

CRACK GROWTH MECHANISM MAPS

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ABSTRACT

Fracture at high temperatures usually has two stages: (i) a region of subcritical crack growth which occurs by a variety of mechanisms depending upon temperature, imposed stress, environment, and material parameters; and (ii) a region of fast fracture that occurs when the slowly growing crack attains a critical length. In this paper we focus our attention on the region of subcritical crack growth since this region can occupy a large portion of the lifetime of a structural material in a high temperature application. Imposed parameters that control fracture in this regime are stress intensity factor, K , and temperature. In response to the imposition of an applied K and temperature, a crack will grow subcritically at some rate until K reaches some critical value and fast fracture ensues. To avoid fracture however, a designer needs to know not only the crack growth rate as a function of temperature and applied loads, but also how this function, or crack growth mechanism, changes with these controlling parameters. Also, materials selection for engineering applications requires some means of comparing the crack growth data for several materials over a range of applied loads and temperatures. In this paper we present a method to display this data and to categorize the various mechanisms of subcritical crack growth at elevated temperatures. In these mechanism maps the experimentally controllable parameters, K and temperature, are plotted as ordinate and abscissa related by contours of constant material response, i.e., crack growth rate. This sort of map emphasizes changes in a material's fracture mechanism as either K or the temperature change, and thereby displays various regions of a material's crack growth response. Data are presented for silicon nitride, a nickel-base superalloy and an aluminum alloy. Mechanisms are proposed to explain the various regions.

KEYWORDS

Creep crack growth; crack growth mechanism; alloy 718; silicon nitride; aluminum alloy; time dependent fracture; high temperature fracture.

INTRODUCTION

Fracture at elevated temperatures can be thought of as occurring in three stages: crack nucleation, crack growth, and rapid fracture. It is now well known that

for most practical purposes, engineering structures begin life with cracks, albeit small, already present in them. Since crack generation is not necessary, it is reasonable to assume that, for these important structures, fracture at high temperatures takes place in only two stages. First, there is a period of slow crack growth which occurs by a variety of mechanisms depending on environment, temperature, imposed loads, and material parameters. For the purposes of this paper, environmental effects are specifically excluded. When the crack has grown to a critical length, the slow stage is ended by the intervention of fast fracture. The onset of fast fracture may result from a change in crack propagation mechanism, or it may be defined simply as the point at which the crack velocity or the change in crack length exceeded some arbitrary value. From the above description of events, it is easy to see why crack velocity and a critical stress intensity factor for fracture, K_{f1} , are more useful for life prediction purposes than the

time-to-fracture of a smooth bar. In the measurement of crack growth rate and toughness, the parameters under the experimentalist's control are the stress intensity factor, K , and the temperature, T . In response to the imposition of a given K and T , a crack will grow with some velocity, v (Fig. 1a). At some critical K , fast fracture will occur (Fig. 1b). These data (crack velocity as a function of K and T) are usually displayed on a graph having K as one axis and v as the other, with contour lines connecting data points obtained at a constant temperature. Such a diagram (Evans and Wiederhorn, 1974) is shown in Fig. 1a. This mode of presentation highlights the K dependence of v . However, as can be seen in Fig. 1a, the isothermal contours exhibit a distinct change in slope; an effect which has been identified with a change in the kinetic mechanism of crack growth as will be discussed below. Since the isothermal contours overlap one another, it is not always possible in K - v space to delineate a region in which only one mechanism of crack growth dominates. Nor does this mode of presentation lend itself to the simultaneous indication of K_f as a function of temperature unless

K_f is associated with a specific velocity. There is another way of presenting the same data which emphasizes the temperature dependence of v . That is to plot v versus T and have contours of constant K . Similar reservations apply to this mode of presentation as were noted for the K - v diagram.

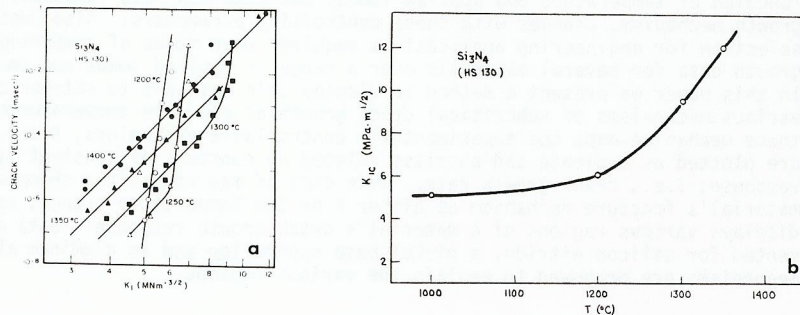


Fig. 1. (a) Crack Velocity as a function of stress intensity factor and (b) critical stress intensity factor as a function of temperature.

