Effect of Interpass Time on the Dynamic Transformation of a Plain C–Mn and a Nb Microalloyed Steel

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(Received on October 14, 2014; accepted on November 23, 2014)

Strip rolling simulations were carried out on a 0.06%C-0.3%Mn-0.01%Si and a 0.11%C-1.0%Mn-0.11%Si-0.03%Al-0.034%Nb steel over the temperature range 1 000°C to 883°C. Pass strains of 0.4 were applied at a strain rate of 1 s⁻¹ with interpass times of 0.5 s, 1 s, 1.5 s, 3 s and 5 s. Two different temperature schedules were employed, namely i) continuous cooling at 6°C/s and ii) isothermal holding. The mean flow stresses (MFS’s) applicable to strip rolling were determined by integration. The flow stress levels and MFS’s decrease when the interpass times are short. When they are long, the flow stress increases with decreasing temperature. These observations indicate that the austenite is transforming dynamically into ferrite and statically into austenite. The nucleation and growth of the ferrite reduce the rolling load and modify the microstructure. The addition of Nb in solid solution delays the occurrence of dynamic transformation (and the retransformation of ferrite back into austenite). The forward nucleation of ferrite occurs displacively while the retransformation back into austenite takes place by a diffusional mechanism.

KEY WORDS: thermomechanical processing; torsion testing; interpass time; phase transformation; dynamic transformation.

1. Introduction

Deformation below the $Ae_3$ and above the $Ar_3$ temperature is known to provoke strain-induced transformation and to produce ferrite grain sizes of 1–3 μm (or even less). However, transformation can also be initiated above the $Ae_3$, as shown by the work of Yada and coworkers in the 1980’s, but under these conditions, it is dynamic and not static. By means of compression tests, they demonstrated that the dynamic transformation (DT) of austenite can be made to take place as much as 166°C above the $Ae_3$ temperature. Strip rolling simulations using interpass times of 1 s were also performed by Yada et al. in 1988. These experiments were again carried out above the $Ae_3$ temperature and led to similar results. They showed that the reverse transformation took place on isothermal holding after deformation. By using a dilatometer, they ascertained that the volume fraction of dynamically transformed ferrite decreased with holding time until all the DT ferrite was completely transformed back into austenite.

Earlier simulations of strip rolling had already indicated that the length of the interpass time can have a significant effect on the rolling load. In 1988, Samuel et al. showed that short interpass times lead to significant reductions in MFS. The rolling simulations of Bowden et al. carried out on microalloyed steels indicated that Nb(C,N) precipitation retarded both dynamic (DRX) and static (SRX) recrystallization. However, in these investigations, no consideration was given to the possibility that austenite was transforming dynamically into ferrite.

More recently, it has again been demonstrated that dynamic transformation can take place during deformation above the $Ae_3$. The critical strain required to initiate this phenomenon has been shown to be around 0.12. These values were calculated using the double differentiation method. By means of rolling simulations, dynamic transformation has also been linked with low rates of MFS increase and even decreases in some cases. Nevertheless, the exact influence of interpass time on this phenomenon and more particularly on the amount and nature of the softening remains to be clarified. In order to investigate these effects, two alloys were selected for the current study: i) a plain C–Mn and ii) a Nb microalloyed steel. Here torsion simulations of strip rolling were carried out with the specific aim of establishing the effects on rolling load of long and short interpass times. The results obtained are described in detail below.

2. Experimental Procedure

A plain C–Mn steel and a Nb microalloyed steel were investigated in the present study. These were supplied in the form of hot rolled plates. The latter were machined into torsion specimens with diameters of 6.3 mm and gauge lengths of 22.2 mm with the cylinder axes parallel to the rolling direction. The chemical compositions of these steels, including their paraequilibrium and orthoequilibrium $Ae_3$ temperatures, are listed in Table 1. The latter were calculated using the FSStel database of the FactSage thermodynamic software.
2.1. Torsion Testing

