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INVESTIGATION OF THE DISLOCATION STRUCTURE AND LONG-RANGE INTERNAL STRESSES DEVELOPING IN AN AUSTENITIC STEEL DURING TENSILE TEST AND LOW-CYCLE FATIGUE

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Abstract

18/10 austenitic stainless steel samples were tensile deformed to different strain values, and fatigued with different plastic strain amplitudes up to failure. In latter case special care was taken to unload the samples either from the tensile or the compressive stress maximum of the hysteresis loop, respectively. The specimens were cut perpendicular and parallel to the load axis, and these surfaces were investigated by high resolution X-ray line profile analysis. The line profiles reveal characteristically asymmetric line broadening as compared to the undeformed initial state. From the line broadening and the asymmetry the dislocation density and the long-range internal stresses prevailing in the cell walls and in the cell interiors have been evaluated. The long-range internal stresses were interpreted on the basis of the composite model of the dislocation cell structure. The results can be used for the different residual life prediction methods.

Keywords: plastic deformation, low-cycle fatigue, X-ray line profile, line broadening.

Introduction

During plastic deformation a heterogeneous dislocation structure arises in most of the materials. According to TEM investigations, dislocation cell structure develops in tensile deformed and low-cycle fatigued austenitic stainless steel [1,2]. In a recent work the long range internal stresses developing during low-cycle fatigue in pure copper polycrystalline samples were determined by evaluating the characteristically asymmetric X-ray diffraction line profiles [3]. It was shown that the directions and magnitude of the long-range internal stresses follow the course of the cyclic stress-strain hysteresis loop.

In the present paper we have studied the evolution of the dislocation density and the magnitude of the long-range internal stresses as a function of the plastic strain viz. plastic strain amplitude in a commercial polycrystalline stainless steel.

The Composite Model of Dislocation Cell Structures

In a dislocation cell structure the dislocation density in the cell walls is much higher than that in the cell interiors, therefore it can be assumed that the cell walls are harder compared to the cell interiors. Such a material can be treated as a composite having the hard cell walls and the soft cell interiors as components. *Fig. 1a* and *1b* show the ideal stress-strain curves of the two components, and that of the composite, respectively. This latter figure can be divided into four parts. During the first part, only elastic deformations happen. In the second part, after reaching the yield point of the softer component, i.e. the yield point of the cell interior (R_{ec}), this component yields plastically, while the harder component, i.e. the cell wall is still in elastic state.

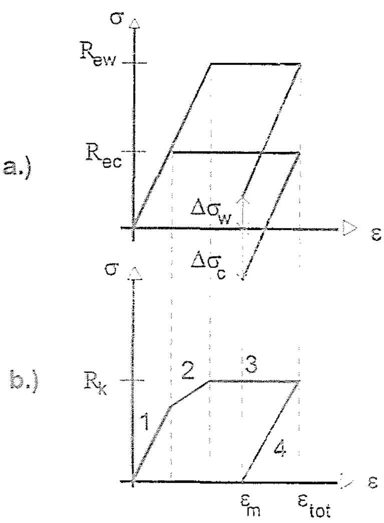

Fig. 1.

This part of the diagram is called microplastic range. After reaching the yield point of the harder component, i.e. the yield point of the cell wall (R_{ew}), the whole material yields plastically, as it can be seen in the third part of the figure. When the composite reaches the required strain value, ϵ_{tot}, the material is unloaded according to the Young moduli of the components, which are supposed to be identical. It can be seen from *Fig. 1a*, that a long-range compressive stress ($\Delta\sigma_c$) remains in the cell interior, and a long-range tensile stress ($\Delta\sigma_w$) remains in the cell wall region.

The residual stresses affect the X-ray line profiles. *Fig. 2* shows a line diagram of a dislocation cell. The arrow shows the previously applied stress, and the two inserts show the tetragonal lattice distortion in the cell wall and the cell interior, respectively. If a Bragg reflection of X-rays is measured from a surface perpendicular to the formerly applied stress (this situation is called axial case), the sub-profile belonging to the cell wall is shifted to the lower angle side, because from this view the lattice parameter of the cell wall is increased.

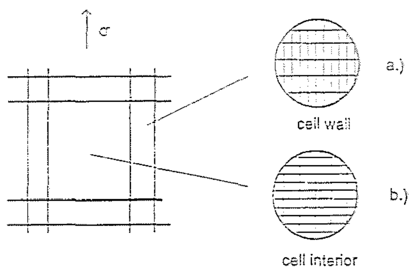

Fig. 2.

Simultaneously, the sub-profile of the cell interior is shifted to the higher angle side, since the lattice parameter from this view is decreased. The resulting X-ray line profile of the composite material in axial case shows a characteristic asymmetry, since the volume fraction of the cell interior is higher than that of the cell wall (*Fig. 3a*).

