Design of Microalloyed Steel Hot Rolling Schedules by Torsion Testing: Average Schedule vs. Real Schedule

Jessica CALVO,1) Laurie COLLINS2) and Stephen YUE1)

1) Department of Mining and Materials Engineering, McGill University, Montreal, Quebec, Canada.
E-mail: jessica.calvo@upc.edu 2) Formerly IPSCO Research & Development. Now at Evraz Inc NA, Regina, Saskatchewan, Canada.

Received on April 22, 2010; accepted on May 27, 2010

The hot torsion simulator has been extensively used as a means to understand the microstructure evolution of different steel grades during hot rolling. The test is suitable to simulate ‘real’ industrial schedules as well as schedules designed to obtain information regarding the intrinsic properties of the materials. For example, it is common to apply ‘average’ schedules, in which deformation per pass, interpass time, strain rate and cooling rate are kept constant, to determine the characteristic temperatures \( A_r^3 \), \( A_r^1 \), and \( T_{nr} \) (start and finish of the austenite transformation and no recrystallization temperatures) of steels. In this work, both a ‘real’ schedule simulating a rolling schedule in a reversing mill and an ‘average’ schedule were applied to a series of Ti and Nb microalloyed steels. In general, the steels exhibited somewhat different behaviours for the different thermomechanical schedules, e.g. the pancaking temperature region is easily detectable after an ‘average’ schedule, while for the ‘real’ schedule some softening can be detected in the pancaking region, which is strongly dependent on the strain and interpass time. Moreover, the paper analyzes a new approach to stress-strain curves, which is used to better understand the sequence of events which take place during rolling and their dependence on rolling parameters.

KEY WORDS: Nb microalloyed steel; hot torsion simulator; hot rolling schedules; pancaking.

1. Introduction

There are two main ways in which hot torsion testing can be used to help design the hot rolling schedules for steel: (i) the determination of the three critical temperatures of hot rolling, i.e. the temperature of no recrystallization \( T_{nr} \), and the start and finish of the austenite transformation \( A_r^3 \) and \( A_r^1 \), respectively; (ii) the simulation of a specific (i.e. ‘real’) industrial hot rolling schedule in order to assess the pass to pass evolution of the microstructure, culminating in generating the final, as-hot rolled structure. As well, for this test, knowledge of the pass to pass flow behaviour is used to determine rolling loads.

Using the critical temperature values generated in the average schedule test, the behaviour of the steel subjected to a real schedule can be rationalised, and ‘new’ schedules can be suggested, if desired.

In this paper, the effectiveness of using an average schedule as an aid to evaluating the real schedule is examined by comparing the results of an average schedule with that of a real schedule. The paper concludes by giving recommendations concerning the design of an average schedule.

2. Materials and Experimental Methods

For this study six experimental low C HSLA steels, whose composition is given in Table 1, were cast and rolled in the CANMET Materials Technology Laboratory in Ottawa, Canada. These steels, which have been used for other investigations, represent in fact different families of microalloyed steels. All six steels have similar carbon and titanium contents (about 0.04% C and 0.02% Ti), and the manganese is also kept approximately constant for all steels. However, the C–Ti and Mo steels are Ti microalloyed steels, whereas the other steels, in addition to Ti, also contain Nb as a microalloying element. The difference between the C–Ti and Mo steels is that the latter steel contains 0.3% Mo in addition to the composition of the C–Ti. Moreover, the effect of Nb as a microalloying addition can be evaluated through the low Nb and high Nb steels which
contain 0.03 and 0.07% Nb respectively. Finally, Cu and Si steels also contain 0.07% Nb but with an additional 0.38% Cu for the former and 0.32% Si for the latter. Both the Si and Cu steels can be compared with the high-Nb steel individually to determine the effects of Si and Cu.

Both compression and torsion tests were carried out to the steels in Table 1. The compression tests were performed in an MTS machine equipped with a radiant furnace mounted on the columns of its load frame. These tests comprised a reheating of the specimens to 1200°C to dissolve the precipitates; according to FactSage\(^{14}\) calculations, this temperature is above the equilibrium formation temperature for Nb(C,N), which can be found in Table 1.\(^{15}\) The reheating was followed by a cooling down to 1000°C where a single hit isothermal test was applied to a strain of 0.7 at a strain rate of 0.1 s\(^{-1}\). These isothermal compression tests were carried out to determine basic hot deformation characteristics, i.e. dynamic recrystallization and precipitation, of the different alloys.