Two contrasting temperature profiles were employed. In the first, the samples were air cooled at about 6°C/s during deformation, while in the second, all straining and holding was done isothermally. The torsion tests were conducted using a servo-hydraulic, computer-controlled MTS torsion machine equipped with a horizontal radiation furnace and a temperature controller. The torque/twist curves were converted into stress-strain curves using the Fields and Backofen\(^{17}\) formula. A thermocouple was welded to each sample to allow the deformation temperatures to be tracked with accuracy. A protective argon atmosphere was used in all tests to minimize oxidation and decarburization during heating. It is important to note that, as the sample diameter increases during free-end torsion testing (while it also shortens), the changes in diameter were taken into account in deriving the flow stresses from the torques.\(^{18}\)

In the continuous cooling schedule, shown in Fig. 1(a), the samples were heated at 1°C/s from room temperature to the austenitization temperature of 1200°C. They were then held for 20 minutes before cooling at 1°C/s down to the roughing temperature of 1100°C. After holding for 60 seconds at this temperature, the samples were strained to \(\varepsilon = 0.4\) at a strain rate of 1 s\(^{-1}\) to simulate a roughing pass. After this, samples were held at this temperature for 90 seconds to allow recrystallization to take place before cooling down to 1000°C. They were then held at this temperature for seven minutes before beginning the 7-pass deformation schedule in which pass strains of 0.4 were applied at a strain rate of 1 s\(^{-1}\). Tests were carried out using interpass times of 0.5 s, 1 s, 1.5 s and 3 s. The samples were deformed while being cooled at approximately 6°C/s. This was accomplished by lifting and opening the upper half of the radiation furnace. The first simulated pass was applied at exactly 1000°C, while the temperatures of the succeeding passes depended on the interpass time, as the cooling rate was held constant. Finally, samples were quenched right after the last pass.

For the isothermal schedule, shown in Fig. 1(b), the same heating and cooling rates, holding times, austenitization and roughing pass temperatures, as well as applied strains and strain rates were employed as in the continuous cooling schedule of Fig. 1(a). Only the finishing temperatures were different: 910°C and 930°C for the plain C–Mn and Nb-microalloyed steels, respectively. These tests were conducted isothermally with interpass times of 0.5 and 5 s. After completing the seventh pass, the samples were quenched in three different ways; i) directly quenched after the last deformation, ii) quenched after 8 s of holding at the test temperature, and iii) quenched after 15 s of holding. This procedure was adopted for assessment of the retransformation behavior.

2.2. Metallography

The torsion samples were cut perpendicular to the longitudinal axis to provide cross-sections for microstructural examination. These were mounted using a hot phenolic mounting material. The mounted samples were ground using 400, 600, 800 and 1200 grit silicon carbide papers lubricated with water. Both 3 and 1 \(\mu\)m diamond suspensions were used for fine grinding before final polishing with a 0.02 \(\mu\)m colloidal silica suspension. The surfaces were etched with 2% nital to reveal the microstructure. This was followed by applying a 10% aqueous metabisulfite (Na\(_2\)S\(_2\)O\(_5\)) solution to provide contrast between the martensite and the ferrite.

3. Results

3.1. Strip Rolling Simulations

The continuous cooling flow curves determined at a strain rate of 1 s\(^{-1}\) on the plain C steel are displayed in Fig. 2. Here interpass times of 0.5 s (Fig. 2(a)), 1 s (Fig. 2(b)), 1.5 s (Fig. 2(c)) and 3 s (Fig. 2(d)) were employed. The temperature of each pass is indicated on each flow curve and it should be noted that all the deformation temperatures are above the \(\text{Ae}_1\)'s for this steel, see Table 1.

In the absence of a phase change, the levels of the flow curves should increase from pass to pass because of the decreasing temperature. For reference, the broken line represents the expected first pass peak stress under these conditions. This was deduced from the first pass flow curve of

| Chemical compositions (mass%) and equilibrium transformation temperatures (°C). |
|-----------------|-----|-----|------|-------|-------|
| C   | Mn | Si | Nb | \(\text{Ae}_\text{yp}\) | \(\text{Ae}_\text{no}\) |
| Plain C–Mn | 0.06 | 0.30 | 0.01 | – | 870°C | 877°C |
| Nb microalloyed | 0.11 | 1.0 | 0.11 | 0.034 | 827°C | 843°C |
rolling simulations conducted at a number of decreasing temperatures to strains of 0.10 (not shown here). This strain is lower than the critical strains for DT and DRX.\textsuperscript{12,13} The slope was then normalized using the relation:

\[
g = \frac{\ln \sigma_p}{\ln \sigma_{0.1}} = \frac{d \ln \sigma_p}{d(1/T)} \quad \text{............. (1)}
\]

where \(\sigma_{0.1}\) and \(\sigma_p\) are the respective flow stresses applicable to applied strains of 0.1 and the peak strain, respectively.

