FROM HETEROGENEOUS TO HOMOGENEOUS NUCLEATION FOR PRECIPITATION IN AUSTENITE OF MICROALLOYED STEELS

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Abstract

This paper studies the influence of strain on precipitate nucleation in austenite for three microalloyed steels with different microalloying element (Nb, V) contents. Precipitation start-time-temperature (PSTT) diagrams have been determined by means of hot torsion tests and nucleation periods have been measured at strains of 0.20 and 0.35, respectively. The increase in the dislocation density caused by the strain has been calculated for both strains, and the driving forces for precipitation have also been calculated. The results show that the influence of the strain on the nucleation time ($t_{0.05}$) is dependent not only on the strain magnitude but also on the driving force for precipitation. When the driving force is high, or low in absolute terms, the influence of the strain, i.e. the increase in the dislocation density, gives rise to a notable reduction in the $t_{0.05}$ value due to heterogeneous nucleation on the dislocation nodes. In contrast, when the driving force is low, or high in absolute terms, the influence of the strain on $t_{0.05}$ decreases considerably and the nucleation of precipitates becomes preponderantly homogeneous. Therefore, the driving force value is responsible for the transition from heterogeneous nucleation to homogenous nucleation.

Keywords: Microalloyed steels; Precipitation; Heterogeneous/homogeneous nucleation; driving forces.

1. Introduction

The most widely used method to achieve grain refinement by the thermomechanical controlled processing (TMCP) of steels is to maximize the number density of nucleation sites in unrecrystallised austenite in order to produce a fine ferrite grain size after phase transformation [1]. It is well known that the static recrystallisation of austenite in microalloyed steels is different before and after strain-induced precipitation. Before, all the elements are in solution and recrystallisation kinetics occur in the same way as in low alloy steels, whereby the various alloying elements contribute to delaying recrystallisation to a greater or lesser degree [2-4]. As the temperature drops, a critical temperature is reached, after which static recrystallisation is momentarily inhibited by the effect of strain-induced precipitates. This momentary inhibition of recrystallisation appears as a plateau on the recrystallised fraction versus time curves [5]. When the end of the plateau is reached, recrystallisation recommences as the coarsening of precipitates consequently reduces the pinning forces against driving forces. After the plateau, the superiority of driving forces for recrystallisation over pinning forces is about two orders of magnitude [6]. In Nb-microalloyed steels the Nb atoms/NbCN precipitates retard the recrystallisation of deformed austenite [7-10]. However, V-microalloyed steels also exhibit grain refinement through intragranular nucleation of ferrite on VN precipitates partly due to low lattice mismatch of VN with ferrite [11,12]. The addition of N in V microalloyed steel stimulates the precipitation of VN particles and increases their volume fraction [13-16].

According to Dutta and Sellars [17] the density of preferential nucleation sites for precipitation in deformed austenite is expected to be sensitive to the density and arrangement of dislocations, and therefore to the conditions of the prior deformation expressed in terms of the strain, strain rate and absolute temperature of deformation. In

addition to the dislocations, grain boundaries are sites known as classic sources of the heterogeneous nucleation of precipitation [18]. The lattice parameter of the precipitate is 20-25% greater than that of the matrix, and a flux of vacancies to the precipitated particles is required in order to accommodate the internal stresses arising from the growth of these particles. Such vacancy fluxes are provided by hot deformation processes, and the dislocation density is also increased, thereby providing an increased number of nucleation sites [19]. Bhadeshia and Honeycombe have also pointed out that grain boundaries and dislocations are highly preferred nucleation sites [20].

Most of the models that predict the nucleation time as a function of the different variables intervening in hot deformation (temperature, strain, microalloying content, strain rate) determine that the nucleation time is inversely proportional to the strain, and give the exponent of the strain a fixed value [17, 21, 22]. Thus, the nucleation time depends on the increase in the dislocation density resulting from the applied strain and this relationship is not affected by other variables like the chemical composition. In other words, for the aforementioned authors, when austenite is deformed, heterogeneous nucleation on the dislocations is preponderant. The present work seeks to demonstrate that strain induced precipitation in the austenite of microalloyed steels is transformed from heterogeneous nucleation on dislocation nodes to homogenous nucleation as the microalloying content rises.

2. Materials and experimental procedure

Three steels were manufactured by Electroslag Remelting (ESR) in a laboratory unit capable of producing 30 kg ingots. The three steels have been chosen from a larger group of microalloyed steels on the basis of their low, medium and relatively high microalloying content, either Nb (steels N1 and N8) or V (steel V8) (**Table 1**).

