

Chapter 2

Scientific Background

Fracture is a problem which troubles structural engineers for centuries. People have been trying to answer when, where, and why the structures fail. Scientists have been trying to investigate the fracture mechanisms of complex components, e.g., the weldment, as the fracture behaviour of weldments influences the crack growth in structures which affects the lifetime and safety of the components. With the rapid development of the finite element method, it is possible to investigate the fracture mechanisms and to predict numerically crack propagation in the materials under investigation. Recently, some scientific activities were performed to investigate the fracture behavior of advanced weldments, for instance, the electron beam welded joints. In the following subchapters, the experimental investigations and numerical simulation activities are presented for the electron beam welded joints. Before explaining the scientific activities, background information is introduced in this chapter to help the reader to acquire the necessary knowledge.

2.1 Electron Beam Welding

Electron beam welding (EBW) is the welding process in which the material is melted and jointed by a high-velocity electron beam, as depicted in Fig. 2.1a. During the welding process, a keyhole can be formed because high density electron beam melting the metal, as described in Fig. 2.1b. The welding speed and the beam current used during the welding process are the major parameters which influence the quality of the welded joints. The EBW is one of the components in High Energy Density Beam (HEDB) welding techniques. Compared to traditional welding processes, like arc welding process, the EBW possesses many benefits. For example, thick joints can be electron beam welded with a single pass whereas multiple passes for the welding process are needed for the arc welding technique. Narrow heat affected zones, little distortion and less residual stresses can be obtained after the welding process due to the high power density during the EBW.

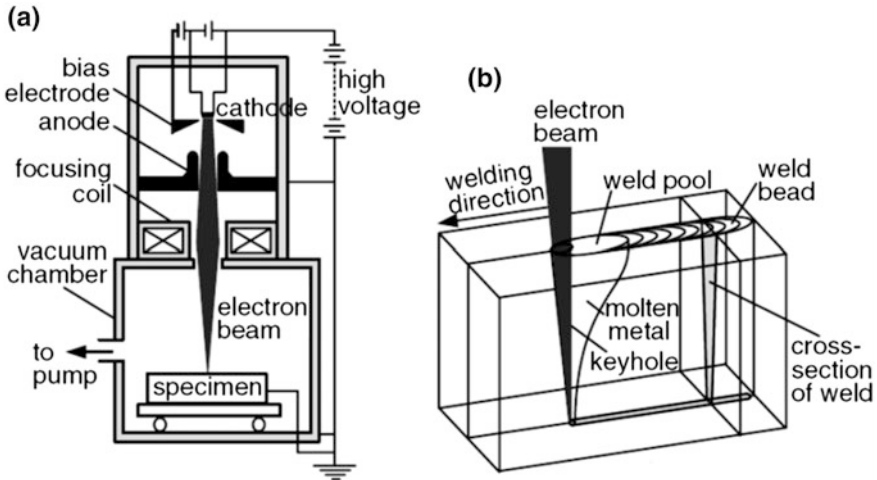


Fig. 2.1 Electron beam welding: **a** process and **b** a keyhole formed during the welding process (Kou 2003). Copyright 2003, with permission from John Wiley & Sons Ltd

Of course, the electron beam welding also provides some drawbacks: the equipment cost for EBW is very high and the requirement of X-ray shielding is inconvenient and time consuming (Kou 2003).

2.2 Fracture Mechanics

2.2.1 The Fracture Mechanics Approach

There are two different judgment ways for the fracture analysis: the energy criterion and the stress intensity approach. The energy approach assumes that crack propagation happens when the energy available for the crack is high enough to overcome the fracture resistance of the material. Griffith (1920) was the first who proposed an energy criterion for fracture, while Irwin (1957) is responsible for developing the approach in the current state. For a crack of length $2a$ in an infinite plate subject to a remote tensile stress (see Fig. 2.2) (Anderson 2005), the energy release rate which is the driving force for the fracture can be expressed as follows:

$$G = \frac{\pi \sigma^2 a}{E} \quad (2.1)$$

where E is Young's modulus, G is the energy release rate, σ is the remotely applied stress and a is half of the crack length.

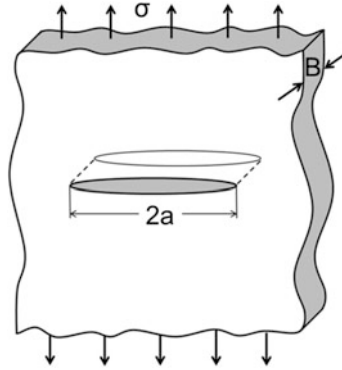


Fig. 2.2 Crack in an infinite plate subject to a remote tensile stress. Reprinted from ref. (Anderson 2005) by permission of Taylor and Francis Group LLC Books

The other approach for the fracture analysis is the stress intensity method. Figure 2.3 shows one element near the tip of a crack in an elastic material and the in-plane stresses on this element is shown (Anderson 2005). Fracture happens at the critical value of K_I , where the entire stress distribution at the crack tip can be calculated with the equations in Fig. 2.3.