¹In ceramics K_f is identified with K_{IC} , but this designation is inconsistent with the current ASTM definition (ASTM, 1978) which does not consider time-dependent crack growth.

There is a third way of presenting the data which we feel gives greater insight into a material's fracture behavior than the two above representations. This representation uses the experimentally controllable parameters, K and T , as the axes of a graph on which contours of constant material response, i.e., crack velocity, are drawn (see Fig. 2). This sort of map emphasizes the dependence of a material's fracture response to both the temperature and the stress intensity factor. It also provides a way of clearly delineating regions of K and T in which one mechanism of crack growth predominates. In analogy to the deformation and fracture mechanism maps developed by Ashby (1972, 1977), we propose to call

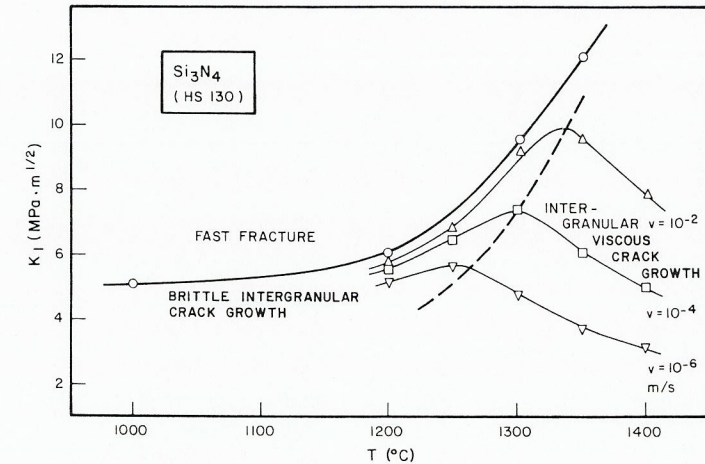


Fig. 2. Crack growth mechanism map for HS-130 silicon nitride.

these diagrams crack growth mechanism maps. In Ashby's maps, the axes are stress and temperature. Contours of constant strain rate or time-to-fracture are drawn in this space which is then subdivided into fields giving the dominant mechanisms of deformation or fracture. Here, we plot contours of constant slow crack growth rate in a K - T space subdivided into fields in which one crack growth mechanism predominates.

In general, the resistance of a material to slow crack growth depends on temperature and stress intensity factor. Any change in the mechanism of cracking or fracture will manifest itself as a change in dependence on these two controlling parameters. On the map, such a change would generally, but not necessarily, be indicated by an abrupt change in slope of the constant velocity contour lines. The loci of such slope changes define boundaries between different mechanisms of slow crack growth or fast fracture as shown in Fig. 2. The boundaries may be sharp or diffuse, but within a set of boundaries exists a region in which one mechanism of slow crack growth or fracture is dominant. In this paper, we discuss these crack growth mechanisms for silicon nitride, Alloy 718, and 2219 aluminum alloy, and present preliminary crack growth mechanism maps for these materials. A more detailed discussion of crack growth mechanism maps for these and other materials will be presented elsewhere (Fields and Fuller, 1981).

CRACK GROWTH MECHANISMS IN SILICON NITRIDE

The crack growth mechanism map for silicon nitride (HS 130) is shown in Fig. 2. It was constructed from the data (Evans and Wiederhorn, 1974) presented in Fig. 1. There are three mechanisms of crack growth indicated on this map: cleavage crack growth, brittle intergranular crack growth, and viscous intergranular crack growth. There are probably other mechanisms, particularly at higher temperatures or lower K_f levels, but these have not been identified yet. A possible example might be the diffusive crack growth observed in many other materials.

