

## Fracture Physics

In this chapter we give a brief account of the theory of fracture, restricting ourselves to brittle fracture. The structure of this chapter is as follows. The first part describes the Griffith criterion of fracture initiation, since this criterion constituted the turning point in our understanding of the basics of fracture. We next consider the question of nucleation both of flaws and of fractures in a solid. This process has recently had a breakthrough and is related to the origin of fracture initiation. The general elastic theory of fracture is considered next – linear elastic fracture mechanics (LEFM) as the only complete theory and the rational extension of the Griffith approach. The analysis of two-dimensional problems, for some of which analytical solutions can be obtained, is emphasized. The last parts deal with two important topics that do not fall within and can not be dealt with by LEFM, namely, dynamic crack propagation and creep. In both of these topics, new advances have been made in recent years and both are therefore a subject matter for contemporary research.

A note of caution is in order. Although the Griffith criterion and the related linear elastic fracture mechanics methods have helped scientists in dealing with static cracks and engineers in building safer structures, the basic dynamics of the fracture problem are as of yet not well understood. This relative ignorance stems from the fact that on the one hand, fracturing is neither an equilibrium nor a near equilibrium process and is therefore not amenable to perturbation procedures, and on the other hand, fracturing, whose principle part occurs at a very restricted area and is highly nonlinear, influences the whole extension of the body in question, and it is therefore not easy to correlate between its different scales.

A full understanding of crack propagation would entail a quantum mechanical calculation of all atoms in a certain vicinity of the crack. Since each atom contains several electrons, a full “many-body” calculation (see, e.g., Fabrocini et al. 2002) should be made. Obviously, even numerically, this procedure is unachievable in the foreseeable future. A less presumptuous attempt would be to consider each atom as a point entity and to calculate fracture propagation by assuming a certain interaction among these entities. Even a classical (rather than a quantum mechanical) calculation would have to assume an approximation of only two or at most three body forces, and it would have to assume, moreover, that these two body forces (say of a Lennard Jones type; see, e.g., Ashcroft and Mermin 1981) can be extended to inter-atomic distances far greater (or smaller) than equilibrium ones. Even under such approximations, the number of atoms and the time period necessary for a useful calculation are as of yet beyond the scope of available computational means. Thus, the maximum number of atoms for which such numerical calculations have been tried is the boasted  $10^9$  (Abraham

et al. 2002). However, for a meaningful dynamical calculation a number about a thousand times higher is needed. As for the time periods needed, these should be scaled by atomic vibrations, which entail at least around  $10^6$  time steps even for such a short time as  $1 \mu\text{s}$ , rendering this task unattainable at the present stage.

There are at present several groups of scientists around the world who try to “connect” the different scales that appear in fracture calculations. The idea is to use a quantum mechanical type of calculation in order to obtain the correct two-body atomic force even for large inter-atomic distances. These forces are to be used only for a relatively small number of atoms near the crack tip and the so called “process zone” to obtain an accurate microscopic (classical) calculation. The latter should yield inter alia the energy needed for crack advance, the stresses and strains near the tip and the appearance and motion of dislocations for non-brittle materials. The boundary conditions for this calculation would be the connection to the major part of the material, which is to be treated in a continuous manner (linear or non-linear fracture mechanics). This ambitious plan is still in its infancy but promises to become a real asset in understanding fracture dynamics in the future.

