Influence of Dwelling Time on Low Cycle Fatigue

K. T. Rie, Institut für Werkstoffkunde und Herstellungsverfahren
der Technischen Universität Braunschweig, Germany
and
H. P. Stüwe, Erich - Schmid - Institut für Festkörperphysik,
Leoben, Austria

Low cycle fatigue is produced by strain amplitudes well above the yield strain. The number of cycles to fracture, N_B, decreases with increasing plastic strain amplitude $\Delta\gamma$ as given by the relation

$$(N_B)^{\alpha}\Delta\gamma = \text{const.} \qquad (1)$$

In view of the wide scatter of experimental results the value of α is still a matter of dispute. The original assumption $\alpha = 1$ (Orowan [1]) has not always been confirmed by subequent research (see, e.g., Coffin [2] where $0.5 \leqq \alpha \leqq 0.7$).

The present anthors have shown that the experimental scatter can be greatly reduced by using a "critical number of cycles", N_{cr}, instead of N_B. N_{cr} is defined as the number of cycles leading to formation of a crack which noticeably weakens the specimen, eliminating thus the random events that influence the growth of the crack before complete fracture. N_{cr} can be taken from a plot of stress vs. integrated strain (fig 1) in the region where the apparent stress significantly deviates from a linear extrapolation. The details of the procedure (experiments were carried out in torsion) have been described elsewhere [3].

This evalutation permits an unambiguous description by equ. (1) (see fig 2) which even includes failure in unidirectional strain (1/4 cycle). It also permits a convenient evalution of the influence of dwelling times Δt between cycles.

Such an influence has been reported previously 4)5). One example (Armco iron at room temperature) is given in fig 3 which shows that some dwelling time between cycles not only shortens specimen life but also increases the peak stress. fig 4 shows that the influence on specimen life can be described by a relation of the type

$$N_{cr}^{\alpha} (\Delta t)^{\beta} = \text{const} \tag{2}$$

In Armco iron β was found to be 0.25. By contrast, no influence of dwelling time was found in high purity aluminium ($\beta = 0$). This seems to suggest that the effect may be caused by the movement of interstitial atoms. Experiments are being continued.

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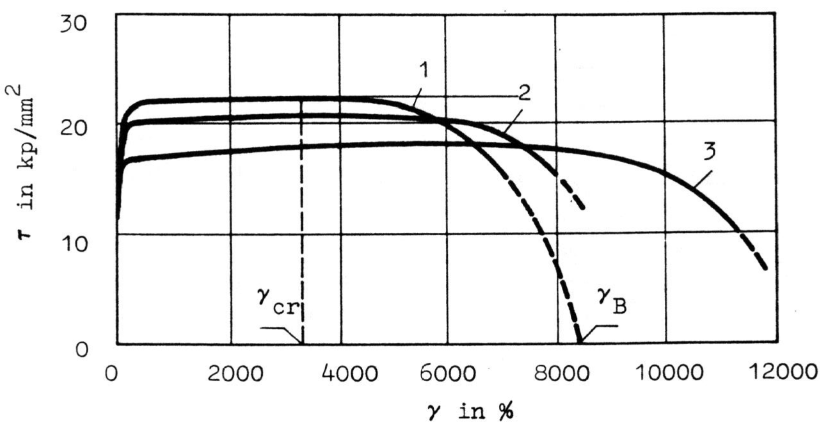
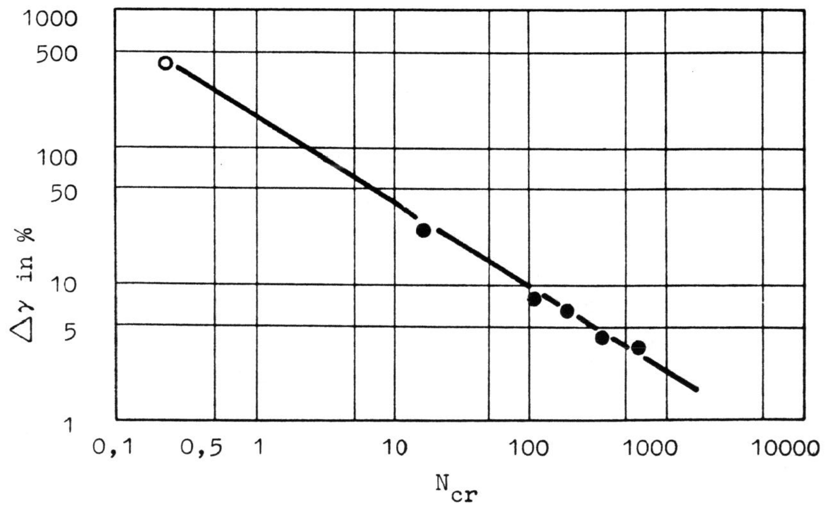
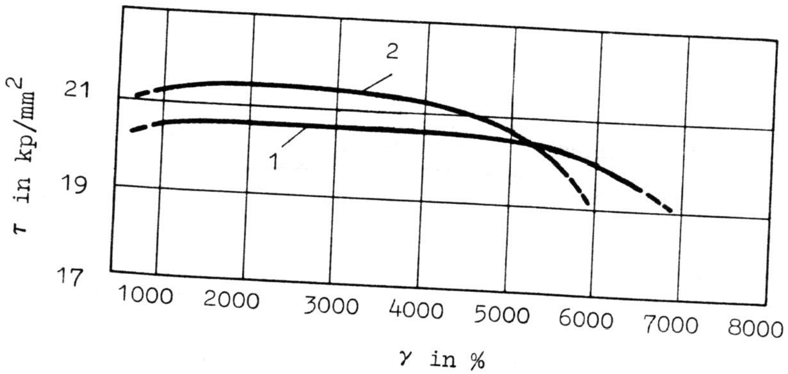


