THE EFFECT OF A PRIOR CYCLIC STRENGTHENING ON THE
SUBSEQUENT FATIGUE BEHAVIOUR OF POLYCRYSTALLINE
COPPER
J. Mendez, P. Violan, M. Ghammouri

The influence of a prior cyclic strengthening on the
subsequent fatigue properties of polycrystalline copper
was investigated at room temperature. To this end
specimens were firstly precycled in vacuum under
different cyclic plastic amplitudes and then
electropolished to eliminate surface damage. The fatigue
behaviour and the surface damage characteristics of
these precycled specimens were established in air under
a constant plastic strain amplitude of 6x10^-4; the results
were compared to that observed on annealed copper
directly cycled at this amplitude.

INTRODUCTION

The study of fatigue damage accumulation in low-high or high-low step
tests is frequently approached by phenomenological and global methods
without separating the role of the different mechanical and physical
factors determining cyclic damage.

In particular, very few work has been devoted to understand the role
of the internal dislocation substructures previously formed by cyclic
deformation on the subsequent fatigue behaviour of the material
especially when submitted to a lower cyclic plastic strain amplitude.
Indeed it can be expected that in this case, strain history effects or
changes in cyclic plastic strain distribution in the bulk could significantly
modify fatigue damage characteristics which are known to control the
fatigue resistance such as the microcrack density, the secondary
microcracks length or the major crack length evolution, with regard to that
established from constant amplitude fatigue tests. A better knowledge of

* Laboratoire de Mécanique et de Physique des Matériaux, URA n°863
CNRS, ENSMA-Poitiers (France).
these strain history induced modifications in fatigue damage would permit to improve life predictions models for multi level testing conditions.

With this aim fatigue experiments have been conducted on polycrystalline copper at room temperature using a high vacuum environment for precycling the specimens; by this way they can be cycled up to large cumulative plastic strain levels with a limited surface microcracking damage. We report in this paper results obtained on smooth precycled specimens from which all surface damage such as microcracks or surface irregularities has been removed to only conserve the dislocation substructure formed in the bulk by precycling and which have been then submitted to a constant cyclic strain amplitude in air.

EXPERIMENTAL METHODS

Annealed cylindrical smooth specimens of an OFHC copper of commercial purity (99.99 %) with a mean grain size of 30 µm were cycled in tension-compression at room temperature in air or in a high vacuum (6x10^{-4} Pa) under plastic strain control. Three plastic strain amplitudes were selected: Δε_p/2 = 6x10^{-4}, 2x10^{-3}, 5x10^{-3}.

The vacuum environment was used to precycle the specimens under different cyclic plastic amplitudes and to different accumulated plastic strains. Indeed, tests in vacuum permit to reach large cumulative plastic strains associated with a low surface damage at the opposite of what is obtained in air (1). An electropolishing of these vacuum cycled specimens is then sufficient to remove all the surface damage and to prepare smooth specimens with different internal dislocation substructures.

These predeformed specimens were then cycled in air under a constant plastic strain amplitude of Δε_p/2 = 6x10^{-4} and their behaviour related to the stress-strain response, the fatigue life or the surface microcracking features were compared to that observed on the annealed specimens.

RESULTS

Fig. 1 shows the cyclic hardening curves (Δε/2-N), stress amplitude vs the number of cycles for annealed specimens cycled in vacuum under constant plastic strain amplitudes. Letters A to F in Fig. 1 indicate the different states of precycling reached for the six precycled specimens investigated in this work. It must be noted that specimens A, D and F have been cycled up to a number of cycles equivalent to the fatigue life obtained in air on the annealed material for each of the three cyclic
plastic strain amplitudes considered here: $6 \times 10^{-4}$, $2 \times 10^{-3}$ and $5 \times 10^{-3}$ respectively.

Data related to specimens A and F concerning either the precycling conditions or the subsequent fatigue life and stress response when cycled under $\Delta \varepsilon_p/2 = 6 \times 10^{-4}$ in air are reported in Table 1.

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Saturation stress $\Delta \sigma_p/2$ (MPa)</th>
<th>Fatigue life $N_p$, cycles</th>
</tr>
</thead>
<tbody>
<tr>
<td>Reference</td>
<td>115</td>
<td>$117,000$ – $125,000$</td>
</tr>
<tr>
<td>A - precycled at $6 \times 10^{-4}$ ($\text{tpcum}=288$)</td>
<td>119</td>
<td>$150,000$</td>
</tr>
<tr>
<td>B - precycled at $2 \times 10^{-3}$ ($\text{tpcum}=12$)</td>
<td>117</td>
<td>$125,000$</td>
</tr>
<tr>
<td>C - precycled at $2 \times 10^{-3}$ ($\text{tpcum}=24$)</td>
<td>120</td>
<td>$110,000$</td>
</tr>
<tr>
<td>D - precycled at $2 \times 10^{-3}$ ($\text{tpcum}=40$)</td>
<td>124</td>
<td>$144,000$</td>
</tr>
<tr>
<td>E - precycled at $5 \times 10^{-3}$ ($\text{tpcum}=16$)</td>
<td>125</td>
<td>$130,000$</td>
</tr>
<tr>
<td>F - precycled at $5 \times 10^{-3}$ ($\text{tpcum}=40$)</td>
<td>137</td>
<td>$127,000$</td>
</tr>
</tbody>
</table>