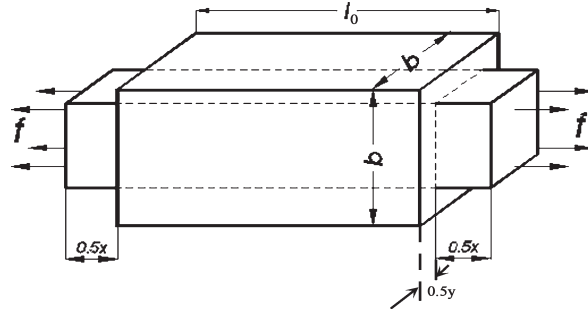
## 1.1 The Griffith Criterion

### 1.1.1 Stress and Strain

Consider the bar in Fig. 1.1 whose initial length and cross section are  $l_0$  and  $A = b \times b$ , respectively. We apply a force in the  $x$ -direction, which causes the bar to elongate by  $x$ . We define  $\sigma_{xx} = f/A$ , as the *stress* operating inside the bar in the  $x$ -direction. Hence, stress has the units of force per area. It is assumed that the bar elongates with the continued application of the force in such a way as to preserve elasticity, namely that it operates like a spring, maintaining a linear relationship between the applied stress  $f/A$  and the elongation,  $x$ ,  $f = kx$  (Hook’s law). This relationship together with Newton’s law  $f = ma$  are the basis of all of elasticity theory (see, e.g., Timoshenko and Goodier 1984). Defining the non-dimensional variable  $\varepsilon_{xx} = x/l_0$  as the strain in the bar (the change in length divided by the initial length), one can write Hook’s law as

$$\varepsilon_{xx} = \frac{\sigma_{xx}}{E} \quad (1.1)$$

**Fig. 1.1.** Schematic stress-strain behavior of a bar ( $l_0 \times b \times b$ ) under uniaxial force  $f$



where  $E$  is Young modulus, related in the case of the bar, to the Hook's constant  $k$  through

$$k = \frac{f}{x} = \frac{\sigma_{xx}A}{\epsilon_{xx}l_0} = \frac{AE}{l_0} \quad (1.2)$$

Under the same stress, there also occurs a (small) reduction of size of the bar in the lateral directions. Thus, if in the  $y$ -direction the decrease of width is denoted by  $-y$  (the minus sign comes to denote a *reduction* of size), the strain in this direction is  $\epsilon_{yy} = -y/b$  and for elastic materials is given by

$$\epsilon_{yy} = -\nu\epsilon_{xx} = -\frac{\nu\sigma_{xx}}{E} \quad (1.3)$$

where  $\nu$  is a non-dimensional constant called the Poisson ratio, which for different materials usually assumes values between 0.1 and 0.5. A similar reduction occurs in the  $z$ -direction ( $\epsilon_{zz}$ ).

An important issue to be used frequently in the following is the value of the elastic energy stored in the material. For the bar problem, suppose we apply the force quasi statically to avoid acceleration and kinetic energy and increase its magnitude from zero to a final value  $F$ . The elastic energy is given by the work needed to create the final state:

$$W = \int_0^{\Delta l} f(x)dx = k \int_0^{\Delta l} xdx = k \frac{\Delta l^2}{2} = \frac{l_0 AE \epsilon_f^2}{2} = \frac{VE \epsilon_f^2}{2} = V \frac{\epsilon_f \sigma_f}{2} = V \frac{\sigma_f^2}{2E} \quad (1.4)$$

where  $\Delta l$  is the final elongation of the bar,  $\epsilon_f = \Delta l/l_0$  is the final strain,  $\sigma_f$  is the final stress ( $= E\epsilon_f$ ) and  $V$  is the initial bar volume. We have therefore obtained the important result that the elastic energy per unit volume ( $W/V$ ) stored in the sample is given by

$$\frac{W}{V} = \frac{1}{2} \sigma \epsilon = \frac{1}{2} E \epsilon^2 = \frac{1}{2} \frac{\sigma^2}{E} \quad (1.5)$$

Although this result was obtained for the specific case of a bar acted upon by a uniaxial force, this result is quite general and we will use it later, e.g., for the Griffith criterion.

In the general case, i.e., for a body of a general shape acted upon by a general distribution of forces and displacements, elasticity theory considers an infinitesimal cube inside the body and the force components per unit area on each side of the cube constitute the stress *tensor*  $\sigma_{ij}$ , where  $i = 1, 2, 3$  and  $j = 1, 2, 3$  denote the  $x$ -,  $y$ - and  $z$ -coordinates, respectively. Thus, for example  $\sigma_{12} = \sigma_{xy}$  denotes the force (per unit area) component in the  $y$ -direction acting on the cube face, which is perpendicular to the  $x$ -direction. This tensor can in general depend both on the location of the cube within the body,  $\mathbf{r}$ , and on time. Therefore,  $\sigma$  is generally written as  $\sigma_{ij}(\mathbf{r}, t)$ . It can be shown that  $\sigma$  is a symmetric tensor, namely