The asymmetry of a Bragg reflection obtained from a surface parallel to the previously applied stress (side case) will be reversed (*Fig. 3b*).

For more information about the composite model of dislocation cell structures the reader is referred to [4, 5].

Experimental

In the present work the high resolution X-ray diffraction measurements were carried out on individual grains of polycrystalline samples. In order to have reasonable scattered intensities the grains of the specimens were grown to about 200–300 μm by recrystallization heat treatment, cf. [6]. Specimens of 8 mm gauge length and 7 mm diameter were tensile deformed at different strain values in an Instron testing equipment, and

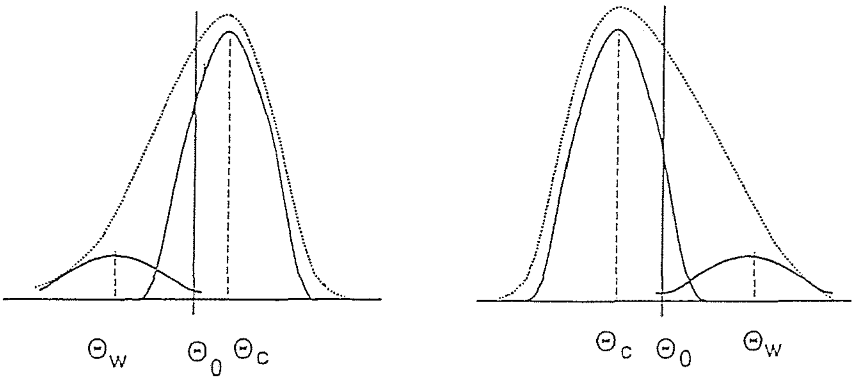

Fig. 3.

cycled in an MTS hydraulic testing machine up to failure. Failure was defined by 10% load drop. In order to have well defined stress states, as far as the residual long-range internal stresses were concerned, the samples were carefully unloaded from the compressive stress maximum of the cyclic hysteresis loop.

The X-ray diffraction experiments were carried out on a high-resolution special double-crystal diffractometer with negligible instrumental line broadening, cf. [7, 8].

The dislocation densities were evaluated by a straightforward Fourier method, cf.[9, 10]. The long-range internal stresses prevailing in the cell wall and in the cell interior materials, induced by plastic deformation, were evaluated on the basis of the composite model of plastic deformation of materials containing heterogeneous dislocation distributions, described above.

Results and Discussion

Fig. 4 shows typical X-ray line profiles of the (002) Bragg reflections obtained on tensile deformed sample surfaces perpendicular to the tensile axis (axial case). The profiles corresponding to the deformed states become broader and show a slight but well defined asymmetry. The tails decay more slowly on the lower angle side than on the higher angle side. Note that these specimens were unloaded from different tensile stress states.

The characteristic asymmetry, which means that in the axial and side cases it has to change sense, is shown in *Fig. 5* for a specimen fatigued by 6505 cycles up to failure at a plastic strain amplitude ε_{pl}=1.0%. The reversal of the asymmetry in the axial and side cases is clearly visible.

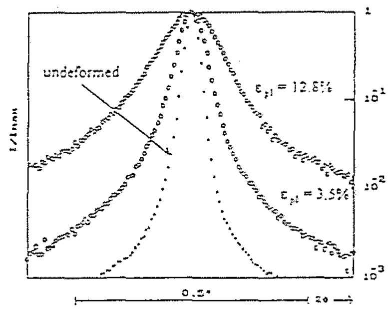

Fig. 4.

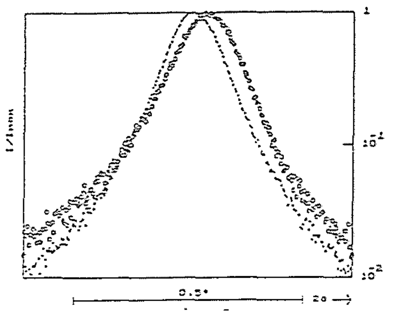

Fig. 5.

The dislocation density, ρ, as a function of the tensile stress for monotonic viz. stress amplitudes for cyclic loading is shown in *Fig. 6*. It is important to mention that the yield stress of the material is subtracted from the applied stresses viz. stress amplitudes. It can be seen that the well-known Nabarro function can be well fitted to the results:

$$\tau - \tau_\epsilon = \alpha G b \sqrt{\rho}.$$

where $\tau - \tau_\epsilon = (\sigma - R_\epsilon)/M$, R_ϵ is the yield stress and $M = 3.06$ is the actual Schmidt factor for the [001] type grains measured in the present case.