To understand the micromechanisms of crack growth, it is helpful to have some idea of the microstructure of HS 130 silicon nitride. It consists mostly of fine silicon nitride grains (1 to 2 μm in diameter) which are hot-pressed with MgO as a densification aid. This MgO, stray SiO_2 , and other impurities can combine to form a glassy second phase which exists as grain boundary films and pockets between the silicon nitride grains (Lange, 1980). The fracture path at and above K_f is rather insensitive to these grain boundaries. The crack tends to cleave the grains and, so, follows a transgranular path.

At lower crack velocities, below K_f , the crack tip becomes more sensitive to the grain-boundary microstructure. The grain boundaries, particularly those with pockets or films of glass become preferred fracture paths. This cracking of the glassy phase is a very brittle process, like cleavage, yet results in an intergranular fracture surface (Fig. 3a). We call this brittle intergranular crack growth. From the map we see that, within this mechanism field, it becomes more difficult (i.e., requires a higher K) to drive the crack forward at a given rate as the temperature increases. This is because glassy components in the grain boundary become less viscous and absorb increasing amounts of energy from the crack tip stress-strain field as the temperature increases. In this mechanism field, the crack velocity may be described by some function of temperature and stress intensity factor: $v=g(K,T)$.

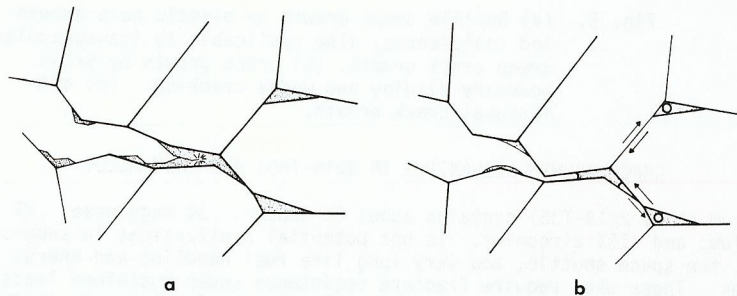


Fig. 3. (a) Brittle intergranular crack growth in which a sharp crack fractures intergranular phase. (b) viscous intergranular crack growth in which crack propagates by the fluid motion and cavitation of molten grain boundary phase.

At higher temperatures, the fluid nature of the grain boundary phase aids crack growth. Cavitation ahead of the crack and grain boundary sliding leading to wedge cracking becomes the dominant crack growth mechanism (Fig. 3b). This

mechanism has a different dependence on K and T than the mechanism of region A: $v=h(K,T)$. The higher the temperature, the more fluid the grain boundary becomes. The more fluid the grain boundary, the easier it is (i.e., requires a lower K) for the grain boundary to slide or cavitate, and the easier it is to propagate a crack at a constant rate. The crack growth mechanism briefly described here has been modelled in detail by Evans and Rana (1980).

The boundary in K - T space between the two mechanisms is defined by $g(K,T)=h(K,T)$. We have determined this boundary by simply drawing a line through the peaks in the constant velocity contours since this appears to be where the mechanism of crack growth changes.

CRACK GROWTH MECHANISM IN ALLOY 718

Alloy 718 is a wrought nickel-base alloy which is used in turbines and other high temperature applications up to about 700°C. It differs from other nickel-base superalloys in that niobium is substituted for much of the aluminum and titanium, and iron is substituted for most of the molybdenum and all of the cobalt. The effect of these differences is to reduce the high-temperature strength while increasing the weldability. The largest and most consistent body of data on crack growth in this material is that of Sadananda and Shahinian (1977, 1978). Mills and Knecht (1976) have measured K_{IC} in this material at various temperatures.

These data have been used to construct the maps shown in Fig. 4. On this map, we have delineated four crack growth mechanism fields. In dynamic fracture, the crack lengthens by ductile crack growth, i.e., a plastic process in which cavities are nucleated on inclusions ahead of the crack tip (Fig. 5a). These cavities