For the infinite plate subject to a remote stress shown in Fig. 2.2, the stress intensity factor can be written as follows:

$$K_I = \sigma\sqrt{\pi a} \quad (2.2)$$

Combining Eqs. (2.1) and (2.2), the relationship between K_I and G is derived as follows:

$$G = \frac{K_I^2}{E} \quad (2.3)$$

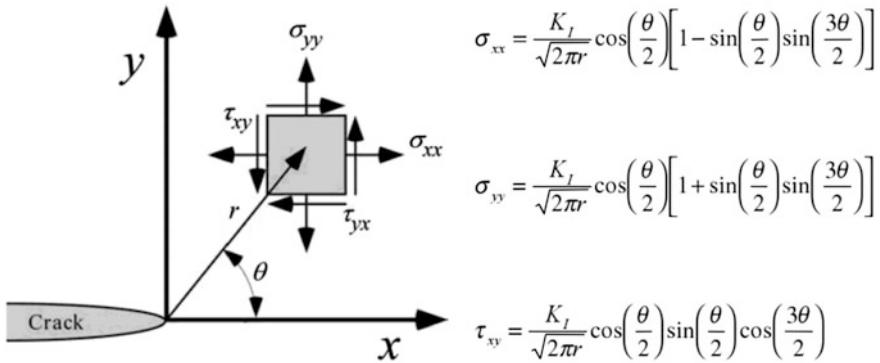


Fig. 2.3 Stresses near the tip of a crack in an elastic material. Reprinted from ref. (Anderson 2005) by permission of Taylor and Francis Group LLC Books

2.2.2 The Brittle Fracture

There are two types of fracture mechanism: ductile and brittle. Ductile fracture applies when the material sustains large plastic strain or deformation before the final fracture. Brittle fracture occurs when a relative small or negligible amount of plastic strain is observed before fracture. Brittle failure results usually from cleavage where separation along specific crystallographic planes is found. At low temperatures, body-centered cubic (Bcc) metals fail by cleavage when the plastic flow is restricted by a limited number of active slip systems (Anderson 2005). When the maximum principal stress reaches the critical one (so called critical cleavage stress σ_c), cleavage fracture will occur (Beremin 1983). Bcc metals, e.g., mild steels will become brittle at low temperatures. For Bcc metals, the force needed to move dislocations is strongly dependent on the temperature and the movement of the dislocations becomes difficult at low temperature. If the maximum principal stress is high enough brittle fracture will happen, otherwise fracture will only happen when a sufficient level of the principle stress is reached (Beremin 1983). According to the discussion of Griffiths (1971), the critical cleavage stress is independent of temperature and can be obtained from notched bend specimens.

The mechanics of fracture progressed from being a scientific curiosity to an engineering discipline, primarily due to what happened to the liberty ships during World War II (Irwin 1957). Many liberty ships built in the early days of World War II ruptured suddenly in 1940s. One of the main reasons is due to the weld, which was produced by a semi-skilled work force and which contained crack-like flaws (Anderson 2005). The weld with initial defects always ruptures suddenly, showing brittle fracture. With the development of weld quality control standards, high quality weldments without initial defects are obtained nowadays.

2.2.3 The J-Integral

The J-integral is a contour integral which was introduced by Rice (1968) as a fracture parameter in a nonlinear elastic material. The mathematical expression of the J-integral is given as follows:

$$J = \int_{\Gamma} \left(W dy - T_i \frac{\partial u_i}{\partial x} ds \right) \quad (2.4)$$

where the strain energy density reads as $W = \int_0^{\epsilon_{ij}} \sigma_{ij} \epsilon_{ij}$, $T_i = \sigma_{ij} n_j$ is the traction vector, u_i are the displacement vector component, and ds is a length increment along the contour Γ , n is the unit vector normal to Γ , while σ_{ij} and ϵ_{ij} are stress and strain tensors, respectively (Fig. 2.4).

As what has been demonstrated by Rice (1968), for a nonlinear material which contains a crack, J is a path independent integral and equal to the energy release rate

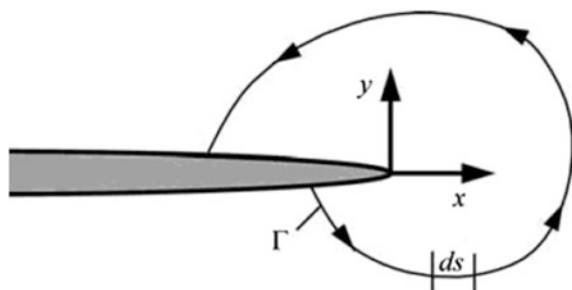


Fig. 2.4 Arbitrary J-integral contour around the tip of a crack. Reprinted from ref. (Anderson 2005) by permission of Taylor and Francis Group LLC Books

G. The J-value was measured firstly experimentally by Begley & Landes (Begley et al. 1972; Landes et al. 1972). According to their method, the same specimens with different initial crack lengths were tensile tested. The force versus displacement curves are shown in Fig. 2.5a. The area under a given curve is equal to U ,

