

Jessica Calvo^{a, b}, Laurie Collins^c, Stephen Yue^d

^aDepartment of Materials Science and Metallurgical Engineering, Universitat Politècnica de Catalunya, Barcelona, Spain

^bCTM Centre Tecnològic, Manresa, Spain

^cEvrax Inc NA, Regina, Canada

^dDepartment of Mining and Materials Engineering, McGill University, Montreal, Canada

Classical controlled rolling of low C steels microalloyed with Ti and Mo

Two low C steels, one microalloyed with Ti and the other with Ti and Mo were subjected to a classical controlled rolling schedule using torsion testing. The specific torsion schedules were a simulation of an industrial schedule adapted to the characteristic transformation temperatures of each steel. It is shown that such steels can develop a pancaked austenite, despite the fact that the alloys do not contain Nb. However, compared to Nb bearing steels, restoration between deformation passes is relatively rapid in the pancaking region. In order to reduce the loss of pancaking, the temperatures of the finishing rolling schedule were shifted to lower temperatures. This appeared to greatly reduce the restoration rates, but the transformation in the two phase region occurred towards the end of the final deformation pass. When comparing the two steels, it could be detected that Mo increases the temperature range for pancaking and reduces the restoration rates between deformation passes in the pancaking region.

Keywords: Ti microalloyed steel; Hot torsion simulator; Pancaking; Thermomechanical processing; Hot rolling schedules

1. Introduction

Thermo-mechanical processing (TMP) and thermo-mechanical controlled processing (TMCP) aim to improve the mechanical properties of hot rolled steel products by control of the microstructure throughout the process [1, 2]. In the case of high strength low alloy (HSLA) steels a fine grained ferritic microstructure can be obtained when transformation is preceded by a pancaking of the austenite [3]. It is generally accepted that precipitation is required for recrystallization to be stopped, promoting strain accumulation in subsequent deformation stages and leading to a pancaked austenite microstructure [4, 5]. More recently, work indicates that dynamic precipitation, specifically, is required to stop recrystallization [6, 7]. Invariably, Nb additions are required to generate the required intensity of precipitation. The temperature at which recrystallization is stopped by precipitation is known as the non-recrystallization temperature (T_{nr}) and it can be pinpointed experimentally by a multipass torsion schedule. Usually, an average torsion schedule, in which each pass has the same strain, strain rate, interpass time and cooling rate, is applied to determine T_{nr} . It is known that low C steels microalloyed with

only Ti as a microalloying addition exhibited very strong indications of recrystallization retardation when they were subjected to average torsion schedules [6, 8–10]. This effect of Ti on recrystallization retardation is evident even for hypostoichiometric steels, i.e. with Ti/N ratios below 3.42 [9, 11]. Regardless of the mechanism, the recrystallization behaviour suggests that classical controlled rolling is possible and thus, hot rolling schedules, consisting of a roughing stage followed by a finishing stage to promote austenite pancaking, can be designed to produce fine grained ferritic steels. The purpose of this work is to investigate the possibility of pancaking in these steels which do not contain Nb. In this case, the pancaking attempt will be done by adapting an industrial schedule to the characteristic temperatures, i.e. T_{nr} , equilibrium transformation temperature (A_{e3}) and continuous cooling transformation temperature (A_{r3}), of a Ti and a Ti + Mo steels.

2. Experimental procedure

For this study two experimental low C HSLA steels microalloyed with Ti were analyzed. The steels were cast and rolled in the CANMET Materials Technology Laboratory in Ottawa, Canada, and their chemical composition is specified in Table 1. Both steels, which have been used for other investigations [6–8, 12–14], have similar carbon and titanium contents (about 0.04 wt.% C and 0.02 wt.% Ti), and the manganese is also kept approximately constant. The difference between the Ti and Ti + Mo steels is that the latter steel contains 0.3 wt.% Mo in addition to the composition of the Ti steel. Mo is usually added to increase the hardenability and promote the formation of non-polygonal ferrite during accelerated cooling in TMCP. Moreover, this element can influence the precipitation of TiN particles [15].

