The Mechanical Behaviour and Microstructures of Interstitial Free Steel Strained in Monotonic or Cyclic Torsion at 1223 K

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(Received on November 4, 2005; accepted on February 6, 2006)

The work hardening behavior of metals under cold cyclic or multiaxial deformation involves lower hardening rates than those caused by monotonic strain to the same total magnitude. The hardening is lower as the strain per cycle decreases. Similar studies under hot working conditions, for a limited number of cycles, also lead to lower hardening, as well as to changes in the shape of the flow curves and to delays in the initiation of Dynamic Recrystallization (DRX) and in the post-processing static softening kinetics. These results point to a decrease in stored energy caused by cyclic straining. A higher number of hot deformation cycles leads to lower steady state stresses ($\sigma_{ls}$) than those corresponding to monotonic deformation ($\sigma_{ls}$). Besides, DRX can be suppressed and replaced by Dynamic Recovery (DRV), and the successive cycles have been associated with the repeated build-up and disintegration of dislocation structures. The present research discusses the hot deformation of an IF steel in the austenitic temperature range, for monotonic torsion or a large number of reversed torsions with various strain amplitudes ($\Delta$). It was confirmed that $\sigma_{ls}$ depends on the relative amplitude of the strain cycles ($\Delta\varepsilon/\varepsilon_p$, where $\varepsilon_p$ is the strain at the peak stress in the monotonic torsion of the material), and that lower values of $\Delta\varepsilon/\varepsilon_p$ eliminate DRX, which is replaced by DRV. The ferritic grain size after deformation into the cyclic steady state stress depends on the value of $\sigma_{ls}$, similarly to results in the literature for monotonic torsion of IF steels.

KEY WORDS: dynamic phenomena; recrystallization and recovery; interstitial free (IF) steels; hot cyclic straining; strain path change.

1. Introduction

Early investigations carried out by Coffin and Tavernelli1) for various metals showed that tension/compression cyclic plastic straining at room temperature led to a saturation stress after a certain number of cycles. The saturation stress decreased as the strain amplitude per cycle was lowered. The authors also observed that cyclic straining could remove some of the hardening associated with prior monotonic deformation, softening the metal towards the strength it would have achieved, for the equivalent amount of cyclic strain in the absence of prior monotonic deformation. Later, Armstrong, Hockett and Sherby2) compressed 1 100 aluminium cubes at room temperature successively in three orthogonal directions, with varying strain amplitudes in each series of compressions. Their results were qualitatively very similar to those described above for tension/compression. The occurrence of a saturation flow stress under multidirectional deformation was attributed to an enhancement of the prevailing recovery processes, in comparison with those in unidirectional deformation. Sub-grains and cell average sizes decreased with increasing unidirectional or multidirectional straining. The sub-grain plus cell concentration increased with strain up to an accumulated strain of 3.0 and then appeared to saturate above this strain for the multidirectional tests. No such saturation, however, was observed for the unidirectional tests. It can be concluded that room temperature, reversed cyclic or multidirectional straining of metals affects the stored energy in the material, in relation to the energy originating from monotonic deformation, for the same levels of total strain. The stored energy decreases as strain amplitude is lowered.

More recently, Davenport and Higginson3) conducted reversed straining experiments at high temperature and for a limited number of cycles. They pointed out that the effects of such straining on the stored energy, recrystallization and grain growth kinetics depend on the microstructure of the metal being deformed (particularly on its stacking fault energy), and on whether particles of precipitates are present. The authors also considered as other important factors the temperature and the strain rate history of deformation, as well as the actual strain path and its magnitude. Zhu and Sellars4–7) showed that the hot working of Al–2%Mg, first in tension to a strain of 0.15 and then in compression to a total strain of 0.30, led to a yield stress on reloading smaller than on loading. This had already been noticed and reported for cold working experiments.8)

Other authors9) investigating the effects of the strain path...
on the static restoration, reported that the time for 50% of the material to recrystallize, \( t_{50} \), for the reverse straining experiments was always larger than for the continuous tests, albeit showing some disagreement on how the value of \( t_{50} \) varied with the accumulated strain. Cowan and co-workers found, for instance, that \( t_{50} \) decreased with increasing accumulated strain regardless of the type of test being conducted (monotonic or reversed). Zhu and Sellars, on the other hand, reported just the opposite. Both groups of investigators, however, attributed these conflicting results to a reduction in dislocation density in the microbands or cell walls in reverse straining experiments as the cumulative strain increased, although unequivocal evidence of this could not be presented.