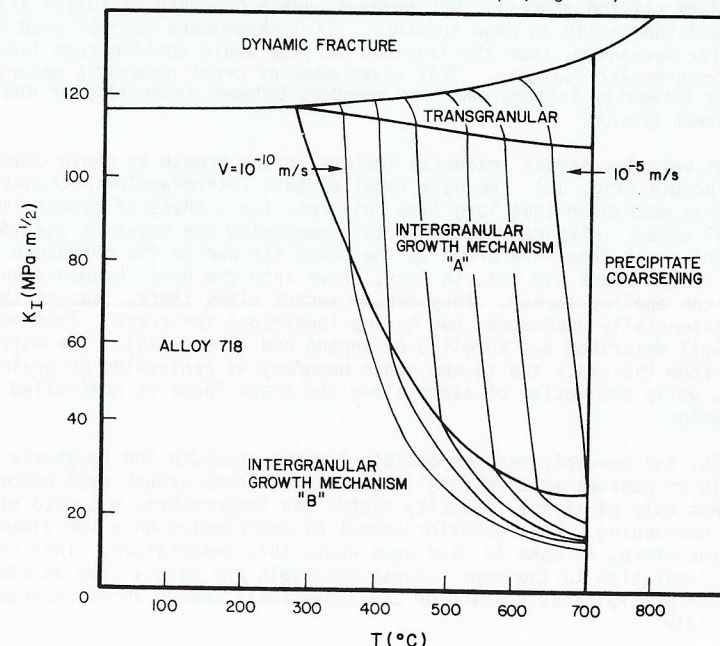


Fig. 4. A crack growth mechanism map for Alloy 718.

grow by conventional dislocation plasticity until they become large enough to interact. Then shear bands form between adjacent cavities and the crack tip. Intense deformation occurs in these bands until the cavities link up to the main crack and thus result in crack extension. At elevated temperatures, this transgranular mechanism becomes significantly time dependent as the cavities can now grow by a dislocation creep process. The appearance of the fracture surface and essential features of this mechanism are not qualitatively different from ductile crack growth at low temperatures. However, the work hardening characteristics and the strain rate sensitivity are certainly altered from their low temperature values. The result is that the intensified stresses at inclusions tend to be lower so that nucleation of cavities at inclusions may be postponed to larger strains. Also, the strain rate sensitivity in creep tends to stabilize flow and thereby postpone the linkage of cavities to the main crack. For these reasons, we distinguish ductile crack growth at creep temperatures from that at lower temperatures, calling it transgranular creep crack growth.

At lower stress intensity factors, a new mechanism of crack extension appears which is labelled Intergranular Mechanism "A". The evidence indicates that grain boundary sliding and wedge cracking ahead of the main crack tip are the important features of this mechanism (Fig. 5b). At elevated temperatures, grain boundary sliding occurs. If the resulting triple junction offset can be accommodated by deformation in the adjoining grains, then no cracking will take place. If, however, a large amount of sliding takes place or the sliding occurs too rapidly then intergranular stresses build up until cavities begin to nucleate at the triple junction. These cavities will form most easily near the crack tip where stresses and grain boundary sliding rates are high. It is not clear exactly how these cavities and the main crack join together. The linkage could occur by a grain boundary sliding process. Or, perhaps, grain boundary diffusion allows the crack tip and the cavity to grow together. If linkage were accomplished by a transgranular mechanism, then the fracture surface would contain both intergranular and transgranular features. This mixed mode of crack growth is observed at high stress intensity factors near the boundary between intergranular and transgranular crack growth.

Finally, at very low stress intensity factors, crack growth by grain boundary diffusion occurs (Fig. 5c). We have labelled this Intergranular Mechanism "B". It is by this mechanism that long time failures, i.e., those of commercial importance, will occur. Physical models of this mechanism are based on the idea that atoms on the crack faces are drawn to the crack tip due to the curvature there. The atoms at the crack tip are, in turn, drawn into the grain boundary under the action of the applied stress. They occupy vacant sites there, leaving the grain boundary essentially unchanged, but having lengthened the crack. This mechanism has been well described and modelled by Chuang and Rice (1973). The motion of the atoms from the crack tip to the grain boundary is controlled by grain boundary diffusion, while the motion of atoms along the crack faces is controlled by surface diffusion.

Above 700°C, the precipitates responsible for the strength and toughness of this alloy begin to coarsen at a non-negligible rate. Crack growth rate becomes a function not only of stress intensity factor and temperature, but also of time or degree of coarsening. This behavior cannot be represented on a two dimensional map and, therefore, no data is displayed above this temperature. This is not an important limitation to the maps because materials are rarely used at temperatures at which the precipitates coarsen or any appreciable change in microstructure occurs rapidly.

