Hot working of high purity Fe-C alloys in the α-range

F Montheillet and J Le Coze
Ecole des Mines (SMS), CNRS UMR 5146, 158 cours Fauriel, 42023 Saint-Etienne Cedex 2, France
montheil@emse.fr

Abstract. The influence of carbon in solid solution on the stress-strain curves of α-iron was investigated using model alloys prepared from high purity iron. Uniaxial compression tests were carried out within the ferritic domain at temperatures between 700 and 880 °C. Oscillating stress-strain curves observed at high temperatures and low strain rates indicate that discontinuous dynamic recrystallization takes place. The macroscopic strain rate sensitivities m and apparent activation energies Q associated with the flow stress are not significantly modified by carbon additions. By contrast, the "mesoscopic" parameters h and r associated with strain hardening and dynamic recovery, respectively, are strongly dependent on the carbon content. Finally, an estimation of the grain boundary mobilities during dynamic recrystallization was carried out from the above rheological data.

1. Introduction
Commercial alloys, especially steels, usually contain a number of addition elements and impurities, which makes it difficult to analyze their respective effects on hot working properties. The aim of the present study was to investigate the influence of carbon in solid solution on the stress-strain curves of iron in the ferritic range.

Pure iron and a series of three Fe-C alloys containing 50, 100, and 200 wt ppm carbon were therefore prepared from electrolytic iron purified by induction melting under high purity hydrogen and argon, in a water cooled silver crucible [1, 2]. The ingots were then processed by hot forging and cold swaging, and cylindrical specimens were machined. Uniaxial compression tests were carried out within the range 700-880 °C at strain rates from \(10^{-3}\) to \(1\) s\(^{-1}\), up to strains of ~0.7. Stress-strain curves are first analyzed through the usual strain rate sensitivity m and apparent activation energy Q in Section 2. In a second step (Section 3), it is shown that parameters h and r derived from the curves using the Yoshie-Laasrariou-Jonas (YLJ) equation are much more sensitive to the carbon content. Finally, in Section 4, a model of discontinuous dynamic recrystallization (DDRX) recently published is used for estimating the associated grain boundary mobilities.

2. Analysis of the stress-strain curves
Fig. 1 shows four sets of flow curves for pure Fe, Fe-50 wt ppm C, Fe-100 wt ppm C, and Fe-200 wt ppm C deformed at 700 °C and various strain rates. A transition occurs from multiple peak behaviour at the lowest strain rate to single peak curves at larger strain rates, which indicates that discontinuous dynamic recrystallization takes place in these materials. This is in agreement with former observations by Glover and Sellars [3], but contrasts with the case of industrial ferritic alloys.
The yield stress $\sigma_y$ increases with the amount of carbon, especially at 0.1 and 1 s$^{-1}$. In the same way, the peak stress $\sigma_M$, and the steady state stress (whenever it is reached) increase monotonically, although the peak strain $\varepsilon_M$ seems to reach a maximum for 100 ppm C, which is likely to reflect lower DDRX kinetics.

Figure 1. Uniaxial compression stress-strain curves for (a) pure iron, (b) Fe-50 ppm C, (c) Fe-100 ppm C, and (d) Fe-200 ppm C at 700 °C and various strain rates. Broken lines indicate the YLJ fits of the curves.

The classical macroscopic rheological parameters $\nu$ (strain rate sensitivity of the flow stress) and $Q$ (apparent activation energy) were determined from the above data. They both remain almost independent of the carbon content with average values $\nu = 0.164$ (700 °C) and $Q = 324$ kJ.mol$^{-1}$ (700 °C and 0.01 s$^{-1}$). Note that this value is quite larger than $Q = 280$ kJ.mol$^{-1}$ reported for self-diffusion in $\alpha$-Fe.

3. Mesoscopic rheological parameters $h$ and $r$

A more detailed description of the above stress-strain curves can be obtained when theoretical fits are applied. A first model was used for the combination of strain hardening and dynamic recovery, viz. the Yoshie-Laasraoui-Jonas (YLJ) equation:

$$\sigma_0 = \left[ \sigma_0^2 - (\sigma_M^2 - \sigma_0^2) \exp\left(-\nu \varepsilon \right) \right]^{1/2}$$  \hspace{1cm} (1)
where \( \sigma_\infty = \alpha \mu b \sqrt{h/r} \) is a virtual steady state flow stress that would occur in the absence of recrystallization, and \( \sigma_e \) the yield stress (\( \mu \) is the elastic shear modulus, \( b \) the Burgers vector, and \( \alpha = 1 \)). The fitting procedure was extended over the strain range \([\varepsilon_e - \varepsilon_L]\), where \( \varepsilon_e \) is the elastic strain, and \( \varepsilon_L \) a given fraction of \( \varepsilon_M \), that is expected to correspond with the onset of DDRX. Here a classical value \( \varepsilon_L = (5/6) \varepsilon_M \) was chosen. The shapes of the curves depend basically on the two mesoscale parameters \( h \) and \( r \), associated with strain hardening and dynamic recovery, respectively. The resulting fits are shown in Fig. 1 (broken lines). To a first approximation, for each of the investigated grades, the strain rate and temperature dependence of \( h \) and \( r \) can be written in the form:

\[
h = h_0 \left( \frac{\varepsilon}{\varepsilon_0} \right)^{m_h} \exp \left( \frac{m_h Q_h}{RT} \right)
\]

(2)

and

\[
r = r_0 \left( \frac{\varepsilon}{\varepsilon_0} \right)^{-m_r} \exp \left( -\frac{m_r Q_r}{RT} \right)
\]

(3)

(where \( \dot{\varepsilon}_0 = 1 \text{s}^{-1} \)), which means that \( h \) becomes larger with increasing strain rate or decreasing temperature, while \( r \) varies in the opposite way.

As a general trend, the addition of carbon solutes in iron increases both \( h \) and \( r \). The strain rate and temperature sensitivities of \( h \) are not significantly modified, whereas \( m_r \) grows and \( Q_r \) strongly drops with increasing carbon content, as shown in Figs 2 a and b. This indicates that dynamic recovery is strongly temperature dependent in pure iron, but much less when dislocation movements are impeded by the presence of solutes.

4. Estimation of the grain boundary mobility \( M_T \)

It has been shown in a recently published paper [4] that the grain boundary mobility pertaining to dynamic recrystallization can be estimated quite easily from experimental data. For that purpose, the (Swift type) power law (PW) equation was used (instead of YLJ) to fit the experimental curves:

\[
\sigma_0 = K \left[ \varepsilon + \left( \sigma_e / K \right)^{1/n} \right]^n
\]

(4)

The latter is based on the mesoscopic differential equation:
\[
\frac{dp}{de} = \frac{H^{\nu+1}}{\rho^\nu}
\]  
(5)
in which \( p \) is the dislocation density, and \( H \) and \( \nu \) are two material parameters associated both with strain hardening and dynamic recovery. Furthermore, \( H \) and \( \nu \) are simple functions of \( K \) and \( n \).

The fitting procedure was conducted as above to determine these parameters from the experimental stress-strain curves. The grain boundary mobility is then given by:

\[
M_\tau = (2\nu + 3) \left( \frac{\nu + 1}{\nu + 2} \right)^{\nu+1} \frac{H^{\nu+1} \varepsilon}{(\sigma_s / \alpha \mu b)^{2(\nu+2)}}
\]  
(6)
where \( \sigma_s \) and \( D_s \) are the steady state flow stress and grain size, respectively, and \( \tau \) is the line energy of dislocations. Since \( D_s \) was not available, it was estimated from \( \sigma_s \) using the well-known Derby relationship \( \sigma_s = k_D / D_s^a \). Values of \( M_\tau \) predicted in this way are displayed in Fig. 3. They again suggest that DDRX kinetics reaches a minimum for 100 ppm C.

![Figure 3. Dependence of grain boundary mobility during DDRX on the carbon content in pure iron.](image)

5. Conclusions
- Specific effects of carbon solutes on the hot working behaviour of \( \alpha \)-iron were brought into evidence from uniaxial compression stress-strain curves, using high purity base model materials.
- The dependence on carbon content of the main rheological parameters associated with (discontinuous) dynamic recrystallization, \( \nu \), \( \varepsilon \), \( h \) (strain hardening), \( r \) (dynamic recovery), and \( M \) (grain boundary mobility), was determined. The latter parameter strongly decreases up to \( \approx 50 \) wt ppm C, and seems to exhibit a minimum for \( \approx 100 \) wt ppm C.
- Microstructural investigations (e.g., steady state grain size measurements) are currently being carried out to support the above observations by their physical counterparts.

References
[1] J. Le Coze, R. Tardy, A. Kobylanski and M. Biscondi. Proc. UHPM-94, Kitakyushu-City, Japan, p. 371 (1994).
[2] C. Desrayaud, S. Girard, J. Le Coze and F. Montheillet. Mater. Trans. 43, 135 (2002).
[3] G. Glover and C.M. Sellars. Met. Trans. 4, 765 (1973).
[4] F. Montheillet and J.J. Jonas. ASM Handbook, vol. 22, Modeling and Simulation: Processing of Metallic Materials, in press.