Original Article

Effect of Nb and Ti addition on activation energy for austenite hot deformation

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ABSTRACT

Constitutive equations are employed in several algorithms to predict load during hot working. The peak stress ($\sigma_p$) can be correlated to the temperature and strain rate at which the metal was deformed using the known expression: $Z = A(\sinh(\omega t))^n$, where, $Z$ is the Zener–Hollomon parameter, given by $Z = \dot{\varepsilon} \exp(Q_{def}/RT)$. The activation energy for hot deformation, $Q_{def}$, magnifies the dependence of the stress on temperature in an exponential way requiring accurate measurement of this quantity. Empirical expressions published in the literature can be used to evaluate $Q_{def}$. Unfortunately, these equations were derived for alloys with relatively high contents of C and N and when applied to the case of IF austenite, they yield unreasonable values. This is so because IF steels have, simultaneously, very low C and N contents and substantial amounts of microalloying elements leading to over prediction of values of $Q_{def}$. This paper investigates, therefore, how the chemical composition of IF alloys can influence the values of the activation energy for hot deformation allowing suitable predictions to be made. Measurements of $Q_{def}$ were carried out for IF steels containing Ti and a combination of Ti-Nb additions. Predicted and measured values of $Q_{def}$ were compared to those calculated from published equations and a corrected expression was derived in order to better fit the experimental results here presented.

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1. Introduction

Constitutive equations relating stresses in the stress–strain curves to strain rates and temperatures have been routinely employed to estimate hot forming loads [1–8]. Among all expressions published in the literature [9–14], probably the most used is the hyperbolic sine function relating $Z$, the Zener–Hollomon parameter, to the stress, $Z = \dot{\varepsilon} \exp(Q_{def}/RT) = A(\sinh(\omega t))^n$. Here, $Q_{def}$ is the activation energy for hot deformation, $A$, $\omega$ and $n$ are constants, the other symbols having their usual meaning. This expression gives a straight line in a log($Z$) vs log($\sinh(\omega t)$) plot irrespective of the level of the stress, whereas, $Z$ expression based on power and exponential laws lose linearity at high and low stresses respectively [2–12]. The peak stress resulting from the occurrence

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of dynamic recrystallization, DRX, and also the steady state stress associated with the nucleation and completion of the first wave of DRX have been successfully modeled by the above expression [15–18]. Once the stress value in the hyperbolic sine function depend on an exponential function of the deformation temperature, and this dependence is magnified by \( Q_{\text{def}} \), it becomes important to accurately measure and predict the activation energy for hot deformation to estimate \( \sigma \) as function of \( Z \) [19].

Some relevant empirical expressions have been published in the literature describing the dependence of \( Q_{\text{def}} \) on the chemical composition for a number of steels [20,21]. Medina and Hernandez’s (MH) [20] model is given by:

\[
Q_{\text{def}} (J/mol) = 267000 - 2536(\%C) + 1010(\%Mn) + 33621(\%Si) \\
+ 35651(\%Mo) + 93681(\%Ti)^{0.592} + 31637(\%V) \\
+ 70730(\%Nb)^{0.565} 
\] (1)

Whilst, Cho and collaborators (CKJ) [21] model is expressed by:

\[
Q_{\text{def}} (kJ/mol) = 297 + 641(\%Nb) + 123(\%Mo) \\
+ (\%Ni) - 111(\%Cr) 
\] (2)

The two research groups worked with a different set of steels. Medina and Hernandez used a series of 18 alloys altering systematically the contents of C, Si, Mn, Mo, Ti, V and Nb, the levels of microalloying elements varying within the usual contents for HSLA steels. The amounts of C covered by Medina and Hernandez, however, were relatively high, ranging between 0.11 and 0.44 in \( \% \) weight. Cho and co-workers, on the other hand, used samples with significantly lower contents of C, around 0.07, Mo varying from 0.15 to 0.28, no V, Ti in all samples, at a constant level of 0.015 and Nb varying from 0.032 to 0.056, all values given in \( \% \) weight. At the same time, the samples investigated in the work by Cho and co-authors presented Ni in between 0.2 and 0.5, and Cr, in the range of 0.3–0.6.