Torsions specimens were prepared with a gauge length of 50 mm and a diameter of 6 mm. The reheating temperature before torsion deformation varied according to whether the steel was microalloyed with V or Nb, as the solubility temperature of the precipitates depends on their nature and on the precipitate-forming element content. To ensure that the testing temperatures corresponded to the austenitic phase, critical transformation temperatures (A_{r3}) were measured by dilatometry at a cooling rate of 0.2 K s⁻¹ (**Table 1**). Once the specimens had been reheated, the temperature was rapidly lowered to the testing temperature, where it was held for a time of no more than 1 min to prevent precipitation taking place before the strain was applied.

Steel	C	Si	Mn	Al	X _i	N	A _{r3} , °C
N1	0.11	0.24	1.23	0.002	Nb=0.041	0.0112	786
N8	0.20	0.20	1.0	0.006	Nb=0.007	0.0056	770
V8	0.37	0.24	1.42	0.012	V=0.120	0.0190	721

Table 1. Chemical composition (mass %), transformation critical temperature (A_{r3} , at 0.2 K/s), being X_i=V, Nb.

The austenite grain size at the reheating temperature was determined by quenching of the specimens applying standard ASTM E-112 (**Table 2**). The parameters of torsion (torque, number of revolutions) and the equivalent parameters of tension (stress, strain) were related according to Von Mises criterion [23].

Steel	ε	$\dot{\varepsilon}$ (s ⁻¹)	RT (°C)	D, μ m
N1	0.20/0.35	3.63	1230	122
N8	0.20/0.35	3.63	1250	140
V8	0.20/0.35	3.63	1200	157

Table 2. Test conditions: strain (ε), strain rate ($\dot{\varepsilon}$), reheating temperature (RT) and austenite grain size (D) at reheating temperature of the steels used.

The testing temperatures varied between 1100 and 800°C for the vanadium steel and between 1150 and 850°C for the niobium steels. In all cases the testing temperatures were set as the recrystallised fraction was determined and the recrystallised fraction curves were drawn, so that the curves finally obtained would include curves where strain-induced precipitation had taken place and curves where it had not, as is discussed below. The applied strains were 0.20 and 0.35, which were insufficient to promote dynamic recrystallisation [24] and the strain rate was always 3.63 s⁻¹ (**Table 2**). The recrystallised fraction (X_d) was determined using the "back extrapolation" method [25,26]. The study of precipitates was carried out using transmission electron microscopy (TEM).

3. Precipitate nucleation

The nucleation rate is obtained from the classic theory of nucleation modified by Zeldovich, Kampmann and Wagner [27-29] as:

$$\frac{dN}{dt} = N_0 Z \beta' \exp\left(-\frac{\Delta G}{kT}\right) \exp\left(-\frac{\tau}{t}\right) \quad (1)$$

where, N_0 represents the number of available sites for heterogeneous nucleation, Z is de Zeldovich non-equilibrium factor, k is the Boltzman constant, β' is the rate at which the atoms are being added to the critical nucleus or atomic impingement rate, T is the absolute temperature, τ is the incubation time and ΔG is the Gibbs energy of formation of a critical spherical nucleus of radius R_c .

The integration of equation (1) would give the number of precipitates (N) per unit of volume. Some authors have proposed new expressions based on this expression which can be applied to deformation induced precipitation in microalloyed steels. The most important reference to predict strain induced precipitation nucleation as a function of hot deformation variables (strain, strain rate, temperature) is perhaps the expression

given by Dutta and Sellars [30] for a time corresponding to 5% of the precipitated volume ($t_{0.05}$), which in practical terms can be taken as the nucleation time for precipitation. These authors state that the density of preferential nucleation sites in deformed austenite is expected to be sensitive to the density and arrangement of dislocations, and therefore to the conditions of the prior deformation expressed in terms of the aforementioned variables. Dutta and Sellars's model was applied to Nb-microalloyed steels and takes into account the Nb-content, strain (ε), strain rate ($\dot{\varepsilon}$) and temperature (T), and the expression is as follows:

$$t_{0.05} = A(Nb)^{-1} \varepsilon^{-1} Z^{-0.5} \exp\left(\frac{270000}{RT}\right) \exp\left[\frac{B}{T^3 (\ln k_s)^2}\right] \quad (2)$$

where, ($Z = \dot{\varepsilon} \exp\frac{Q_d}{RT}$) is the Zener-Hollomon parameter. The good approximation of the activation energy (Q_d) to the energy for self-diffusion in Fe_γ 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 [31]. Nb is the percentage of this element, A and B are constants and k_s is the supersaturation ratio parameter.

