only two basic material properties, namely the yield strength and the Charpy energy.
- The *Standard Option* needs the fracture toughness and yield and tensile strength. It distinguishes between materials with and without Lüders plateau. Elements of the ETM and of the British R6 procedure are included.
- The *Mismatch Option* deals with strength mismatched welded joints are treated, again based on R6 and the ETM.
- The *Stress-Strain Defined Option* requires the complete stress-strain curve and the fracture toughness of the material. Strength mismatched situations can also be analysed.
- The *J-Integral Analysis* includes also the use of the CTOD and is based on FE analyses of J and δ.
- Finally, the *Constraint Option* deals specifically with low-constraint cases with special reference to the δ_5 technique.

A second European project, the *Fitness-for Service Network* (FITNET) comprised about 50 organisations. It includes several modules: a fracture module based on the above mentioned SINTAP options and modules for fatigue crack extension, fatigue life, corrosion and creep [103].

6 Models of the Process Zone

The idea of a "process zone" at the crack tip, where material degradation and separation occur, is quite old and does not come into conflict with classical fracture mechanics. The concept is that a continuum field exists around this zone which may be K- or J-dominated. Dugdale [21] assumed that the stresses cannot exceed the yield strength, σ_0, at the crack tip, see Fig. 1a. Barenblatt [4] considered a zone of material degradation, where the stresses $\sigma(x)$ become zero at the crack tip, see

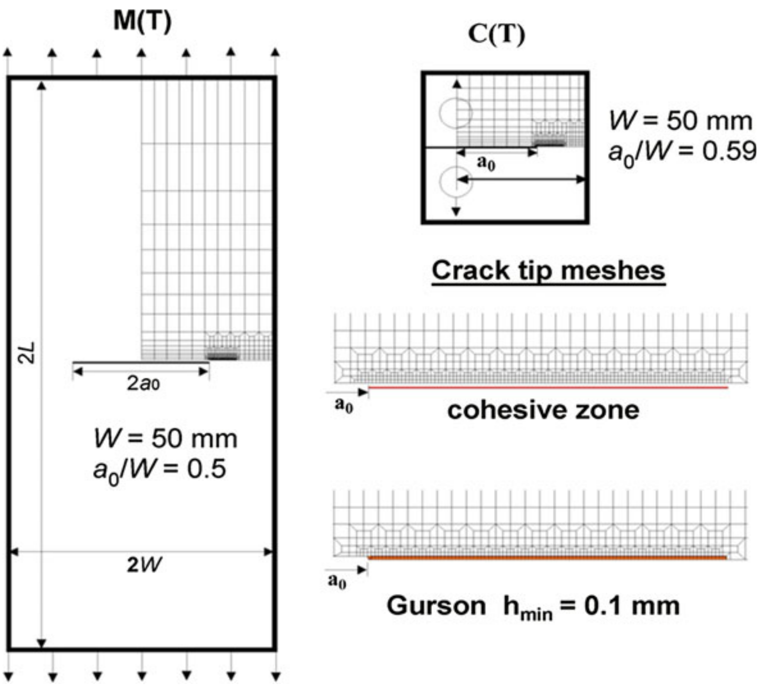


Fig. 12 FE-models of M(T) and C(T) with cohesive zone and damage zone (GTN model [29, 56]), respectively, in the ligament

Fig. 1b. This stress distribution could not be measured, however, and it needed models of damage to calculate it.

With the concept of a “process zone” ahead of the crack tip, uncoupling of remote plastic work, R_{pl}, and local separation energy, Γ_c, in Eq. (10) can be realised in numerical models, if according to Barenblatt’s idea [4] specific elements are introduced, where material degradation and separation occur [88, 89], Fig. 12. The simulations require a constitutive description of the material behaviour in the process zone, which can mirror the local loss of stress carrying capacity. In general, two alternatives are used: Micromechanically based damage models or phenomenological cohesive models.

6.1 Damage Models

The development of damage mechanics began in 1958 when Kachanov [43] published the first paper introducing a damage variable for creep failure of metals, which is nearly the same year as the birth of fracture mechanics identified with Irwin’s paper [40]. The concept of “continuum damage mechanics” (CDM) was picked up again in the eighties [44], particularly in France [51], extended to fatigue and ductile fracture [69] and generalised within the framework of thermodynamics of irreversible processes.