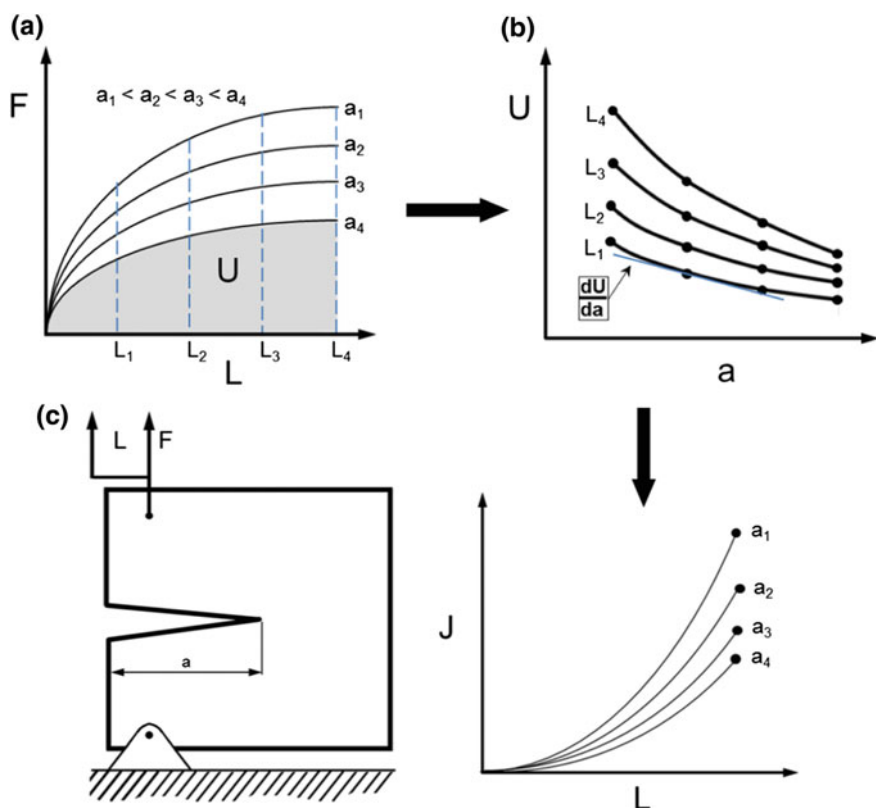


Fig. 2.5 Schematic of experimental measurements of J . Reprinted from ref. (Anderson 2005) by permission of Taylor and Francis Group LLC Books

which is the energy absorbed by the specimen. The U versus crack length curves at different displacements are shown in Fig. 2.5b and the J -values versus displacement at various crack lengths are presented in Fig. 2.5c (Anderson 2005). The J -integral can be calculated as a function of the area under the force versus displacement curve when an edge cracked specimen is adopted:

$$J = -\frac{1}{B} \left(\frac{\partial U}{\partial a} \right) \quad (2.5)$$

where B is the thickness of the specimen and U is the energy absorbed by the specimen.

For linear elastic material behavior, the relation between the crack tip opening displacement (CTOD) δ and J is given as follows:

$$J = m\sigma_0\delta \quad (2.6)$$

where m is a material dependent constant which is influenced by the stress state and σ_0 is the yield stress.

2.3 Constitutive Damage Models

Ductile damage in crystalline solids originates from the nucleation, growth, and coalescence of microvoids (Tvergaard 1989). The voids nucleate mainly at second phase particles (Tvergaard 1989). In the early study of void growth in ductile materials, various models exist focusing on the description of void growth alone. The surrounding material is taken to be rigid perfectly plastic and a state of plane strain is assumed with a prescribed strain rate.

2.3.1 The Rice and Tracey Model

Rice and Tracey (Rice et al. 1969) performed micromechanical studies which focused on the growth of a single void in an infinite elastic-plastic solid. The rate of variation of void radius can be expressed as follows:

$$\frac{dr}{r} = 0.283 \exp\left(\frac{3\sigma_m}{2\sigma_0}\right) d\epsilon_{eq}^{pl} \quad (2.7)$$

where r is the radius of the void, σ_m is the mean stress, σ_0 is the yield stress and ϵ_{eq}^{pl} is the equivalent plastic strain of a perfectly plastic matrix material. This model does not consider the interaction between neighbouring cavities.

For a strain hardening material, the σ_0 value in Eq. (2.7) can be replaced by the von Mises equivalent stress σ_{eq} , and the revised Eq. (2.8) is as follows:

$$\frac{dr}{r} = 0.283 \exp\left(\frac{3\sigma_m}{2\sigma_{eq}}\right) d\varepsilon_{eq}^{pl} \quad (2.8)$$

Huang (1991) assumes that the growth rate of a spherical void in an infinite perfectly plastic matrix is underestimated by Eq. (2.8) and they introduced a new equation to explain the variation of the void radius:

$$\frac{dr}{r} = \begin{cases} 0.427(\eta)^{\frac{1}{4}} \exp\left(\frac{3}{2}\eta\right) d\varepsilon_{eq}^{pl}, & \frac{1}{3} \leq \eta \leq 1 \\ 0.427 \exp\left(\frac{3}{2}\eta\right) d\varepsilon_{eq}^{pl}, & \eta > 1 \end{cases} \quad (2.9)$$

where $\eta = \frac{\sigma_m}{\sigma_{eq}}$ represents the stress triaxiality. The equation does not consider the influence of void coalescence. Moreover, material softening due to damage has not been considered in the Rice and Tracey model.

2.3.2 The Rousselier Model

Within a thermodynamic framework, Rousselier (1987, 2001) developed a damage model. In the Rousselier model, damage is defined by a variation of the void volume fraction originating from second phase particles under tensile loading conditions. The general yield condition of the Rousselier model is as follows:

$$\Phi = \frac{\sigma_{eq}}{1-f} + Df\sigma_k \exp\left(\frac{\sigma_m}{\sigma_k(1-f)}\right) - R(p) = 0 \quad (2.10)$$

where σ_{eq} is the von Mises equivalent stress, σ_m is the hydrostatic stress, f is the void volume fraction (initial value f_0), σ_k and D are material constants, p is the cumulated plastic strain and $R(p)$ is true stress-true plastic strain curve of the material. The von Mises yield surface and the yield surface modified by the Rousselier Eq. (2.10) can be found in Fig. 2.6 (Seebich 2007).

