Axial Contraction during Torsion Simulation of Steel Rolling at High Temperatures

J. J. JONAS and L. S. TÓTH

McGill University, Montreal, H3A 2B2 Canada. E-mail: john.jonas@mcgill.ca
1) Université de Metz, Ile du Saucy, Metz 57000 France.

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The deformation textures determined in previous experiments on the free-end torsion testing of pure copper (99.95 %) bars are reviewed. At room temperature and 125°C, lengthening is observed up to shear strains of 11 or 12. By contrast, at 200 and 300°C, shortening takes place after initial shear strains of about 10 and 5, respectively. Simulations are carried out using polycrystal plasticity methods with the aim of reproducing the experimental textures. It is shown that the lengthening generally observed at ambient temperatures is entirely attributable to the characteristics of dislocation glide in fcc metals. By contrast, the “shortening” that takes place at elevated temperatures cannot be reproduced in this way and requires introduction of the concepts of the nucleation and growth of new grains. Comparison of the experimental results and the predictions of the simulations leads to the conclusion that the very large axial contractions frequently reported during the torsion simulation of steel rolling schedules are due to dynamic recrystallization.

KEY WORDS: torsion testing; rolling simulation; austenite deformation; dynamic recrystallization; axial shortening; axial compressive stresses; texture changes.

1. Introduction

Torsion testing is commonly used for the simulation of rolling schedules because the very large accumulated strains imposed industrially can be readily attained. Furthermore, the temperature changes associated with rolling can be fairly easily simulated and interpass times as short as 10 ms can be imposed. Although the maximum strain rates that are readily controllable do not much exceed \(10^{-3}\) s\(^{-1}\), extrapolation methods are available to extend the observed trends to higher values. One of the remaining difficulties with this technique is that axial stresses are developed under “fixed end” conditions of testing, whereas length changes are generally observed when the experiments are carried out under “free end” conditions.

These length changes have been referred to as the Swift effect since the classical “free end” experiments of this author in 1947.\(^1\) Swift observed that most metals (with the notable exception of lead) “lengthen” at room temperature, while samples undergo shortening when tested at elevated temperatures. The lengthening of samples is readily explained by the usual methods of polycrystal plasticity. This arises because simple shear (torsion) produces lattice rotations (and therefore non-random textures) such that the mean slip plane of the dislocations is gradually inclined upwards in the direction of shear. This inclination is responsible for the increase in length of the sample.\(^2\) Note, however, that at the start of straining, and as long as the initial texture of the sample is random, there is neither lengthening nor shortening. This observation relies on the symmetry of the sample texture, which requires the number of grains belonging to “lengthening” orientations to be exactly equal to those belonging to “shortening” orientations. The departure from this simple state of affairs caused by lattice rotation and the progressive increase in the number of “lengthening” grains is then responsible for the lengthening phenomenon.\(^3\)

By contrast, it is much more difficult to account for specimen shortening. Because this only takes place at high homologous temperatures (and is absent in aluminum), it is of course tempting to attribute it to dynamic recrystallization. The possibility that the shortening of torsion samples is indeed caused by this phenomenon is examined in this paper. If testing is carried out under fixed end conditions, lengthening is replaced by the development of compressive axial forces and shortening by that of tensile forces. It is shown here that the texture changes associated with dynamic recrystallization are responsible for axial shortening and by extension for the development of tensile axial forces.

2. Experimental Observations

When rolling simulations are being carried out, typical accumulated equivalent strains can reach 3.9. (This represents a reduction in the bar thickness from 50 mm down to 1 mm). When the “three critical temperatures” of hot rolling (\(T_{nr}\), \(A_{r3}\) and \(A_{r1}\)) are being evaluated, still larger total strains are often applied. Under these conditions, torsion specimens can undergo considerable shortening, with the axial contraction attaining values as high as 50%, see
Because of the phase change that takes place in conventional austenite, the experimental textures that play such an important role in anisotropic plasticity have not been determined. For this reason, we will rely here on the textures determined on relatively pure (99.95%) copper tested in a “free-end” machine at North Carolina State University.