First of all, continuous cooling torsion (CCT) tests were performed to approach the T_{nr} temperature of the steels. With this purpose, an average 21 pass schedule, with a constant strain of 0.2 per pass, strain rate of 1 s^{-1} and interpass times of 30 s, was applied to the samples on cooling. The *MFS* (mean flow stress), calculated for each pass according to Eq. (1) [16], was plotted vs. $1000/T$. Further details of the thermomechanical schedules and testing conditions were reported by Calvo et al. [8].

$$\bar{\sigma}_{eq} = \frac{1}{\varepsilon_b - \varepsilon_a} \sum_{i=a}^b \frac{\sigma_{i+1} + \sigma_i}{2} \times (\varepsilon_{i+1} - \varepsilon_i) \quad (1)$$

Table 1. Chemical composition of the steels (wt.%).

	C	Mn	Si	Ti	Mo	Al	N
Ti steel	0.041	1.47	0.06	0.020		0.030	0.0086
Ti + Mo steel	0.041	1.44	0.06	0.017	0.32	0.023	0.0091

In Eq. (1), $\bar{\sigma}_{eq}$ is the mean flow stress and $\varepsilon_b - \varepsilon_a$ is the equivalent strain of the pass of interest.

After the average schedule, other multipass torsion schedules were applied to the steels in the current investigation, based on the industrial hot deformation schedule in a reversing mill for Nb-bearing HSLA steels [6]. One of the differences between Nb microalloyed steels and the steels in Table 1 is the value of the T_{nr} , i. e. depending on the Nb content, Nb microalloyed steels can exhibit T_{nr} above 1000 °C, whereas the Ti microalloyed steels exhibit T_{nr} around 850 °C to 900 °C [9]. These T_{nr} temperatures lie significantly below the finishing stage of the industrial schedule for Nb microalloyed steels, so the existing industrial schedule was redesigned for each steel simply by shifting the roughing stage to lower temperatures, enabling the finishing stage to begin below the corresponding T_{nr} obtained from the CCT tests. In Fig. 1, the characteristics of the thermomechanical cycles followed by the samples of Ti and Ti + Mo steels in this study are represented by the black line. In the same figure, the current basic cycle is compared to the characteristics of the industrial thermomechanical cycle previously applied to HSLA steels [6], which is represented in grey. Thermomechanical cycles have 13 deformation passes, 7 in the roughing region (above T_{nr}) and 6 in the finishing region (below T_{nr}). In the industrial schedule for Nb microalloyed steels, the roughing stage starts at 1150 °C and finishes at 1070 °C and the finishing stage starts, after a long interpass time, at 970 °C and finishes at 840 °C, above the A_{r3} transformation temperature of the steels. This basic industrial schedule was adapted for the Ti and Ti + Mo steels with regard to their T_{nr} temperatures, obtained from the average CCT tests described previously. The adapted schedules will be characterized by the start and finishing temperatures of the roughing stage (T_{Rs} and

T_{Rf}), and the start and finish temperature of the finishing stage (T_{Fs} and T_{Ff}). Cooling rates had to be modified in order to have enough time during the finishing stage to carry out the 6 deformation passes with their corresponding interpass times (industrial conditions were maintained for interpass times) within the temperature range defined by the T_{nr} and A_{r3} temperatures of the steels. In fact, two different cooling rates during the finishing stage were applied, leading to two different T_{Ff} temperatures for each steel.

In addition to the schedules simulating a controlled rolling of the steels, following the principle of continuous cooling compression test [8], a continuous cooling continuous torsion (CCCT) test was applied in this work to detect the transformation temperatures from austenite to ferrite after each given thermomechanical cycle. This test consisted of a single deformation pass, which was applied at a strain rate of 0.0015 s⁻¹ during cooling at a cooling rate of 0.5 K · s⁻¹. The characteristics of the CCCT test are also detailed in Fig. 1.

3. Results

3.1. Average schedule

In Fig. 2, the MFS vs. $1000/T$ curves for the Ti and Ti + Mo steels, obtained by applying an average rolling schedule, are represented. The T_{nr} temperature can be determined as the point where the slope of the curve changes in the high temperature range. At temperatures above T_{nr} the slope is lower, because S-REX (static recrystallization) is taking place between passes. At temperatures below T_{nr} , recrystallization is partially inhibited between passes, because precipi-