From cavity growth measurements (Rousselier 1987, 2001) and theoretical considerations, the parameter D is considered as a material independent parameter. For most materials this value can be set to $D = 2$. The initial void volume fraction, f_0 , depends on the volume fraction of all inclusions in the material (voids plus non-metallic particles). The f_0 is suggested to be determined from metallographic investigations on the polished material surface. For steel, if the metallographic experiment is not available, f_0 can be estimated from the chemical composition thanks to Franklin's equation (Franklin 1969) in which the manganese (Mn), sulphide (S), and oxide (O) inclusions are considered as inclusions:

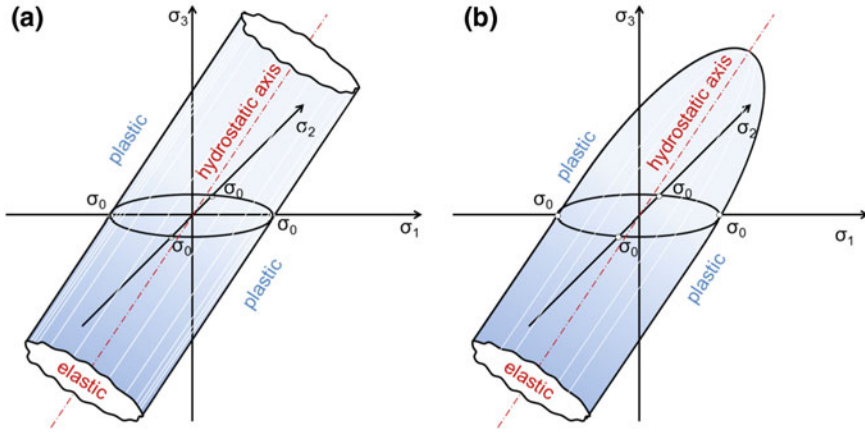


Fig. 2.6 **a** Von Mises yield surface and **b** Rousselier yield surface (Seebich 2007)

$$f_0 = 0.054 \left[\%S(\text{wt}) - \frac{10^{-3}}{\%Mn(\text{wt})} \right] + 0.055\%O(\text{wt}) \quad (2.11)$$

According to the discussion of Rousselier (2001), the first try for σ_k is set equal to the value of two thirds of the equivalent stress (σ_{eq}) for the smooth round specimen under tensile test when the final fracture happens. Based on the previous experiences at MPA (Seidenfuss 1992), σ_k is assumed to be 445 MPa for the current investigated steels (10MnMoNi55, S355NL, et al.). When the critical void volume fraction f_c is reached, the stress carrying ability of the material will be loosed completely. Damage mechanical studies of tensile tests have shown that $f_c = 0.05$ is a reasonable value (Seidenfuss 1992; Mohanta 2003; Weber et al. 2007). The Rousselier model was used successfully to study the ductile fracture behaviour of homogenous materials (Kusmaul et al. 1995; Uhlmann et al. 1999; Schmauder et al. 2002; Schmauder and Mishnaevsky 2009) and inhomogeneous electron beam welded steel joints (Tu et al. 2011, 2013) at IMWF/MPA Stuttgart.

The ductile crack propagation in precracked specimens (compact tension [C(T)]) depends on the element size l_c , element type, symmetries, mesh geometry, etc. (Rousselier 2001). Before the numerical application, the l_c -value should be calibrated first. The parameter l_c is calibrated on notched round specimens, then the same l_c -value is adopted to study the crack propagation of C(T)-specimens. This is the so called mesh dependence of the Rousselier model, where the numerical simulation result is affected by the element size l_c . The classical Rousselier model is called the local Rousselier model where the material damage process at a certain point has a close relationship with the stress and strain field at the same point. Samal developed a nonlocal damage law trying to overcome the drawbacks of the local Roussellier model (Samal et al. 2008, 2009) where the damage growth law is defined in terms of the local void volume fraction but keeping a local definition for

strain. This is the new trend for the Rousselier model as well as for the GTN model. However, a large amount of debugging time is needed to develop the user defined code and to derive results calculated from the non-local Rousselier model which is limiting the application of the non-local damage model.

2.3.3 The Gurson-Tvergaard-Needleman (GTN) Model

Like Bishop and Hill (1951), the widely well-known porous ductile material model is that developed by Gurson (1977). The material simulated by the Gurson model behaves as a continuum since the voids appear through their influence on the global flow behaviour. The yield condition of the Gurson model is written as:

$$\Phi = \left(\frac{\sigma_{eq}}{\sigma} \right)^2 + 2f \cosh\left(\frac{\sigma_m}{2\sigma} \right) - 1 - f^2 = 0 \quad (2.12)$$

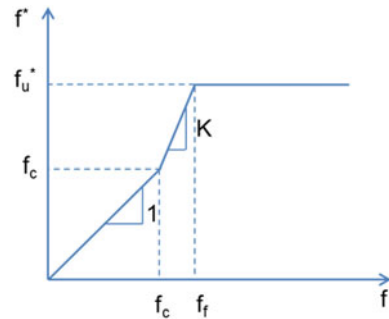
where σ_{eq} is the von Mises equivalent stress, σ is the flow stress for the matrix material of the cell, f is the void volume fraction and σ_m is the mean stress. The implication of this analysis is that the voids are assumed to be randomly distributed, so that the macroscopic response is isotropic.