Giving support to the high temperature findings reported above, Bartolome and co-authors reported experiments in which one reversed torsion was applied to a Nb microalloyed steel, in order to investigate the effect of strain reversal on the mechanical response and on the dynamic (DRX) and static (SRX) recrystallization kinetics. The experiments were conducted in a temperature range allowing the occurrence of some strain induced precipitation, so that the effect of strain reversal on the kinetics of precipitation could also be studied. The flow curves were obtained from torsion tests conducted at 1323 K and the monotonic curve showed a peak strain, \( \epsilon_p \), of the order of 0.4. The samples were, initially, deformed to \( \epsilon = 0.25 \), (below the critical strain for the initiation of DRX, \( \epsilon_c \)), and then strain was reversed. The yield stress upon reversal was lower than that before reloading, and was followed by a yield plateau. Further deformation led to an increase in the stress up to a peak and, eventually, after sufficient deformation was given to the sample, a steady state stress, \( \sigma_{ss} \), was achieved after the full occurrence of DRX. The remarkable feature presented in their experiments was that the full reversal of straining produced a significant delay in the initiation of DRX. It is well known that this initiation is related to the stored strain energy and to the number and distribution of possible sites for nucleation. The authors suggested that the measured delay in the onset of DRX is associated with a decreased stored energy, caused by differences in the dislocation densities of material deformed monotonically or cyclically, since no large discrepancies in the number of nucleation sites was proven to be significant in their experiments.

The tests carried out by Bartolome and co-authors showed, additionally, that the kinetics of static recrystallization (SRX) were also affected by differences in the strain path undertaken by the material. They compared the kinetics of SRX for samples deformed according to three different strain paths: a single deformation of 0.25, a deformation of 0.25 applied in one direction followed by a reversed deformation of 0.1 and a deformation of 0.25 given in one direction followed by a full 0.25 deformation given in the opposite direction. The authors measured the times spent to achieve a volume fraction of 40% of recrystallization, \( t_{40} \). They found \( t_{40} = 60 \) s for the single 0.25 deformation sample, 160 s for the +0.25/−0.10 deformation sample and 32 s for the +0.25/−0.25 specimen. Assuming then \( t_{40} \) is proportional to \( \epsilon^{−2.6} \), it can be demonstrated that the expected value of \( t_{40} \) would be around 10 s for the sample deformed to a total absolute strain of 0.5. That is, achieving a total strain of 0.5 via forward and reversed torsion delayed recrystallization and increased \( t_{40} \) from the expected 10 s to 32 s, postponing therefore the kinetics of SRX. Conversely, the expected strain that would produce \( t_{40} \) around 32 s would be \( \epsilon \approx 0.32 \), considerably smaller than 0.5, demonstrating that the stored strain energy accumulated locally on reversed torsion tests is smaller than expected. The same effect was verified for the +0.25/−0.1 deformation experiment.

The results reported above deal either with a large number of cycles applied at room temperature or with a small number of deformations in simple tension/compression or reversed torsion tests. On the other hand, some results have been reported for a large number of hot cyclic torsions for copper at 673 K and a strain amplitude \( \Delta \varepsilon = 0.24 \), and at 773 K and \( \Delta \varepsilon = 0.025, 0.05, 0.10, 0.2 \) and 0.4, as well as for an Interstitial Free (IF) steel deformed at 973 K and \( \Delta \varepsilon = 0.015, 0.05, 0.10, 0.2 \) and 0.6 (in the ferritic temperature range) and at 1223 K and \( \Delta \varepsilon = 0.03, 0.09 \) (in the austenitic temperature range). The overall results for copper are very similar to those reported by Coffin and Tavner et al. for room temperature straining, already discussed in this introduction. In addition, cyclic torsion eliminates the DRX stress peak observed under monotonic torsion of copper, for strain amplitudes below a fraction of the peak strain for DRX. Under such circumstances, it was demonstrated that DRX is replaced by DRV, involving the disintegration and build-up of dislocation structures for successive torsion cycles. A saturation stress was reached after a certain number of cycles, and the level of this stress decreased as strain amplitude was lowered. A relationship of the saturation stress with the relative strain amplitude (strain amplitude as a fraction of the peak strain) was established. Optical microscopy was performed for copper deformed at 773 K, both monotonically and cyclically, with strain amplitudes of \( \Delta \varepsilon = 0.025 \) and 0.10, always to a total accumulated effective plastic strain of 2.5. Monotonic torsion led to DRX and grain refinement. Cyclic torsion with the lower strain amplitude (\( \Delta \varepsilon = 0.025 \)) caused no observable microstructural changes in comparison with the undeformed material, whereas the higher amplitude (\( \Delta \varepsilon = 0.10 \)) seems to have triggered an incipient DRX process, leading to coarse, blocky grains.

