Influence of the Size and Volume Fraction of TiN Particles on Hot Strength and Dynamic Recrystallisation in Structural Steels

S. F. MEDINA, M. I. VEGA,1) M. GÓMEZ and P. P. GÓMEZ

National Centre for Metallurgical Research (CENIM-CSIC), Av. Gregorio del Amo 8; 28040-Madrid, Spain.
E-mail: smedina@cenim.csic.es 1) FCC Construction S.A., Research & Development Department, c/ Acanto 22; 28045-Madrid, Spain.

(Received on March 14, 2005; accepted on June 7, 2005)

The low solubility of titanium nitrides (TiN) in austenite is taken advantage of in structural steels to control the evolution of the microstructure in the hot rolling process and in the heat affected zone (HAZ) in processes involving the application of heat, such as welding. However, the quantitative influence of the precipitation state of these particles on hot strength and dynamic recrystallisation kinetics, given by the precipitate size distribution and the precipitated volume, is practically unknown. The present work studies the influence of various Ti and N compositions, which give rise to different precipitation states at the reheating temperatures, on the aforementioned phenomena. The influence of the precipitation state on hot strength has been quantified by changes in the peak stress and the activation energy. A maximum has been obtained for the activation energy which corresponds to a Ti/N ratio of approximately 1.3. The model used to predict the flow curve and dynamic recrystallisation kinetics has been improved, extending it to include microalloyed steels containing Ti.

KEY WORDS: Ti/N ratio; precipitation state; peak stress; activation energy; dynamic recrystallisation.

1. Introduction

During hot deformation the behaviour of steels with titanium as the unique microalloying element, forming nitrides or carbides, is highly different to that of other microalloyed steels with niobium or vanadium. The reason for this is that while niobium or vanadium precipitates are dissolved at the reheating temperatures, titanium nitrides (TiN) have very little solubility in austenite and their complete dissolution is not possible unless the reheating temperature is very high, close to the melting point of steels, or in some cases, depending on the chemical composition, at temperatures corresponding to the liquid state.1–5)

This characteristic makes Ti and N ideal elements for controlling the austenite grain size in processes involving the application of heat, for instance in the heat affected zone in welding.5,7) For this reason, their addition is indispensable in structural steels to weldability requirements.

The role of TiN particles in austenite grain size control is well known and the mechanism by which they exert their influence may be explained by Zener and Gladman expressions for pinning forces,8–11) which relate the precipitate size, precipitated volume and austenite grain size. The results published to date agree that a Ti/N ratio below the stoichiometric ratio (3.42) yields good results in to control austenite grain size, especially when precipitation takes place in the solid state as finer precipitates are obtained.12–14)

Previous studies on the influence of Ti on the hot flow stress during hot deformation in steels15,16) established a quantitative dependence of the activation energy on the Ti content in the steels, without considering the precipitation state distribution, i.e. the precipitate size and precipitated volume. This obviously represented an approximation, since it did not take into consideration the pinning forces exerted by the precipitates and concentrated all the effect of Ti as a function of its total content in the steels. In other words, the effect of the pinning forces on the climb of dislocations in hot deformation and on their annihilation during static recrystallisation, respectively, was replaced, for reasons of simplification, by the effect of Ti solute drag.

The mechanism that governs the plastic flow of austenite is principally the climb of edge dislocations, which depends in turn on the diffusion of vacancies. The most important function of the climb during hot deformation is to help dislocations to overcome obstacles to slip. In other words, the deformation is primarily accomplished by slip, but the factor controlling the amount of slip is the climb of dislocations over obstacles.15) The movement of vacancies is obviously due to the diffusion of atoms which occupy substitutional sites, and in the case of austenite this includes all the alloying elements except carbon, which occupies interstitial sites. The climb of dislocations will be hindered when they meet nanometric size TiN particles. The strengthening contribution of the hard titanium nitrides appears to increase with decreasing particle size, and shows remarkably good theories based on dislocation bypassing.9)
forces \( (F_d) \) for dynamic recrystallisation (DRX), the process would probably be halted and the stress–strain curve would not present a maximum, being an asymptotic curve towards a constant stress value. Nevertheless, unlike in static recrystallisation (SRX), the kinetics of dynamic recrystallisation, which can be expressed by the recrystallised fraction as a function of the strain, depend on other variables such as the temperature, strain rate, austenite grain size, and an intrinsic magnitude of the material related with the mechanism that regulates the process, namely the activation energy.15,18)