$$\sigma_{ij} = \sigma_{ji} \quad (1.6)$$

The force per unit area acting on a general surface perpendicular to a unit vector  $\hat{n} = n_1\hat{x} + n_2\hat{y} + n_3\hat{z}$  ( $\hat{x}, \hat{y}, \hat{z}$  are unit vectors in the  $x$ -,  $y$ - and  $z$ -directions respectively) is called the *traction*  $\mathbf{t}$  and its components are given by

$$t_i = \sum_{j=1}^3 \sigma_{ij} n_j \quad (= \sigma_{i1}n_1 + \sigma_{i2}n_2 + \sigma_{i3}n_3) \quad (1.7)$$

Next, displacements from equilibrium position are defined such that under the action of forces, a point within the body initially (with no force) located at the position  $(x, y, z)$  is displaced to the position  $(x + u_1, y + u_2, z + u_3)$  by the action of the forces. The displacement vector is thus denoted by  $\mathbf{u} = u_1\hat{x} + u_2\hat{y} + u_3\hat{z}$ . Newton's equation of motion (force = mass  $\times$  acceleration) is transferred into the infinitesimal cube to yield (per unit volume)

$$\sum_j \frac{\partial \sigma_{ij}}{\partial x_j} + f_i = \rho \frac{\partial^2 u_i}{\partial t^2} \quad (1.8)$$

where the first term on the left denotes the effect of the surface forces (by measuring their difference in the  $i$ -direction between the two sides of the cube, while  $f_i$  is the  $i^{\text{th}}$  component of the so-called "body force" if such a force exists (e.g., gravity when it is relatively large compared with surface forces and can not be ignored) and  $\rho$  is the density (mass per unit volume) at the position  $(x, y, z)$ . Equation 1.8 is the basic equation of motion of an elastic medium and all further linear elastic considerations are conducted in order to solve it. An important special case of Eq. 1.8 is static equilibrium, where all movements are zero. In that case,

$$\sum_j \frac{\partial \sigma_{ij}}{\partial x_j} + f_i = 0 \quad (\text{static equilibrium}) \quad (1.9)$$

or, when no body forces exist,

$$\sum_j \frac{\partial \sigma_{ij}}{\partial x_j} = 0 \quad (\text{static equilibrium, zero body force}) \quad (1.10)$$

In order to solve Eq. 1.8, one should have a way to connect  $\sigma$  with  $u$ . This is done via the generalized Hook's law in the following way. Firstly, a strain tensor  $\varepsilon_{ij}$  is defined as

$$\varepsilon_{ij} = \frac{1}{2} \left( \frac{\partial u_i}{\partial x_j} + \frac{\partial u_j}{\partial x_i} \right) \quad (1.11)$$

As can be seen,  $\varepsilon$  is also a symmetric tensor, i.e.,  $\varepsilon_{ij} = \varepsilon_{ji}$ . The most general Hook's law can now be written as a linear relation between  $\sigma$  and  $\varepsilon$ ,

$$\sigma_{ij} = \sum_{k,l} C_{ijkl} \varepsilon_{kl} \quad (1.12)$$

where  $C_{ijkl}$  is a constant tensor of the 4<sup>th</sup> rank replacing the Young modulus of Eq. 1.1. It is a material property. For isotropic homogeneous materials,  $C_{ijkl}$  has only two components  $\lambda$  and  $\mu$  (called the Lamé constants) different from zero and the law can be written as

$$\sigma_{ij} = 2\mu\epsilon_{ij} + \lambda\delta_{ij}\sum_k \epsilon_{kk} \quad (1.13)$$

where  $\delta_{ij}$  is the Kronecker  $\delta$ ,

$$\delta_{ij} = \begin{cases} 1 & \text{if } i = j \\ 0 & \text{if } i \neq j \end{cases} \quad (1.14)$$

For example,  $\sigma_{xz} = \sigma_{13}$  equals  $2\mu\epsilon_{13}$  while  $\sigma_{zz} = \sigma_{33} = 2\mu\epsilon_{33} + \lambda(\epsilon_{11} + \epsilon_{22} + \epsilon_{33})$ . The Lamé constants are connected to the Young modulus and the Poisson ratio  $\nu$  through

$$\nu = \frac{\lambda}{2(\mu + \lambda)} \quad \text{and} \quad E = \frac{\mu(2\mu + 3\lambda)}{\mu + \lambda} \quad (1.15)$$