As well, torsion tests were carried out using a servo-hydraulic, computer controlled MTS machine equipped with a radiant furnace. Two different thermomechanical schedules, which are shown schematically in Figs. 1(a) and 1(b), were applied to the samples. The first schedule, ‘average’ schedule in Fig. 1(a), is based on a schedule for a Steckel mill, which is essentially a reversing mill. The main parameter that is affected by mill specificity is the interpass time, which here is taken as 30 s and pass strains for this type of mill are around 0.2, strain rate was 1 s\(^{-1}\) and the cooling rate was 1°C/s throughout the whole deformation schedule. This schedule was designed to generate values for the three critical temperatures of hot rolling, hence the test was conducted down to below the austenite transformation temperature, in particular the schedule consisted in 21 passes which started at 1150°C and ended at 550°C.

A second schedule, the ‘real’ simulation of the rolling schedule in Fig. 1(b), was designed based on a real industrial schedule. This schedule is obviously more complex, because strains per pass and interpass times are different for each pass. In the simulation, strain and interpass times were the same as in the industrial schedule, but as a simplification for testing, the cooling rates were kept constant within roughing (1°C/s) and finishing (0.5°C/s). The temperatures at each pass were close to what is performed in the mill. The other difference in this schedule compared to the ‘average’ one is that there is a considerable time between roughing and finishing in order to finish in the pancaking region.

From the torsion simulation the complete flow curves for all passes can be represented together as shown in Fig. 2(a) for the high Nb steel after an ‘average’ schedule. This representation allows the identification of different regions which indicate an evolution of the microstructure. Another conventional way to identify the main microstructural changes taking place during rolling is through the representation of the mean flow stress (MFS) for each pass as a function of inverse absolute temperature, which is shown in Fig. 2(b). MFS was calculated according to Eq. (1)\(^{15}\):

\[
\sigma_{eq} = \frac{1}{\varepsilon_b - \varepsilon_a} \sum_{i=1}^{n} \frac{\sigma_{i,b} + \sigma_{i,a}}{2} \times (\varepsilon_{i+1} - \varepsilon_i) \quad (1)
\]

The representation in Fig. 2(b) shows, in a greater detail than Fig. 2(a), the different stages. For example, at the beginning of the curve, region I, an increase of the stress as the temperature decreases is evident. Moreover, as the temperature is further decreased there is a change in the slope, region II, indicating a more pronounced increase of the

| C  | Mn  | Si  | Ti  | Mo  | Cu  | Nb  | Al  | N   | Nb(C,N)\(^{1}\) (°C) |
|----|-----|-----|-----|-----|-----|-----|-----|-----|----------------------|
| C-Ti | 0.041 | 1.47 | 0.06 | 0.020 | 0.030 | 0.0086 | -   |
| Mo  | 0.041 | 1.44 | 0.06 | 0.017 | 0.32  | 0.023 | 0.0091 | -   |
| Low Nb | 0.042 | 1.47 | 0.07 | 0.020 | 0.30  | 0.030 | 0.031 | 0.0088 | 1083 |
| High Nb | 0.039 | 1.51 | 0.08 | 0.018 | 0.28  | 0.068 | 0.011 | 0.0095 | 1189 |
| Cu  | 0.049 | 1.59 | 0.07 | 0.020 | 0.28  | 0.38  | 0.069 | 0.040 | 0.0090 | 1190 |
| Si  | 0.042 | 1.50 | 0.32 | 0.019 | 0.30  | 0.068 | 0.024 | 0.0080 | 1179 |

\(^{1}\) Formation start temperature of Nb(C,N) calculated with FactSage\(^{14}\).
stress with decreasing temperatures. The temperature at which the slope of the MFS vs. 1000/T curve changes corresponds to the no recrystallization temperature ($T_{nr}$). From a metallurgical point of view, static recrystallization (S-REX) takes place between passes at temperatures above $T_{nr}$. However, this softening mechanism is impaired when precipitation appears, at temperatures below $T_{nr}$, because precipitates pin austenite grain boundaries. The decrease of the stress which takes place at relatively low temperatures, region III, is related to the austenite to ferrite transformation and region IV corresponds to the end of the transformation and work hardening of the new phases.