The aim of this work has been to study the quantitative influence of the precipitation state of TiN particles, defined by the precipitated volume and size distribution, on the maximum hot strength of austenite and on DRX. In this way it has been attempted to clarify, from a quantitative viewpoint, the influence of TiN particles on the aforementioned magnitudes.

2. Experimental Procedure

Six steels were manufactured in a vacuum induction furnace from electrolytic iron with high purity. The steels may be classified in two groups according to their nitrogen content (Table 1). The first includes three steels with 40 ppm nitrogen (steels S1, S2 and S3) and the second has three steels with 80 ppm nitrogen (steels S4, S5 and S6). Steels S1 and S4 were not alloyed with titanium in order to serve as references in the evaluation of the Ti content’s influence. The titanium and nitrogen contents of the steels guarantee the occurrence of the precipitation of titanium nitride particles in the solid state, since the solubility temperatures are less than 1 500°C, which is approximately the melting temperature of the studied steels. Hot deformation tests were performed in a completely automated torsion machine equipped with test programming and data acquisition software. The test specimens had a gauge length of 50 mm and a radius of 3 mm. During testing the specimens were protected with an argon stream in order to prevent their oxidation. The magnitudes of torsion, the torque and number of revolutions have been transformed into equivalent stress and strain using Von Mises criterion.9) For obtaining flow curves the specimens were initially heated to an austenitisation temperature of 1 300°C for 10 min. The temperature was then lowered to the testing temperature (<1 min) and was maintained at that temperature for a time of no more than two minutes (1 100, 1 000 and 900°C, respectively), applying an equivalent strain of 2.7 which was sufficient to promote the DRX of the steel. The tests were carried out at three different strain rates, 1.45, 3.63 and 5.80 s\(^{-1}\), respectively.

The study of precipitates was performed using Transmission Electron Microscopy (TEM) and Scanning Electron Microscopy (SEM). TEM was used to analyse the finer precipitates by means of the carbon replica extraction technique. SEM was used to analyse the coarser precipitates. The studied precipitation states corresponded to the austenitisation temperature of 1 300°C for 10 min.

3. Results and Discussion

3.1. Analysis of Precipitates

In the analysis of precipitates a size distribution was observed which varied from a few nanometres up to micrometers. For this reason it was necessary to divide them into two groups: fine (size <100 nm) and coarse (size >100 nm). The size selected to distinguish between fine and coarse precipitates was determined bearing in mind that the pinning forces are very weak for precipitates larger than 100 nm.9) Identification of the observed precipitates as titanium nitrides (TiN) was achieved by electron diffraction and X-ray dispersion. Figure 1 presents two TEM micrographs of the precipitation state of steels S2 and S5 at a temperature of 1 300°C, respectively. Figure 2 shows two SEM micrographs for steels S3 and S6, respectively, in which larger precipitates that exceed a size of 1 µm can be

| Table 1. Chemical composition (wt%) of the steels used. |
|---|---|---|---|---|---|---|---|---|
| Steel | C | Mn | Si | P | S | Ti | Al | N | Ti/N |
| S1 | 0.12 | 1.25 | 0.34 | 0.014 | 0.015 | 0 | 0.047 | 0.0040 | 0 |
| S2 | 0.13 | 1.23 | 0.32 | 0.011 | 0.013 | 0.021 | 0.045 | 0.0046 | 4.56 |
| S3 | 0.13 | 1.16 | 0.31 | 0.011 | 0.013 | 0.047 | 0.045 | 0.0043 | 10.93 |
| S4 | 0.10 | 1.13 | 0.30 | 0.007 | 0.012 | 0 | 0.032 | 0.0080 | 0 |
| S5 | 0.12 | 1.18 | 0.29 | 0.007 | 0.012 | 0.018 | 0.034 | 0.0080 | 2.25 |
| S6 | 0.11 | 1.09 | 0.25 | 0.005 | 0.012 | 0.031 | 0.035 | 0.0083 | 3.73 |