Table 1 – Cyclic stress amplitude at saturation and fatigue life for annealed and precycled copper specimens cycled at $\Delta \varepsilon_p/2 = 6 \times 10^{-4}$ in air.

In the case of specimen A, which has been precycled at $6 \times 10^{-4}$, cyclic hardening continuous to follow the same curve as in vacuum during 150,000 cycles that is for a number of cycles very close to the fatigue life of the reference annealed material.

Fig. 2 illustrates the cyclic behaviour of specimens B, C and D previously precycled under $\Delta \varepsilon_p/2 = 2 \times 10^{-3}$ to different accumulated plastic strain levels. For these specimens the cyclic behaviour under $\Delta \varepsilon_p/2 = 6 \times 10^{-4}$ is characterized by a progressive softening which starts after about 50 cycles to lead to a quasi-saturation state after 3000 cycles. However a strain memory effect is observed on the precycled specimens since the stress-strain relationship corresponding to the annealed material is never exactly recovered. As it can be noted in Fig. 2 for specimens B, C and D, this increase in the flow stress depends on the accumulated plastic strain level; nevertheless, for all the three specimens precycled at $2 \times 10^{-3}$ the memory effect remains minor.

However as it appears in Fig. 3, the strain history effects cannot be ignored for specimens prestrained at $5 \times 10^{-3}$ particularly for those cycled up to a high number of cycles; indeed, for specimen F the flow stress at
the imposed cyclic plastic strain amplitude of $6 \times 10^{-4}$ is increased by a factor 1.3 with respect to the annealed material.

This behaviour is associated with the dislocation substructures formed during precracking in the specimen F, which are composed of equiaxed cells in the proportion of 80 per cent. This structure, which is observed to remain unchanged during cycling in air at $6 \times 10^{-4}$, is therefore significantly different from the substructures directly formed at $6 \times 10^{-4}$ in the annealed material and which are composed of veins, walls and channels and in a short majority of elongated cells (2).

On the other hand, whatever the stress amplitude reached at saturation, all the precycled specimens exhibit the same fatigue resistance as the reference annealed one (see Table 1).

Interrupted tests conducted on precycled specimens show that no significant differences appear on the cyclic prestrained samples concerning the major crack length evolution (2). However an increase in the surface microcrack density can be observed at failure on the precycled specimens. This is illustrated in Fig. 4 by histograms giving the distribution in number and surface length for the population of microcracks observed at failure on the annealed specimen and on the specimen C previously cycled under $\Delta \varepsilon_p/2 = 2 \times 10^{-3}$.

Although the number of cycles to failure of these specimens are very similar, 117 000 cycles and 110 000 cycles respectively, the microcrack density for the specimen C is twice higher (67 microcracks/mm$^2$ instead of 31 microcracks/mm$^2$). It can be noted that this increase in the number of surface microcracks concerns especially the very small cracks with a length lower than 100 $\mu$m.

CONCLUSIONS

It has been shown using electropolished specimens previously precycled in vacuum that strain history effects on the flow stress levels can be observed in copper for high-low fatigue step tests which are associated with the dislocation substructures formed under high strain amplitudes. These effects are accompanied by an increase in the number of microcracks initiated during the second part of the step test. However modifications induced in the internal microstructures have no influence on the fatigue life of the predeformed specimens when they are cycled at a lower strain amplitude after repolishing. These results suggest that the internal dislocation substructure and the associated flow stress level have a weak influence on the rate of initiation and growth of the main crack which appears to be controlled in copper by the cyclic plastic strain amplitude (3).
REFERENCES


Figure 1: Cyclic behaviour in vacuum of annealed copper and conditions of precycling.
Figure 2: Cyclic behaviour at $6\times10^{-4}$ in air of specimens B, C and D precycled at $2\times10^{-3}$.

Figure 3: Cyclic behaviour at $6\times10^{-4}$ in air of specimens E and F precycled at $5\times10^{-3}$.

Figure 4: Number of the surface microcracks per mm² at failure vs. their surface length. (a) annealed specimen, (b) specimen B precycled at $2\times10^{-3}$. 