The differences in chemical compositions of the steels used by both research groups led, inevitably, to different elements multiplying factors in the models to evaluate \( Q_{\text{def}} \). The most important of these discrepancies, perhaps, is how Nb additions affect \( Q_{\text{def}} \). Owing to the Nb exponent equal to 0.565, MH model predicts a saturation effect on \( Q_{\text{def}} \) with increasing Nb, whereas CKJ model, with Nb exponent equal the unit, considers a linear increase on \( Q_{\text{def}} \) with changes in Nb and no saturation. As a consequence, both models predict different values of \( Q_{\text{def}} \) and \( \sigma \) as function of \( Z \) for a given chemical composition, especially for high Nb content steels where the differences increase. For instance, for a HSLA base chemical composition of 0.05\%C–0.05\%Nb, the predicted values would be 291 and 329 kJ/mol for the MH and CKJ expressions, respectively. These differences are further exacerbated for the case of IF austenite since it presents simultaneously very low C levels and relatively medium to high microalloying contents. Predictions for a 0.002\%C–0.1\%Nb composition, for instance, give \( Q_{\text{def}} \) 297 and 361 kJ/mol respectively, from MH and CKJ models.

The purpose of the present work was to measure the activation energy for hot deformation of a Ti and a TiNb stabilized IF steel evaluating the effects on \( Q_{\text{def}} \) of separate additions of Ti and Nb, and to compare the measured \( Q_{\text{def}} \) obtained here with the predicted values from the MH and CKJ models. Finally, a more effective expression was proposed in order to appropriately estimate \( Q_{\text{def}} \) for IF steels.

### 2. Experimental technique

The chemical compositions of the IF steels employed in this work are shown in Table 1. Samples for torsion experiments, 15 mm length and 10 mm in diameter, were cut at ¼ of the thickness of a hot rolled billet of 100 x 100 mm cross section. The specimens were machined with their axis parallel to the billet rolling direction. The experiments were carried out in a computer controlled servo-hydraulic machine equipped with an infrared furnace. All tests were performed in argon atmosphere and the temperature was controlled by a cromel-alumel thermocouple welded to the sample surface. The specimens were initially pre-heated at 1473 K (1200 °C) for 900 s (15 min). Torque and angular displacement were converted to equivalent stress–strain curves using the following equations [22]:

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

and

\[
\varepsilon = \frac{1R}{\sqrt{3}l} \theta \] (4)

Here, \( R \) and \( L \) are the sample radius and length, respectively, and \( I \) and \( \theta \) are the measured torque and angular displacement.

Non-isothermal testing was conduct to determine the values of \( T_{nr} \) and \( A_{r3} \), using the technique described by Boratto et al. [23]. After preheating the samples were cooled at a

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### Table 1 – Chemical composition of steels employed in the present work. All numbers are given in % weight, except for C and N with numbers given in ppm weight. Base IF, with no microalloying elements, was used as reference steel. The table also shows values of \( T_{nr} \) and \( A_{r3} \) measured for the alloys employed in this work. Base IF and Ti IF steels presented no \( T_{nr} \) differently from the TiNb IF steel. The values of \( A_{r3} \) decreased consistently as a function of microalloying additions the TiNb IF steel displaying the lowest value of \( A_{r3} \).