Fig. 1. TEM image of fine TiN precipitates corresponding at 1 300°C×10 min: (a) Steel S2; (b) Steel S5.
Figures 3(a)–3(b) show an electron diffraction image and an energy dispersive X-ray analysis spectrum, respectively. The lattice parameter determined from the electron diffraction image reveals a f.c.c. cubic lattice for steel S3 with a value of 4.232 Å, which is identified, in accordance with the reference value of lattice constant, as a titanium nitride.9)

The sizes of the fine and coarse precipitates in each steel were measured. In the case of fine precipitates, the size measurements were carried out on a population of approximately 300, and in some cases more than 400. In this way a frequency diagram was obtained for each steel, like those shown in Figs. 4(a)–4(b), which refer to histogram of steels S2 and S5, respectively. The distribution can be seen to be lognormal in both cases, i.e. it follows a logarithmic law in the form of a Gauss bell. Medina et al. have reported a broad study on the precipitation state of these steels, where the size distribution can be appreciated in each case.20)

The weighted average size of the fine and coarse precipitates, respectively, was measured for Ti steels used (Table 2). On the other hand, the Ti and N contents were calculated, both in solution and in precipitation state, along with the volumetric fractions precipitated at the reheating temperature of 1300°C (Table 3). The calculations were made taking into account the solubility product of the TiN particles1) and the expressions to determine the precipitated volume, respectively.21)

Small precipitates will grow by Oswald ripening according to Lifshitz, Slyozov and Wagner’s expression.22) It has been demonstrated that this expression accurately predicts precipitate growth as a function of the temperature and holding time.20)

3.2. Austenite Grain Size and Pinning Forces

Austenite grain size was determined at 1300°C for 10 min, in the same conditions as reheating of the specimens for the determination of flow curves. The specimens
were quenched by a pressurised water stream through the interior of a tube, achieving a high quenching severity and thus guaranteeing the complete transformation of austenite into martensite. These specimens were also used for the analysis of precipitates, as described in the last paragraph.

The austenite grain size was determined in accordance with ASTM standard E-112. The linear intercept mean length and the average grain size were noted, both being deduced from the ASTM number. The former has the advantage that it can be checked with direct measurements made using the optical microscope. The results obtained are shown in Table 3. More extensive studies, performed on the same steels, have shown that the plotting of the grain size versus Ti/N ratio presents a parabolic shape, whatever the reheating temperature, with a minimum corresponding approximately to a Ti/N ratio of 2, and that this law is interrupted when the temperature is excessively high, giving rise, as in the case of steel S2, to a considerable quantitative jump in the grain size compared with its immediate predecessors in the value of Ti/N, namely steels S5 and S6, and thus diverging from the parabolic shape.20) This jump was due, as is noted in Tables 3 and 4, both to an increase in the average precipitate size and a reduction in the precipitated volume. Although the Ti/N ratio does not cease to be a technical parameter upon which the grain size to a large extent depends, it is closely related with the amount of Ti in solution which is in equilibrium with the TiN particle and therefore, according to the expression of Lifshitz, Slyozov and Wagner, with the precipitate size.20)

The pinning forces opposing the grain growth and the progress of recrystallisation were determined. It is known that the finer precipitates are those that act on grain growth control, and for this reason the value of r inserted in the expressions of pinning forces corresponds to the distribution of fine precipitates. On the other hand, the different expressions used to calculate the pinning forces assume the particles to be spherical, and for this reason it was necessary to convert the average TiN particle size, corresponding to the side of a square, \( l \), to the radius of a circle, \( r \), by means of the approximate equality of areas, so that \( r = l/\sqrt{4/\pi} \). Given that most TiN particles present a rectangular shape in the micrographs obtained by TEM, the average side of their equivalent square was obtained as \( l = \sqrt{l_1 l_2} \), where \( l_1 \) and \( l_2 \) are the lengths of the sides, respectively.