These experiments were carried out at room temperature, as well as at 125, 200, and 300°C, i.e. at homologous temperatures of 0.22 to 0.42. This range is formally equivalent to temperatures of 113 to 482°C for steel. While the latter is somewhat low for austenite (hot) rolling, the high purity of the copper rendered it particularly susceptible to dynamic recrystallization. Nevertheless, 300°C was the highest temperature that enabled the deformation texture of the copper to be retained at room temperature. It was also high enough to permit the initiation and propagation of dynamic recrystallization. It should be noted that “solid bar” torsion specimens were used, because of the experimental difficulties involved in the deformation of thin-walled tubes (these tend to collapse during testing). Such solid bar specimens can be expected to undergo less shortening than thin-walled tube samples deformed to the same shear strain because of the restraining effects of the inner layers of the solid bar.

Some of the experimental textures observed at room temperature and at 300°C are reproduced in Fig. 2 in ODF form. Here, only the 0, 15, 30 and 45° sections are shown and the 60 and 75° sections normally required for a full representation of textures of monotonic symmetry are omitted for reasons of simplicity. The texture developed at room temperature at a shear strain of $\gamma=11$ is illustrated in Fig. 2(a) and those pertaining to $\gamma=4$ and $\gamma=11$ at 300°C in Figs. 2(b) and 2(c). A key to the ideal orientations evident in the diagrams is provided in Fig. 2(d).

It is clear from Fig. 2(a) that the C {100} component is the most intense at room temperature, reaching a level of just above 5 times random at $\gamma=11$. It is also a feature of this texture that the ideal orientations are “rotated” away from their symmetry positions by several degrees. By contrast, when the temperature is increased to 300°C, Fig. 2(c), the most intense components change to the A/A– {111}, with levels of more than 6 and 5 times random, respectively. The presence of a rotated cube component (350) (along the top axis) can also be seen.

While these differences may appear small, the texture transitions away from the C and towards the A/A–, together with the changes in the tilts, are of particular importance with respect to axial effects. The length changes observed in these tests are illustrated here in Fig. 3. It can be seen that lengthening begins rapidly and then slows down at room temperature (and at 125°C), but that it continues unabated to shear strains of 12. By contrast, at 300°C, there is lengthening up to a strain of about 5 (about 10 at 200°C), after which shortening is clearly evident. It is the aim of the present paper to propose a physical explanation for the tex-
ture transitions and axial effects and to demonstrate how the occurrence of dynamic recrystallization at elevated temperatures can produce, not only the texture differences, but also convert the axial effects from lengthening to shortening (in free-end testing) and from compression to tension (in fixed-end testing).

3. Computer Simulations

3.1. The DRX Model

In this work, the dynamic recrystallization model introduced by Jonas and Tóth and Tóth and Jonas for fcc materials was employed. Although a more refined version of the original approach (using the self-consistent model and the 'volume transfer scheme') was developed for bcc materials by Hildenbrand et al., here the Taylor viscoplastic polycrystal code was used. This is because the latter permits relaxation of the axial strains in simple shear, which is more appropriate when length changes of the aggregate are being simulated.

The present model permits the simulation of both oriented nucleation and growth ('ONG') and selective growth ('SG'). The stored energy plays important roles in both ONG and SG during DRX, for which purpose the Taylor factor was used as a controlling parameter. Although the Taylor factor reflects only the instantaneous plastic work rate of the deformation process, it was shown by Bacroix et al. that the instantaneous work rate is approximately proportional to the stored energy (which is the actual driving force). Here the Taylor factor \( M \) is defined as

\[
M = \frac{\sum s \dot{\gamma}_s}{\tilde{\varepsilon}} \quad \text{...........................(1)}
\]

where \( \dot{\gamma}_s \) is the shear rate in slip system \( s \), \( n \) is the number of slip systems, and \( \tilde{\varepsilon} \) is the von Mises equivalent strain rate of the crystal.