The results for the IF steel at 1223 K, \( \Delta \varepsilon = 0.03 \) and 0.09 (in the austenitic temperature range) are quite similar to those reported for copper, concerning both stress–strain curves and optical microscopy. For the case of cyclic torsion under the higher strain amplitude (\( \Delta \varepsilon = 0.09 \)), the evidence of incipient DRX stems from the presence of some small grains embedded in a coarse grained structure. The experiments with the IF steel deformed at 973 K and \( \Delta \varepsilon = 0.015, 0.05, 0.10, 0.2 \) and 0.6 (in the ferritic temperature range) led to conclusions which were also very similar to the above results, considering the observed stress–strain curves. An important difference, however, comes from the fact that the IF steel does not undergo DRX at 973 K. As a consequence, the cyclic torsion only affects the saturation stress, without a change of restoration mechanism from DRX to DRV. Optical microscopy was performed for samples deformed monotonically and cyclically with strain amplitudes of \( \Delta \varepsilon = 0.05, 0.2 \) and 0.6, up to a total strain of 2.5.
Monotonic deformation led to elongated grains and the formation of a banded structure. Cyclic torsion with the lowest strain amplitude (Δε = 0.05) reproduced the microstructure of the undeformed material; the increase in strain amplitude (to Δε = 0.2 and 0.6) led to structures of increasing complexity, tending to that produced by monotonic torsion. One can thus conclude, for IF steel deformed both in the ferritic and in austenitic temperature ranges, that cyclic reversed torsion causes an increased level of DRX and retards DRX, which may however develop in an incipient way as the strain amplitude per cycle increases.

The objective of this research is to present further experimental results and to model the effect of hot cyclic torsion on (a) the dynamic softening of a material undergoing DRX and (b) the resulting grain size. The experiments were performed on an IF steel deformed in the austenitic temperature range, due to its technological importance, since most of industrial rolling is carried out under this condition.

2. Experimental Techniques

2.1. Materials and Mechanical Testing

The composition of the IF steel employed in this research was 0.005C, 0.15Mn, 0.014Si, 0.013P, 0.007S, 0.085Ti, 0.011V, 0.063Al and 0.0049N, all numbers given in weight %. The amount of Ti in stoichiometric equilibrium was estimated from the expressions.

\[
%\text{Ti}_{\text{stoichiometric}} = \frac{14 \times \text{[N\%]} + 32 \times \text{[S\%]} + 12 \times \text{[C\%]}}{48}
\]

%\text{Ti}_{\text{excess}} = 0.085 - %\text{Ti}_{\text{stoichiometric}}

For the present alloy, the calculated value for the equilibrium concentration of Ti is 0.0473 % and the excess concentration of Ti is 0.0337 %, all C and N being, therefore, captured by the Ti available in the steel.

All mechanical tests were carried out by torsion in a servo-hydraulic computer controlled machine with a microprocessor controlled infrared furnace. In order to homogenize the microstructure, all specimens (cylinders 6.4 mm in diameter and 15.0 mm in length, with end grips) were initially annealed at 1 223 K (950°C) for 200 s and then cooled down to room temperature in the furnace. Temperature was measured by a thermocouple in contact with the surface of the gauge length of the specimens. A quartz tube involved the sample throughout the test, and a protective argon atmosphere was injected in the tube to avoid sample oxidation. The measured torque (\(\Gamma\)) and plastic twist angle (\(\theta\)) data were converted to effective stress, \(\sigma\), and plastic strain, \(\varepsilon\), following equations:

\[
\sigma = \frac{3.3\sqrt{3}\Gamma}{2\pi R^3}
\]

\[
\varepsilon = \frac{\theta R}{\sqrt{3} L}
\]

where \(L\) and \(R\) are the gauge length and radius of the specimen, respectively.

All reported strain results consider only the plastic twist component of the total applied twist angle (\(\theta\)). The elastic twist component was evaluated by drawing a parallel to the elastic initial portion of the loading through the values of applied torques of interest.

Some samples were cooled down to room temperature for microstructure observation, at specific levels of strain. Cooling was carried out by injection of a stream of water in the quartz tube immediately after straining. On cooling, transformation always led to a fully ferritic microstructure. A thermocouple attached to the specimen showed that all specimens were cooled from 1 223 K (950°C) to 473 K (200°C) in about 6 s.

2.2. Determination of the Ar3 and Ar1 Temperatures

In order to establish the testing temperature for continuous and reversed torsion tests, the values of Ar3 and Ar1 were measured by plotting the mean flow stress versus the inverse of the absolute temperature. In this way, samples were pre-heated at 1 423 K (1 200°C) for 600 s and then cooled down at a rate of 1 K/s during which 20 passes of 0.3 equivalent strain and 0.1 s\(^{-1}\) equivalent strain rate were applied. Deformation of the samples was carried out at time intervals of 60 s so that recovery and recrystallization could occur between passes, avoiding any effects of strain accumulation on the values of Ar3 and Ar1. The stress strain curves so obtained displayed no Tnr (Temperature of no recrystallization), as shown in Fig. 1, and the measured values of Ar3 and Ar1 were 1 210 K (937°C) and 1 153 K (880°C), respectively.