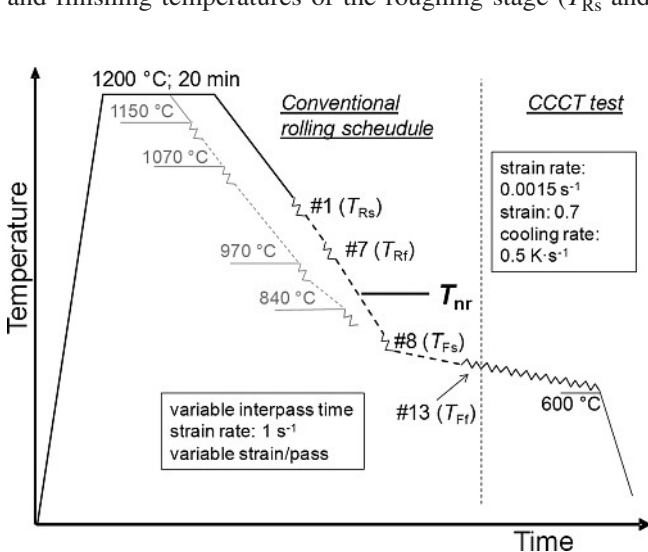


Fig. 1. Thermomechanical schedule of the simulation of the hot rolling schedule and the CCCT test.

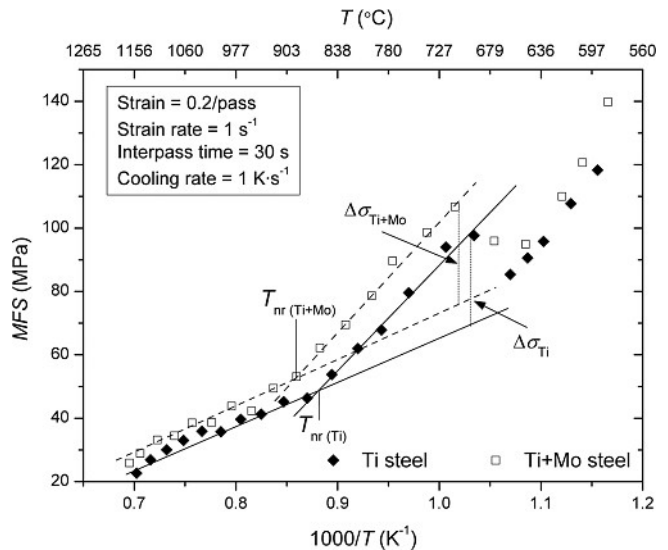


Fig. 2. MFS vs. $1000/T$ for the Ti and Ti + Mo steel, illustrating the T_{nr} and residual stress determination.

pitiation starts taking place, pinning the austenite grain boundaries. This translates into an increase in the slope of the curve and a stress accumulation, $\Delta\sigma$, which is around 30 MPa for both steels. This stress accumulation is bigger than the values reported in the literature for similar compositions and testing conditions and comparable to those of slightly hyperstoichiometric steels [9].

Some other changes in the slope of the curve are found in the low temperature range, which can be associated with the austenite to ferrite transformation. However, since there is a gap of 30 °C between each point of the graph, it is considered that the CCT test does not offer enough resolution to pinpoint A_{r3} and A_{r1} from CCT tests.

The T_{nr} temperatures obtained from the average schedule CCT tests and the continuous cooling austenite to ferrite transformation temperatures, A_{r3} , obtained from a continuous cooling compression (CCC) test in previous work [8], are specified in Table 2. Given the characteristics of CCC tests, the values of A_{r3} in the table are equivalent to what other authors have described as A_{r3}^d (A_{r3} temperatures obtained under dynamic conditions to encompass the effect of deformation on transformation kinetics) [17]. The equilibrium austenite to ferrite transformation temperatures, A_{e3} , calculated using the thermodynamic software FactSage® [8] are also included in the table. These temperatures will be the basis to design the adapted pancaking schedules for the Ti and Ti + Mo steels.

3.2. Adapted industrial schedules

According to the characteristic transformation temperatures in Table 2, the industrial pancaking schedules for Nb microalloyed steels were adapted to the Ti and Ti + Mo steels by shifting the roughing and finishing stages to lower temperatures, in order to situate the T_{nr} temperatures within the interpass period between the two stages. In the adapted schedules, the strains and interpass times were kept identical to the ones in the industrial schedule. However, the cooling rates in the finishing stage had to be slowed down because the gap between the T_{nr} and A_{r3} of the strained austenite, in which finishing rolling should be applied, was less than 100 °C for the Ti steel and slightly higher for the Ti + Mo steel. Two different schedules were applied for each steel, (i) high finish temperature of the finishing stage

and (ii) low finish temperature of the finishing stage. The second schedules were designed to avoid the excessive softening observed before the last finishing pass. The temperatures which define the different adapted schedules, i.e. T_{Rs} , T_{Rf} , T_{Fs} and T_{Ff} , are specified in Table 3.

