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# Growth of very long "short cracks" initiated at holes

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ABSTRACT

The initiation and growth behavior of very long microstructurally short fatigue cracks formed at circular holes is described. Very long here means cracks which are several millimeters or even centimeters long. Microstructurally short refer to the fact that these cracks, in spite of their physical length, are still smaller than the grain size of the material and thus exhibit the characteristic features of such cracks. Growth retardation or even halt at grain boundaries and fluctuating crack growth rates can readily be observed with the naked eye by employing a experimental technique which allows one to increase the grain size of Al1050 Aluminum alloy until the centimeter scale by applying a series of mechanical and heat treatments. Once the thermo-mechanical treatment is completed and the desired grain size obtained, a circular notch is machined on each specimen, and the samples are subjected to fatigue loading. With this method, interactions between cracks and microstructural barriers can be studied with an unprecedented level of ease and detail. An interesting observation is that the location of the crack initiation point along the hole contour varies greatly with the ratio between the hole diameter and the grain size: for large ratios, the initiation point is located close to the point corresponding to the maximum circumferential stress (the horizontal symmetry axis in our case), but for smaller ratios, however, the point of crack initiation moves markedly away from the symmetry axis.



## 1. Introduction

The problem of the characterization of the crack tip stress and strain fields for fatigue problem has a long history. The inauspicious beginnings of the characterization of fatigue crack growth rate of long cracks by the stress intensity factor range as described in the Paris law is a well-known episode in fracture mechanics lore. There, a single parameter embodies the physics of the problem and crack growth rate can be expressed as a unique function of $\Delta K$ irrespective of the geometry of the component or the type of load applied. Dimensional analysis arguments (Navarro and de los Rios [1]) show that fatigue crack growth rate can indeed be expected to depend on $\Delta K$ alone when the size of the plastic zone relative to the crack length does not change. Consider a fatigue crack growing in an infinite medium (see Fig. 1, taken from [1]). As it is well known, the growth of a crack by fatigue depends on the plasticity generated at the tip of the crack. The dimensional analysis should then consider the length $a$ of the crack and a new characteristic length $c$ representing the extent of the plastic zone. Then, the rate of crack growth can be shown to be given by

$$\frac{da}{dN} = C \cdot \Delta K^m \cdot \left(\frac{a}{c}\right)^n \tag{1}$$

But, of course, for long cracks and for the values of applied stress in which LEFM is considered to hold, $\tau/\tau_y < 0.3$, the Dugdale relation

$$\frac{a}{c} = \cos\left(\frac{\pi}{2}\frac{\tau}{\tau_y}\right) \tag{2}$$

gives values of between 0.9 and 1 for the ratio $a/c$, and therefore, neglecting this variation above, the Paris law is obtained

$$\frac{da}{dN} = C^* \cdot \Delta K^m \tag{3}$$

When plasticity is more extensive, the term $a/c$ should be taken into account and the parameter $\Delta K$ on its own does not describe physical similarity either. Attention is also drawn to the idea that the Dugdale-type relationship describes the self-similar growth of the crack across an ideal homogeneous body, where there is no distinguishable microstructural feature that may be used to define any intrinsic unit of length. It is only in these conditions that single parameter characterizations of fatigue crack growth would be applicable.

Microstructurally short fatigue crack growth is thus a prime example of a situation where a single parameter characterization of the crack stress and strain fields does not seem to be entirely appropriate. It is a very important problem from the practical point

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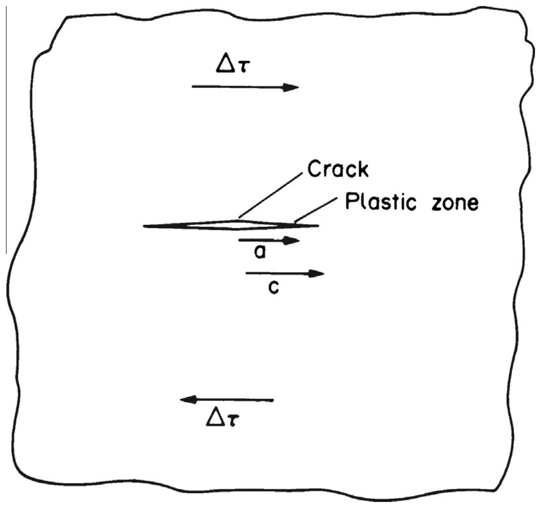