Now, the right hand side of Eq. 1.8 can be differentiated with respect to  $x_i$  and  $x_j$  to relate it through Eq. 1.11 with  $\epsilon_{ij}$ , and the left-hand side of Eq. 1.8 in its turn can be transformed by Eq. 1.12 or 1.13 to be given in terms of  $\epsilon_{ij}$  (as well). In this way, the ensuing equation will have similar variables on both sides. There is, however, a problem arising from the connection between the displacements and the strain (Eq. 1.11). Although the displacement  $\mathbf{u}$  (a vector with three components) uniquely defines  $\epsilon_{ij}$  (a symmetric tensor with six components), the  $\epsilon_{ij}$  does not uniquely define the displacements. There is a redundancy. In order to insure a unique  $\mathbf{u}$ , the strain components must satisfy an additional criterion (called the compatibility equations, see also Sect. 1.3.3.1)

$$\frac{\partial^2 \epsilon_{ij}}{\partial x_k \partial x_l} + \frac{\partial^2 \epsilon_{kl}}{\partial x_i \partial x_j} = \frac{\partial^2 \epsilon_{ik}}{\partial x_j \partial x_l} + \frac{\partial^2 \epsilon_{jl}}{\partial x_i \partial x_k} \quad (1.16)$$

The mathematical complexity of this set of equations (Eqs. 1.8–1.16) both for the dynamic but even for the static cases makes exact analytic solutions for elastic problems almost unobtainable. Exceptions are problems that due to their symmetry can be reduced to simpler forms. The most common solvable problems are “two-dimensional” (2D) ones, where the influence of the third dimension is ignored. Two types of such problems are of importance. In the first, we deal with a thin plate, where it is assumed that the thickness is much smaller than the other dimensions and can be ignored (plane stress); and in the second a very thick specimen is considered where the third dimension is ignored (plane strain) under the condition that *no change* of the applied force occurs across the thickness of the plate and the whole system is homogeneous and therefore solutions do not change with the coordinate in this direction.

### 1.1.2

#### Energy Considerations and the Griffith Criterion

Fractures occur in solids at much lower stresses than those which can be deduced from simple estimations. For example, if we assume that the solid breaks when the distance between its atomic planes become elongated by, say,  $1/10$  of the equilibrium distance, this would mean that  $\epsilon \sim 1/10$  and  $\sigma = E\epsilon \sim E/10$ . Most solids break under loads that

are two or three orders of magnitude lower than this value. This discrepancy was resolved by Griffith (1920). In principle his argument is that the stress is not transferred to the solid homogeneously. Rather, due to the existence of flaws in the body (and almost all solids contain flaws, unless grown in a very special method – controlled whisker procedure), the stress becomes concentrated at the tip of such a flaw. Hence, strains and stresses at this location become much larger than those prevailing in the rest of the material – leading to fracture there. In fact, it is the longest flaw (favorably oriented), which sets the limit for the critical stress, the stress above which fracture would occur, as is shown presently.

Consider (Fig. 1.2), a large plate of thickness  $b$  and cross section  $A$ , which is operated on by a constant force  $f$  or a stress  $\sigma_0 (= f/A)$  and in which there exists a flaw (fracture) of length  $c$ . The fracture evidently has caused the elastic energy in its vicinity to be relieved. We can estimate the volume wherefrom the elastic energy was relieved to be approximately equal to the half cylinder shown.