The flow curves of each pass of the torsion schedule for both steels are shown together in Fig. 3. It can be seen that there are three families of curves, which have been properly identified in Fig. 3a. One set is clustered at relatively low flow stresses and encompass the roughing stage. In this group, the first pass, applied at the highest temperature, corresponds to the curve exhibiting the lower stress values and the smallest amount of total strain. The successive passes, up to pass 7, exhibit increasing stress and amount of strain

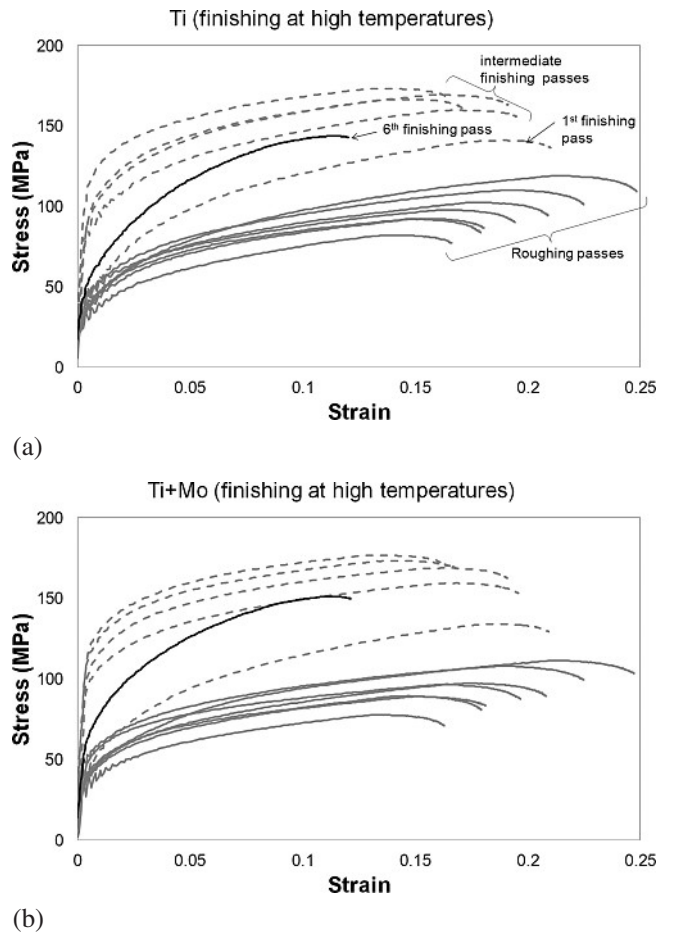


Fig. 3. Representation of the stress–strain curves of the controlled finishing at high temperatures for the (a) Ti and (b) Ti + Mo steels. Continuous grey lines correspond to the stress–strain curves of the roughing passes and dashed grey lines correspond to the finishing passes, except for the last finishing pass which is represented by the black line.

Table 2. Transformation temperatures (°C) [8].

	T_{nr}	A_{r3}	A_{e3}
Ti	840 ± 15	762	840
Ti + Mo	897 ± 6	771	850

Table 3. Characteristic temperatures of hot torsion schedules (in °C).

	Pass #13	T_{Rs}	T_{Rf}	T_{Fs}	T_{Ff}
Ti	High temperatures	1010	928	823	807
	Low temperatures	1010	928	823	760
Ti + Mo	High temperatures	1040	958	853	820
	Low temperatures	1040	958	853	790

levels. A second set, which is clustered at relatively high flow stresses, encompass all the finishing passes apart from the first and last finishing passes. The first and last finishing passes make up the third set of curves, which fall in between the flow stress levels of the two other groups, and also exhibit significantly higher work hardening capacity.

In order to inhibit the strong softening which takes place before the last finishing pass, the finishing stages were carried out at lower temperatures, with a last pass applied at temperatures slightly above the A_{r3} of the strained austenite in Table 2, obtained through CCC tests [8]. The stress-strain curves obtained after this new rolling schedule are represented in Fig. 4. In this case, the behaviour is similar for both steels at the passes simulating the roughing. However, the finishing passes exhibit higher strengths as a result of the lower temperatures at which these are applied. Moreover, the last finishing pass exhibits less restoration than for the rolling schedule with the finishing stage carried out at higher temperatures, represented in Fig. 3.