Here the final pass temperatures were 967°C (Fig. 2(a)), 949°C (Fig. 2(b)), 936°C (Fig. 2(c)) and 884°C (Fig. 2(d)). It is evident that the stress levels actually decrease at short interpass intervals (Figs. 2(a) and 2(b)) and increase less than expected at longer times (Figs. 2(c) and 2(d)). For example, in Fig. 2(a), the level drops by approximately 10% in going from the 1st to the 2nd pass and by about 15%
when the 1st pass is compared with the last pass. There is also a noticeable change in the shapes of the curves (the 2nd to 7th display peaks). These changes indicate that strain accumulation is taking place, leading to dynamic and meta-
dynamic recrystallization. Similar conclusions apply to
Figs. 2(b) and 2(c).

It has been shown recently that two distinct softening mechanisms are contributing to the decreasing flow stress levels: these are dynamic transformation and dynamic
recrystallization.10 As mentioned above, the strains at which these are initiated have been identified using the double differ-
entiation method.13,14 In all cases, the critical strain for
dynamic transformation is reached before the one for
dynamic recrystallization. In the case of Fig. 2(d), the stress levels increase with decreasing temperature, but less rapidly than called for by the first pass values expected in the absence of DT and DRX and extrapolated to lower temperatures. The significance of these trends in the stress levels will be considered in more detail after presentation of the metallographic results.

Similar test conditions and parameters were used to produce the Nb steel flow curves displayed in Fig. 3. Interpass
times of 0.5 s (Fig. 3(a)), 1 s (Fig. 3(b)), 1.5 s (Fig. 3(c)) and 3 s (Fig. 3(d)) were again employed and straining was also applied at 1 s⁻¹. Once again, all the temperatures indicated are above the Ae₃. Here the stress levels decrease after the 2nd pass in Fig. 3(a) and after the 4th pass in Fig. 3(b), but not as sharply as in the case of the plain C steel, Fig. 2. The stress levels in Figs. 3(c) and 3(d) display somewhat lower rates of increase than those associated with the 1st pass flow stress peaks extrapolated to lower temperatures (broken lines). Note that the majority of the pass tempera-
tures are above 900°C, so that most of the Nb is expected to remain in solid solution. Even in the case of the 3 s inter-
pass intervals (Fig. 3(d)), where the last two pass tempera-
tures are 898 and 883°C, little carbonitride precipitation is expected to have occurred because of the limited time available in the vicinity of 900°C and below.

3.2. Isothermal Rolling Simulations

Although the continuous cooling rolling simulations described above are closer approximations of industrial roll-
ing, the characteristics of DT can be determined more readily by carrying out tests under isothermal conditions. In this way, the effects of DT and dynamic recrystallization can be distinguished more clearly. For this purpose, the present two alloys were subjected to fixed temperature rolling simulations with seven pass strains of 0.4 being applied at a strain rate of 1 s⁻¹. The results for interpass times of 0.5 s and 5 s are presented in Figs. 4(a) and 4(b) for the plain C–Mn steel and in Figs. 5(a) and 5(b) for the Nb microalloyed steel. The horizontal broken lines in Figs. 4(a) and 4(b) denote the flow stress levels applicable to 930°C, which is 60°C above the paraequilibrium temperature of the plain C steel. When short pass intervals are employed, as shown in Fig. 4(a), the stress levels decrease by about 25% from the initial to the final pass. When the interpass time is increased to 5 s (Fig. 4(b)), a somewhat smaller stress drop of about 20% is observed. These drops can be directly associated with the volume fraction of DT ferrite formed, as will be discussed in more detail in the sections that follow.