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Ti</th>
<th>Nb</th>
<th>N</th>
<th>( T_{nr} ) °C</th>
<th>Ar3, °C</th>
<th>Ar3, °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>TiNb IF</td>
<td>24</td>
<td>1.70</td>
<td>0.17</td>
<td>0.006</td>
<td>0.002</td>
<td>0.014</td>
<td>0.11</td>
<td>19</td>
<td>945</td>
<td>840</td>
<td>855</td>
</tr>
<tr>
<td>Ti IF</td>
<td>28</td>
<td>0.15</td>
<td>0.01</td>
<td>0.013</td>
<td>0.009</td>
<td>0.083</td>
<td>–</td>
<td>38</td>
<td>–</td>
<td>905</td>
<td>915</td>
</tr>
<tr>
<td>Base IF</td>
<td>89</td>
<td>0.13</td>
<td>0.01</td>
<td>0.008</td>
<td>0.010</td>
<td>–</td>
<td>–</td>
<td>22</td>
<td>–</td>
<td>910</td>
<td>900</td>
</tr>
</tbody>
</table>

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cooling rate of 1 K/s (1 °C/s). Successive deformations of 0.3 equivalent strain were applied to the sample at time intervals of 30 s. The first deformation was delivered at approximately 1453 K (1180 °C) and the last at about 873 K (600 °C), covering temperatures typical of hot rolling in the austenite and a reasonable range of temperatures in the ferrite, similar to those employed in warm rolling of IF steels.

Isothermal testing was performed to obtain stress–strain curves at constant strain rate and at several temperatures. Samples were pre-heated as described above and then cooled down to testing temperature at a rate of 1 K/s (1 °C/s). All samples were kept for 90 s (3 min) at the testing temperature prior to twisting promoting temperature equalization throughout the specimen. In the case of the tests conducted at 1473 K (1200 °C), the homogenization procedure was not necessary and deformation started right after a pre-heating time of 900 s (15 min). Tests were performed at temperatures of 1123, 1173, 1223, 1273, 1323 and 1373 K (850, 900, 950, 1000, 1050 and 1100 °C) and strain rates of 0.1, 1, 3 and 10 s⁻¹. All samples were deformed to an equivalent strain of 7, allowing the occurrence of full DRX and the achievement of the steady state in all tests.

3. Results and discussion

Table 1 shows the values of $T_m$ and $A_r$ measured and $A_e$ calculated by software CEQCSI® for the alloys employed in this work. In the absence of any microalloying elements, Base IF presented no $T_m$ and an $A_r$ value of 910 °C, close to the $A_e$ found for this alloy (900 °C) and the γ → α equilibrium transformation temperature for pure Fe. Similarly, the Ti IF displayed no $T_m$ and a close value of $A_r$. Fig. 1 shows the CEQCSI computation, based on the Ti IF chemistry, for the phases and precipitates formed in equilibrium condition as function of the temperature. As indicated in Fig. 1a, TiN is formed during the solidification of steel (1535 °C) and not being fully dissolved on heating prior to deformation, as well-informed by the literature [24,25]. These particles are relatively large not affecting the kinetics of static recrystallization importantly enough to prevent its occurrence. The entire N goes to these particles and the austenite becomes free of this element (Fig. 1b and d). $\text{Ti}_4\text{C}_2\text{S}_2$, on the other hand, usually forms at lower temperatures still in the austenite or on transformation to ferrite [26]. In Ti IF steel, due to the relative high Ti content (0.083%), the appearance temperature for this precipitate is around 1280 °C. Consequently, the solubility temperature applied here are not enough to dissolve neither TiN nor $\text{Ti}_4\text{C}_2\text{S}_2$ in Ti IF steel. In the same way that N, the C precipitates in $\text{Ti}_4\text{C}_2\text{S}_2$ particles and the austenite of Ti IF steel becomes free of both elements in temperatures below nearly 1100 °C (Fig. 1c and d). On the other hand, the Ti content in the alloy is much higher than the stoichiometric regarding TiN and $\text{Ti}_4\text{C}_2\text{S}_2$ formation. Thus, 475 ppm of Ti remains in solid solution in austenite until the phase transformation, afterward this Ti continuing in ferrite as substutional element. Therefore, the presence of Ti either as carbonitrides particles or in solid solution did not delayed recrystallization enough to cause an accumulation of deformation between torsions not leading to the occurrence of $T_{nr}$, as advised by Barbosa et al. [27].