The pinning force exerted by a number of particles per unit of area \( N_s \) on the grain boundary is given by:

\[
F_p = \pi r \gamma N_s \] .... (1)

where \( \gamma \) is the interfacial energy per unit of area (0.8 J/m²) and \( r \) is the average radius of the second phase particles.

For the calculation of pinning forces consideration was made of the expressions of Gladman’s rigid boundary and flexible boundary models, respectively, which are differentiated in the way of calculating \( N_s \) :23–25)

a) Gladman Expression

\[
F_p = \frac{3\pi r^2 \gamma}{2r} \] .... (2)

b) Rigid Boundary Model

\[
F_p = 6\pi r^2 \gamma \] .... (3)

c) Flexible Boundary Model

\[
F_p = 3\pi r^2 \gamma / 2 \] .... (4)

Given that only the fine precipitate fraction has been
taken into account in the calculation of pinning forces, it has been assumed, in accordance with Gao and Baker,²⁶) that this fraction is approximately 50% of the total precipitated volume. The results (Table 4) show that the flexible boundary model yielded greater pinning forces than those calculated in accordance both with the rigid boundary and Gladman.

The greatest pinning forces corresponded to steels S5 and S6, due especially to the smaller average sizes of the precipitates.

### 3.3. Hot Deformation Strength and Influence of the Ti/N Ratio

Flow curves were determined at different temperatures and strain rates. Figures 5 and 6 show flow curves obtained for the indicated tested conditions. In particular, the smaller grain size of steels S5 and S6 compared with the other steels has led DRX to start at a lower strain, since the grain boundaries are preferential nucleation sites for the new recrystallised grains.

The maximum stress value, known as the peak stress, indicates the maximum hot strength of the steel. Table 5 shows the peak stress in increasing order for each of the tested conditions. As can be seen, steels S5 and S6 are those that presented the highest peak stress values in all the straining conditions, which means that these steels will present the higher hot deformation strength and thus the rolling forces to be applied must in turn be greater.

In contrast with what may seem logical, i.e. the higher the microalloying element content (steel S3) the greater the strength, it has been seen that this is not the case. As may be observed, steels S5 and S6, with Ti/N ratios lower than or close to the stoichiometric ratio (3.42) and with the smallest average precipitate sizes, in the distribution of both fine precipitates and coarse precipitates, are those that presented the highest peak stresses.

Another way of measuring the influence of Ti and N on the hot deformation strength is through the activation energy ($Q_d$). The reason for this can be found in the Zener–Hollomon equation¹⁵) which relates peak stress with activation energy. On the other hand, it has been seen that the activation energy is a parameter that is connected with the deformation mechanism and its value is related with the greater or lesser difficulty that the nanometric precipitates pose to the climb of dislocations. For this reason, the precipitation state of the TiN particles not dissolved at the austenitisation (reheating) temperatures must show their influence on the value of $Q_d$.

The value of $Q_d$ was determined for each steel in accordance with the Zener–Hollomon expression, given by,

$$Z = \dot{\varepsilon} \exp \left( \frac{Q_d}{RT} \right) = A (\sinh \alpha \sigma)^n$$

where, $\dot{\varepsilon}$ is the strain rate and $\sigma$ is the equivalent stress.

Medina et al. determined experimentally and in the corresponding statistical treatment that the most probable values for $\alpha$ and $n$ in low alloy and microalloyed steels are 0.0118 MPa⁻¹ and 4.458, respectively.¹⁵)

Figures 7 and 8 show different examples of the determination of $Q_d$ corresponding to steels S3 and S5. It is seen that the value of $Q_d$ given by the slope multiplied by $n$ and $R$ (universal gas constant) does not depend on the austenite grain size or strain rate, coinciding with other results reported previously.⁵,²⁷) It is noted that the peak stress ($\sigma_p$) has been used as the stress in the Zener–Hollomon parameter and not the steady state stress ($\sigma_{ss}$). In tests carried out with high values of $Z$, in other words, at high strain rates and relatively low temperatures, close to $\Lambda$, it is advisable to use $\sigma_{ss}$ instead of $\sigma_p$ as the former is reached at relatively low strains ($\varepsilon_{ss}$) and the in-