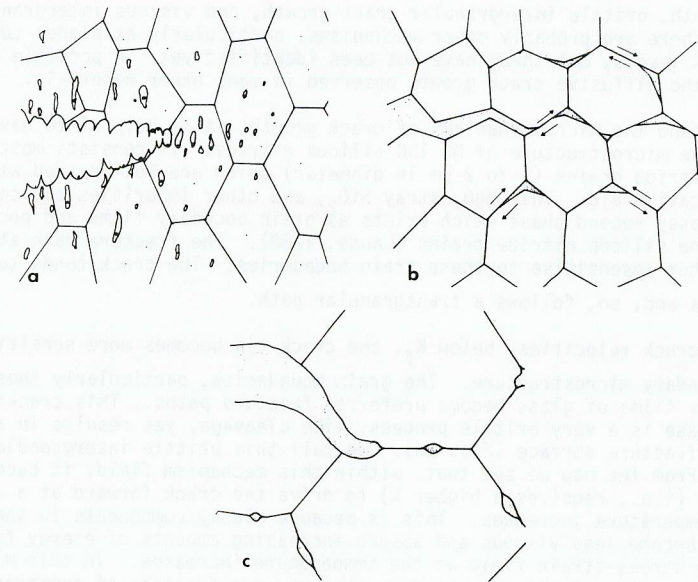


Fig. 5. (a) Ductile crack growth by plastic hole growth and coalescence, also applicable to transgranular creep crack growth. (b) crack growth by grain boundary sliding and wedge cracking. (c) diffusional crack growth.

CRACK GROWTH MECHANISMS IN 2219-T851 ALUMINUM ALLOY

Aluminum alloy 2219-T851 contains about 6% copper, .3% manganese, .2% iron, .1% vanadium, and .15% zirconium. It has potential applications in supersonic aircraft, the space shuttle, and very long life fuel handling and energy supply systems. These uses require fracture resistance under sustained loads at temperatures which are high for aluminum alloys (>100C.). Crack growth in this particular alloy has been studied quite thoroughly at Alcoa and some results are presented in two papers by Kaufman and co-workers (1974, 1976). A map has been constructed from the data obtained from this work and is shown in Fig. 6. As for silicon nitride and alloy 718, there is a region of dynamic fracture which occurs above a "fracture K_I " (Kaufman's terminology). Fracture K_I is that stress intensity

factor at which slow stable crack growth is replaced by unstable, rapid fracture; it is what we call K_f in this paper.

In common with Alloy 718, there is a temperature above which the microstructure changes too rapidly to make possible a sensible presentation of unique crack velocities as a function of K and T . Also, as in Alloy 718, there is a ductile

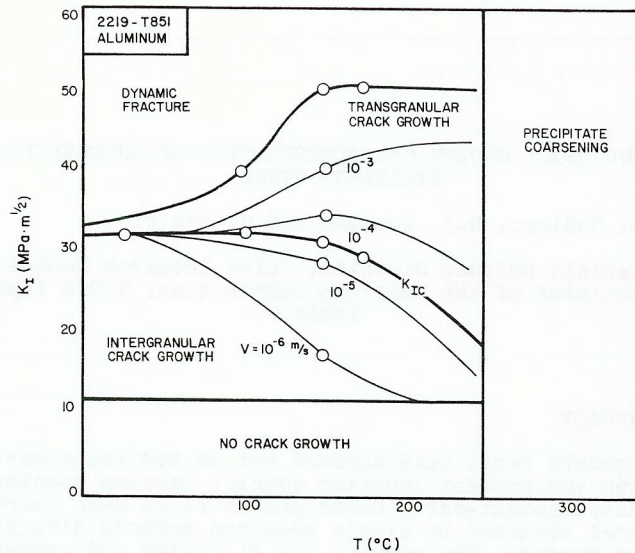


Fig. 6. A Crack growth mechanism map for aluminum alloy 2219.

transgranular crack growth mechanism and an intergranular crack growth mechanism. Transgranular crack growth takes place by the time-dependent or time-independent plastic enlargement of cavities that have formed around inclusions, and the subsequent linkage by shear of these cavities to the main crack (Fig. 5a).

The boundary between transgranular and intergranular crack growth in this aluminum alloy is given by K_{IC} as a function of temperature. K_{IC} , as defined by ASTM (1978) and as used by Kaufman, is not the point of unstable crack growth and fracture in these aluminum alloys. This unstable point is denoted "fracture K_I " by Kaufman. K_{IC} is, however, the stress intensity factor at which significant crack growth is first detected, but crack velocity may still be quite low. Considerably higher loads may be supported by the specimen (at least temporarily) than those determined from the K_{IC} . Kaufman noted that at and above K_{IC} , the crack path was predominantly transgranular and followed clustered constituent particles. The sustained load crack path at lower stress intensities was predominantly intergranular.