In TiNb steel, the lower Ti addition (0.014%) likewise allow the Nb,Ti(C,N) high temperature precipitate formation in
around 1425 °C, in this case in solid phase (Fig. 2a). It is important to note that Nb,Ti(C,N) starts as a Ti–N rich precipitate and evolve to a composition richer on Nb and C only below 800 °C (Fig. 2b), a comparable result to that found by Akamatsu et al. [28]. Ti4C2S2 is expected to appear in austenite in temperatures less than 1020 °C. The lower Ti4C2S2 formation temperature theoretically leads to an increase in nucleation rate and a reduction in growing kinetics, what can allow these particles to be finer in TiNb IF than Ti IF steel. Fig. 2b–d shows that N and C are completely removed from austenite due to Nb,Ti(C,N) and Ti4C2S2 formation, and the same happens with Ti that reduces gradually its content in austenite been closed to zero at the beginning of phase transformation. Despite of high Nb (0.11%) in TiNb IF, the low C content reduces the driving force for Nb precipitation, so most of the Nb remains in solid solution in austenite until the γ → α transformation.

Microalloying elements addition it is known to delaying recovery and preventing static recrystallization to commence, leading to a strain accumulation that can be caused or by fine particle precipitation in austenite or by the solute drag effect [21,29,30]. Tr, of approximately 1218 K (945 °C) was measured for TiNb IF alloy, which can be result of (a) fine Ti4C2S2 precipitates formed in temperatures closed to Tr, or (b) solute drag effect due to near 1100 ppm of Nb in solid solution. However, it is evident here that more in deep investigation is needed in order to explain better the Ti/Nb metallurgical phenomena influencing Tr in interstitials free steels. As expected, Nb also delayed transformation of austenite to ferrite, decreasing further, as shown in Table 1, Ar3. Most of the stress–strain curves in this work were obtained from tensile tests at temperatures ranging from 1373 to 1223 K (1100 to 950 °C), well above Ar3. However, as Ar3 for the TiNb IF steel was so low, tests were also conducted at 1173 and 1173 K (850 and 900 °C) for this alloy.

Fig. 3a shows the stress strain curves obtained for the TiNb IF alloy tested at the strain rate of 1 s−1 and a range of temperatures. All curves display a characteristic peak stress, σp, indicating that DRX was already in progression in all cases. The specimens tested here were deformed to total equivalent strains around 7.0, in order to guarantee full DRX even for the lowest temperatures. The peak strains increased from approximately 1 for the sample tested at 1323 K (1050 °C) to about 1.5 for the experiment conducted at 1173 K (850 °C). Likewise, the strain to achieve the steady state stress in stress–strain curves (corresponding to 100% DRX) increased from 3 at the higher test temperature to 5 at the lowest temperature. Fig. 3b shows stress–strain curves for Base IF, Ti IF and NbTi IF steels deformed at a constant strain rate of 1.0 s−1 and test temperature of 1273 K (1000 °C). All samples show the characteristic peak stress resulting from the occurrence of DRX, the peak strain increasing from around 0.5 for Base IF to 0.8 for the Ti IF steel and to 1.2 for the NbTi IF steel. The strain for the initiation of the steady state stress also increases with microalloying content going from 1.8 for the Base IF alloy, to 3 for the Ti IF and 4.2 for the TiNb IF steel respectively. Clearly, the presence of microalloying elements in the IF austenite delays the onset of DRX and also retards the kinetics with which this recrystallization reaction proceeds.

Fig. 3 – (a) Stress–strain curves for samples of TiNb IF alloy deformed at a constant strain rate of 1.0 s⁻¹ at several temperatures: 1123 K (850 °C), 1223 K (950 °C) and 1323 K (1050 °C). All samples show the characteristic peak stress resulting from the occurrence of DRX. The experiments were conducted to an equivalent strain of 7 in order to allow full DRX, leading to a steady state stress in all cases. (b) Stress–strain curves for Base IF, Ti IF and TiNb IF steels deformed at a constant strain rate of 1.0 s⁻¹ and test temperature of 1273 K (1000 °C). All samples show the characteristic peak stress resulting from the occurrence of DRX, the peak strain increasing from around 0.5 for Base IF to 0.8 for the Ti IF steel and 1.2 to the TiNb IF steel. The strain for the initiation of the steady state also increases with the microalloying content in the steels.