Similar trends are evident in the isothermal flow curves
of the Nb steel displayed in Fig. 5. In this case, however, the stress levels are higher than the anticipated values because of the work hardening that is retained from pass to pass. In these tests, the same testing conditions were employed as in Fig. 4, but the temperature was decreased to 910°C, which is about 80°C above the relevant Ae₃. Here the rate of flow stress decrease (after the third pass) is slightly greater when the interpass time is 0.5 s (Fig. 5(a)) than when it is 5 s (Fig. 5(b)). The effect of interpass time on DT as well as that of the addition of Nb will be taken up in more detail below when the mean flow stress and optical microstructure results are presented and discussed.

4. Discussion

4.1. Mean Flow Stress

The mean flow stress (MFS) in rolling depends primarily on strain, strain rate, temperature of deformation, interpass time and composition. It can also provide considerable insight into the occurrence of metallurgical phenomena such as recrystallization, carbonitride precipitation and the retention of work hardening. In the present case, the MFS’s are of particular use with respect to assessing the extent to which DT ferrite is formed during a given simulation. Here the areas under the flow curves of Figs. 2 and 3 were determined by integration and the MFS’s obtained in this way are presented in the form of Boratto diagrams in Figs. 6 and 7 for the plain C and Nb steels, respectively. For reference, the paraequilibrium and orthoequilibrium Ae₃ temperatures are shown as broken vertical lines in Fig. 6 and listed in the caption to Fig. 7. As mentioned above, all the strains in these simulations were applied in the austenite region of the phase diagram.

In Fig. 6, the broken line represents the effect of temperature on the MFS in the absence of DT and DRX. These were identified experimentally by performing first pass rolling simulations to strains of ε = 0.10 (i.e., prior to the initiation of DT) at a series of decreasing temperatures. An approach similar to that described in section 3.1 was used to derive the temperature dependences of the mean flow stress. In this way, the effects of DT and DRX on the MFS were eliminated and only the effect of decreasing the temperature is being presented.

It can be seen that the MFS values increase slightly from the 1st to the 2nd pass due to strain accumulation. The longer the interpass time, the less work hardening is retained due to the effects of static recovery and recrystallization during the interval of unloading. When the interpass time is as short as 0.5 s, the MFS drops continuously after the second pass, despite the decreases in temperature. In the case of the
1 and 1.5 s intervals, the drops are smaller and there is a slight increase with decreasing temperature in the latter case. These values are well below those associated with the material undergoing straining in the absence of DT and DRX. Only when the interpass time is increased to 3 s do the MFS values begin to approach those called for by the decreases in temperature. When the last pass MFS’s of the four simulations are compared with their corresponding expected values, the shortest interpass time value displays a shortfall of about 15% as opposed to a deficit of only 5% for the 3 s interval. It is clear that as the interpass time is increased, the MFS gradually approaches its expected value.

The addition of Nb has a dramatic effect on the MFS versus inverse temperature plot, see Fig. 7. The presence of Nb in solution retards recovery between passes, with the result that there is much more strain accumulation than in the plain C steel. At the shortest interpass time, the large amount of second-pass retained work hardening is followed by appreciable MFS decreases in the succeeding passes. The broken line again represents the expected MFS behavior in the absence of DT and DRX. With respect to the 0.5 s and 1 s simulations, it can be seen that the MFS decreases significantly after the 2nd pass. In the cases of the 1.5 s and 3 s interpass intervals, the rates of the MFS increases are below those expected. This is consistent with the results obtained on the plain C steel in which shorter intervals generate greater MFS decreases than longer times.

4.2. Effect of Interpass Time

The flow curves of the isothermal rolling simulations of Figs. 4 and 5 were analyzed and the results obtained in this way are presented here as Figs. 8(a) and 8(b) for the plain C and Nb steel, respectively. It can be seen from Fig. 8(a) that there is more retained work hardening in the second pass (about a 10% MFS increase) after an interpass time of 0.5 s than when a 5 s interval was used. This is because there is less time for static recovery so that more dislocations are retained after short interpass times. Although the 2nd pass MFS value is higher in the 0.5 s simulation, the final pass MFS is lower. This is because more ferrite is formed during short interpass time rolling and there is less time for the ferrite to retransform into austenite under these conditions. This interpretation is supported by the optical microstructures presented in the next section.