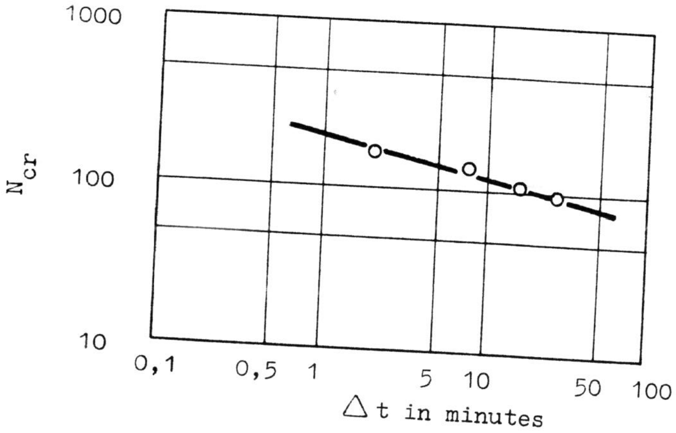
1) Peak stress τ vs. Integrated strain γ of Armco iron for three strain amplitudes $\Delta\gamma$ (1:$\Delta\gamma$ = ± 8.28%; 2:$\Delta\gamma$ = ± 6.25%; 3:$\Delta\gamma$ = ± 3.11%)



2) Critical number of cycles N_{cr} vs. strain amplitude $\Delta\gamma$
○: unidirectional shear
●: alternating shear



3) Peak stress τ vs. integrated strain γ for dwelling times 0 (curve 1) and 27 minutes (curve 2).



4) Critical number of cycles N_{cr} vs. dwelling time Δt.

Corrosion Fatigue Crack Initiation in Cu and Cu 7. 8% Al

D. Duquette, P. Andresen and H. Masuda
Materials Division
Rensselaer Polytechnic Institute
Troy, New York 12181

Introduction

The origin of corrosion fatigue cracking has generally been associated with either pitting[1], various forms of stress assisted dissolution[2,3] or localized film rupture[4]. To date most studies of corrosion fatigue have concentrated on the examination of cycles to failure data or on post failure observation of cracking. An important exception to this type of examination has been the work of Forsyth in high strength aluminum alloys[5,6].

The present study outlines an investigation on the effect of slip character on corrosion fatigue crack initiation by comparing the surface characteristics of a wavy slip material (Cu) with a planar slip material (Cu-7.8% Al) exposed to cyclic stresses and aggressive environments.

Experimental Procedure

Axial tension-tension fatigue tests in air, and in 0.5N NaCl solution were conducted at 30 Hz on smooth surface cylindrical specimen of Cu and a Cu-7.9% Al alloy. Prior to testing, the materials were alternately cold worked and hydrogen annealed to produce a grain size of approximately 0.15 mm. This heat treatment resulted in a yield strength of 3.0 kg/mm^2 and a tensile strength of 24.4 kg/mm^2 for the Cu and a yield strength of 11.2 kg/mm^2 and a tensile strength of 44.1 kg/mm^2 for the alloy. Specimen surfaces were electropolished prior to testing.

Environmental tests were conducted utilizing a lucite chamber hermetically sealed to the machine pull rods. Pull rods and specimen grips were constructed of glass fiber reinforced epoxy resin to eliminate galvanic effects. Small anodic currents were applied to the Cu specimens in order to induce environmental cracking. Crack initiation was determined by replicating the surfaces of specimens during interrupted testing and observing the replicas in an optical microscope.