In all cases, analyses established on the work-hardening rate (\(\dot{\varepsilon} = \partial \varepsilon / \partial \varepsilon\)) were performed to determine relevant points of the stress–strain curves. The critical strain for the initiation of DRX (\(c_c\)) was measured from the inflections in plots of \(\dot{\varepsilon} / \varepsilon\) or from the minimum in \(-\partial \varepsilon / \partial \varepsilon\) \(\sigma\) curves, as reported by Poliak and Jonas [31,32], the values laying approximately at 0.7–0.8\(\varepsilon_p\), for the Base IF alloy, dropping to about 0.5–0.6\(\varepsilon_p\) for the IF steels, consistent with other values reported in the literature [12,33–35].

Fig. 4 shows the dependence of the peak stress on the reciprocal of the test temperature for samples deformed at strain rates of 0.1, 1, 3 and 10 s⁻¹ for (a) NbTi IF, (b) Ti IF and (c) Base IF alloys, respectively. The slopes, a function of the activation energy for hot deformation, are fairly independent of strain rate, increasing with microalloying content. The activation energy for deformation can be calculated from the hyperbolic sine relation

\[
Q = 2.3nR \left[ \frac{\partial [\log(\sinh(\alpha \varepsilon))]}{\partial (1/T)} \right]
\]

(5)

The value of \(\alpha\) was taken as 0.012 MPa⁻¹ as suggested in the literature [36,37] and \(n\) was calculated for a given temperature from the slope of a \(\log(\dot{\varepsilon})\) vs \(\log(\sinh(\alpha \varepsilon))\) plots. The constant \(A\) in the hyperbolic sine equation could then be evaluated. The values of \(Q_{\text{act}}\), \(n\) and \(A\) measured in this work are shown in
Table 2 together with values predicted using the MH and CKJ models.

For the Base IF alloy $Q_{def}$ was measured as 272 kJ/mol, very similar to the value predicted by the MH model, 267 kJ/mol, and close to the self diffusion activation energy for Fe in austenite. The value of 297 kJ/mol was predicted by the CKJ model, almost 30 kJ/mol higher than the measured in the experiments conducted here. This discrepancy apparently results from the incapacity of the CKJ model to separate the effect of Ti from that of Nb on $Q_{def}$. On the other hand, the measured $Q_{def}$ for Ti IF steel was 293 kJ/mol, close to the predicted values by both MH and CKJ models. Ti IF steel, with addition of 0.083% in weight of Ti, presented an increase of about 20 kJ/mol in $Q_{def}$ in comparison to Base IF. The MH model predicts an increase of the order of 21 kJ/mol for the concentration of Ti present in the alloy, whereas, the CKJ model predicts no increase, maintaining $Q_{def}$ as 297 kJ/mol. This is not surprising since this model, as pointed out, is unable to account for the effect of Ti additions on $Q_{def}$. The addition of Ti and Nb to TiNb stabilized IF steel led to a measured $Q_{def}$ of 305 kJ/mol similar to that predicted by the MH model, 302 kJ/mol, and fairly different from that estimated from the CKJ expression. The Ti Nb steel tested here has 0.014%Ti and 0.11%Nb. According to the MH model $Q_{def}$ increases from 267 kJ/mol to about 290 kJ/mol as a result of the Ti addition and then to 302 kJ/mol as a result of the addition of Nb. The CKJ model does not anticipate any effect of Ti additions on $Q_{def}$ and predicts Nb concentrations acting linearly on the value of the activation energy. In this way, $Q_{def}$ goes from 297 to 368 kJ/mol as a result of a 0.11%Nb addition. This discrepancy between predicted and measured values can be possibly explained by the manner the effect of Nb additions on $Q_{def}$ is considered in both models. As already mentioned, the MH model takes the effect of Nb on $Q_{def}$ in such a way that a saturation of this effect is attained as the content of Nb increases. The CKJ model, on the other hand, considers the effect of Nb on $Q_{def}$ as a linear function of the concentration of this element in the alloy. As the concentration of niobium in Nb IF steel is significantly higher than that usually found in HSLA or in the low C alloys used to obtain Eqs. (1) and (2), this combination of Nb linear dependence and high contents of Nb may have caused an overestimation of $Q_{def}$ by the CKJ model. It may be then concluded that values around 270, 290 and 305 kJ/mol seem to be reasonable for activation energies for hot deformation in the case of Base IF, Ti IF and TiNb IF steels, disallowing here the apparently overestimated value of 368 kJ/mol indicated by CKJ model that predicts Nb concentrations acting linearly on the value of the activation energy and have no saturation profile like MH model.