3. Results

3.1. Isothermal Flow Curves at 1000°C

As noted above, isothermal tests were performed at 1000°C to overview austenite hot deformation behaviour of these steels. As shown in Fig. 3, the steels behave similarly up to strains of about 0.25; beyond this strain, 3 ‘families’ of steels can be identified.

3.1.1. C–Ti and Mo Steels

Although the steady state flow stress of the Mo steel is significantly greater than that of the C–Ti steel, these steels have similar well defined peak stresses suggesting relatively rapid dynamic recrystallization kinetics. All the other steels show much less well defined peaks, probably because the effects of any dynamic recrystallization are being offset by extensive dynamic precipitation of Nb(C, N).

3.1.2. High Nb and Cu Steels

The flow curves at 1000°C are exactly the same, indicating that there is no effect of Cu under this test condition. Moreover, the steady state flow stresses are higher than those of the C–Ti and Mo steels, probably due to dynamic precipitation of Nb(C, N) restricting restoration and/or adding a precipitation strengthening component.

3.1.3. Low Nb and Si Steels

The flow curves of these two steels have identical steady state flow stresses, even though the Si steel has a higher Nb level and a higher Si level, both of which may be expected to increase the flow strength. It is true that there is a significant solid solution strengthening effect of Si in ferrite but it is likely that this benefit is not transferred to austenite at 1000°C, since solid solution effects generally decrease with increasing temperature. As mentioned above, the influence of Nb in increasing the steady state flow stress of the family Sec. 3.1.2. steels is likely due to precipitation characteristics. Therefore, it is possible that the Si addition has increased the level of precipitation, as suggested by Calvo et al., leading to the Si steel having the same ‘effective’ Nb concentration as that of the low Nb steel.

From these curves, three steel families have been identified, with respect to hot rolling behavior. It is thus anticipated that three characteristic behaviours will emerge for the average and real schedules.

3.2. Average Schedule

The usual way to analyse any multipass experiment, as mentioned above, is through the MFS vs. 1000/T curves, as shown in Fig. 4 for all the steels. At very low temperatures, the austenite transformations are taking place, and these will be referred to later on. Before this, the behaviour in the austenite region will be described.

In general, the MFS values of the austenite are very close to each other at high temperatures, since the influence of solid solutions is relatively low in austenite. These MFS values deviate in austenite at lower temperatures mainly due to differences in recrystallization kinetics. These kinetics are embodied in the $T_{nr}$ values, listed in Table 2. Thus,
the C–Ti steel, which has the lowest $T_{nr}$ value, exhibits the lowest MFS for a given temperature because the austenite has not workhardened as much as the other steels. In this diagram, the Mo and C–Ti steels do not seem to behave as closely as in the isothermal test, but can still be grouped together as being the steels with the lowest $T_{nr}$ values. The other steels are closely grouped, but can still be grouped as in the initial isothermal analysis, again on the basis of $T_{nr}$.

With regard to austenite transformation behaviour, the start and finish of the transformation are similar for all steels with the exception of the Cu steel, which appears to have a much lower transformation temperatures.

### 3.3. Real Schedules

As for the average schedules, the analysis begins with the MFS vs temperature plots in Figure 5. Note that the real hot rolling schedule finishes at 1 000/T values of about 0.875, whereas the average schedule reaches values >1. One consequence of this is that the MFS differences cover a lower range; hence the range of the $y$-axis has been reduced to capture more detail. Another consequence is that the schedule finishes more or less at the $T_{nr}$'s of both the Mo and C–Ti steels, hence these steels exhibit no $T_{nr}$ in the real schedule, and can be grouped together. Because the temperature gap between roughing and finishing falls in the region of the $T_{nr}$'s of the other steels, the $T_{nr}$'s cannot be precisely determined, the steels cannot be delineated on the basis of $T_{nr}$, and therefore appear to be one family. Moreover, the transformation behaviour due to the real schedule is not captured.