Results and Discussion

Fig. 1 shows the effect of aerated 0.5N NaCl on the fatigue life of Cu-7.9% Al and shows that, at high applied stresses fatigue life is relatively unaffected by environment. At lower applied stresses however, the endurance limit for a given number of cycles is significantly reduced, the endurance limit being reduced by approximately 25% at 10^6 cycles to failure and 30% at 10^7 cycles to failure. Similar experiments conducted in $0.5\underline{N}$ Na_2SO_4 showed no significant differences from fatigue tests conducted in laboratory air. Qualitatively the surfaces of specimens tested in $0.5\underline{N}$ NaCl appeared to be more severely corroded than specimens exposed to $0.5\underline{N}$ Na_2SO_4 solutions.

Fig. 2 compares fatigue crack initiation in air and in $0.5\underline{N}$ NaCl at 26 kg/mm^2 (N_f in air = $5\text{x}10^5$ cycles; N_f in NaCl = $2.3\text{x}10^5$ cycles). In air the fatigue crack shown initiated in a short grain boundary segment and propagated along the surface of the specimen by slip band cracking (Stage I). It should be noted however, that fatigue crack initiation in air is more often slip band initiated and surface propagation is entirely associated with the slip structure. In $0.5\underline{N}$ NaCl on the other hand, fatigue crack initiation is invariably grain boundary initiated and surface propagation occurs entirely by grain boundary cracking. Fig. 2 also indicates that slip bands become more prominent and widely spaced in $0.5\underline{N}$ NaCl solutions.

In contrast to the Cu-Al, Fig. 3 shows that the fatigue life of copper is unaffected by $0.5\underline{N}$ NaCl under freely corroding conditions, but that the application of small anodic currents to the copper surface results in a reduction of fatigue life with the reduction increasing at larger currents. Fig. 4 shows that, as in the case for the Cu-Al alloy freely exposed to $0.5\underline{N}$ NaCl, applied anodic currents result in a change in the crack initiation mode for pure copper from transgranular to intergranular. It was also noted that prior to crack initiation, anodic dissolution resulted in preferential grain boundary attack. The application of small anodic currents (50μ A/cm^2) appears to enhance the intensity of surface slip offsets while the application of larger anodic currents (600μ A/cm^2) resulted in a "leveling" of the surface with only selected slip offsets being observed.

The results presented in this study show that the crack initiation process is highly dependent on environmental conditions. For the Cu-Al alloy, crack initiation shifts from a largely transgranular mode of crack initiation and surface propagation to an entirely intergranular mode when the material is exposed to an aqueous sodium chloride environment. Copper, which exhibits a more homogeneous type of slip deformation is unaffected by the combined exposure to cyclic

stress and sodium chloride solution. Howe anodic currents results in a decrease in f crack initiation mode from an entirely tra surface propagation event to an entirely in surface growth mode. For both materials co grain boundary grooving and an increased cr These data are in marked contrast to previou low carbon steels where corrosion fatigue cr was shown to be initiated entirely in the transgranular mode[2]. Transgranular corrosion fatigue cracking has also been reported for other alloys while intergranular cracking has been reported only under severe stress corrosion cracking conditions for aluminum alloys[7].

The reduction in fatigue life associated with environmental attack of cyclically deformed Cu and Cu-Al may be attributed to a number of interconnected phenomena. For example, the observation that relatively mild corrosive conditions enhance surface slip offset delineation suggests a preferential reaction of the corrosive environment with emerging slip steps resulting in the broadening of the normal intrusion and extrusion effect. Under more severe conditions the slip offsets are in fact obliterated (Fig. 4c). According to this model the stress intensity associated with intrusion-extrusion pairs in the metal surface is thus relieved. At the same time, preferential attack of grain boundaries results in stress concentrations at these locations and subsequent crack initiation. A similar model for intergranular crack initiation in pure materials subjected to large cyclic stresses has also been proposed[8].

The differences in behavior between these materials and previously reported materials apparently lies in their high purity and the mechanism for fatigue crack initiated in air. In steels and other engineering alloys exposed to corrosive environments general attack rather than grain boundary attack is normally observed. Thus the expected site for crack initiation remains at emerging slip steps or intrusion-extrusion pairs.