<table>
<thead>
<tr>
<th>Steel</th>
<th>Radius of particle (nm)</th>
<th>Precipitated volume 50%</th>
<th>Gladman $\gamma f/2\varepsilon_r$ (MPa)</th>
<th>Model of rigid boundary $3 f/2\varepsilon_r$ (MPa)</th>
<th>Model of flexible boundary $3 f/2\varepsilon_r$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S2</td>
<td>12.98</td>
<td>0.88 x 10⁻³</td>
<td>0.008</td>
<td>0.010</td>
<td>0.116</td>
</tr>
<tr>
<td>S3</td>
<td>36.79</td>
<td>1.25 x 10⁻³</td>
<td>0.004</td>
<td>0.005</td>
<td>0.052</td>
</tr>
<tr>
<td>S5</td>
<td>7.79</td>
<td>1.21 x 10⁻³</td>
<td>0.019</td>
<td>0.024</td>
<td>0.240</td>
</tr>
<tr>
<td>S6</td>
<td>7.96</td>
<td>2.02 x 10⁻³</td>
<td>0.030</td>
<td>0.039</td>
<td>0.330</td>
</tr>
</tbody>
</table>

Table 5. Increasing order of peak stress ($\sigma_p$) for steels used.

<table>
<thead>
<tr>
<th>Strain rate ($\dot{\varepsilon}$), s⁻¹</th>
<th>Test temperature, °C</th>
<th>Increasing magnitude of $Q_d$</th>
</tr>
</thead>
<tbody>
<tr>
<td>5.80</td>
<td>900</td>
<td>S3=S1=S2=S4=S5=S6</td>
</tr>
<tr>
<td>3.63</td>
<td>1000</td>
<td>S2=S1=S3=S4=S6=S5=S5</td>
</tr>
<tr>
<td>1.45</td>
<td>1100</td>
<td>S1=S2=S4=S6=S5=S5=S6</td>
</tr>
<tr>
<td></td>
<td>1000</td>
<td>S3=S2=S1=S4=S6=S5=S5</td>
</tr>
<tr>
<td></td>
<td>1000</td>
<td>S3=S2=S1=S4=S6=S5=S6</td>
</tr>
</tbody>
</table>

Fig. 5. Flow curves for steels S1, S2 and S3.

Fig. 6. Flow curves for steels S4, S5 and S6.
crease in the specimen temperature is still small. However, the steady state is reached at considerably greater strains where the temperature has undergone a significant increase and therefore the stress–strain curve cannot be considered isothermal, and the specimen experiences an increase in temperature and thus a certain reduction in $\sigma_{\text{ss}}$.15,16)

The values found for $Q_d$ are shown in Table 6, and are very similar to others reported for steels containing Ti as a microalloying element.15) The good approximation of these values to the energy for self-diffusion in Fe indicates that the mechanism which governs the plastic flow of austenite is principally the climb of edge dislocations, which depends in turn on the diffusion of vacancies affected by the presence of particles and solutes.15,28,29)

With the aim of measuring the influence of the precipitation state, or in its defect the Ti/N ratio, on the activation energy, the increase in the activation energy ($\Delta Q_d$) of each steel containing Ti compared with its reference steel (without Ti) was plotted versus the Ti/N ratio (Fig. 9). In order to allow the plotting of the parabola by regression of the data, with a logarithmic scale for the abscissa axis, it was necessary to suppose that the reference steels S1 and S4 had a Ti content of 0.001 and 0.002, respectively, which in practical terms is equivalent to a zero content, since these amounts have no influence on the peak stress and on the DRX, and consequently on the value of $Q_d$. The equation obtained by regression for the parabola was as follows,

$$
\Delta Q_d (\text{J/mol}) = -7.678 \ln \frac{\text{Ti\%}}{\text{N\%}}^2 + 3.990 \ln \frac{\text{Ti\%}}{\text{N\%}} + 20.638
$$

(6)