3.3. CCCT tests

CCCT test were carried out by applying a torsion pass during cooling after the different hot rolling schedules. Usual-

ly, the stress is supposed to increase as the temperature is reduced, however, when a phase transformation takes place this behaviour is interrupted and the transformation temperatures can be obtained at the points where the trend changes. For the Ti and Ti + Mo steels, the CCCT curves after hot rolling schedules with finishing stages at high and low temperatures can be observed in Fig. 5. In this case, the CCCT curves in black, corresponding to rolling schedules with high finishing temperatures, show a slight increase in the strength followed by a drop related to the beginning of the austenite to ferrite transformation. This behaviour indicates that the last pass of the rolling was carried out in the austenite region. On the other hand, the CCCT curves obtained by deforming the samples after the whole rolling schedule with low finishing temperatures (grey curves), show a continuous softening since the beginning, on cooling. This means that the transformation had already started when the last finishing pass was applied.

According to these tests, the transformation temperature for the strained austenite (A_{r3}^d) for the Ti steel is around 780 °C and around 800 °C for the Ti + Mo steel. The differences between these transformation temperatures and the ones obtained from the CCC test (Table 2) lie in the bigger amount of stress accumulation, which further accelerates the transformations [18]. This has to be considered when

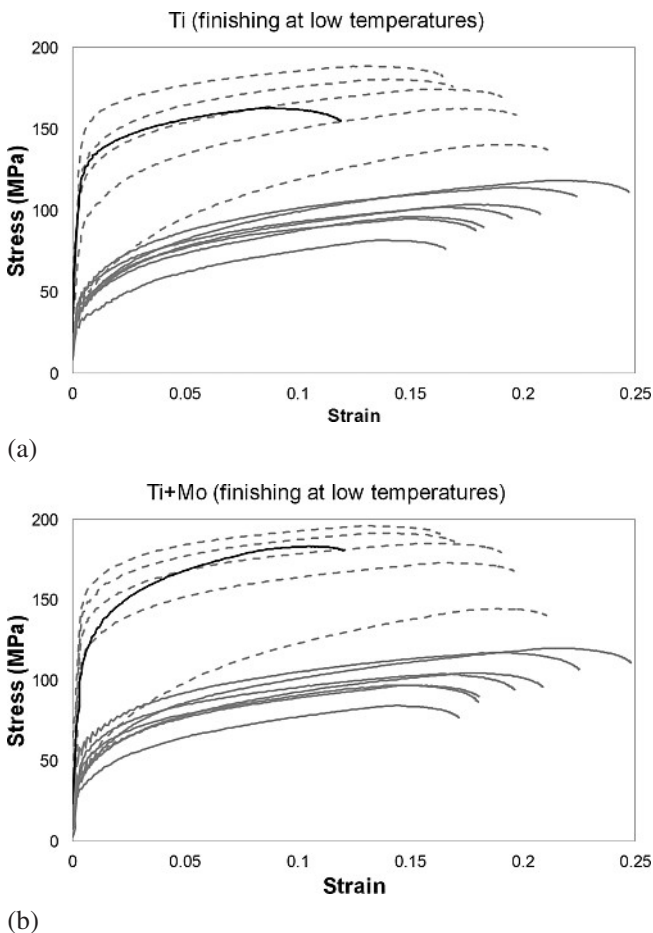


Fig. 4. Representation of the stress-strain curves of the controlled rolling finishing at low temperatures for the (a) Ti and (b) Ti + Mo steels. Continuous grey lines correspond to the stress-strain curves of the roughing passes and dashed grey lines correspond to the finishing passes, except for the last finishing pass which is represented by the black line.