2.3. Torsion Testing

Two types of torsion experiments were conducted in the present paper: continuous (monotonic) torsion and interrupted reversed or cyclic torsion testing. Continuous torsion tests were performed under the strain rates of 0.01, 0.1 and 1.0 s\(^{-1}\), at the temperatures of 1 223, 1 273, 1 323 and 1 373 K (950, 1 000, 1 050 and 1 100°C). For reasons that will be explained below, an additional test was performed at 1 223 K and a strain rate of 0.5 s\(^{-1}\).

The present work intends to explore the influence of the strain path on the levels of the stress–strain curves when they are compared to the stresses measured from a continuously deformed sample, that is, with no interruptions or any shearing reversals. The ideal testing conditions require the
occurrence of full DRX without an oscillating steady state stress ($\sigma_{ss}$) and slow static softening during the time interruptions between shearing reversals. The latter is necessary because static softening could mask possible effects caused by the straining reversals. Therefore, the lowest testing temperature, 1223 K (950°C), was chosen to perform most tests. Examining the results displayed in Fig. 2(a), it can be seen that the stress–strain curves of tests carried out at strain rates above 0.01 s$^{-1}$ display no oscillation in the steady stress values. However, the curve measured at a strain rate of 0.1 s$^{-1}$ indicates that full DRX is achieved at almost the end of deformation, which is also not convenient for the intended experiments to be carried out here. Lowering the strain rate to 0.5 s$^{-1}$ decreased the minimum strain at which DRX was completed to an approximate value of 1.8. The transition between the peak stress ($\sigma_p$) and $\sigma_{ss}$ in this curve was, however, relatively smooth and spreading out over a strain interval of 1.1, from $\varepsilon$ ranging from 0.7 to 1.8. This, as a consequence, would lead to a high number of passes in the interrupted strain testing in order to achieve the accumulated strains comparable to the minimum necessary to reach $\sigma_{ss}$ in the continuous test. Therefore, the experiment performed at a strain rate of 0.1 s$^{-1}$ seemed to be the best compromise gathering all the relevant features such as full DRX at relatively small strains (approximately $\varepsilon_{ss}=1$), a peak strain $\varepsilon_p=0.6$ and minimum static softening between strain interruptions. This was then the condition chosen to perform the base experiments for this work. It will be shown later that monotonic tests under such conditions lead to a critical strain for the beginning of dynamic recrystallization ($\varepsilon_c=0.38$). The first three values of $\Delta\varepsilon$ are below the critical strain for the beginning of dynamic recrystallization ($\varepsilon_c=0.38$). The fourth strain amplitude is similar to $\varepsilon_c$. The value of $\Delta\varepsilon=0.54$ is close to the peak strain ($\varepsilon_p=0.6$) and the final value of $\Delta\varepsilon$ (0.86) is between the peak strain and the strain for the full development of dynamic recrystallization ($\varepsilon_{ss} \approx 1.0$).

2.4. Static Softening during Interrupted Reversed Torsion Testing

The time interval spent between reversals in this work was of the order of 0.1 to 0.3 s. This may allow some static softening to occur in between deformation steps masking any effects derived exclusively from strain path changes. Therefore, measurements of the fraction of static softening were carried out as shown in Fig. 3. Static softening was obtained from double deformation torsion testing$^{17}$ with time intervals up to 3 s, ten times longer than the usual
maximum time intervals spent in the reversed torsion experiments. It can be seen that softening is negligible and almost nil up to 0.5 s, at the test condition. There is then an increase in the softening fraction to 10% at times of the order of 1 s. Softening then stops and the curve shows, apparently, a small plateau which is followed by an increase in the fraction of softening up to 30% at 3 s. The curve was extrapolated using an Avrami type equation to full softening taking about 100 s to completion. It can be concluded that no relevant static softening could have occurred during the time intervals spent in the strain reversals, since they were always smaller than 0.3 s.

2.5. Metallography

Samples were taken from a longitudinal section of the torsion specimens, at a depth of about 150 μm from the surface, for optical and electron scanning microscopy. Preparation of samples for optical microscopy involved grinding down to paper graded 1 000 mesh and polishing with diamond paste sizes 9, 3, 1 and 0.25 μm. Etching was carried out with 6% nital for a period varying from 20 to 120 s.