During oriented nucleation followed by growth, the new crystal has the same orientation as its parent and its growth is assumed to be controlled by the following relationship:

\[
\frac{\Delta V}{V} = R_{\text{ONG}}^{\max} \left( \frac{M - M_{\min}}{M_{\max} - M_{\min}} \right) \quad \text{........................(2)}
\]

In this relation, \( \Delta V/V \) is the new volume fraction created by ONG during a given increment of strain, \( R_{\text{ONG}}^{\max} \) is the parameter that controls the maximum rate of ONG, \( M_{\min} \) is the minimum Taylor factor (equal to \( \sqrt{3} \)) and \( \delta \) is the half-width of the gaussian distribution (0.789 in the simulations).

In the second DRX mechanism, selected nuclei grow that are in coincidence site lattice positions with respect to the parent grain. For this purpose, 40° rotations around (111) axes of the fcc lattice were adopted. As there are 4 such axes in a cubic crystal, some form of variant selection was required to prevent randomization of the texture. As in the original model, the particular (111) plane chosen (to which the selected (111) axis was perpendicular) carried the maximum crystallographic shear due to dislocation glide. The plane of interest was identified with the help of the crystal plasticity model, which provided the slip activity of all slip systems. Positive and negative rotations around the particular (111) axis were equally considered for the generation of nuclei. These nuclei were then allowed to grow according to the following relation:

\[
\frac{\Delta V}{V} = R_{\text{SG}}^{\max} \left( 0.4 + 0.6 \frac{M - M_{\min}}{M_{\max} - M_{\min}} \right) \quad \text{.............(3)}
\]

Here \( R_{\text{SG}}^{\max} \) is the parameter that controls the maximum rate of growth by SG and \( M_{\max} \) is the maximum value of the Taylor factor (equal to \( \sqrt{18} \)). This mechanism of DRX leads to the growth of grains with orientations that differ from that of the parent grain. It also increases the number of grain orientations. The latter development is difficult to deal with when a large number of strain increments is employed, as in the present case. To solve this difficulty, a special simulation technique was applied, which is described below.

3.2. Simulation Method

The model polycrystal was composed of 500 grain orientations that were selected to represent the initial texture of the copper samples, which was not perfectly random. In the DRX simulations, a deformation texture was first established by twisting to an initial shear of \( \gamma = 0.5 \), the strain considered necessary to initiate the DRX process under the experimental conditions. After that strain, both plastic flow and DRX were modeled in sequential steps as follows. In the first cycle, four plastic strain increments (of \( \Delta \gamma = 0.05 \)) were applied using only dislocation glide. Then allowance was made for DRX by changing the volume fractions of the new grain orientations using Eq. (2) for ONG and Eq. (3) for SG. During SG, the number of grain orientations is tripled because each initial grain orientation produces two new crystals in different locations in Euler space. The increased population of grains was then reduced to 500 by calculating the new orientation distribution function (ODF) of the texture and re-discretizing it. For this purpose, the discretization technique of Tóth and Van-Houtte was used. With this last step, the second cycle was completed. The two cycles were repeated as many times as necessary to produce the large strains of interest. The parameters for ONG and SG were tuned so that both the simulated textures and length changes were in reasonable agreement with the experimental observations. The following values led to the

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**Fig. 3.** Length changes observed in the experiments of Ref. 5).
best results: $R_{\text{ONG}}^{\text{max}} = 1.2, R_{SG}^{\text{max}} = 0.2$. 

3.3. Texture Predictions

The textures predicted using the above model are displayed in Figs. 4 and 5. The former applies to a shear strain of $\gamma = 11$ in the “absence of dynamic recrystallization”, where only dislocation glide is taking place. (This is the simulated room temperature texture.) Simulations were carried out for two kinds of boundary conditions: i) by relaxing “all” normal strain components (as in the DRX simulations), and ii) by imposing the measured lengthening strain (with equal contractions in the radial and circumferential components) together with the shear strain. The results pertaining to strains of $\gamma = 4$ and $\gamma = 11$ obtained from the dynamic recrystallization model are reproduced in Figs. 5(a) and 5(b), respectively. The $\gamma = 4$ ODF represents the deformation texture as modified by the nucleation and growth of new grains. Here, nucleation is considered to be initiated at a shear strain of $\gamma = 0.5$ and sufficient growth has taken place for lengthening to be arrested at a shear strain of $\gamma = 2$. Beyond this strain, shortening continues unabated.