One of the problems of interpreting the results of a real schedule simulation in terms of microstructural evolution is that the process variables of each pass are different. For example, in the MFS plot (Fig. 5) the flow stress of the last pass tends to be much lower than the penultimate pass in all cases, even though the temperature has decreased. Looking at the rolling schedule, one immediate observation is that the strain of the final pass is the lowest out of all the passes, and this will lead to a different MFS since the area under a typical (e.g. power law) flow curve is not linearly related to the flow strain.

One way around this problem of varying pass strain is to look at the pass to pass flow curve evolution of the real schedule. The usual way of plotting this is shown in Fig. 2(a), and at least has the merit of clearly indicating the $T_{nr}$ if one exists. An alternative way of presenting flow curves is as shown, for all the steels, in Fig. 6. There is more detail in this presentation simply because the full flow curve for each pass is directly compared. However, the differences in strain at each pass can be somewhat distracting. In this schedule, for these steels, there is no sign of dynamic recrystallization, hence ‘cropped’ versions of these graphs can be plotted, as shown in Fig. 7. These have the benefit of avoiding the distraction of the differing strains per pass. Moreover, in the absence of dynamic recrystallization, the initial part of the flow curve is enough to indicate the pass by pass microstructural evolution of the steels.

From either the ‘full strain’ diagrams or the ‘cropped’ ones, it is clear that the yield strength of the final pass is lower than the yield stress of the penultimate pass. The interpass time before the final pass is much longer than for the other passes, which suggests that some restoration has occurred, even for the Nb bearing steels which have been subjected to several passes below the $T_{nr}$. With increasing strain, the flow stress reaches, or exceeds the flow stress of the penultimate pass, due mainly to the lower temperature of the final pass. Thus, in this case, the low MFS values of the final pass in each steel seem to be a true indication of the microstructural condition (i.e. static restoration has taken place prior to the final pass) as opposed to being an artefact of the MFS calculation.

It is also remarkable from Figs. 6 and 7 that pass 8, which is the first pass of the finishing stage, presents similar yield strength to the roughing passes. This is particularly interesting for the steels containing 0.7% Nb, i.e. high Nb, Cu and Si steels, because these steels present a high $T_{nr}$ (see Table 2). However, the big change in yielding appears in pass 9, which is carried out at slightly lower temperatures than the previous one. This seems to indicate that pass 8 is
a transition pass which provides ‘fresh’ dislocations to nucleate precipitates. The critical pass is retarded as Nb contents are decreased, i.e. the big change in the yield strength for the low Nb steel appears from pass 9 to pass 10 and for the Mo steel the biggest change appears between pass 11 and 12. No significant increment in the yield strength is detected for the C–Ti steel. Given the dependence of the variation in the yield strength with the Nb contents, it seems reasonable to think that these increases of the yield strength, after a certain pass, i.e. after a certain temperature, are an indication of $T_{nr}$.

4. Discussion

From the analyses of the average schedule and real schedules, it is clear that both types of test need to be conducted in order to generate, in the case of the average schedule, the material properties relevant to hot rolling, and, in the case of the real schedule, to determine the microstructural evolution during hot rolling. The theme of this discussion revolves around the design of the average schedule in order to generate values that are applicable to the real schedule.

One problem, which has been considered elsewhere (7), is the transformation behaviour after hot rolling. The average schedule approach has two limitations: (i) multistage deformation is executed to temperatures lower than the transformation temperatures in order to reveal the required temperatures. Since most real schedules finish well above the $A_{13}$, the transformation temperatures from the average schedule may not be applicable to the real schedule. However, in the case where the real schedules incorporate considerable pancaking (e.g. strains of > 1), higher levels of pancaking may not significantly affect the transformation behaviour; (ii) the resolution of the transformation temperatures is determined by the interpass time. One way of getting around this problem in the average schedule is to stop the multipass deformation schedule at the finish rolling temperature in the real schedule, and then execute a continuous cooling torsion pass to encompass the transformation temperatures at
a slow strain rate. Thus the amount of deformation leading to the transformation is much reduced, and the temperature resolution is continuous as opposed to incremental.