The original yield condition of the Gurson model was modified by Tvergaard (1982a) and Tvergaard and Needleman (1984) and the equation of the so called Gurson-Tvergaard-Needleman (GTN) model is as follows:

$$\Phi = \left(\frac{\sigma_{eq}}{\sigma} \right)^2 + 2q_1 f^* \cosh\left(\frac{q_2 \sigma_m}{2\sigma} \right) - 1 - q_3 f^{*2} = 0 \quad (2.13)$$

where q_1 , q_2 and q_3 are parameters introduced by Tvergaard which can improve the accuracy of the predictions of the Gurson model (Tvergaard 1982b). The function f^* is a function of f as shown in Fig. 2.7 which represents the accelerated damage originated from the void coalescence when a critical volume fraction f_c (Tvergaard and Needleman 1984) is reached:

Fig. 2.7 The function of the modified void volume fraction $f^*(f)$



$$f^* = \begin{cases} f & f < f_c \\ f_c + k(f - f_c) & f_c < f < f_f, \\ f_u^* & f \geq f_c \end{cases} \quad k = \frac{f_u^* - f_c}{f_f - f_c} \quad (2.14)$$

According to this model, a crack appears when the current f value reaches the final void volume fraction, f_f . That means the material loses its stress carrying ability completely when the ultimate void volume fraction f_u^* is reached, where $f_u^* = 1/q_1$. Zhang thought the value f_c can be determined from unit cell modeling results (Zhang et al. 2000).

It is assumed that the rate of the void volume fraction consists of the growth of existing voids and the nucleation of new voids, as explained by Needleman and Rice (1978) and Chu and Needleman (1980):

$$\dot{f} = \dot{f}_{\text{growth}} + \dot{f}_{\text{nucleation}} \quad (2.15)$$

The void growth is given according to the following equation:

$$\dot{f}_{\text{growth}} = (1 - f)\dot{\varepsilon}_{kk}^{\text{pl}} \quad (2.16)$$

where $\dot{\varepsilon}_{kk}^{\text{pl}}$ is the plastic volume dilatation rate and the matrix is assumed behaving plastically incompressible.

The nucleation of new voids is given as follows:

$$\dot{f}_{\text{nucleation}} = A\dot{\varepsilon}_{kk}^{\text{pl}} + B\left(\dot{\sigma}_m + \frac{1}{3}\dot{\sigma}_{kk}\right) \quad (2.17)$$

The influence of the void nucleation is either strain controlled ($B = 0$) or stress controlled ($A = 0$). The void nucleation rate for the strain controlled nucleation situation, as suggested by Chu and Needleman (1980) is as follows:

$$\dot{f}_{\text{nucleation}} = A\dot{\varepsilon}_{kk}^{\text{pl}}, A = \frac{f_n}{s_n\sqrt{2\pi}} \exp\left[-\frac{1}{2}\left(\frac{\varepsilon_{\text{eq}}^{\text{pl}} - \varepsilon_n}{s_n}\right)^2\right] \quad (2.18)$$

where f_n is the void volume fraction of the void-nucleating particles, ε_n is the mean strain when void nucleation occurs and s_n is the corresponding standard deviation of the nucleation strain, which follows a Gaussian distribution.

For stress controlled void nucleation, the void nucleation rate is shown in the equation as follows:

$$\dot{f}_{\text{nucleation}} = B\left(\dot{\sigma}_m + \frac{1}{3}\dot{\sigma}_{kk}\right), B = \frac{f_n}{s_n\sqrt{2\pi}} \exp\left[-\frac{1}{2}\left(\frac{\sigma_m + \frac{1}{3}\sigma_{kk} - \sigma_n}{s_n}\right)^2\right] \quad (2.19)$$

where $\sigma_m + \frac{1}{3}\sigma_{kk}$ is an approximate value of the maximum normal stress acting on the particle/matrix interface (Needleman 1987; Xia 1996), σ_m is the mean stress for nucleation.

With the development of computer power, the GTN model is used to investigate the fracture behavior of homogeneous materials, e.g., steel StE690 (Springmann and Kuna 2005), X100 pipeline steel (Tanguy et al. 2008), X70 ferritic–pearlitic steel (Rivalin et al. 2001a, b) and Al2024 sheets (Chabanet et al. 2003). The GTN model was also used to study the fracture behaviour of inhomogeneous structures (Østby et al. 2007a, b). In recent years, the GTN model has found wide application in the fracture analysis of welded joints. The GTN model was used by Needleman (Needleman et al., 1999) and Tvergaard (Tvergaard and Needleman 2000, 2004) in studies of conventional fusion welded joints. After the initial study of the fracture behaviour of a laser welded joint (Çam et al. 1999; Santos et al. 2000), the GTN model was adopted to study the ductile fracture behaviour of a laser welded joint at the Helmholtz-Zentrum Geesthacht Centre for Materials and Coastal Research (Nègre et al. 2003, 2004; Cambrésy 2006). The GTN model was used to predict crack propagation when different initial crack positions were considered. It has been proved that the GTN model can be successfully used to study the fracture mechanisms of an inhomogeneous laser welded joint (Nonn et al. 2008). Meanwhile, the GTN model was used to study the ductile fracture behaviour of friction stir welded AA2024 joints by Nielsen (2008). Nielsen focuses on studying the ductile damage development in an FS-welded aluminium joint under tensile loading normal to the weld line. A study of the effect of varying the distribution of the volume fraction of second phase particles, from which voids are assumed to nucleate, was carried out as well.