It should be noted that expression (6) does not predict the increase in the activation energy as a function of the Ti/N ratio in every case, since this will depend on the precipitation state and in turn on the Ti and N contents, and on the temperature and holding time. Nevertheless, its significance is highly indicative, and supposes that the precipitation state may in principle be expressed by the Ti/N ratio and that the maximum value obtained for $\Delta Q_d$ given by the derivative of Eq. (6), corresponds to a Ti/N value close to 1.3. The coefficients of expression (6) would obviously change if the reheating temperature were different, although it is foreseeable that the maximum value of $\Delta Q_d$ would be obtained for the same Ti/N ratio, or very similar. It is also foreseeable that with higher Ti and N contents, but maintaining a ratio of close to 1.3, the values obtained for $\Delta Q_d$ would be somewhat higher, since the pinning forces would also be considerably higher. Nevertheless, this increase in the Ti and N contents would be limited to the formation of precipitates taking place in the solid state.

The values found for $Q_d$ are very low and this means that even in the case of the finest precipitation, which corresponds to steel S3, or in the case of steel S6, which has the highest pinning forces, the precipitation state of the TiN particles is not capable of notably increasing the peak stress. Furthermore, it is also not capable of appreciably delaying the start of DRX, and the grain size is the variable that has most influenced the bringing forward or the delaying of the start of DRX.

The magnitude that best represents the precipitation state is $F_p$, whose expression includes the average precipitate size and the precipitated volume. However, the graphic representation of $\Delta Q_d$ versus $F_p$, as can be seen, does not show a clear tendency due to the role that the precipitates may

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**Table 6.** Values of experimental activation energy ($Q_d$), increment values for deformation ($\Delta Q_d$) and theoretical values predicted by expressions (6) plus (9).

<table>
<thead>
<tr>
<th>Steel</th>
<th>Experimental $Q_d$ (kJ/mol)</th>
<th>Experimental $\Delta Q_d$ (kJ/mol)</th>
<th>Predicted $Q_d$ (kJ/mol)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S1</td>
<td>282</td>
<td>0</td>
<td>279</td>
</tr>
<tr>
<td>S2</td>
<td>285</td>
<td>3</td>
<td>288</td>
</tr>
<tr>
<td>S3</td>
<td>270</td>
<td>-12</td>
<td>265</td>
</tr>
<tr>
<td>S4</td>
<td>278</td>
<td>0</td>
<td>278</td>
</tr>
<tr>
<td>S5</td>
<td>300</td>
<td>22</td>
<td>296</td>
</tr>
<tr>
<td>S6</td>
<td>292</td>
<td>14</td>
<td>289</td>
</tr>
</tbody>
</table>

---

**Fig. 7.** Sinh($\sigma$) against the reciprocal of the temperature for steel S3.

**Fig. 8.** Sinh($\sigma$) against the reciprocal of the temperature for steel S5.

**Fig. 9.** Increment of activation energy ($\Delta Q_d$) for every steel with regard to its reference steel.
play as nucleation sites for the recrystallised grains.

### 3.4. Driving Forces for DRX

The flow curves obtained in this work show that DRX takes place in every case. In all the curves a maximum stress value is reached (\(\sigma_p\)), after which the stress decreases until reaching a constant value (\(\sigma_{st}\)). This means that the driving forces of recrystallisation must be higher than the pinning forces. With the aim of verifying this hypothesis, the recrystallisation forces \((F_r)\) were calculated for the six steels using the following expression,\(^{30}\)

\[
F_r = \mu b^2 \Delta p / 2
\]

where \(\mu\) is the shear modulus \((4 \times 10^4 \text{ MN/m}^2)\), \(b\) is Burger’s vector \((2 \times 10^{-10} \text{ m})\) and \(\Delta p\) is the variation in the density of dislocations associated with the movement of the recrystallisation front in the deformed zone. Although this equation does not directly reflect the influence of the temperature, its influence is contemplated in \(\Delta p\).

On the other hand, the value of \(\Delta p\) during hot working is related with the increase in flow stress,\(^{31}\)

\[
\Delta \sigma = 0.2 \mu b \sqrt{\Delta p}
\]

Using the flow curves, \(\Delta p\) should be measured up to the start of dynamic recrystallisation, which takes place at a strain of somewhat less than \(\varepsilon_p\), as will be seen below. Nevertheless, as \(\varepsilon_p\) is the maximum stress reached and and it represents the highest dislocation density, it seems more logical to use this value as the stress more representative in order to calculate \(\Delta p\) in the Eq. (8) for dynamic recrystallisation.