This volume is

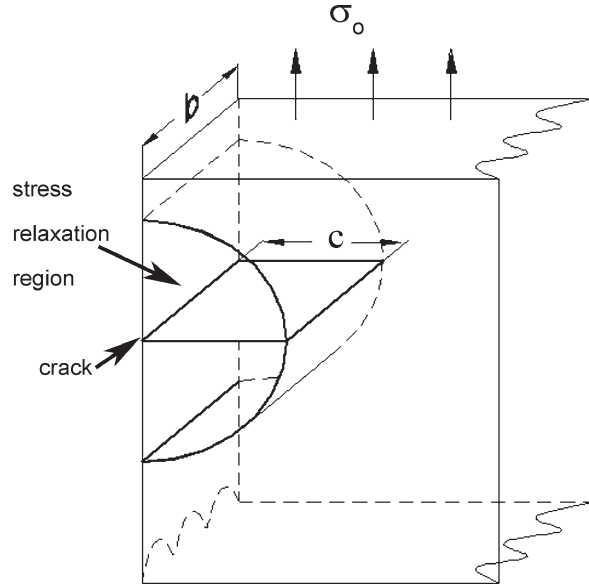
$$V \sim \frac{1}{2} \pi c^2 b \quad (1.17)$$

And according to Eq. 1.5, the total relieved energy is

$$W_E \sim V \frac{\sigma^2}{2E} = \frac{\pi \sigma_0^2}{4E} c^2 b \quad (1.18)$$

Therefore, should the fracture propagate by an additional increment  $\Delta c$ , the change in relieved energy would be

**Fig. 1.2.**  
A through crack in a semi infinite slab and the (schematic) region in which energy is relaxed



$$\Delta W_E \sim \frac{\pi b}{2E} \sigma_0^2 c \Delta c \quad (1.19)$$

By the way, the energy per unit length  $\Delta W_E / \Delta c$  is usually denoted by the (not very fortunate) term “energy release rate”  $G$  (in honor of Griffith).

According to Griffith, the energy thus released from the elasticity stored reservoir should go in its entirety (or almost so) to create the areas of the new addition to the fracture surface. If we assume that the energy needed for the creation of a new unit of fracture area,  $\Gamma$ , is a material constant (it is in fact half of the energy needed to break the chemical bonds between the two newly created surfaces), then the energy required for an increment  $\Delta c$  of the fracture is

$$\Delta W_s = 2\Gamma b \Delta c \quad (1.20)$$

The factor of two comes because two new surfaces are created above and below the increment. Energy balance entails that the change in the released increment of elastic energy be equal to the increment of surface energy, i.e.,

$$2\Gamma b \Delta c \sim \frac{\pi b}{2E} \sigma_0^2 c \Delta c \quad \text{or} \quad \frac{\Delta W_E}{\Delta c} = \frac{\Delta W_s}{\Delta c}$$

yielding

$$\sigma_0^2 c \sim \frac{4}{\pi} \Gamma E \quad (1.21)$$

A similar calculation applies for an internal flaw (crack). Assume that this crack is circular of radius  $c$ . The volume from which the elastic energy was released is

$$V = \frac{4}{3} \pi c^3$$

and therefore in this case

$$W_E = \frac{4}{3} \pi c^3 \frac{\sigma_0^2}{2E}$$

where  $\sigma_0$  is the remote stress perpendicular to the flaw. The surface energy needed to create this flaw is  $W_s = 2\pi c 2\Gamma$ . Griffith's incremental energy balance,

$$\frac{\partial W_E}{\partial c} = \frac{\partial W_s}{\partial c}$$

yields

$$\sigma_0^2 c = 2E\Gamma \quad (1.21')$$

Although the method used here to obtain Eq. 1.21 or 1.21' is somewhat crude, the result, except perhaps for the coefficient, is accurate. This is Griffith's condition. Let us first calculate the value of  $\sigma_0$  needed to initiate a crack and compare it to the homogeneous estimate of  $E/10$  we had before. The decrease with respect to  $E/10$  of  $\sigma_0$  needed for fracture commencement can be seen as a measure of the enhancement of stress at the crack tip relative to  $\sigma_0$ . From Eq. 1.21,

$$\sigma_0 = \frac{E}{q}; \quad q \approx \sqrt{\frac{Ec}{\Gamma}} \quad (1.22)$$

Therefore, the magnitude of  $q$  is a measure of stress increase (relative to  $\sigma_0$ ) at the tip of a flaw of length  $c$ .