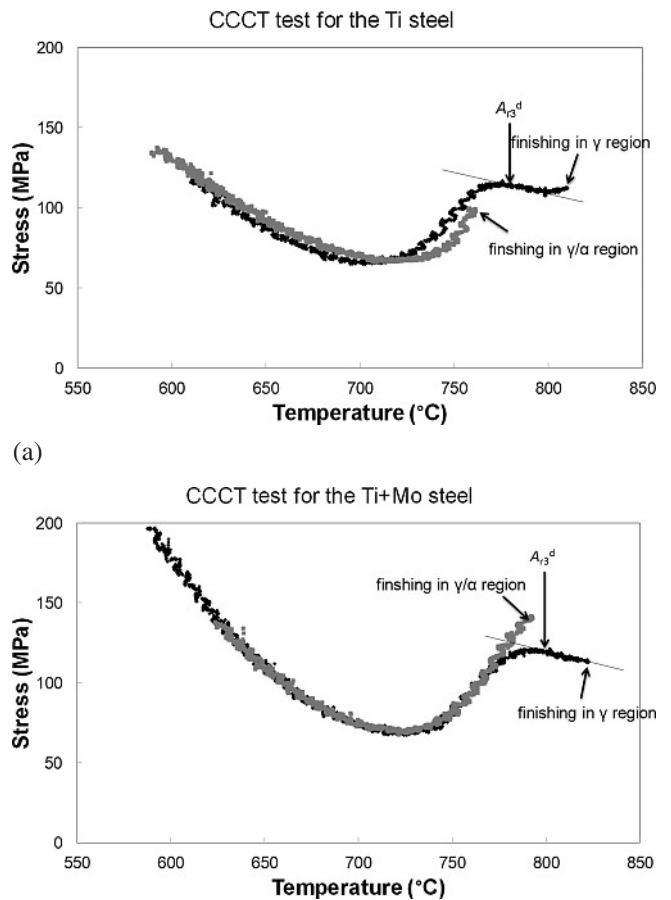


Fig. 5. CCT curves after the rolling schedule at high finishing temperatures (black line) and low finishing temperatures (grey line) for the (a) Ti and (b) Ti + Mo steels.

designing rolling schedules which have to finish in the single austenite region.

4. Discussion

According to the results in Figs. 2 to 4, the Ti and Ti + Mo steels present the typical behaviour of steels that are being pancaked, i.e. the first pass of the finishing stage presents a yield strength similar to the yield strength of the material during the roughing, and, after that, there is a continuous increase in the yield strength of consecutive deformation passes in the finishing stage. This increase is usually related to the stress accumulation which takes place as a consequence of the retardation of recrystallization by the Zener drag effect of precipitates [19]. In the first pass of the finishing, the yield strength is low because there has been no strain accumulation and precipitation has not taken place. In fact, precipitation takes place induced by deformation and, given the strain hardening rate of the first finishing pass, it seems reasonable to consider that precipitation appears concurrently with deformation, i.e. it is dynamic precipitation. In fact, the dynamic nature of precipitation has been proven by Mousavi et al. [7] for an Nb microalloyed steel. The Ti and Ti + Mo steels exhibit an analogous behaviour which would be indicating that pancaking is possible for these steels, although the nature of the precipitates which promote pancaking is different to the precipitation in Nb microalloyed steels.

Usually, Ti is added to induce the precipitation of TiN, which takes place at very high temperatures and is seen as an effective way to inhibit grain growth during thermomechanical processing, particularly during reheating. However, the evidence of static recrystallization retardation in Ti microalloyed steels, which has been shown in this work and in other studies [9–11], could be due to two mechanisms: (i) solute drag effect of Ti in solid solution or (ii) the pinning effect of precipitates. For a hypostoichiometric composition, such as the ones of the Ti and Ti + Mo steels, it is not likely to have enough Ti in solid solution to retard static recrystallization by solute drag, although some authors have reported that quantities as small as 0.002 wt.% of dissolved Ti could have this effect [20].

Thermodynamic calculations using FactSage[®] were performed for the compositions of the Ti and Ti + Mo steels, and the results are shown in Fig. 6. According to these calculations, TiN would start forming the onset of solidification of the steel, and TiC and AlN are the only precipitates which could form in the finishing temperature range, promoting pancaking. According to the calculations, TiC and AlN would form at the expense of the TiN. This is not likely to be taking place under real conditions because at low temperatures the kinetics of dissolution of TiN precipitates would be very slow. However, it could still be possible to find Ti in solid solution from the incomplete precipitation of TiN during the cooling from the melt [21]. Recrystallization could therefore be retarded by either precipitation or by the solute drag effect of any Ti in solid solution. Even though finding out the exact mechanism which is promoting the T_{nr} behaviour of the Ti and Ti + Mo steels is not the objective of this work, this should be considered with regard to the properties of the final product. In particular, if recrystallization retardation is caused by solute drag,

rather than Zener drag of the precipitates, this could have a positive effect on the toughness, which is usually negatively affected by precipitation.