The activation energy for hot deformation is essential in the definition of how the maximum stress depends on Z, as pointed out earlier in this study. Figs. 5a–c shows the dependence of the maximum stress on the Zener–Hollomon parameter. The square symbols represent experimental points obtained in this work. The full line means the value of maximum stress calculated using Eq. (6):

$$\sigma_p = \frac{1}{\alpha} \left[ \sinh^{-1} \left( \frac{Z}{A} \right)^{1/n} \right]$$

(6)

with $Q_{def}$, n and A measured in the present work. The dotted line represents the value of the peak stresses calculated according to Eq. (6) employing, however, $Q_{def}$, n and A estimated from the HM model. A comparison of predicted values of $\sigma_p$ from the CKJ model with MH model or the results from this study is not possible since the A values in CKJ model are specific and refer to chemical compositions used by the authors in the reference work [21]. As it can be seen, the values of maximum stresses predicted by the MH model are always higher than the measured in this work, for a given value of $Z$. This might look initially surprising. The maximum stress, however, is a function of $Q_{def}$, n, A and $A_{MH}$. From these constants, the measured and the predicted values for two of them, $Q_{def}$ and n, are fairly similar. The constant $\alpha$ was used as 0.012 MPa$^{-1}$ for both experimental results and the MH model predictions. Therefore, the disagreement shown in Fig. 5 comes from diverging A values obtained in this work. Clearly, measured and predicted A values are of the same order of magnitude for the TiNb IF alloy, as shown in Table 2. However, the MH model predicts a faster drop in A values as the microalloying content of the steels decreases. The model predicts, in the case of Ti IF steel, an A value of 0.8 × 10$^{-10}$ s$^{-1}$, whereas the measured value was 2.7 × 10$^{-10}$ s$^{-1}$, an order of magnitude higher. The same trend is further stressed for the case of Base IF, that is, the predicted value is 0.2 × 10$^{-10}$ s$^{-1}$ and the measured 1.4 × 10$^{-10}$ s$^{-1}$. Clearly, the levels of C and Nb used in IF steels exceed those employed by Medina and Hernandez in their experimental work undermining the ability of the model to predict A values for such compositions. Therefore, a correction on how the MH predicts A values if maximum stresses for IF steels are to be reasonably modeled.

A slight correction on the MH model is here proposed in order to decrease the discrepancies just reported for the case of IF steels. Clearly, an extensive investigation on $Q_{def}$, n and A values as a function of the chemical composition for a large set of IF steel compositions is not the purpose of this work, therefore a simple adjustment to the MH model was felt here necessary to account for the differences shown in Fig. 5. According to Medina and Hernandez, A is given by

$$A_{MH}(s^{-1}) = (12.2 + 65.6(%C) − 49.1(%Nb)) \exp(7.1 \times 10^{-5}Q_{def})$$

(7)
As the $Q_{\text{def}}$ and $n$ values measured here in this work and those predicted by the MH model were fairly similar, a new term was added to the original MH equation correcting the predictions for the peak stresses of IF steels as shown below. The new term separates the influences of Ti and Nb on the value of $A$, whilst preserving the original expression proposed by Medina and Hernandez. In summary, the value of $A$ could be calculated as

$$A = A' \times A_{\text{MH}}$$

$$A' = 8.2 \exp(-10.1[\%\text{Ti} + \%\text{Nb}])$$

Fig. 5 – The dependence of the maximum stress on the value of $Z$ for tests conducted at several temperatures and strain rates: (a) TiNb IF, (b) Ti IF and (c) Base IF. The calculated maximum stresses depend on the values of $Q_{\text{def}}$, $n$ and $A$.