**Figure 10** shows the value of \(F_r\) for each steel, calculated in accordance with Eq. (7) versus the deformation temperature and for a strain rate of 3.63 s\(^{-1}\). As the temperature rises the driving forces decrease as a consequence of the reduction in the peak stress and thus in the density of dislocations. Comparing the results for the different steels it is seen that the highest \(F_r\) values correspond to steels S5 and S6 as a consequence of their greater peak stress.

In the same way, **Fig. 11** shows the value of \(F_r\) versus the strain rate at a deformation temperature of 900°C. An increase in the strain rate causes an increase in the peak stress and in the density of dislocations, and thus in the driving force. This figure also shows that the greatest driving forces correspond to steels S5 and S6.

The values reached for \(F_r\) in all cases exceed the pinning forces (Table 4) by more than two orders of magnitude, which demonstrates that in these steels where precipitation is not strain-induced, recrystallisation always progresses and is not hindered by the presence of the precipitates. However, in microalloyed steels with Nb or V where the precipitation is strain-induced, very fine precipitate distributions are reached which may cause the delay or even completely inhibit dynamic recrystallisation.\(^{32}\)

### 3.5. Models of Dynamic Recrystallisation and Flow Curves

Medina et al. reported a model to predict the kinetics of the dynamic recrystallisation process and the flow curves of low alloy and microalloyed steels.\(^{15,16,33,34}\) The expressions that constitute this model allow the recrystallised fraction to be determined as a function of all the external variables that intervene in hot deformation, such as the strain, strain rate and temperature, and of the internal or intrinsic variables of austenite, such as the chemical composition and austenite grain size. These expressions are set out in the Appendix.

The model is based on Avrami’s law which expresses the recrystallised fraction \((X_d)\) as a function of the strain once the critical strain has been reached \((\varepsilon_p)\), which is given by,

\[
X_d = 1 - \exp \left( -k \left( \frac{\varepsilon - a \varepsilon_p}{\varepsilon_p} \right)^m \right)
\]

valid for values of \(\varepsilon > a \varepsilon_p\), \(a = 0.95\), and where \(k\) and \(m\) are given as a function of the dimensionless parameter \(Z/A\) deduced from Eq. (5).

The most original aspect of the reported model would be the introduction of an expression for \(Z/A\) obtained by regression of the experimental data and which, in view of its importance in the present work, is transcribed below,\(^{15}\)

\[
Q_d (J/mol) = 267 000 - 2535.5(C\%) + 1010(Mn\%) + 33 620.7(Si\%) + 35 651.3(Mo\%) + 31 673.5(V\%) + 70 729.8(Nb\%)^{0.50565(...(10)
\]

where the brackets represent the weight percentage of each element.

Expression (10) was deduced for tests performed with a total of 18 steels and when all the alloying elements, including V and Nb, were in solution, since the reheating
temperatures were sufficiently high (> 1 200°C) to dissolve the precipitates. The strain rates between 0.54 and 5.22 s⁻¹ prevented the occurrence of precipitation during deformation.

It should be noted that if Eq. (9) is extrapolated to pure iron in the austenitic phase (Feₐ), Qₐ would yield a value of 267 kJ/mol, which coincides approximately with the energy for self-diffusion.¹⁵

Therefore, Eq. (10) serves to quantify the drag solute effect. However, as has been mentioned above, Ti is precipitated at the reheating temperatures and thus the value of Qₐ for these steels is determined by adding to expression (10) the increase given (∆Qₐ) by expression (6). The calculated values of Qₐ are presented in Table 6, and their comparison with the experimental values shows the good approximation that is obtained with these expressions.

The set of equations used to predict the flow curves and at the same time the dynamic recrystallisation kinetics are included, as has already been mentioned, in the Appendix, where they have been noted following approximately their order of calculation.