3. Results

3.1. Continuous Torsion Testing Stress–Strain Curves

Figures 2(a) to 2(d) show the stress–strain curves for samples deformed in continuous torsion testing at several temperatures and strain rates. The stresses increased up to a peak stress, σ_pe, and then decreased until a steady state, σ_st, was reached. This shape of the stress–strain curve indicates the occurrence of DRX in all tests. As expected, the values of the peak stress and peak strain increased as the temperature decreased or the strain rate increased. Samples tested at 0.01 s⁻¹ showed oscillating values of the σ_pe related to grain size growth during recrystallization whereas all other curves displayed no such oscillations.

As already mentioned, the temperature of 1 223 K and the strain rate of 0.1 s⁻¹ were chosen as the base conditions for all tests in the present research. Figure 4(a) highlights the stress–strain curve obtained under these conditions, whereas Fig. 4(b) shows the work hardening rate (θ) versus stress (σ) curve. The value of the peak stress can be measured at the point of the θ–σ curve at which the value of θ is nil and, in this case, σ_c is 102 MPa. The value of the peak strain can then be measured from Fig. 4(a), producing ε_p = 0.6. The values of the critical stress and critical strain at which DRX is initiated can also be determined from the θ–σ curve, as seen in Fig. 4(b). The value of the critical stress, ε_c, is obtained at the first point where the slope of the θ–σ curve changes from decreasing to increasing values and then the critical strain, ε_c, can be determined from the stress–strain curve shown in Fig. 4(a). In the present work, these values were σ_c = 95 MPa and ε_c = 0.38, respectively.

3.2. Reversed Torsion Testing Stress–Strain Curves

Figures 5(a) to 5(d) show the results obtained for some of the reversed straining torsion testing, for the small strain amplitudes (Δε = 0.03 and 0.09) and for the high strain amplitudes (Δε = 0.54 and 0.86). The stress–strain curves for the intermediate levels of strain amplitude (0.20 and 0.38) display features intermediate to those for low and high strain amplitudes. For comparison purposes, all figures display the continuous stress–strain curve plotted as a dotted line. The steady state stress values achieved for the envelope curve of the reversed torsion, σ_st, are lower than the same values shown for the continuous curve, σ_st, for the low strain amplitudes. As the strain per pass increases from 0.03 to 0.09, the value of σ_st increases, approaching σ_pe for the continuous curve, although remaining yet considerably lower than the latter value. As the value of the strain per reversed pass is increased further, σ_st approaches that of the continuous curve, as shown in Figs. 5(c) and 5(d). In these cases, however, there are straining cycles which tend to be above the value of the monotonic steady state stress (σ_st). This is most probably associated with the interruption and/or delay of the dynamic recrystallization (DRX) and the restructuring of the dislocation arrangements not yet affected by the DRX, caused by the first strain inversion. As described in the literature, this inversion leads to a delay in the DRX started in the first cycle, associated with an ini-
tial “plateau” in the stress–strain curve, where the previous dislocation structure is rearranged, which is then followed by an increase in the flow stress and a final, new, DRX peak. In Fig. 5(c), the initial strain of 0.54 leads to incomplete DRX. The strain reversal would delay this DRX and provoke a rearrangement only of the dislocations in the non-dynamically recrystallized fraction of the material, causing the stress “plateau” in the second strain cycle. It seems, however, that the strain amplitude in this cycle was not sufficient to complete the dislocation rearrangement, since no increase in stress was detected. On the other hand, such hardening is observed in the third straining cycle, where the final stress climbed above the steady state stress under monotonic torsion ($\sigma_{ss}$), as a new DRX peak is begun. The fourth straining cycle in Fig. 5(c) would then correspond to a situation similar to that in the second cycle. The partial DRX at the end of the third cycle would be delayed by the strain reversal in the fourth cycle and an initial stress “plateau” is again detected. It is noteworthy that no appreciable difference in the flow stress at the end of the second cycle and at the beginning of the third cycle in Fig. 5(c) is observed, suggesting that the dislocation structure in both situations is similar. This is not the case between the first and second cycles and between the third and fourth cycles, where a substantial dislocation rearrangement probably occurs.

The results in Fig. 5(d) can be interpreted similarly to Fig. 5(c), but the initial higher strain (0.86) leads to a smaller fraction of material which has not undergone DRX than in the case of Fig. 5(c). The first strain reversal delays the progress of the DRX in the first cycle, and initially causes only the rearrangement of the smaller fraction of the unrecrystallized material. There is a stress drop, but a low strain is necessary for the dislocation rearrangement (due to the smaller fraction of unrecrystallized material in comparison to the situation in Fig. 5(c)), and work hardening is quickly restarted, similarly to the situation in the third cycle of Fig. 5(c). The third cycle in Fig. 5(d) displays an initial stress close to the stress at the end of the second cycle, as in the case of the second and third cycles of Fig. 5(c). As should be expected, the third cycle displays a lower work hardening rate than the second one, and the stress is raised above the monotonic steady state stress ($\sigma_{ss}$) as one approaches a new DRX peak.

