Evolution of Dislocation Structures and Deformation Behavior of Iron at Different Temperatures:
Part I. Strain Hardening Curves and Cellular Structure

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The deformation behavior of iron has been investigated at different temperatures by means of tension tests. There exist two temperature ranges for deformation. In the low-temperature range \((T < 293 \text{ K})\), the flow stress \(\sigma\), the work-hardening rate \(\Theta\) at \(\varepsilon = 0.06\), and the yield stress \(\sigma_y\) decrease with increasing temperature, but in the higher temperature range \((T \geq 293 \text{ K})\), \(\sigma\) and \(\Theta\) at the same strain increase while \(\sigma_y\) decreases more slowly. The change of dislocation density, with temperature, at \(\varepsilon = 0.06\) exhibits the same tendency as that of the flow stress. The strain-hardening rates decrease almost linearly with increasing stress up to necking in the low-temperature range, except the initial strain range. At the higher temperature range, the hardening rates decrease linearly with stress only at the early stage of deformation, but above certain strains, the decreases become more gradual; that is, the \(\Theta-\sigma\) curves deviate from the linear region. The evolution of dislocation structure has also been observed by transmission electron microscopy (TEM). The results show that a substructural transition takes place in the nonlinear range of \(\Theta-\sigma\) curves. In the linear decreasing region of strain-hardening curves, the deformation is controlled by the uniformly distributed dislocations or cell multiplication prevails. However, in the nonlinear region of \(\Theta-\sigma\) curves, cell multiplication seems to be balanced by cell annihilation.

I. INTRODUCTION

In order to describe the deformation behavior of crystalline materials, it is necessary to have some knowledge of their dislocation structure evolution, which depends on many variables, such as plastic strain rate, deformation temperature, etc. Recently, Kocks and Mecking\(^{[2]}\) pursued a phenomenological approach to macroscopic plasticity of metals that appears very useful in describing the work-hardening behavior at the early stages of deformation. This approach is based on the assumption that the kinetics of plastic flow are determined by a single structure parameter (average dislocation density) representing the current structure. The dependence of the flow stress \(\sigma\) on the plastic strain rate \(\dot{\varepsilon}\) and the absolute temperature \(T\) at a given structure of the material is given by the kinetic equation

\[
\sigma = \sigma(\rho, \dot{\varepsilon}, T) \quad [1]
\]

For a complete description of plastic behavior, the kinetic equation is complemented with an evolution equation which describes the variation of the structure parameter with strain \(\varepsilon\) at given strain rate and temperature:

\[
d\rho/d\varepsilon = f(\rho, \dot{\varepsilon}, T) \quad [2]
\]

According to the Kocks and Mecking approach, the structure parameter evolves toward a saturation value as the deformation progresses, and the flow stress tends to a saturation or steady-state value \(\sigma_s\). At constant strain rate \(\dot{\varepsilon}\) and deformation temperature, the strain-hardening rate \(\Theta\), \(d\sigma/d\varepsilon\), is found to decrease linearly with flow stress; this behavior can be simply expressed as

\[
\Theta = \Theta_0 (1 - \sigma/\sigma_s) \quad [3]
\]

Though the main characteristics of the large-strain behavior are doubtless suitably described by the conclusions of Kocks and Mecking, actual flow curves could be more complex at large strains. For most of the published large-strain flow curves,\(^{[3]}\) the strain-hardening rate does not lead directly to a saturation stress \(\sigma_s\). Instead, the decrease in the strain-hardening rate slows down rather abruptly at the medium stages of deformation in different modes. The reason for this transition was discussed in terms of the TEM observation with aluminum,\(^{[4]}\) and it was found that it is correlated with the substructure transition from a stage of cell multiplication to another in which the cell multiplication rate seems equal to the cell annihilation rate.

It is obvious that the deformation structure in body-centered cubic (bcc) metals at low temperature differs from that in face-centered cubic (fcc) metals,\(^{[5]}\) since the mobility of screws in bcc metals is much smaller than that of nonscrews due to the higher Peierls stress,\(^{[6]}\) which decreases rapidly with increasing temperature. However, the question of whether there exists a fundamental difference in work-hardening behavior between bcc and fcc metals, particularly for polycrystalline metals, under large strains cannot be completely answered.

In this article, the evolution of dislocation structure in iron with increasing strain at different temperatures was observed by TEM, and the reason for the transition on \(\Theta-\sigma\) curves was discussed in terms of the TEM observations made on the same samples.

II. EXPERIMENTAL PROCEDURE

The iron used in this investigation was provided in the form of about 16-mm-diameter rod in the hot-rolled state.
with the chemical composition in Table I. The tensile specimens that were machined from the rod had a 6-mm diameter and 30-mm gage length. The specimens were annealed at 500 °C for 20 minutes and water-quenched. The resulting average grain diameter was about 150 μm. Tensile deformation was performed on an Instron machine at a constant strain rate of 4.2 \times 10^{-4}/s. To investigate the deformation behavior at different temperatures, the annealed specimens were strained at 173, 213, 253, 293, 373, and 440 K, respectively. Furthermore, at a given temperature, the specimens were deformed to different strains and water-quenched if the deformation temperature was higher than room temperature. The dislocation structures in these specimens deformed to different strains were then observed with TEM.

Transmission electron micrographs were taken on transverse sections of the tensile specimens. The extraction of TEM thin foils was performed by first cutting disks. The disks were then ground on 600-grit silicon carbide to a minimum thickness of 0.2 mm and finally electrochemically thinned on a Struers Tenupol II using an electrolyte of 19 parts CH₃COOH and 1 part HClO₄ at 285 K. Electropolishing was performed at 70 V. The foils were examined in a JEOL-JEM-200A and a JEOL-JEM-2000EXII operating at 200 KV utilizing a double-tilt stage. Three foils were typically examined from each deformed specimen. Dislocation density was determined using Ham's intercept method and two-beam conditions were used to image the dislocations. The cell size was described with cell diameter measured with a Mop (semiautomatic structure analysis system).

III. RESULTS

The true stress-strain (σ-ε) curves up to necking and the corresponding work-hardening (Θ-σ) curves for the iron at various temperatures and a strain rate of 4.2 \times 10^{-4}/s are shown in Figures 1 and 2, respectively. The influence of temperature on the true stress σ and work-hardening rate Θ at ε = 0.06 and yield stress σ₀ are plotted in Figure 3. It can be seen that with increasing temperature, σ and Θ at ε = 0.06 and σ₀ decrease almost linearly in the low-temperature range (T < 293 K), while in the higher temperature range (T ≥ 293 K), σ and Θ at ε = 0.06 increase and σ₀ decreases more slowly (Figures 1 and 3). The stress-strain curves at 373 and 440 K exhibit serrated flow, and the sharp yield point disappears (Figure 1). This is in the so-called blue-brittle region. The work-hardening curves (Θ-σ) in the serrated flow region at these two temperatures are calculated with

![Fig. 1—True stress-strain (σ-ε) curves at various temperatures.](image1)

![Fig. 2—Strain-hardening rate Θ vs flow stress σ at various temperatures.](image2)

![Fig. 3—Flow stress σ and strain-hardening rate Θ at ε = 0.06, square root of dislocation density \( \sqrt{\rho} \) at the same strain, and yield stress σ₀ vs temperature.](image3)