2.4 Cohesive Zone Model (CZM)

The concept of a cohesive zone model was introduced by Dugdale (1960) and Barenblatt (1962). Dugdale and Barenblatt assume that the crack consists of two parts: the stress-free part and the parts loaded by cohesive stresses. In 1960, Dugdale introduced a strip-yield model where he assumed that the cohesive stress is equal to the yield stress and the material is supposed to perform as elastic-ideally plastic (Schwalbe et al. 2009), as shown in Fig. 2.8. The model presented by Barenblatt adopted a cohesive law to describe the decohesion of atomic lattices

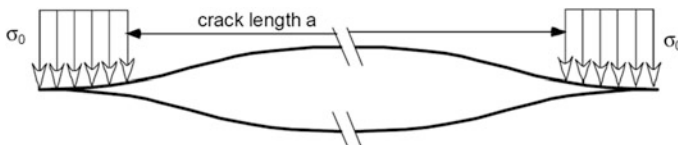


Fig. 2.8 The Dugdale model (Dugdale 1960; Schwalbe et al. 2009)

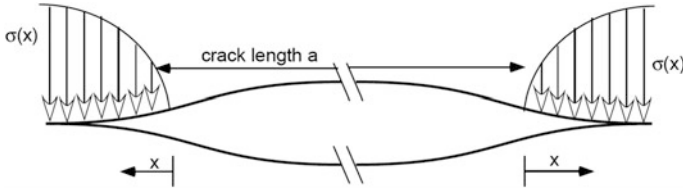
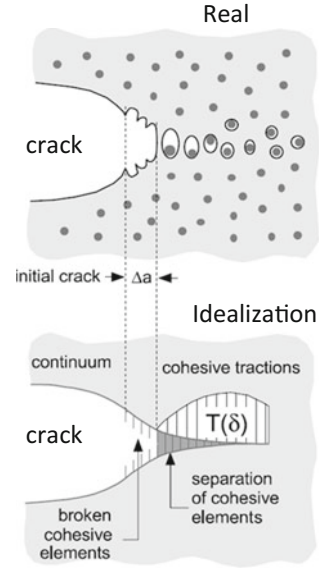


Fig. 2.9 The Barenblatt model (Barenblatt 1962; Schwalbe et al. 2009)

Fig. 2.10 Depiction of the ductile failure process by a cohesive zone model (Cornec et al. 2003). Copyright 2003, with permission from Elsevier



(Schwalbe et al. 2009) where the stresses in the ligament of the crack follows a prescribed distribution $\sigma(x)$ which is related to the material, as shown in Fig. 2.9. The special zone where damage occurs is called the cohesive zone. As shown in Fig. 2.10, under loading conditions, the void initiation, growth and coalescence process of a ductile material is described by the damage of the cohesive element. The damage of the cohesive element is described by a traction-separation law (TSL) which consists of three cohesive parameters: the cohesive strength T_0 , the critical separation length δ_0 and the cohesive energy Γ_0 . The cohesive strength T_0 is the maximum stress obtained when the crack initiates and the cohesive energy Γ_0 is the energy for the total separation of a unit area of the material. When the cohesive strength T_0 and the cohesive energy are known, the third parameter δ_0 can be calculated with the following equation:

$$\Gamma_0 = \int_0^{\delta_0} T(\delta) d\delta \quad (2.20)$$

Therefore, there are only two independent cohesive parameters: the cohesive strength T_0 and the cohesive energy Γ_0 . For mode I loading, when the normal component of the separation reaches δ_0 , the cohesive element fails completely where the stresses become zero. The cohesive energy can be calculated by Eq. 2.20 and is assumed equal to the J-integral value at initiation of ductile crack extension (J_i) as discussed by Brocks (2004).

Following this idea of the cohesive zone model, different TSLs were proposed in the past to investigate the ductile and the brittle fracture behaviour of a number of materials. The first application of the cohesive model in the FE simulation was performed by Hillerborg et al. (1976) to describe the brittle fracture behavior of a concrete beam. The linear decreasing TSL introduced by Hillerborg is shown in Fig. 2.11a. The equation of the linear decreasing TSL is as follows:

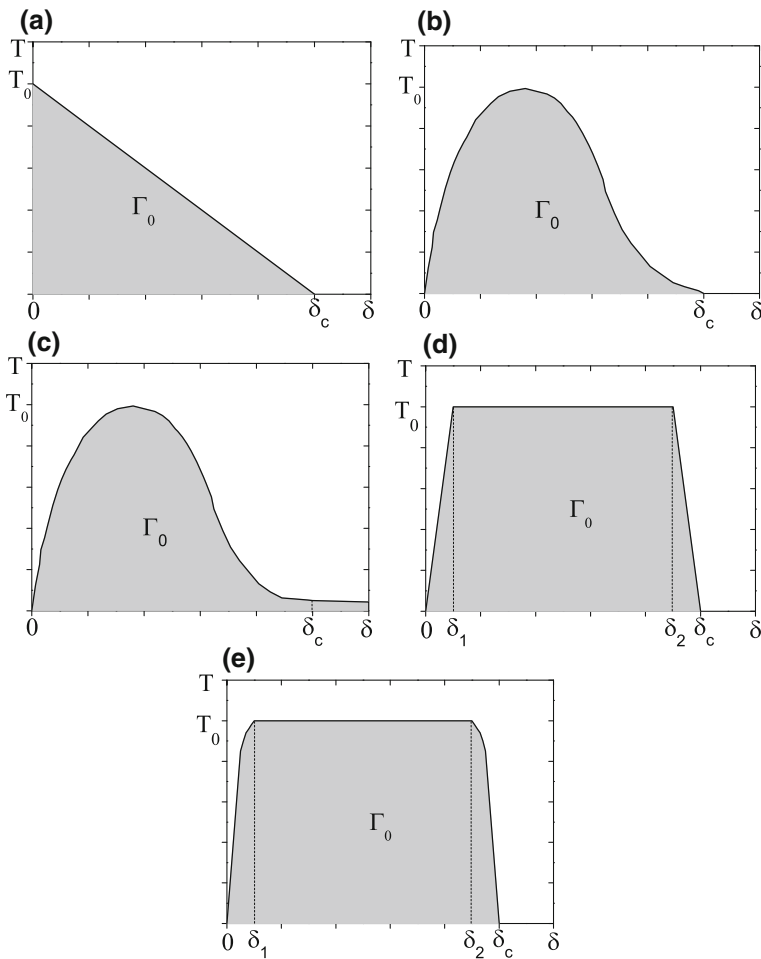


Fig. 2.11 Shape of traction separation laws according to: **a** Hillerborg (1976), **b** Needleman (1987), **c** Needleman (1990), **d** Tvergaard and Hutchinson (1992), and **e** Scheider (2001)

$$T(\delta) = T_0 \left(1 - \frac{\delta}{\delta_0} \right) \quad (2.21)$$

This kind of TSL is normally adopted to study the behavior of brittle material.

Needleman introduced two TSLs to describe the decohesion behavior of a ductile material. The polynomial and the exponential TSLs are shown in Fig. 2.11b–c. A polynomial shaped TSL (Needleman 1987) can be found in Fig. 2.11b, where the equation of $T(\delta)$ is given as:

$$T(\delta) = \frac{27}{4} T_0 \frac{\delta}{\delta_0} \left(1 - \frac{\delta}{\delta_0} \right)^2 \quad (2.22)$$

The function $T(\delta)$ with an exponentially shaped TSL (Needleman 1990) can be depicted as follows:

$$T(\delta) = T_0 e^z \frac{\delta}{\delta_0} \exp \left(-z \frac{\delta}{\delta_0} \right) \quad (2.23)$$

where $e = \exp(1)$ and $z = 16e/9$.

For materials which show a more ductile behavior, Tvergaard and Hutchinson (1992) proposed a trapezoidal shaped TSL to describe the fracture behavior which is shown in Fig. 2.11d. They introduced two additional parameters, δ_1 and δ_2 in the equation, the initial cohesive stiffness $K_{nn} = T_0/\delta_1$ is constant before $T(\delta)$ reaches T_0 and the formulation of $T(\delta)$ is explained in the equation as follows:

$$T(\delta) = T_0 \begin{cases} \left(\frac{\delta}{\delta_1} \right) & \delta < \delta_1 \\ 1 & \delta_1 < \delta < \delta_2 \\ \left(\frac{\delta_0 - \delta}{\delta_0 - \delta_2} \right) & \delta_2 < \delta < \delta_0 \end{cases} \quad (2.24)$$

This TSL was later modified by Scheider (2001) where the initial cohesive stiffness varies. This TSL proposed by Scheider was used to study the fracture behavior of laser weldments of which the TSL function is as follows:

$$T(\delta) = T_0 \begin{cases} 2 \left(\frac{\delta}{\delta_1} \right) - \left(\frac{\delta}{\delta_1} \right)^3 & \delta < \delta_1 \\ 1 & \delta_1 < \delta < \delta_2 \\ 2 \left(\frac{\delta - \delta_2}{\delta_0 - \delta_2} \right)^3 - 3 \left(\frac{\delta - \delta_2}{\delta_0 - \delta_2} \right)^2 + 1 & \delta_2 < \delta < \delta_0 \end{cases} \quad (2.25)$$

The shape influence of the TSL on the damage behavior of a material has not been reported uniformly so far. Some authors concluded the shape of the TSL has only tiny influence (Yuan et al. 1996; Tvergaard and Hutchinson 1992) on the fracture behavior of the material. However, in recent publications (Scheider and

Brocks 2003), the shape of the TSL has been found having a severe influence on the damage behaviour.

After the successful application of the cohesive zone model in different homogeneous structures (Siegmund and Needleman 1997; Chen et al. 2003, Lin et al. 1998; Li and Siegmund 2002), the cohesive zone model has been applied to investigate the fracture behavior of complex structures, e.g., welded joints (Anvari et al. 2006; Lin 1998b; Lin et al. 1999; Scheider 2001; Tu and Schmauder 2016). This proves that the cohesive model is able to describe the damage behaviour of homogeneous structures and inhomogeneous welded joints.