The attractiveness of CDM is its unified framework. Constitutive equations of a damaged material are derived from the same formalism as for a non-damaged material except that the stresses, σ_{ij}, are replaced by the effective stresses", $\tilde{\sigma}_{ij}$, which is called the principle of strain equivalence. In the simplest case of isotropic damage, effective stresses result from

$$\tilde{\sigma}_{ij} = \frac{\sigma_{ij}}{1 - D}, \quad 0 \leq D < 1, \tag{23}$$

where D is a scalar damage variable. In analogy to the plastic strain rates as derived from a yield potential (normality rule), the evolution of damage is derived from a generalised dissipation potential, Φ,

$$\dot{D} = \lambda \frac{\partial \Phi}{\partial Y}, \tag{24}$$

which is a convex function of the conjugate stress variables σ_{ij}, R, ξ_{ij}, Y. $R\left(\bar{\varepsilon}_p\right)$ is the uniaxial flow stress of isotropic hardening, ξ_{ij} are the back stresses of kinematic hardening, and Y is an equivalent "damage stress" as dual state variable to D, also called energy density release rate [51]. The latter is capable of being misunderstood and confused with Griffith's elastic energy release rate, however.

This is a nice theoretical framework consistent with thermodynamics but does not answer the question wherefrom to get the dissipation potential. The latter requires a physically based description of the micromechanical damage processes in a material. The apparent theoretical consistency of CDM attempts mechanics people, in particular, to permanently develop new damage models, sometimes without any precise perception of "damage" and apparently without ever having seen a "real" material.

Alternatively, damage models based on the micro-mechanisms of ductile rupture [93, 96], namely the nucleation, growth and coalescence of voids (Fig. 13), were developed. The mechanism of void growth in a plastic material had been analysed in the late 60s, already [53, 68], and the essential influence of the hydrostatic stress was well known. Remarkably however, this physical understanding [13] had little effect on the discussions on "constraint effects" within the fracture mechanics community, which by the majority and particularly in the USA preferred "non-singular stresses" like T or Q as triaxiality parameters [6, 20, 60, 61] and bashed damage models as poor mechanics.

This is all the more incomprehensible since a fundamental model of ductile damage based on a yield potential for dilatant plasticity by Gurson [29] originates from the USA and was applied to simulations of ductile rupture [56, 57] around the same time as Rousselier's "local approach" [69] in France. The model of Gurson, Tvergaard and Needleman (GTN model) has been favoured in Europe predominantly [10, 92] and particularly promoted by the Fraunhofer Institute in Freiburg. It needed a second conversion of the apologist of a one-parameter J-approach [86],

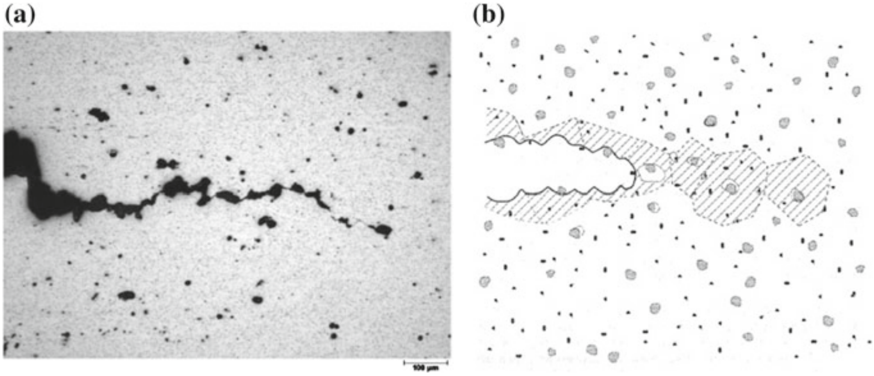


Fig. 13 Ductile crack extension by void nucleation, growth and coalescence: **a** experimental observation for an Al alloy, **b** schematic of a process zone with "cells"

then a two-parameter *J*-*Q*-approach [60, 61] to popularise the GTN model in the USA, which necessarily had to come along with a special new name "computational cell" [101] or "cell model" [24, 25].

Numerical models can employ the constitutive equations of continuum damage either for the complete structure, thus allowing for arbitrary directions of crack growth and not even needing an initial crack [56, 98] or just in one row of elements along the ligament considered as "process zone" as in Fig. 12 [88, 89], thus resembling a cohesive zone model. Significant void growth is commonly restricted to one row of elements.