Fig. 6 – The dependence of the maximum stress on the value of $Z$ for tests conducted at several temperatures and strain rates: (a) NbTi IF, (b) Ti IF and (c) Base IF. The calculated maximum stresses depend on the values of $Q_{\text{def}}$, $n$ and $A$. In this case, a slightly modified $A$ value in relation to that proposed by the MH model was used here so that the disagreement between predicted and measured values was minimized.

Fig. 6a–c shows the agreement between the experimental results, represented by square symbols and straight lines, with the predictions obtained from the corrected MH model. However, it is unequivocally recognized here that more deep study is needed in order to clarify better the influences of microalloying elements on the activation energies for hot deformation and, as a consequence, on the maximum stresses of IF austenite.
4. Conclusions

The present results lead to the following conclusions:

1. In Ti IF steel TiN is formed during the solidification of steel (1535 °C) and cannot be fully dissolved on heating prior to deformation. Ti₄C₂S₂, on the other hand, usually forms at lower temperatures still in the austenite or on transformation to ferrite. However, the appearance temperature for this precipitate is around 1280 °C. Consequently the solubility temperature applied here are not enough to dissolve neither TiN nor Ti₄C₂S₂ in Ti IF steel.

2. In TiNb IF steel the formation of Nb,Ti(C,N) precipitates in around 1425 °C, in this case in solid phase. Nb,Ti(C,N) starts as a Ti-N rich precipitate and evolve to a composition richer on Nb and C only below 800 °C. Ti₄C₂S₂ is expected to appear in austenite in temperatures less than 1020 °C.

3. Qaday was measured as 272 kJ/mol for the Base IF alloy, very similar to the value predicted by the MH model, 267 kJ/mol, close to the self diffusion activation energy for Fe in austenite. The value of 297 kJ/mol was predicted by the CKJ model, almost 30 kJ/mol higher than the measured in the experiments conducted here. This discrepancy apparently results from the incapacity of the CKJ model to separate the effect of Ti from that of Nb on values of Qaday.

4. The measured Qaday for Ti IF steel was 293 kJ/mol, close to the predicted values by both MH and CKJ models, although for different reasons. The MH model predicts an increase of the order of 21 kJ/mol for the concentration of Ti present in the alloy, whereas, the CKJ model predicts no increase, maintaining Qaday as 297 kJ/mol, apparently disregarding any effect Ti additions may have on the value of Qaday.

5. Addition of Ti and Nb in TiNb IF steel led to a measured Qaday of 305 kJ/mol similar to that predicted by the MH model, 302 kJ/mol, and fairly different from that estimated from the CKJ expression. The discrepancy between predicted and measured values was reasoned in terms of how the effect of Nb additions on Qaday is considered in both models.

6. Values around 270, 290 and 305 kJ/mol seem to be reasonable for activation energies for hot deformation in the case of Base IF, Ti IF and TiNb IF steels. It reflects the activation energy for Fe in austenite in case of Base IF and the Ti and Nb addition to the base alloy.

7. The activation energy for hot deformation is essential in the definition of how the maximum stress depend on Z, as pointed out in this study. The values of maximum stresses predicted here by the MH model were always higher than those experimentally measured, for a given value of Z. The model predicts a faster drop in A values as the microalloying content of the steels decreases. Therefore, a correction on how the MH model predicted A was corrected for the case of IF steels and reasonable modeling of the maximum stresses as a function of Z could be achieved.

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Conflicts of interest

The authors declare no conflicts of interest.