MICROSTRUCTURAL EFFECTS ON CREEP AND FATIGUE

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ABSTRACT

Both creep and fatigue are highly complex phenomena involving many factors. The roles of microstructure are even more complex and often conflicting. This paper discusses the roles of microstructure in low temperature phenomena first, that is, fatigue at room temperature and below, then elevated temperature creep, and finally fatigue again but at elevated temperatures. At low temperatures fatigue crack initiation, microcrack propagation, the stress intensity range threshold for macrocrack propagation including closure stresses, and mid-range macrocrack propagation are discussed in sequence. The discussion of creep and high temperature fatigue emphasizes the influence of microstructure on damage accumulation at grain boundaries and microstructural instability during service.

FATIGUE AT ROOM TEMPERATURE AND BELOW

The fatigue failure process may be divided into several stages: 1) initiation of one or more microcracks, 2) propagation or coalescence of microcracks to form one or more macrocracks and 3) propagation of one or more macrocracks to final failure. A very early concept in the science and art of fatigue of metals is that the microstructure became "tired" from cyclic loading and the metal failed by a cyclic "aging" process (1). The accumulation of microstructural damage to cause initiation of a fatigue crack is highly localized and the "aging" must be highly localized. Theories of fatigue failure based on general strain or energy accumulation throughout the specimen cannot be correct.

Fatigue cracks usually initiate near or at singularities which lie on or just below the surface. Such singularities may be sharp
changes in cross-section, pits, inclusions, embrittled grain boundaries, etc. However, even when the surfaces of metals are highly polished, the metal is flaw free and no stress concentrators are present, a fatigue crack may still form. Localized regions of plastic deformation develop under continued cycling until they become sufficiently severe to initiate one or more fatigue cracks. Such cracks, when initiated, are very small. The initially observed fatigue cracks have become smaller and smaller as the resolving power of the microscopes used to study them have become smaller. With TEM of surface replicas taken with a load applied to open the crack, cracks as small as 0.1 \( \mu m \) have been observed along slip bands (2). Observation of such small fatigue cracks is quite general (3). Figure 1 (4) is an example showing two tiny cracks adjacent to a slip band in OFHC copper cycled 100 times at a plastic strain amplitude of 5 \( \times 10^{-4} \).

How do such small fatigue cracks form? One suggestion is that they form by accumulation of vacancies generated by the to and fro dislocation motion (5). These might be imagined to diffuse along dislocation pipes until they reach the surface forming a pit where the dislocation emerges. Since tiny fatigue cracks also form at 4.2°K (6), such a mechanism or any thermally activated process cannot have general applicability.

Some success has been achieved with a model (4) based on vacancy type dislocation dipole accumulation in the slip band. A crack is proposed to initiate when the accumulated displacement in the dipole pile-up equals the critical strain for fracture of a perfect metal. The model predicts that the crack should initiate near the slip band matrix boundary, as observed with OFHC copper. The resulting equation for the cycles to initiation is

\[
N_i \approx \frac{0.4}{\pi(1-\nu)} \frac{\Delta r^2 k}{(\Delta \tau - 2k)(\Delta \tau)^2} \frac{b}{a} \frac{(1-R)^2}{f^*} \frac{m}{\Delta \tau^2}
\]

where \( \nu = \) Poisson's ratio, \( k = \) resistance stress to dislocation motion, \( \Delta \tau = \) stress range, \( b = \) Burger's vector, \( a = \) half slip band length, \( R = \) stress ratio, \( m = \) orientation factor, e.g., Taylor factor, and \( f^* = \) damage accumulation efficiency factor, i.e., average ratio of dislocations stored per cycle to dislocations stored during first half cycle.

In this equation \( k, a, \) and \( f^* \) depend on microstructure. As the yield strength or \( k, \) which is a function of microstructure, increases, an increase in \( N_i \) is predicted as generally observed. The grain size influences \( N_i \) not only through \( k \) but also through \( a \) since \( a \) is expected to increase with grain size. Lin et al. (4) estimated \( f^* \) from the stored energy and also from the back stress obtained from hysteresis loops. If the dislocation motion is