A serious problem of damage models as established in the constitutive framework of "simple materials" is their feature of localisation, which gave reason to severe objections against the significance of the respective numerical results from the viewpoint of numerical mathematics. Actually, the results of FE simulations are mesh dependent, and the bearing load of a structure decreases with decreasing element height in the ligament. At the least, it is argued, that the postulation of convergence with reducing mesh size is violated, but some people deny simulations with softening materials any significance at all. Again, it is a question of the basic perception of models one has, and whether models and numerical methods are confused with "reality". Continuum mechanics in general is an approximate description of real matter, since the assumption $(\Delta x, \Delta y, \Delta z) \rightarrow 0$ conflicts with the micro-structure of materials. This has no impact on the solution for hardening materials but becomes an issue for softening. The constitutive equations are relations between stresses and strains in *solid elements*, representing the micro-structure of the material in an *average sense*. In order to obtain physically meaningful results for the dissipation rate, Γ_c, in the process zone, a length parameter, h_0, has to be introduced, which depends on the average spacing of void nucleating particles and the hardening behaviour of the metallic matrix. The respective relation is established by an energy equivalence for a representative volume element (RVE) or "cell" (see Fig. 13b). The work of separation, ΔU_{sep}, per incremental crack extension, Δa, is

$$\Gamma_c = \frac{\Delta U_{\mathrm{sep}}}{B\Delta a} = \frac{1}{B\Delta a}\int\limits_{V_0}\left(\int\limits_{t}^{t+\Delta t}(1-f)\bar{\sigma}\,\mathrm{d}\bar{\varepsilon}^{\mathrm{pl}}\right)\mathrm{d}V = \bar{u}_{\mathrm{sep}}\,h_0, \tag{25}$$

with f and $\bar{u}_{\mathrm{sep}}$ being the average void volume fraction and the average strain energy density, respectively, in an RVE of volume $V_0 = h_0 B\Delta a$ [7]. If "local" constitutive equations are applied, which do not contain an intrinsic length scale, the height of the finite elements, h_0, in the ligament has to be considered as a characteristic material parameter [5, 91]—which is devil's notion for some FE experts. If a "finite" element is considered as a mathematical entity for solving boundary value problems in continuum mechanics, the perception of an element size as material parameter sounds weird, indeed. But if it is regarded as a representation of an RVE it appears more natural. In order to point out this difference, Shih introduced the term "computational cell" [24, 25, 101] instead of finite element. There are other concepts for mending the "pathological" mesh dependence like non-local approaches or gradient theories which please the theoretical requirements better than introducing the finite element size as material parameter. These approaches need special subroutines exceeding the capabilities of commercial FE codes, however, and they encounter new problems.

An additional problem with the application of damage models is their commonly large number of parameters, which are supposed to represent micromechanical properties but nevertheless difficult to identify [5] and to verify their uniqueness. An initial euphoria that all parameters of the GTN model can be determined from tensile test data turned out to be unrealistic since stress triaxiality plays an important role and is too low in a tensile specimen. Altogether, the application of damage models is still mostly a preserve of experts.

6.2 Cohesive Models

Cohesive models (CM) describe various kinds of decohesion processes, see Fig. 14, by a relation between (normal) surface tractions, σ_{n}, and respective material separation, δ_{n}, i.e. the traction-separation law (TSL) or cohesive law. For this, particular surface elements are introduced at the boundaries of solid elements along a pre-defined crack path as shown in Fig. 12. The constitutive relation of the interface elements represents the effective mechanical behaviour due to the physical processes, for instance micro-void nucleation, growth and coalescence in a ductile material. Commonly, the cohesive law is defined by two parameters, a cohesive strength (CS), σ_{c}, and a critical separation, δ_{c}, Fig. 15, or, alternatively, a separation energy (SE), Γ_{c}, which simply represents the area under the traction-separation law.

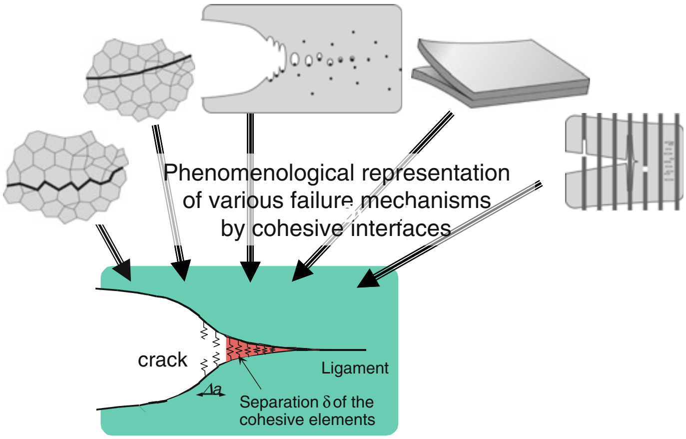


Fig. 14 Cohesive model as phenomenological representation of various decohesion processes in materials

$$\Gamma_c = \int_0^{\delta_c} \sigma_n(\delta_n)\mathrm{d}\delta_n. \tag{26}$$

The cohesive model can be regarded as a renascence of Griffith's concept of a surface energy. The significant differences, however, are that