**Fig. 1.** A crack in an infinite body.

of view, for there are many situations where the life of the component is "decided" when the crack length is of the order of the grain diameter. This is particularly the case in long-cycle fatigue problems when the load is close to the fatigue limit. It is now well established that the fatigue limit in metallic materials is really a threshold condition for the propagation of very small cracks. Fatigue cracks have been reported to initiate at persistent slip bands (PSBs), grain boundaries, pores and non-metallic inclusions, see the two recent reviews by Chan [2] and by Sangid [3], where initiation mechanisms and the role of microstructure thereof are discussed at length. When the length of a advancing crack is comparable to major microstructural features, its growth rate is very sensitive to the distribution of phase and grain boundaries ahead of it. Evidence of grain or phase boundaries retarding or even permanently arresting the growth of short cracks has been reported in many materials including aluminum alloys [4], titanium alloys [5] a nickel base superalloy [6] and carbon steels [7]. At stresses below the fatigue limit, cracks start growing fast but then they stop and become non-propagating [8,9]. At stresses just above the fatigue limit, cracks decelerate and may temporarily halt a number of times, but they do not stop growing altogether. Later on they accelerate and finally reach a regime of apparent continuous propagation [4,10–14]. Microscopic observations have identified the locations of minimum crack growth as microstructural barriers to slip propagation such as grain or phase boundaries. The crack decelerates on approaching the grain boundary until a new slip band is initiated in the neighboring grain, along which the crack will propagate next. This nucleation process of slip bands has to be repeated afresh in each grain [15].

The study of the fatigue limit of notched components is another case of interest. Here the problem is one of short crack growth, with the additional difficulty that the stress field through which the crack is growing has usually a very steep gradient. It has been found that in sharp notches it may be possible for a crack to grow through a few grains and then to become non-propagating [16–19]. This suggests that the critical event in notch fatigue may not be the initiation of the crack itself, but rather the relative capacity of this crack to overcome successive microstructural barriers when the stresses that are driving it diminish rapidly. Typical non-propagating cracks found in sharp notched specimens of carbon steel have lengths of the order of a few tens of microns [17].

In the growth of microstructurally short cracks in metallic materials self–similarity is not preserved. The actual physical size of the crack is not as important as the relative size of the crack with respect to the characteristic microstructural dimension [20–22]. The size of the crack at any point in time must be measured in terms of how much of the grain it has traversed or how many grains it already spans.

We have devised [23,24] a simple experimental technique whereby all this can be studied with an unprecedented level of ease and detail. The innovative aspect of this technique is the use of specially developed test coupons with grain sizes of a few millimeters – or even centimeters – and the use of low magnification USB cameras by means of which the crack growth process and the interactions with the microstructure can easily be registered and examined. Digital image correlation techniques have also been employed to enhance the technique. In this paper we have used this new technique to analyze the behavior of very long "short" fatigue cracks formed at circular holes. Very long here means cracks which are several millimeters or even centimeters long. Holes can be made as small as to fit inside a single grain of the metal or as big as to cover a very large number of grains. An intriguing result is the observation that the location of the point of crack initiation along the contour of hole seems to vary with the ratio between the hole radius and the grain size.

## 2. Experimental procedure

### 2.1. Thermo-mechanical treatments

Our studies show that it is possible to produce a substantial increase in the grain size of commercially pure aluminum sheets by means of a combination of two thermal treatments and an intermediate moderate cold working. The process is very simple and easy to control. It is quite feasible, through the correlations derived between the final grain size and the control parameters, to set the values of these parameters in order to obtain any desired grain size fixed in advance (within a certain range). The technique has been shown to produce highly consistent and repeatable results [23,24].

Aluminum 1050 99.5-H24 is used in sheets 4.0 mm thick. Chemical composition (% wt): 99.56; Cu: 0.08; Fe: 0.2; Si: 0.1. The aluminum sheets are cut into pieces of 45 × 300 mm, in parallel to the lamination direction. A tubular furnace is used for the thermal treatments (Carbolite model 215GHA12). The aim of the first thermal treatment is to obtain a deformation-free equiaxial structure; different treatments were tried, changing the recrystallization temperature, the heating rate and the period at constant temperature. After observing the resulting microstructures, a heating rate of 2.6 °C/min was chosen, from room temperature to 550 °C. Then the temperature was kept constant during 5 h, followed by air cooling. It has been found that if this temperature is maintained for more than 5 h, the surface grains become slightly larger than the inner grains, and this induces a non-uniform deformation across the specimen thickness during the following stage (mechanical treatment). Next, cold working is performed in a MTS 810 servo-hydraulic machine. The strain level applied at this stage will determine the size of the grains after the second recrystallization. The following deformation percentages were chosen: 0, 8%, 11%, 14% and 18%. Zero corresponds to the material undergoing only the first recrystallization. The third stage is again a temperature ramp from room temperature to $T = 550$ °C at a rate of 2.6 °C/min; then, the temperature is to be kept constant during 15 h and then increased again up to 575 °C and maintained there during 1 h. At this stage new crystals growing at the expense of the old ones are formed. Finally, the specimens are air-cooled.

As it has been said above, the final grain size obtained depends critically on the level of plastic strain applied during the cold working. Fig. 2 shows an example of the possible microstructures obtained and the corresponding level of total deformation applied,