Therefore, the possibility of pancaking microalloyed steels with no Nb in their composition, could lead to the design of new hot rolling schedules aiming to obtain fine-grained ferritic steels based on a Ti microalloyed composition. However, the pancaking of these steels presents several characteristics which should be taken into account. First of all, the T_{nr} temperature of Ti microalloyed steels is lower than the T_{nr} of Nb microalloyed steels. This, in addition to the fact that strain accumulation during pancaking increases A_{r3} , leads to a narrow range of temperatures at which the pancaking could be carried out. Furthermore, higher strain rates, which can also take place in the industrial process, could additionally increase the transformation temperatures.

Moreover, both the Ti and Ti + Mo steels exhibit fast restoration rates during finishing, as observed in Fig. 7, which represents the mean flow stress, calculated according to Eq. (1), vs. $1000/T$ curves for the C–Ti and Mo steels after the two different finishing schedules (in the graphs, finishing in the single phase region is equivalent to finishing at high temperatures and finishing in the 2-phase region is equivalent to finishing at lower temperatures). In fact, this representation is a common way to identify microstructural changes taking place during hot rolling. The drop in the MFS for the last finishing pass (indicated by big signs in the graph) is due both to softening and to the smaller strains applied during this pass. In fact, the effect of softening with regard to the finishing rolling schedule is more evident when the stress–strain curve of the last pass is compared to the ones for the previous passes in Figs. 3 and 4. An interesting aspect, which is revealed by the curves in Fig. 7, is that softening is also very pronounced during the interpass between roughing and finishing. It is considered that the change in the slope of the MFS vs. temperature indicates a change in the deformation mechanisms of roughing and finishing. In the case of the Mo steel, apart from the MFS of the first and last finishing passes, the MFS of the other finishing passes are above the ones predicted by extrapolating the MFS vs. temperatures to $1000/T$ values of the finishing stage, which is similar to the behaviour exhibited by Nb mi-

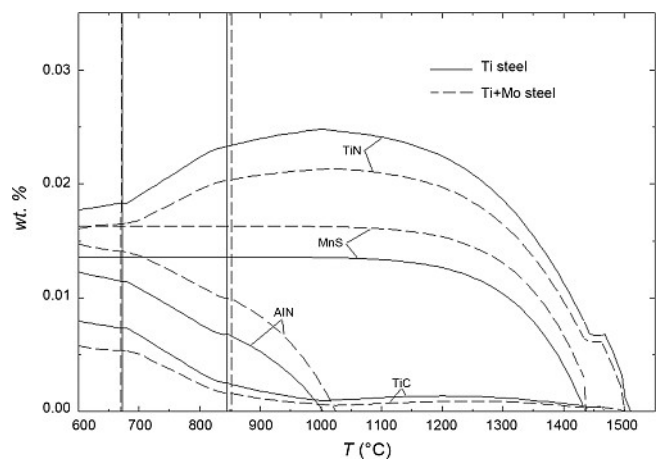


Fig. 6. Precipitation curves from thermodynamic calculations using FactSage[®] software.

coalloyed steels [6]. However, for the Ti steel only the *MFS* of the penultimate finishing pass falls above the extrapolation, which indicates that softening was very pronounced during the interpass between roughing and finishing. This softening is less pronounced for the Ti + Mo steel, which is favorable to pancaking.

In terms of feasibility of performing such hot rolling in an industrial mill, it is instructive to compare the mean flow

stresses of a high Nb steel subjected to a simulated classical hot rolling schedule to the *MFS* of Mo steel with the pancaking schedule used in this work (Fig. 8). It can be seen that the rolling loads are generally somewhat higher for the Mo steel for both roughing and finishing. It is, of course, possible to rough at higher temperatures, but this may lead to coarsening of the recrystallized austenite grains, which may limit subsequent ferrite grain refinement.

It should be taken into account that the industrial TMP cycle simulated in this study corresponds to a reversing mill. This cycle is characterized by long interpass times, which become longer after each deformation pass, as the product reduces its thickness and increases its length. However, other industrial schedules, for example in tandem mills, do not require long interpass times during finishing and therefore recrystallization retardation in Ti and Ti + Mo steels would be more effective.