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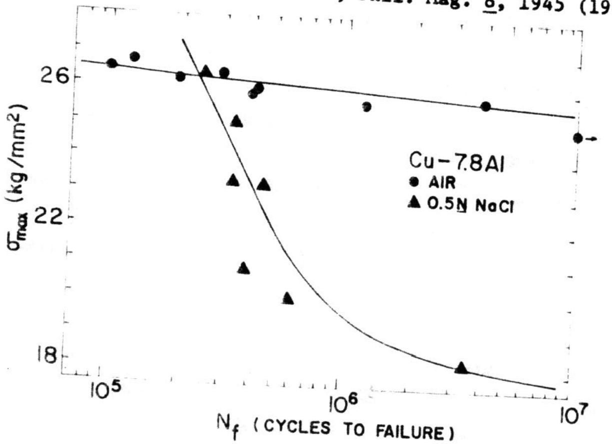

Fig. 1
Effect of maximum stress on fatigue life of Cu-7.8 Al in air and in 0.5N NaCl

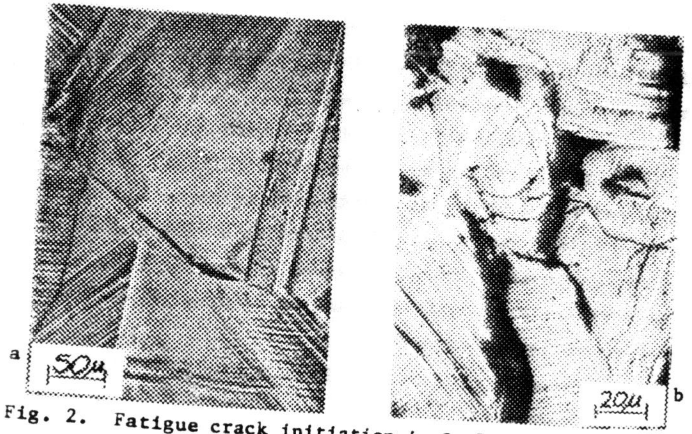

Fig. 2. Fatigue crack initiation in Cu-7.8 Al in

a) air at 66% of life, and

b) 0.5N NaCl at 15% of life.

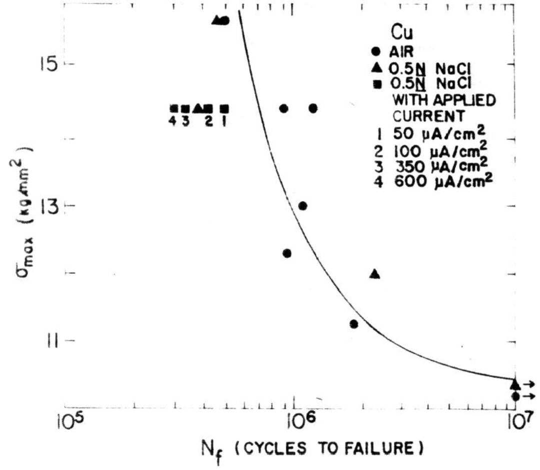

Fig. 3

Effect of maximum stress on fatigue life of copper in air and under applied anodic currents in 0.5N NaCl.

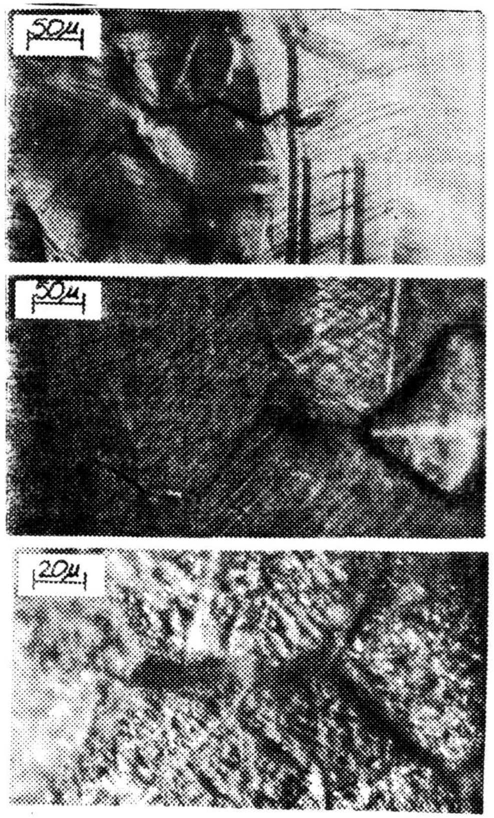

Fig. 4

Fatigue crack initiation in copper in

a) air at 27% of life,

b) in 0.5N NaCl with applied anodic current of 50µ A/cm² at 77% of life, and

c) in 0.5N NaCl with applied anodic current at 47% of life.