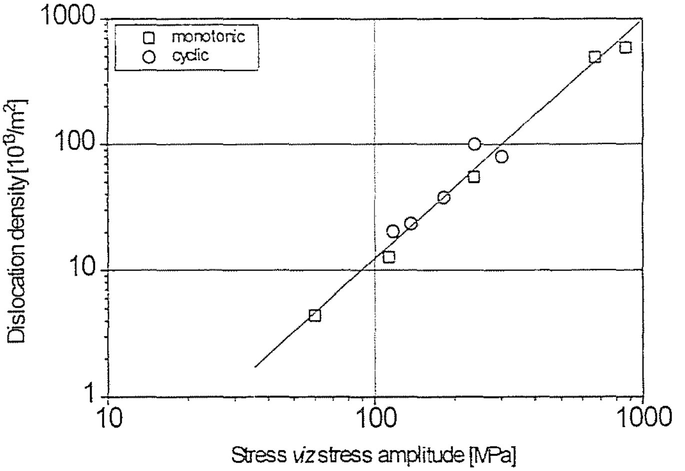

Fig. 6.

The figure shows that the $\rho - (\sigma - R_e)$ data follow the same linear behaviour for tensile deformation and fatigue in a double logarithmic plot. The straight line has a slope of 1/2, and α has been obtained to be 0.27, in good agreement with data for fcc metals, cf. [11,12].

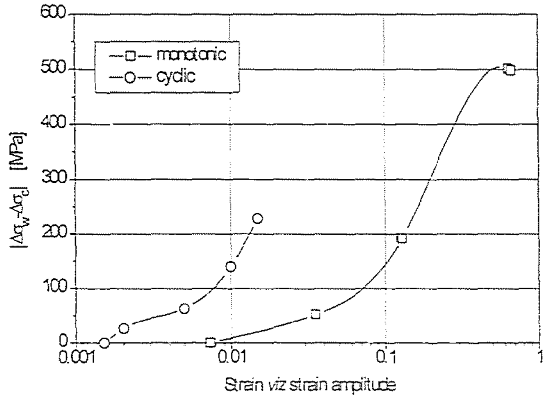

Fig. 7.

The magnitudes of the residual long-range internal stresses, $|\Delta\sigma_w - \Delta\sigma_c|$, as a function of plastic strain viz. plastic strain amplitude are shown in *Fig. 7*, while *Fig. 8* shows these quantities as a function of the applied stress viz. stress amplitude. It can be seen that the values of $|\Delta\sigma_w - \Delta\sigma_c|$ at the same plastic strain viz. strain amplitude are much higher for cyclic deformation than for monotonic loading.

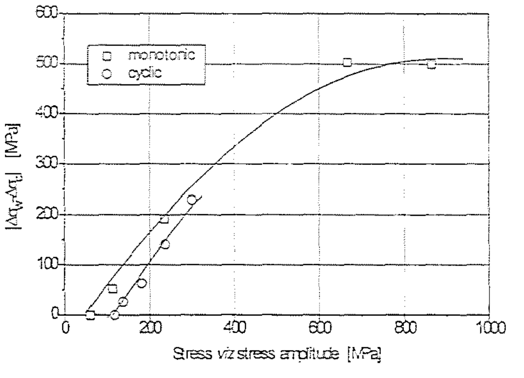

Fig. 8.

This behaviour is due to the fact that the values corresponding to cyclic deformation are attained after cycling the material into saturation where the dislocation density is relatively high. This dislocation arrangement with high density is polarized by a relatively small value of plastic strain, which in case is the plastic strain amplitude. The quantity $|\Delta\sigma_w - \Delta\sigma_c|$ is a measure for the polarization of the dislocation structure. The large difference in this value at the same plastic strain viz. strain amplitude is indicating that in the case of cyclic deformation high dislocation densities are polarized by small plastic strain amplitude values, whereas in the case of monotonic loading small dislocation densities are polarized by somewhat larger values of plastic strain amplitudes.

The residual stresses, $|\Delta\sigma_w - \Delta\sigma_c|$ are plotted versus stress viz. stress amplitude for monotonic and cyclic deformation, respectively, in *Fig. 8*. The figure shows that at the same values of stresses the residual stresses corresponding to cyclic loading are smaller than that corresponding to monotonic loading. This indicates that it is more difficult to polarize the dislocation structure obtained by fatigue than the one obtained after monotonic loading. The dislocation structure produced by fatigue seems to be harder at the same value of the dislocation density than the one obtained by monotonic loading. This behaviour is in accordance with the fact that the fatigued samples fracture at considerably smaller values of dislocation densities than the tensile deformed specimen. The dislocation structure produced during fatigue is more brittle than that obtained by tension, when measured according to the density of dislocations.

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