### 3.3. Changes in Grain Size during Continuous and Interrupted Reversed Torsion Testing

Figures 6(a) to 6(d) show the microstructure of the ferrite transformed from the IF austenite cooled down to room temperature at several values of strains during testing in continuous torsion. Figure 6(a) displays the initial microstructure for comparison purposes. Figure 6(b) shows the microstructure for a sample deformed up to a strain of 0.3, smaller than the critical strain for the initiation of DRX. The grains are relatively coarse and of the order of $35 \mu m$, and a mixture of large and fine grains is also observed, denoting that full DRX had still not occurred at this point. Figure 6(c) shows, on the other hand, a homogeneous grain size distribution revealing that the sample was already deformed well into the steady state of the continuous stress–strain curve. The grain size here is of the order of $20 \mu m$, considerably smaller than that presented after a straining of 0.3.

A similar effect was found for the reversed torsion experiments, as seen in Figs. 7(a) to 7(f). All samples were cooled down at strains well into the steady state. Figures
7(a) and 7(b) show that applying small strains per pass, \( \Delta \varepsilon = 0.03 \) and 0.09, led to relatively homogeneous coarse ferrite grain sizes of 55 \( \mu m \) and 42 \( \mu m \) respectively. The same occurred when \( \Delta \varepsilon \) is incremented to 0.20 (Fig. 7(c)), presenting the grain size of 39 \( \mu m \) and a microstructure similar to that obtained for the continuous deformation experiment interrupted at a strain of 0.3. As \( \Delta \varepsilon \) was increased to 0.38 (Fig. 7(d)), just at the initiation of DRX, the ferrite grains showed a more inhomogeneous distribution of sizes and small grains appear surrounded by larger ones, display-
ing an average size of 34 μm. The same situation is observed for the sample deformed with $\Delta \varepsilon = 0.54$ (Fig. 7(e)), larger than the critical one for the initiation of DRX but still considerably smaller than that needed to achieve the $\sigma_{ss}^c$. The grain size is now around 27 μm. At a strain of 0.86, the accumulated strain in the sample was well within the steady state regime and the ferrite grain sizes became more uniform and considerably smaller than those seen in Figs. 7(a) to 7(e). The grain refining effect was similar to that found for the continuous deformation experiment, leading to an average grain size of the order of 20 μm.

4. Discussion

4.1. Effect of Strain Reversals on the Dynamic Softening of IF Austenite

As reported elsewhere,$^{11}$ DRX was observed in the monotonic torsion of copper at 773 K under a strain rate $\dot{\varepsilon} = 0.1$ s$^{-1}$, promoting extensive grain refinement, as expected. However, cyclic torsion stress–strain curve of copper tested at 673 K, under a strain rate expected. However, cyclic torsion stress–strain curve of copper tested at 673 K, under a strain rate $= 0.1$ s$^{-1}$ and strain amplitude $\Delta \varepsilon = 0.24$ did not display the characteristic DRX peak.$^{10}$ TEM evidence indicated the absence of DRX even after a large amount of cyclic torsion, and the sole observed restoration mechanism was dynamic recovery (DRV). Furthermore, cyclic torsion of copper under the above-mentioned experimental conditions involved the successive disintegration and build-up of dislocation microbands. The accumulated strain energy was not, therefore, sufficient to trigger DRX. This result was consistent with those reported for strain path changes under cold working. Such restructuring caused enhanced levels of DRV, eliminating the DRX observed under monotonic straining. The similarity of copper and of austenite concerning their structure and stacking fault energy is a strong indication that the experimental results discussed above for copper can be applied to the present results for hot torsion of austenitic IF steel. Therefore, it is believed that, for the lower strain increments employed in this research (0.03, 0.09, 0.20), the DRX prevailing for monotonic torsion is completely eliminated under cyclic torsion up to strains much higher than those needed to trigger DRX under monotonic torsion. The restoration mechanism should involve only the repeated disintegration and build-up of a microband structure, similarly to copper. A consequence of this is the low number of small grains in the observed microstructures after cyclic torsion into the saturation stress range (see Figs. 7(a), 7(b) and 7(c)). On the other hand, the corresponding microstructures for higher strain amplitudes (Figs. 7(d), 7(e) and 7(f), for strain increments of 0.38, 0.54 and 0.86 respectively) display an increasing amount of small grains, caused by increasing levels of DRX occurring simultaneously with the DRV associated with cyclic torsion. One concludes that the effect of cyclic torsion on the restoration mechanism under hot working conditions of austenite depends on the magnitude of strain increment. Low increments can completely eliminate DRX, promoting DRV as the sole restoration mechanism. Increasing strain increments introduce more and more DRX, causing the presence of localized small grains.