Another somewhat unexpected issue arises, in this real rolling schedule, from the long interpass time from the end of roughing to the start of finishing, used to ensure that the steel drops below the $T_{nr}$. An analysis of the effect of this on the real schedule is shown in Fig. 8. Here, the roughing and finishing stages are being analyzed on the basis of the effect of temperature on flow strength. A simple approach is to consider the effect of temperature on flow strength to be approximately linear. Thus, if there is a difference between the deformation mechanisms of roughing and finishing, this will be manifested in a change of the slope of the MFS vs. temperature of each rolling regime. As expected, the Nb containing steels exhibit a much steeper slope in finishing, than roughing, and the C–Ti and Mo steels exhibit similar slopes. What is unexpected, in the C–Ti and Mo steels, is that the finishing MFS values fall below those predicted by extrapolating the MFS vs. temperature relation to the 1000/T values of the finishing stage. At present, the reason for this is not entirely clear. The immediate possibility is grain coarsening, although these alloys contain Ti, which should prevent this. However, restoration was also observed in all steels in the interpass before the final pass, so perhaps grain coarsening after roughing is possible, at least for the C–Ti and Mo steels.

Whatever the reason for this phenomenon, with regard to design of an average schedule, perhaps a more appropriate average schedule would include the real interpass time between roughing and finishing.

5. Conclusions

When using torsion testing, the average schedule approach combined with a real simulation is an excellent combination with which to examine and analyse an industrial steel hot rolling schedule. In order to make the average schedule more appropriate, long interpass times between roughing and finishing should be incorporated. Once the finishing temperature of the real schedule has been reached in the average schedule a continuous cooling, slow strain...
rate torsion test should be conducted to determine the transformation temperatures. This will minimise the influence of the average schedule deformation and improve the temperature resolution of the transformation temperatures. However, average schedules should also be performed down to the transformation temperatures in order to capture low temperature $T_n$’s of ‘weakly’ microalloyed steels.

Acknowledgements
The authors thank NSERC for the financial support.

REFERENCES
1) D. Q. Bai, S. Yue, W. P. Sun and J. J. Jonas: Metall. Mater. Trans. A, 24A (1993), 2151.
2) M. I. Vega, S. F. Medina, M. Chapa and A. Quispe: ISIJ Int., 39 (1999), 1304.
3) M. Gómez, S. F. Medina, A. Quispe and P. Valles: ISIJ Int., 42 (2002), 423.
4) L. P. Karjalainen, T. M. Maccagno and J. J. Jonas: ISIJ Int., 35 (1995), 1523.
5) L. N. Pussegoda and J. J. Jonas: ISIJ Int., 31 (1991), 423.
6) A. Najafi-Zadeh, S. Yue and J. J. Jonas: ISIJ Int., 32 (1992), 213.
7) F. Siciliano, Jr. and J. J. Jonas: Metall. Mater. Trans. A, 31A (2000), 511.
8) T. M. Maccagno, J. J. Jonas, S. Yue, B. J. McCrady, R. Slobodian and D. Deeks: ISIJ Int., 34 (1994), 917.
9) L. E. Collins: Int. Symposium Niobium 2001, Minerals, Metals and Materials Society (TMS), Warrendale, PA, (2001), 527.
10) C. Roucoules, P. D. Hodgson, S. Yue and J. J. Jonas: Metall. Mater. Trans. A, 25A (1994), 389.
11) J. Calvo, A. Elwazri and S. Yue: 3rd Int. Conf. on Thermomechanical Processing of Steels, AIM, Italy, (2008).
12) J. Calvo, A. Elwazri, D. Bai and S. Yue: 7th Int. Pipeline Conf., Calgary, (2008).
13) J. Calvo, A. Fatehi, A. Elwazri and S. Yue: Rio Pipeline 2007, Rio de Janeiro, (2007).
14) FSStel database, www.factsage.com
15) J. Calvo, I.-H. Jung, A. M. Elwazri, D. Bai and S. Yue: Mater. Sci. Eng. A, 520 (2009), 90.
16) T. Gladman: The Physical Metallurgy of Microalloyed Steels, The Institute of Metals, London, (1997).
17) X. Lou, J. Calvo, A. M. Elwazri, D. Q. Bai and S. Yue: Conf. of Metallurgists, Winnipeg, (2008).