2.5 ARAMIS System

ARAMIS is an optical 3D deformation analysis system which is based on an image evaluation technique to capture the surface deformation of a sample under load. The ARAMIS system can be used to analyze, calculate and document the surface displacements and surface strains at each deformation step (ARAMIS 2008). Some applications of the ARAMIS system monitoring the deformation of tested samples are shown as follows. For example, the ARAMIS system was used to record the true stress-strain behavior of flat specimens of PA6/elastomer composites (Huang et al. 2011). During the tensile test process, the effective length and diameter change (Δl and Δd) is recorded by the ARAMIS system; see Fig. 2.12a. The sample is viewed by two CCD cameras which record the surface deformation (see Fig. 2.12b). The ARAMIS system recorded the deformation process of flat specimens and middle cracked tension [M(T)] specimens extracted from Al6013-T6 laser beam welded and friction stir welded joints at the Helmholtz-Zentrum

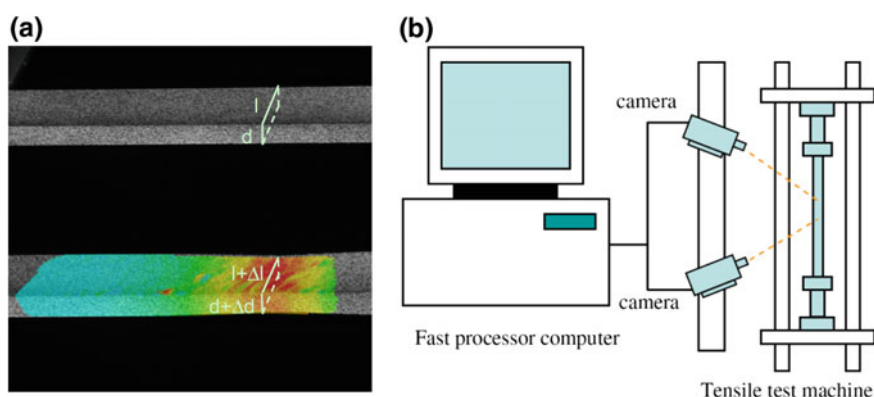


Fig. 2.12 **a** Measurement of the effective length (Δl) and diameter changes (Δd); **b** sketch of the tensile test combined with the ARAMIS system (Huang et al. 2011). Copyright 2012, with permission from Elsevier

Geesthacht Centre for Materials and Coastal Research (Seib 2006). The strain distribution of steel adhesive joints is shown, e.g., by Sadowski et al. (2010). As shown on the website of GOM company (GOM 2014), the ARAMIS system is also able to monitor the crack propagation of Single Edge Notched Bend (SENB) specimens. The crack propagation during three point bending tests on the clay brick panel was shown by Graziani et al. (2014). These works have confirmed that the ARAMIS facility is qualified for 3D deformation measurements and the applications in the fracture mechanics field.

2.6 Synchrotron Radiation-Computed Laminography (SRCL)

With the development of phase contrast methods (Nugent et al. 1996; Cloetens et al. 1996; Paganin et al. 2002), 3D imaging with Synchrotron radiation-computed tomography (SRCT) became possible. The SRCT technique can help to understand the structure within the material in a non-destructive way. However, SRCT is particularly suitable for the investigation of the damage evolution of bulk materials with the thickness of 1 mm, it is difficult to image the local microstructure when the sample size significantly exceeds the field of view (1.5 mm) of the detector, such as in the case of a flat specimen (Shen et al. 2013). Synchrotron radiation-computed laminography (SRCL) has been developed as a non-destructive three dimensional (3D) imaging method for flat specimens (Helfen et al. 2005). In comparison to the SRCT, SRCL provides better scanning results (Helfen et al. 2011) resulting from the change of the laminographic angle ($\theta = 90^\circ$ for SRCT, $\theta < 90^\circ$ for SRCL), which is the angle between the rotation axis and the incident of the X-ray beam. Figure 2.13 shows the comparison of the schematic views of a typical SRCT setup (Fig. 2.13a) and SRCL setup (Fig. 2.13b). The SRCT is particularly used to scan and image the stick-like samples (Fig. 2.13a) where the SRCL is developed to image the laterally extended specimen, like the flat specimen, as shown in Fig. 2.13b.

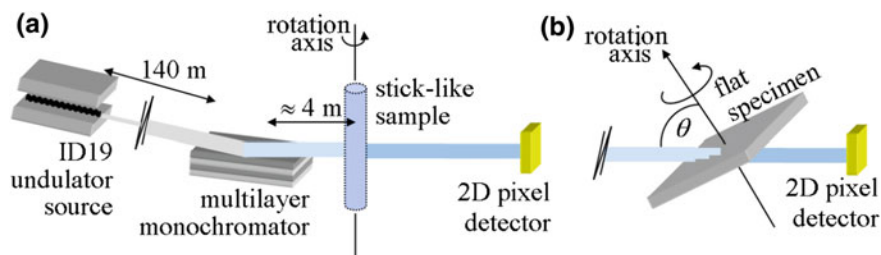


Fig. 2.13 Schematic views of a typical **a** SRCT setup in comparison to the **b** SRCL setup (Morgeneyer et al. 2013). Copyright 2012, with permission from Springer Science + Business Media

With the development of the SRCL technique at ANKA (Synchrotron Radiation Facility at the Karlsruhe Institute of Technology (KIT)), an increasing number of scientific publications are found in literature. The SRCL technique was adopted to investigate the crack initiation of Al2139 (Morgeneyer et al. 2009, 2013), the crack initiation and propagation in Al6061 aluminum alloy sheets (Shen et al. 2013) and the damage evolution in a polymer composite (Xu et al. 2010; Laiarinandrasana et al. 2012; Cheng et al. 2013). These scientific investigations confirm that the SRCL technique is qualified for 3D imaging of the damage evolution of different materials.

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Numerical Simulation and Experimental Investigation of
the Fracture Behaviour of an Electron Beam Welded
Steel Joint

Tu, H.

2018, XVII, 171 p. 190 illus., 163 illus. in color.,

Hardcover

ISBN: 978-3-319-67276-2