It has been previously shown for copper$^{11}$ that the saturation stress ($\sigma_{ss}^c$) in cyclic torsion increases with the ratio between the strain increment and the peak strain ($\Delta \varepsilon / \varepsilon_{ss}$, also known as relative strain). For the case of ferritic IF steel$^{12}$ no DRX is observed during its hot monotonic or cyclic torsion. However, once again the saturation stress increases with strain increment, but there is now a relationship between $\sigma_{ss}^c$ and $\Delta \varepsilon / \varepsilon_{ss}$, where $\varepsilon_{ss}$ is the necessary strain for the onset of a saturation stress under monotonic torsion.

For the present experiments, Fig. 5 shows that, as $\Delta \varepsilon$ increases, the corresponding levels of the saturation stress, $\sigma_{ss}^c$, approach and may even partially surpass the steady state stress under monotonic torsion ($\sigma_{ss}$), which is approximately 90 MPa. For practical purposes, one can consider that, even for the high strain amplitude range, $\sigma_{ss}^c \approx \sigma_{ss}^c$. Figure 8 shows that $\sigma_{ss}$ increases with the relative strain ($\Delta \varepsilon / \varepsilon_{ss}$), approaching the saturation stress for monotonic torsion ($\sigma_{ss}$ dotted line, at approximately 90 MPa). The situation is thus similar to that described for copper and for IF steel deformed in the ferritic range. The present experimental data can be fitted to an equation of the type.

$$
\sigma_{sat} = \sigma_0 + (\sigma_{ss} - \sigma_0) \left[ 1 - \exp \left( - C \left( \frac{\Delta \varepsilon}{\varepsilon_{ss}} \right) \right) \right]^{m} \cdots (6)
$$

The value of $\sigma_0$, $C$, and $m$ that gave the best fit to the experimental data (to a R$^2$ of 0.99) were 35 MPa, $4.4 \pm 0.3$ and $0.80 \pm 0.05$, respectively. Figure 8 indicates that the value of $\sigma_{ss}^c$ is approximately 90 MPa for relative strains higher than about 0.63. This is not surprising since the critical strain for the initiation of DRX is in the range of 0.6 to 0.8 of the peak strain.$^{51}$ It thus seems that once DRX has been initiated in cyclic torsion, the levels for $\sigma_{ss}$ and $\sigma_{ss}^c$ become approximately the same.

On the other hand, the smaller values of $\sigma_{ss}^c$ for relative strains lower than 0.63 can be correlated to a corresponding smaller average dislocation density, $p_{ss}^{\rho}$, associated with an enhanced dynamic recovery caused by the reversing strains. This reduction in the mean dislocation density could not have resulted from static softening between strain cycles since this was measured and found to be negligible. The decrease in $\sigma_{ss}^c$ with respect to $\varepsilon_{ss}$ when $\Delta \varepsilon$ becomes smaller than $\varepsilon_{ss}$ is reasonably large and independent of the total accumulated plastic strain. This is supported by the torsion findings of Bartolome$^9$ and others when they reported a lower yield stress measured on reloading, followed, as de-
formation continued, by a small stress plateau and then by an increase in the stress as deformation proceeded. In this way, it is concluded that the reduction in the value of \( \sigma_{av} \) measured for the reversed torsion experiments at low strain amplitudes resulted mostly from a decrease in the average dislocation density, in relation to the corresponding density for monotonic straining, for that particular deformation condition.

### 4.2. Effect of Strain Reversals on Ferrite Grain Size

On cooling, IF austenite with the composition employed in this paper transformed into ferrite, no matter which cooling rate was used, indicating a very rapid kinetics of phase transformation. Ferrite grain size depends on cooling rate and prior deformation of the austenite grains. Cooling rates were kept approximately constant for the samples tested here. It is thus reasonable to assume that the sizes of the ferrite grains are related to the sizes and shapes of the prior austenite grains. This reasoning explains the ferrite grain sizes and morphology shown in Fig. 6, for continuous straining. The ferrite transformed from a sample deformed to an accumulated strain lower than the critical strain to the initiation of DRX produced, in the present investigation, equiaxed ferrite grains of a certain size (see Fig. 6(b)). On the other hand, ferrite transformed from austenite deformed to an accumulated strain larger than that to the initiation of the steady state stress, that is, larger than the strain needed for the completion of DRX, also led to equiaxed grains, but with size substantially smaller than the size of the grains originating from straining below the onset of DRX (see Fig. 6(d)). Austenite grains deformed to an effective strain between the critical strain to initiate DRX and the critical strain to initiate steady state produced ferrite grains with a mixture of small and large grains (see Fig. 6(c)) reflecting the originally heterogeneous austenite structure.