A similar analysis was carried out for the Nb isothermal simulation and the results obtained are illustrated in Fig. 8(b). In this case, strain accumulation in the second pass increases the MFS by 30% in both the short and long interpass simulations. Here there is much less recovery between passes than in the case of the plain C steel. In the succeeding passes, more work hardening is retained after the 0.5 s than after the 5 s intervals, leading to higher MFS’s in the 3rd, 4th and 5th passes. Conversely, there is much more dynamic softening during straining, so that the pass 7 MFS’s are only slightly lower (0.5 s interpass time), or not perceptibly (5 s

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**Fig. 9.** Optical microstructures of the plain C steel subjected to the 930°C 7-pass, 0.5 s interpass time rolling simulation of Fig. 4. The samples were quenched: a) immediately after straining, b) after holding for 8 s, and c) after holding for 15 s. Light regions are ferrite while dark regions are martensite (prior austenite). The DT ferrite is seen to retransform gradually into austenite during holding.

**Fig. 10.** Optical microstructures of the Nb steel subjected to the 910°C 7-pass, 0.5 s interpass time rolling simulation of Fig. 5. Here the quenching conditions of Fig. 9 were employed: a) immediately after straining, b) after holding for 8 s, and c) after holding for 15 s. There is less reversion to austenite than in the plain C steel.
interpass time) than the pass 6 MFS’s. This indicates that more ferrite is being produced (and retained) when the interpass times are short than when they are long.

4.3. Retransformation of DT Ferrite

In torsion simulations, the applied strain increases linearly from zero along the axis to a maximum at the surface of the sample. For this reason, the volume fraction of DT ferrite that is formed increases from the axis to the surface. The optical micrographs presented in Figs. 9 and 10 were all taken 0.3 mm from the surface. Those of Fig. 9 for the plain C steel are from the isothermal simulation of Fig. 4 using pass intervals of 0.5 s and represent the microstructure: i) right after the last pass (Fig. 9(a)), ii) after 8 s of holding (Fig. 9(b)), and iii) after 15 s of holding (Fig. 9(c)). The martensite appears dark while the ferrite is light. The microstructure of Fig. 9(a) consists essentially of ferrite (about 80–90%), produced by applying an accumulated strain of $\varepsilon = 2.8$. When the sample was held for 8 s before quenching, Fig. 9(b), the volume fraction of DT ferrite decreased to approximately 50–60%. Finally, when the holding time was increased to 15 s, this fraction was further reduced to 10–20%. This confirms that when the holding time is increased, more and more ferrite retransforms to austenite. These microstructures support the interpretation of the MFS plots proposed above.

The optical microstructures relating to the isothermal simulations performed on the Nb steel (Fig. 5) are illustrated in Fig. 10. These samples were subjected to the 910°C 7-pass, 0.5 s interpass time rolling simulation of Fig. 5. Here the samples were quenched: i) right after straining (Fig. 10(a)), ii) after 8 s of holding (Fig. 10(b)), and, iii) after 15 s of holding (Fig. 10(c)). It is of interest that there is little or no difference in the volume fraction of ferrite. This is in agreement with the MFS observations of the isothermal simulation showing the convergence of values in the last two passes.

In the plain C steel, the decrease in the proportion of ferrite formed after holding for 15 s is caused by the reverse transformation of DT ferrite back into austenite. The nucleation of DT ferrite during rolling is known to occur in a displacive manner while the reverse transformation takes place by a diffusional mechanism.20,21) In this way, the amount of retransformation will depend on the length of the interpass interval as well as the composition. It can be seen from the results obtained here that the presence of Nb in solution significantly retards both the forward as well as the backward transformation rates by a solute drag effect.22–24)

The presence of DT ferrite in Figs. 9 and 10 was confirmed using EBSD techniques of phase identification, as illustrated here in Fig. 11. Inverse pole figure (IPF) plots related to the orientations of the ferrite formed are shown on the left side of the diagram while the phase distribution is displayed on the right-hand side. Here the red regions are ferrite while the black ones are martensite (prior austenite). EBSD images associated with the C–Mn samples of Figs. 10(a) (directly quenched) and 10(c) (15 s of holding) are displayed in Figs. 11(a) and 11(b), respectively. These samples were subjected to the 930°C 7-pass, 0.5 s interpass time rolling simulation of Figs. 4 and 5. The C–Mn samples were quenched: Fig. 11(a) - immediately after straining, and Fig. 11(b) - after holding for 15 s. The Nb steel sample of Fig. 11(c) was quenched after 15 s of holding. In the right hand figures, the ferrite phase is red and the prior austenite is black.