It can be seen that, while there are some differences between Figs. 2(a) (experimental) and 4 (simulation), the main features of the experimental torsion textures are well reproduced by the simulations of the deformation textures. This is especially true for the texture presented in Fig. 4(b), which corresponds exactly to the strain path of the experiment. Although the C component is too strong in the simulation, this can be attributed to the Taylor conditions, according to which the evolution in grain shape is disregarded. When the form effect is taken into account, as in self-consistent polycrystal simulations, more realistic simulated textures can be obtained.11)

Concerning the simulations carried out under dynamic recrystallization conditions, comparison of Figs. 2(c) (experimental, 300°C) and 5(b) (simulation, DRX) can be seen to justify the assumptions regarding the operating slip systems and the DRX model. It should be clear that when the occurrence of DRX is taken into account, the grain shape does not evolve as dramatically as under glide conditions, so that the Taylor description of deformation is a better approximation.

These comments do not, on the other hand, apply to a comparison of Figs. 2(a) (the experimental deformation texture) and 5(b) (the simulated DRX texture). This is because the introduction of dynamic recrystallization (in the model) has led to the sharp reduction in the intensity of the C component and its replacement by the A/A– and B/B–, as observed experimentally. Another effect of DRX is the ‘stabilization’ of the texture, which is clear from a comparison of Figs. 5(a) and 5(b). The introduction of nucleation and growth is consistent with numerous metallographic observations of dynamic recrystallization.

4. Discussion

An important aspect of the simulations involves the predicted axial strains. These are illustrated in Fig. 6, where it is evident that, in the absence of dynamic recrystallization, the specimen continues to lengthen without interruption up to the strains of $\gamma = 11$ or 12 employed here. Note that these observations apply to “solid bar” samples, which undergo less axial strain per unit shear strain than thin-walled tube samples. By contrast, when allowance is made for the initiation of dynamic recrystallization (in the model) has led to the sharp reduction in the intensity of the C component and its replacement by the A/A– and B/B–, as observed experimentally. Another effect of DRX is the ‘stabilization’ of the texture, which is clear from a comparison of Figs. 5(a) and 5(b). The introduction of nucleation and growth is consistent with numerous metallographic observations of dynamic recrystallization.
usual state of affairs comes about because dynamic recrystallization involves a steady state of deformation in which work hardening is balanced by the softening attributable to grain boundary migration. The motion of the boundaries, in turn, removes the dislocations introduced by straining. DRX also replaces the "lengthening" grains introduced by the lattice rotation due to dislocation glide by a larger number of "shortening" grains.

When torsion testing is carried out under fixed-end conditions, compressive axial stresses are developed at room temperature and tensile axial stresses at austenite deformation temperatures. Although not shown here in detail, the compressive stresses can be readily reproduced by the usual methods of crystal plasticity based solely on dislocation glide. By contrast, the appreciable "tensile" stresses generally observed at elevated temperatures can only be reproduced by models that allow for the occurrence of dynamic recrystallization.

5. Conclusions

(1) The length changes occurring under free-end conditions of torsion testing can be appropriately simulated using various crystal plasticity models. For lengthening, only dislocation glide events (with suitable grain interaction conditions) need to be incorporated in the model. However, in order to reproduce axial contraction, allowance must be made for dynamic recrystallization. This includes both the oriented nucleation and selective growth aspects of the phenomenon.

(2) The calculations described here can account for the axial shortening strains of as much as 50% that have been reported by Boratto\(^4\) and other workers.

(3) Although not reported here in detail, simulations of a similar type based on the occurrence of DRX can reproduce the axial tensile forces that develop when austenite is subjected to high temperature torsion testing under "fixed-end" conditions.

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