- though Γ_c is supposed to be a "surface" energy, the respective physical separation process occurs in a volume of finite, though commonly small thickness, in reality, Eq. (25), and
- the CS, σ_c, is an additional independent, phenomenological parameter, representing the maximum tensile stress which can be sustained by the material microstructure.

Cohesive laws can also be established for mixed mode separation processes, which will require an additional assumption on the interaction of tensile and shear modes [70]. The TSL requires significantly fewer material parameters than damage models, and numerical simulations based on cohesive models are less susceptible to convergence problems. A vital advantage compared to the continuum models of damage is that they do not show pathological mesh dependence and do not require the introduction of an additional length parameter via the FE mesh, since they are established as a relation between stresses, σ, and displacements, δ, instead of stresses and strains, ε. A major drawback is their restriction to pre-defined crack paths along the boundaries of solid elements.

The TSL is purely phenomenological and cannot be measured directly, in general. Various relations have been proposed in the literature, see Fig. 15 and overview in [9]. Since it represents micro-mechanical processes of material

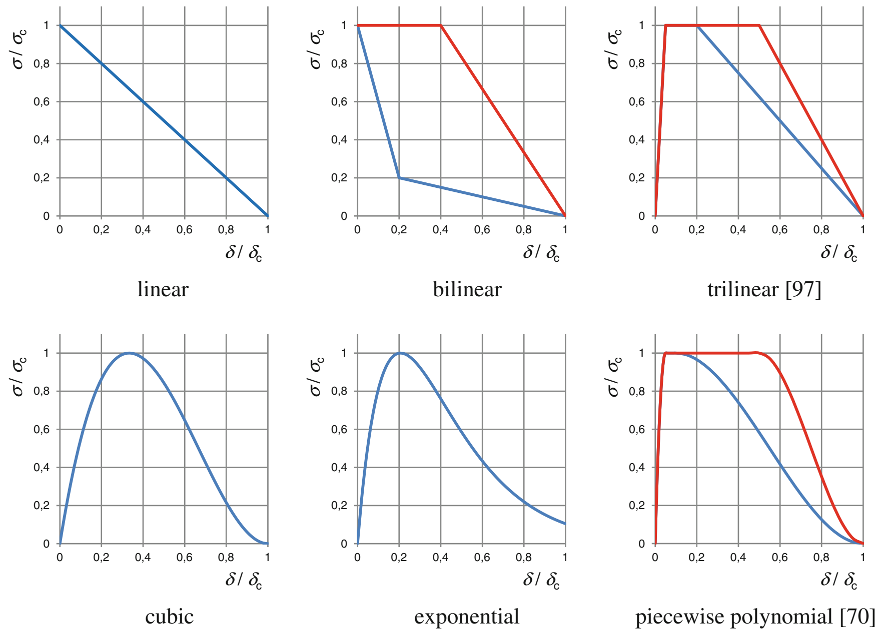


Fig. 15 Various traction-separation laws in the literature (overview from [9])

degradation and damage, respective micro-mechanical models may help to identify it. Processes of void growth have been investigated by numerical simulations of RVEs or "unit cells" containing a void [14, 45], which are assumed to represent a typical periodic microstructure of ductile materials. The TSL for ductile rupture by Scheider [70], which is, among others, depicted in Fig. 15, has been supported by micromechanical analyses [7, 88].

Numerical studies have also demonstrated that material separation based on void growth and coalescence depends on the stress triaxiality. This observed local "constraint" effect is quite evident: higher triaxiality causes an increase of the "fracture stress", i.e. a higher CS, and a decrease of ductility, i.e. a lower SE as in Fig. 5a, see [7, 88]. The particular micromechanical process of void growth and coalescence governing ductile rupture thus exhibits a local constraint effect, which adds to global constraint effects on the overall plastification of the structure. As the (local) SE is very small compared to the (global) plastic work per crack extension, however, Fig. 5b, the effect of triaxiality on the cohesive parameters is commonly negligible and CS and SE can be regarded as material constants from an engineering point of view [89]. The geometry dependence of J_R curves can be accurately predicted by FE simulations employing cohesive elements, Fig. 16.