Let us examine soda lime glass as an example. The relevant constants for glass are  $E \sim 6.2 \times 10^{10}$  Newton  $\text{m}^{-2}$  and  $\Gamma \sim 0.54$  Joule  $\text{m}^{-2}$ , hence for glass  $q \sim 3.4 \times 10^5 \sqrt{c}$ . This relation is plotted in Fig. 1.3. It is seen that our first estimate of  $q$ ,  $q \sim 10$  is obtained only for flaw lengths of the order of atomic distances, while for usual flaw lengths of the order of 0.1 mm,  $q$  is of the order of several thousands.

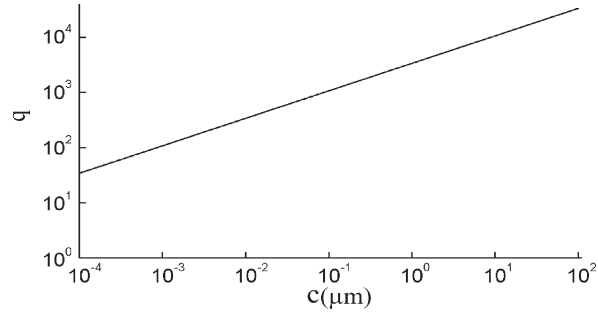
According to the Griffith criterion, Eq. 1.21, therefore, the important experimental variable "responsible" for crack initiation is the product  $\sigma_0^2 c$  of  $\sigma_0 \sqrt{c}$  and not  $\sigma_0$  itself. When this variable reaches a certain threshold, which depends on the specific material and on the geometry of the loading system, fracture should commence. This threshold according to Eq. 1.21 depends on the Young modulus and the surface energy of the fracturing sample. As will be shown next (Sect. 1.3), a more elaborate and accurate treatment of linear elasticity (called linear elastic fracture mechanics (LEFM) an established branch of mechanical engineering nowadays) yields the same result as Eq. 1.21, up to a change of coefficient, for uniaxial constant loading. This LEFM treatment has been used for many years now to achieve solutions for engineering static safety problems and has proven extremely useful for that purpose. Basically, it is an extension of the Griffith approach. As we shall point out, however, this approach, so successful for regular static cracking problems, is inadequate when dealing with the following two important issues:

1. Fracture development under low stresses in very long time periods (the so-called "creep" situation).
2. Dynamic crack propagation.

We dedicate a section for each of these problems.

**Fig. 1.3.**

The amount  $q$  by which the stress  $\sigma_0$ , needed to cause fracture, is decreased from  $E$  as a function of the Griffith-fracture length  $c$