Some of the model’s expressions, such as A, B, B' and εₕ have been newly determined for the studied steels, since the previously reported expressions were determined for commercial steels and those considered in the present work were manufactured from electrolytic steel. The impurities contained in commercial electric steels (S, Cu, Mo, P) cause a slight increase in the peak stress compared with the steels studied here. The new expressions differ very slightly from the old ones,¹⁵,¹⁶,³³,³⁴ but yield considerably better predictions of the flow curves for these steels.

Figures 12 and 13 present examples of the model’s application in the determination of flow curves. Attention is drawn to the good approximation between the theoretical and experimental flow curves. The part of the curve between the peak stress and the steady state corresponds approximately to the dynamic recrystallisation kinetics. In the present model, dynamic recrystallisation begins when a εₕ value equal to 0.95εₕ is reached, a value that has been identified as the most probable in the tests performed with eighteen steels and different deformation conditions (strain rate, deformation temperature and reheating temperature), although somewhat higher than the ratio found by other authors.²⁷,³⁵,³⁶

4. Conclusions

(1) In microalloyed steels an increase in the titanium content does not necessarily mean that the deformation strength also increases.

(2) A relationship has been found between the precipitation state and the maximum strength of the steels during hot deformation. The influence of the precipitation state is seen to affect the value of the activation energy, which is maximum when the Ti/N ratio reaches a value close to 1.3.

(3) The increase obtained both for the peak stress and for the activation energy in steels with the finest precipitate distribution is relatively small and barely represents an increase in the rolling forces.

(4) In the steels containing Ti the driving forces of dynamic recrystallisation are approximately two orders of magnitude greater than the pinning forces, and therefore the finer precipitation state (steels S5 and S6) is not able to impede the progress of dynamic recrystallisation.

(5) The most notable influence of Ti is the grain refinement at the reheating temperatures and therefore the reduction in the critical strain necessary for dynamic recrystallisation to begin.

(6) The model that is used predicts flow curves and dynamic recrystallisation kinetics with good approximation.

Acknowledgement

Financial support of this work by the ECSC is gratefully acknowledged.

REFERENCES

8) C. S. Smith: Trans. AIME, 175 (1948), 15.
increase in comparison with the reference steel.

\[ A(s^{-1}) = \exp(6 \times 10^{-5}Q) \]  

(b) Flow Curves

For the hardening and dynamic recovery region, the following expression is given:

\[ \sigma_z = B[1 - \exp(-C\varepsilon)]^p \]  

For the reduction of stress for strains greater than \( \varepsilon_p \) due to the effect of dynamic recrystallization is given by:

\[ \Delta \sigma = B' \left[ 1 - \exp \left( -k \left( \frac{\varepsilon - \varepsilon_p}{\varepsilon_p} \right)^m \right) \right] \]

\[ \varepsilon > \varepsilon_p; \quad a = 0.95 \]

The flow curve is given by the resulting expression \( (\sigma_z - \Delta \sigma) \) and the different parameters are as follows:

\[ B(MPa) = \left( 9.826 + 0.649 \ln \frac{Z}{A} \right)^2 \]  

\[ C = 3.920 \left( \frac{Z}{A} \right)^{0.0592} \]  

\[ p = 0.345 \exp \left( 0.0139 \sqrt{\frac{Z}{A}} \right) \]  

\[ B'(MPa) = 29.5 \left( \frac{Z}{A} \right)^{0.146} \]  

\[ k = 0.597 \exp \left( 1.233 \left( \frac{Z}{A} \right)^{-0.5} \right) \]  

\[ m = 1.09 \exp \left( 0.0264 \sqrt{\frac{Z}{A}} \right) \]

Peak strain is given by:

Steeels S1, S2, S3 and S4,

\[ \varepsilon_p = 0.102D^{0.223} \left( \frac{Z}{A} \right)^{0.164} \]  

Steeles S5 and S6,

\[ \varepsilon_p = 0.102D^{0.223} \left( \frac{Z}{A} \right)^{0.116} \]

(c) Dynamic Recrystallisation Kinetics

From expression (9), XRD is given as follows:

\[ X_d = 1 - \exp \left( -k \left( \frac{\varepsilon - \varepsilon_p}{\varepsilon_p} \right)^m \right) \]