In the present circumstances, the rapid quenching of the sample did not allow the occurrence of SRX after both monotonic and cyclic straining. However, monotonic torsion into the steady state stress leads to DRX and a resulting stable austenitic grain size which in turn will transform to a fine ferrite grain size, as shown in Fig. 6(d). Reversed torsion at low amplitudes eliminates DRX, turning DRV into the dominant restoration mechanism. This is associated with the successive disintegration and build-up of dislocation structures and a resulting lower level of stored energy, in comparison to monotonic torsion. The cyclic character of straining does not lead to any substantial level of austenite grain elongation, actually eliminating the consideration of this factor as important in the determination of the transformed ferrite grain size. One concludes that the final ferrite grain size after cyclic torsion of austenite, in the absence of DRX and SRX, depends exclusively on the stored energy during deformation, which is related to the prevailing saturation stress (\( \sigma_{ss} \)). Increasing strain increments causes an increase in \( \sigma_{ss} \) and the associated stored energy, and should lead to finer ferrite grain size. This is clearly shown in Figs. 7(a), 7(b) and 7(c)

Further increases in strain increments lower the restoration caused by DRV and trigger more and more DRX, leading to the presence of an increasing number of small dynamically recrystallized grains. The resulting ferrite microstructure should thus display small grains originating from the DRX grains, embedded in a ferrite matrix whose grain size should decrease as strain increment is increased. This is the situation shown in Figs. 7(d), 7(e) and 7(f).

Previous work published in the literature \(^{19}\) supports the above interpretation of the phenomena. It was found that the mean flow stress of the last passes of deformation in torsion correlated to the ferrite grain sizes of IF steels microalloyed with Ti, Nb and Ti–Nb according to

\[
d = k \sigma^{-0.125} \tag{7}
\]

The analysis presented in Fig. 9 confirms qualitatively the above description of the grain size dependence on the flow stress. For reversed torsion, \( k = 5.45 \times 10^5 \mu \text{m MPa}^{-1} \) and \( \sigma = -1.17 \). However, it should be noticed that, for the same flow stress, monotonic torsion leads to a somewhat lower grain size than reversed torsion, as highlighted in the figure.

### 5. Conclusions

The cyclic torsion of an IF steel at 1 223 K and 0.1 s\(^{-1}\) leads to steady state stresses (\( \sigma_{ss} \)) that increase with the relative amplitude of the strain cycles (\( \Delta \varepsilon / \varepsilon_p \), where \( \varepsilon_p \) is the strain at the peak stress in monotonic torsion of the material), \( \sigma_{ss} \) is lower than the corresponding steady state stress under monotonic torsion (\( \varepsilon_{ss} \)), up to \( \Delta \varepsilon / \varepsilon_p \) values associated approximately with the critical strain for dynamic recrystallization (DRX) initiation.

For \( \sigma_{ss} \), values below \( \sigma_{ss} \) the prevailing softening mechanism is dynamic recovery (DRV). Under such circumstances, the lower values of \( \sigma_{ss} \) in relation to \( \sigma_{ss} \) are associated with a lower average density of dislocations (\( \rho_{ss} \)) than that generated by monotonic hot torsion (\( \rho_{ss} \)). This is caused by the repeated build-up and disintegration of dislocation structures for the successive strain cycles. Increasing values of \( \Delta \varepsilon / \varepsilon_p \), above the level associated only with DRV, trigger increasing amounts of DRX, whose initiation is delayed to higher strains, in relation to the situation for monotonic torsion.

The lowering of the relative strain per cycle (\( \Delta \varepsilon / \varepsilon_p \)), causes an increase in the final average ferritic grain size (\( d_{ss} \)) of the material, after deformation into the steady state stress. This grain size is associated with the originating austenitic grain size (\( d_{ss} \)), and depends on the cyclic steady state stress (\( \sigma_{ss} \)), which is a measure of the stored energy in the material and thus of the average density of dislocations.
The situation is qualitatively similar to that observed for monotonic torsion, where the final ferritic grain size depends on the observed average flow stress.

Acknowledgments

The authors acknowledge the financial support given to this research by Brazilian Council for Scientific Research and Technological Development, CNPq, through the PRONEX Program, contract nr.317/96. Thanks are also due to CDTN/CNEN (Centro de Desenvolvimento de Tecnologia Nuclear/Comissionao Nacional de Energia Nuclear), for annealing the specimens.

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