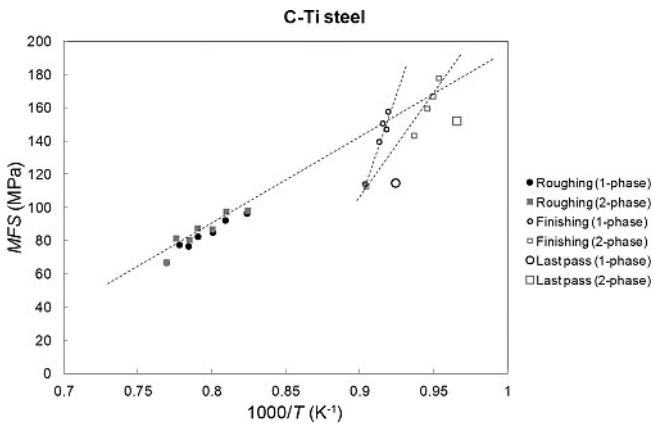
5. Conclusions

In general, the analysis of the *MFS* vs. $1000/T$ curves after a torsion simulation of an industrial rolling schedule show that Ti microalloyed steels could be pancaked. If the temperature range for finishing and softening behaviour is considered, it can be concluded that the Mo steel exhibits better behaviour in terms of pancaking than the C–Ti steel. However, even for the Mo steel the temperature range between T_{nr} and A_{r3} is narrower than for Nb microalloyed steels, and it also exhibits a higher tendency to softening. Industrially, these outcomes would indicate that pancaking of Ti microalloyed steels would only be reached for finishing schedules in which cooling rates could be slow in order to warrant that all finishing passes fall between T_{nr} and A_{r3} . Moreover, interpass times should be as short as possible to avoid softening.

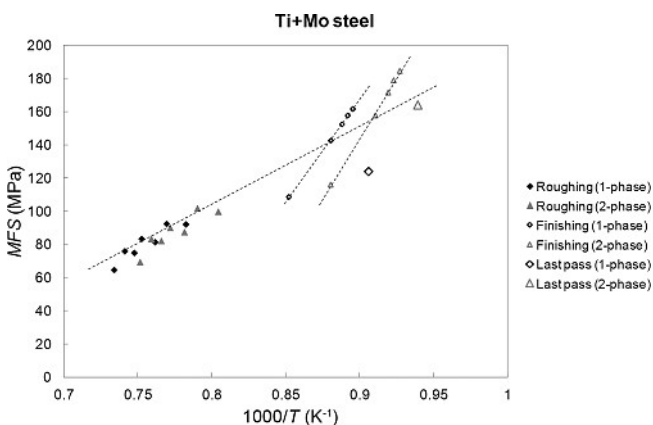
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(a)



(b)

Fig. 7. *MFS* vs. $1000/T$ after the rolling schedules with the finishing at high temperatures (1-phase region) and low temperatures (2-phase region) for the (a) Ti and (b) Ti + Mo steels.

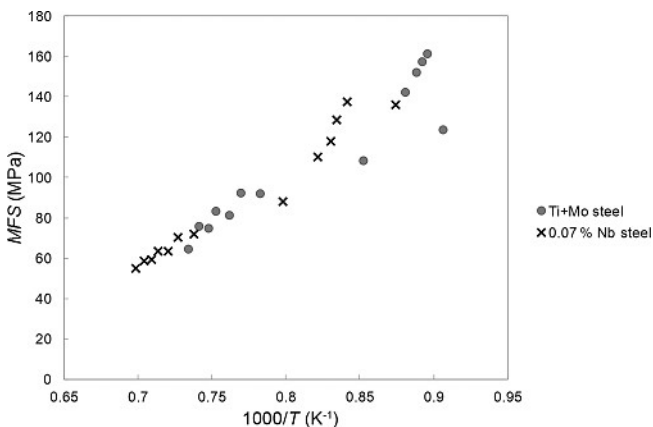


Fig. 8. Comparison of *MFS* vs. $1000/T$ for the Ti + Mo steel and an 0.07 % Nb microalloyed steel.

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Correspondence address

Dr. Jessica Calvo
 Av. Diagonal 647
 Pabellón E
 08028, Barcelona
 Spain
 Tel.: +34 934054154
 Fax: +34 934016706
 E-mail: jessica.calvo@upc.edu