Fig. 11. EBSD micrographs of transverse cross-sections of the C–Mn and Nb steels subjected to the 7-pass, 0.5 s interpass time rolling simulation of Figs. 4 and 5. The C–Mn samples were quenched: Fig. 11(a) - immediately after straining, and Fig. 11(b) - after holding for 15 s. The Nb steel sample of Fig. 11(c) was quenched after 15 s of holding. In the right hand figures, the ferrite phase is red and the prior austenite is black.
rolling simulation of Fig. 4. Here the volume fraction of ferrite is significantly higher in the directly quenched sample than in the sample held for 15 s. This confirms the interpretation of Fig. 9 where the decrease in ferrite volume fraction was attributed to the reverse transformation.

Similar analyses were performed on the Nb steel. The phases in the sample of Fig. 10(c) (15 s of holding) are positively identified here as Fig. 11(c). This sample was subjected to the 7-pass, 0.5 s interpass time rolling simulation at 910°C of Fig. 5. Here again the formation of DT ferrite is confirmed, although there was little difference in ferrite volume fraction between the directly quenched sample (not shown here) and the one held for 15 s after deformation. This result supports the interpretation of the optical micrographs of Fig. 10 according to which the presence of Nb in solid solution was concluded to retard the reverse transformation. This will be shown in more detail in a future paper, where measurements of the volume fractions of DT ferrite formed in each pass and subsequently lost by retransformation will be presented.

4.4. Industrial Implications

The present results indicate that DT takes place during each pass of strip rolling and that the interpass interval affects the volume fraction of ferrite formed and retained above the \(\Delta Ae_2\). This then explains why the rolling load and MFS do not increase with decreasing temperature in strip mills. The present results are also consistent with the observations of earlier researchers\(^{2,3,10,23-28}\) on dynamic transformation. Thus it appears that the lack of increase in rolling load cannot be attributed solely to DRX. The dilatation associated with ferrite formation can also be seen in this way to be responsible for the increase in volume flow rate (but not mass flow rate) as a bar passes through a mill.

A further application of the present work involves the carbon partitioning that takes place during ferrite formation. By moving into the prior austenite, it can be responsible for the formation of undesirable amounts of martensite on the runout table. The present observations also indicate that the introduction of water cooling directly after the last mill stand (i.e. after reduced delays) is likely to reduce the amount of retransformation that takes place and affect the mechanical properties in this way.

5. Conclusions

(1) The rolling simulations carried out on the plain C steel reveal that the MFS decreases significantly when short interpass times are used and increases less rapidly than expected when this interval is increased in length. These trends are directly attributable to the dynamic transformation of austenite to ferrite within the austenite phase field, i.e. above the \(\Delta Ae_2\). The microstructures observed using optical microscopy indicate that more DT ferrite forms and is retained when the interpass intervals are shorter than when longer intervals are used.

(2) The simulations carried out on the Nb steel reveal that the MFS decreases substantially from the 2nd to the succeeding passes when short interpass times are used and displays a low rate of increase when this interval is increased in length. These trends can be directly associated with the formation of DT ferrite from pass to pass during the simulation. It appears that the shorter interpass times both produce more DT ferrite and permit less reversion to austenite than the longer times. The optical micrographs indicate that ferrite forms during straining and that most of it are retained during unloading.

(3) Comparison of the behavior of the plain C and Nb steels reveals that Nb addition retards both the forward as well as the reverse transformation. As a result, more DT ferrite is formed in the plain C steel than in the Nb steel.

(4) Most of the DT ferrite formed in the plain C steel is able to retransform into austenite after holding for 15 s. By contrast, the reverse transformation is severely retarded in the Nb steel. As a result, there is little or no retransformation after 15 s of holding under the present conditions.

Acknowledgements

The authors acknowledge with gratitude funding received from the McGill Engineering Doctoral Award (MEDA) program and the Natural Sciences and Engineering Research Council of Canada.

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