The intergranular mechanism of crack growth in 2219 as determined by Kaufman is similar to Intergranular Mechanism "A" in alloy 718 (Fig. 5b). That is, the high stress field ahead of the crack causes grain boundary sliding. When the displacements due to this sliding cannot be accommodated by deformation, cracks form ahead of the dominant crack tip at grain boundary triple junctions (wedge cracks) and at particles or steps on a two-grain boundary. Kaufman and Low (11) found

that the grains on either side of these partially cavitated boundaries then completely separate and join the main crack by a conventional plastic hole growth mechanism within the precipitate-free zone of the grain boundary. The result is a completely intergranular crack path containing a certain fraction of shallow dimples as well as the original larger creep cavities. This mechanism of cavity linkage may be applicable to intergranular crack growth in Alloy 718.

Finally, it was found that below about $13 \text{ MPa} \cdot \sqrt{\text{m}}$, no creep crack growth was observed. In general, all materials will have a thermodynamic minimum stress intensity factor below which no crack growth is possible. This is the Griffith limit. In the case of a grain boundary crack, the limit is

$$K_{\min} = \sqrt{(2\gamma_s - \gamma_{gb}) \times E/(1-\nu^2)} \quad (1)$$

where γ_s and γ_{gb} are the surface and grain boundary energy, respectively; E is Young's modulus and ν is Poisson's ratio. Evaluating this expression for aluminum, using $\gamma_s = 1.1 \text{ J/m}^2$ (Jones, 1971) and $\gamma_{gb} = 0.625 \text{ J/m}^2$ (Hondros, 1969), indicates that no crack growth would be expected below about $0.4 \text{ MPa} \cdot \sqrt{\text{m}}$. The observed threshold of $13 \text{ MPa} \cdot \sqrt{\text{m}}$ is much higher and must be due to some other, as yet undetermined, cause.

USE AND LIMITATIONS OF THE MAPS

Two areas in which crack growth mechanism maps may be used effectively are in life-time prediction and material selection. The life of a component which fails by crack growth and fracture may be determined from the following equation:

$$\int_0^{t_f} dt = t_{(f)} = \int_{a_0}^{a_c} \frac{1}{g(K,T)} da \quad (2)$$

where a_0 is the initial crack length. The critical crack length for rapid fracture, a_c , is determined from the fracture K and the applied load. If there are several mechanisms of crack growth, the above integration must be performed in a piece-wise fashion:

$$t_{(f)} = \int_{a_0}^{a_{g/h}} \frac{1}{g(K,T)} da + \int_{a_{g/h}}^{a_c} \frac{1}{h(K,T)} da \quad (3)$$

where $a_{g/h}$ is determined by the boundary in K - T space between mechanism g and mechanism h , and the applied load. Crack growth mechanism maps are a logical way of locating the boundaries between different mechanisms.

The effect of a multiaxial stress state on crack growth may depend sensitively on the mechanism by which this growth takes place. For example, ductile transgranular crack growth, which relies on the deviatoric stress for its driving force, responds differently to a biaxial stress state than, say, diffusive crack growth or cleavage crack growth which rely on the normal stress for their driving force. The maps indicate the mechanism by which crack growth is occurring and thus allow the designer to use the appropriate multiaxial crack growth criterion for the case at hand.

In the area of materials selection, the maps can be used to quickly compare several materials over a wide range of temperatures and stress intensity factors. This comparison can be made between materials having slightly different compositions or heat treatments, and so the maps could be useful in materials development.

In conclusion, the following two limitations must be noted. It is assumed here that crack velocity is a unique, steady-state function of stress intensity factor and temperature. Thus, transient effects are specifically excluded. Secondly, the material parameters involved in the fracture process must not change with time or crack growth. Therefore, precipitate coarsening and degradation or work hardening of the material far from the crack tip are not taken account of in the maps. Nevertheless, the maps are a logical and useful mode of presentation of fracture and crack growth data and theories. They provide a broad and selfconsistent picture of how materials resist fracture.

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