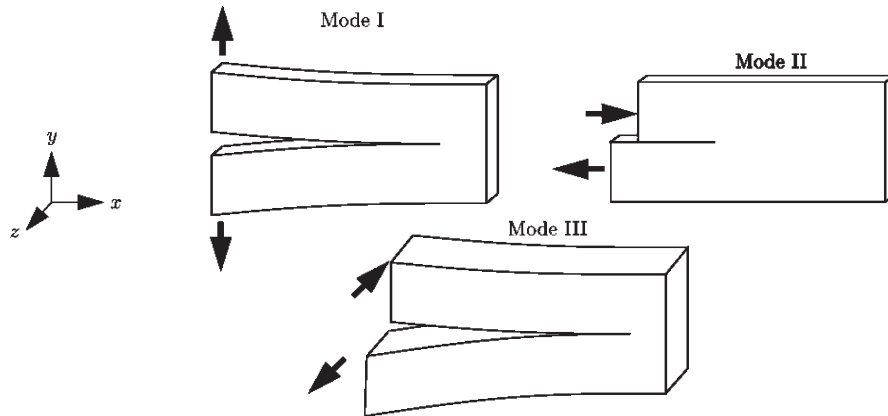


Fig. 1.4. The three fracture modes

### 1.1.3 Fracture Modes

Cracks can appear in solids under stress only in three different geometrical situations, called modes, or in situations that are combinations of these modes. Figure 1.4 shows these three geometries called mode I, mode II and mode III. Almost all engineering treatments of fracture problems are developed in a mode I geometry, which is the prevalent mode of fracture. The presence of mode II and III in addition to mode I is manifested by changes of crack directions as in the appearance of lancets, striae or plumes, undulations etc. In geological cracks, we discern between “joints” that are created in a mode I geometry and “folds” that are created under the other mode types. In the description of all three modes (Fig. 1.4), the fracture surface is in the  $x$ - $z$ -plane and the fracture propagates in the  $x$ -direction. In mode I, stress (tension or compression) is applied in the  $\pm y$ -direction and the displacements  $u$  also occur in the same direction. Modes II and III are shear modes. In mode II, the shearing stress and the displacements are in the same direction ( $\pm x$ ) as that of the crack propagation. Mode III fracture is also called an “antiplane” mode, since stress and displacements occur in the  $\pm z$ -directions, out of the plane of the fracture and normal to the plane of the specimen.

## 1.2 Nucleation

### 1.2.1 General

Let us first define the three different nucleation processes that are used here (see also Sect. 2.2.2 for additional applications of the term “nucleation”):

1. “Flaw nucleation” is the process or processes by which a seed for the subsequent creation of a propagating fracture is created. The seed is usually a flaw in the otherwise “perfect” or homogeneous solid, which facilitates the appearance of a crack in its continuation (see Sect. 1.1.2).

2. “Microcrack nucleation” is the process or processes by which a flaw (which is itself not a microcrack) develops small cracks at the positions of highest stress concentrations such that
  - a the combined structure, flaw plus small cracks, can be considered a microcrack for subsequent developments,
  - b this combined structure is stable, and
  - c it has not reached the critical (Griffith) condition for propagation.
3. “Crack nucleation” is the process or processes by which a *critical crack* (i.e., a crack fulfilling the Griffith criterion for propagation) is created.

Crack nucleation can be “homogeneous”, i.e., with no flaws present (a rare occurrence) or “heterogeneous”, where a flaw distribution already exists in a solid. Note that in the literature the term “crack nucleation” is used to describe both flaw- and crack-nucleation processes. The terms homogeneous and heterogeneous are also used to describe materials. Thus a homogeneous material is an ideal material, which is made up of the same chemical substance, is devoid of grains and is completely devoid of flaws. The only existing material that is almost completely homogeneous is a glass whisker, i.e., a glass fiber grown under very strict conditions of purity. Since no flaws (or almost no flaws) are created in the production phase,  $q$  of Eq. 1.22 is of the order of ten and this material fractures only under very high stresses. Its fracturing is also “homogeneous”, namely all parts of it separate at the same time and the material integrates into dust under a large blast. Heterogeneous brittle materials can be divided into three groups according to the degree of severity of the heterogeneity. The least heterogeneous material is regular glass, containing flaws of all kinds (see below). Next in line are rocks or ceramic materials having the same chemical composition. These materials are grainy materials containing grain boundaries and “triple points” as inherent flaws. This group further divides into isotropic (i.e., having the same properties in any direction) and anisotropic. The most heterogeneous (third) group of materials are ceramics or rocks which consist of substances having different chemical compositions. These are usually both grainy and anisotropic. Homogeneous (heterogeneous) nucleation is a nucleation process in a homogeneous (heterogeneous) material.

We use the term “flaw” here as a general term for defect. This is the usual meaning of the term in the fracture literature, while, e.g., in solid state physics the common term is “defect”. A flaw can be two-dimensional in the form of a microcrack, or three-dimensional, such as a pore, an inclusion, a vacancy, a dislocation etc. The term “Griffith flaw” is used in the literature generally to describe a flaw in the form of a microcrack (the common form of flaw in glass). A three-dimensional flaw can be transformed to a microcrack by developing actual microcracks at points where high stress concentrations exist (critical or singular points; see below). Note that regular glass, besides Griffith flaws, can contain other flaws (voids etc.) generated during its formation stages or by external environmental degradation. In general, the appearance of flaws in a solid or on its surfaces can be induced by the activity of two mechanisms, “external agents” operation or internal processes. The first mechanism includes surface abrasions and erosion, corrosion by chemical substances, radiation and impact effects, etc. The operation of each agent is subject matter for extensive research. A short review can be found in Lawn (1993, Chap. 9). We shall not consider here the flaws that develop during the production phase and include pores, vacancies, grain boundaries, inclusions etc. (see *ibid*). In this section, we will concentrate on the internal

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