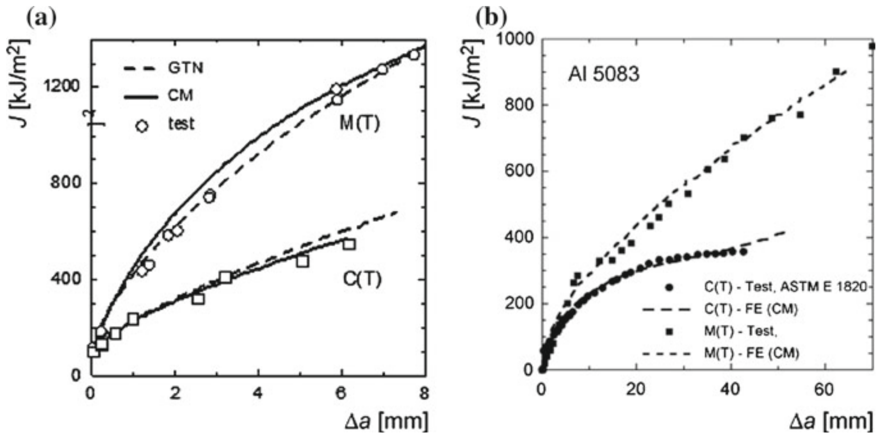


Fig. 16 J_R curves of C(T) and M(T) specimens, tests and FE simulations with cohesive model: **a** steel DIN StE460, side-grooved specimens, plane strain [89], **b** Al 5083, 3 mm thick panels, plain stress [8]

7 Conclusions

The history of fracture mechanics is an example of a more esoterical science looked at with scepticism in some parts of the world. Special resistance was observed in the service load fatigue community which relied on tests only. However, the new paradigm finally transformed into a versatile tool box for the development, design and operation of modern engineering structures. Pressure on high exploitation of structural mass and on economical operation without compromising safety led to wide acceptance of fracture mechanics. The concept of fracture mechanics has become an indispensable ingredient in numerous industrial codes. After several decades of development, it is now well established, demonstrated by standardisation of test methods and establishment of assessment procedures, sometimes in industry-specific form.

An interesting aspect in the development of fracture mechanics is the local distribution of the most important activities. In the beginning, the major driving forces were observed in the US. LEFM was driven in that country by failures in the aerospace area, where safety is of utmost importance. A very important aspect in the development of LEFM is again its application to the aerospace area: Fatigue crack propagation analysed with fracture mechanics made it possible to quantify the residual life time of a cracked component. A similar technological pressure emerged from the upcoming nuclear industry, driving the J-integral methods in terms of test procedures and assessment schemes. Westinghouse and some university groups for theoretical basics were key players.

The activities in the UK were historically earlier, here devoted primarily to welded joints, in particular for the developing offshore industry. The Welding

Institute and CEGB were the most prominent driving forces. CEGB was unique insofar as here the research institutes were immediately responsible for safe operation of their power plants. This way, experience in operation had a short way into the research groups, and vice versa. Whereas in the US the focus was on the *J*-integral, the UK worked on the CTOD. Both worlds were ultimately united as alternatives of equal rights in test standards and in assessment methods.

In Germany, major contributions have been provided by research centres such as Bundesanstalt für Materialprüfung (BAM) in Berlin, Fraunhofer-Institut für Werkstoffmechanik in Freiburg and GKSS Research Centre in Geesthacht, and University institutes in Karlsruhe, Magdeburg and Freiberg.

It should be noted that the present mature position of classical fracture mechanics would not have been possible without the contribution of numerical methods which have been gradually developed to high standards to meet the demands of understanding the experimentally observed phenomena of fracture and to transfer fracture mechanics material parameters to structural behavior. However, the heydays of classical fracture mechanics are now over, the major problems and their solutions are laid down in standards and codes, and the amount of research has substantially decreased. Only a few activities are still visible, such as application to new materials or material combinations. Even teaching fracture mechanics is about to vanish. Mechanical modelling in terms of damage models and cohesive zone models emerges as a more demanding tool, capable of dealing with problems inaccessible by classical fracture mechanics.

In conclusion, the history of fracture mechanics is an example of technology driven progress in science which finally found its way into wide-spread practical application. It also shows that being open to new ideas and working on them with enthusiasm does not only provide scientific merits, more importantly, it pays in terms of technological leadership, in contrast to more conventional traits.

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