A new model for strain induced precipitation based on expression (2) has recently been published for any microalloyed steel and the following equation has been proposed [32]:

$$t_{0.05} = A \varepsilon^{-\beta} \dot{\varepsilon}^{-r} D^s \exp\left(\frac{Q_d}{RT}\right) \exp\left[\frac{B}{T^3 (\ln k_s)^2}\right] \quad (3)$$

Equation (3) has been proposed because the results showed that the exponent of the strain (β) is not a constant and is not equal to 1 but is dependent on the chemical composition, particularly on the microalloying element content. It was also seen that the

austenite grain size (D) influences the parameter $t_{0.05}$. Furthermore, as will be seen below, the determination of the supersaturation ratio parameter (k_s) was not a constant, and the mathematical expression found shows that it depends on the chemical composition of the steel, in particular the microalloying type and content. The use of nineteen steels with different Nb, V and Ti contents allowed values and expressions to be found for the different parameters in equation (3) making it possible to predict with good approximation the experimental values found for $t_{0.05}$ in any microalloyed steel.

In both equations (2) and (3) the role of ε is to increase the dislocation density and thus the number of nodes in the dislocation network, $N_0=0.5\Delta\rho^{1.5}$ included in equation (1), being $\Delta\rho=(\Delta\sigma/0.2\mu b)^2$ the variation in the dislocation density associated with the recrystallisation front movement in the deformed zone at the start of precipitation [5], $\Delta\sigma$ is the difference between the flow stress and yield stress at the deformation temperature, b is the Burgers vector and μ is the shear modulus. The exponent of the strain in equation (3) was determined experimentally, having obtained the following expression [32]:

$$\beta = 1.96 \left[1 - \exp \left(- 3.994 \times 10^{-2} \left(\frac{1}{w} \right)^{0.813} \right) \right] \quad (4)$$

where, w is the microalloying element content (wt%).

Equation (4) means that the strain starts to influence the precipitation kinetics when the microalloying element content is less than a certain amount, which in practical terms could be approximately 0.5 (wt%). At the same time, the maximum value of β should be 1.96 (wt%). Note that the parameter β is a function of the microalloying content without intervention of the N or C content. This is due to the fact that the diffusion coefficient of Nb and V are several orders of magnitude smaller than those of N and C, and therefore are those that govern precipitate formation [33].

4. Experimental results

4.1. Precipitation start-time-temperature (PSTT) diagrams for two strains

The recrystallised fraction, determined by applying the back extrapolation method, was drawn against time for each testing temperature. The shape of the recrystallised fraction versus time curves were similar for the microalloyed steels used, it being observed that some curves display a plateau caused by the formation of precipitates which momentarily inhibit the progress of recrystallisation [34-37]. The plateau is caused by strain-induced precipitation, as occurs in all microalloyed steels, and the start and end of the plateau are identified approximately with the start and end of strain-induced precipitation, respectively. While the start of the plateau seems to coincide with good exactness with the start of strain-induced precipitation, the end of the plateau may coincide with important growth in the average size of precipitates, which become incapable of inhibiting recrystallisation [38]. The plateau is not unlimited, i.e., precipitation does not permanently inhibit recrystallisation, and recrystallisation again progresses until is complete, following a graphic plot similar to that recorded before the formation of the plateau. As an example, Fig. 1 shows the recrystallised fraction of the steel N1 for a strain of 0.35.

The recrystallised fraction versus time curves were used to plot start of precipitation start-time-temperature (PSTT) diagrams for both strains. The points defining the start of the plateau were taken to plot the curves for the start of precipitation (P_s). In this way, SPTT diagrams were obtained for the two strains of 0.20 and 0.35. Figures 2-4 correspond to steels N1 and N8 and V8, respectively. At the moment when precipitation starts, whatever the temperature (P_s curve), it is assumed that the precipitated fraction corresponds to a value of 5%.

The reheating temperature was 1230°C for steel N1, 1250°C for steel N8 and 1200°C for steel V8. In the last case it was seen that VN type particles were precipitated, and therefore the temperature of 1200°C was sufficient to place them in solution. In the two steels containing Nb, the precipitated particles were carbonitride types, and for this reason one of the prototypes with the highest probability of formation, $\text{NbC}_{0.7}\text{N}_{0.2}$, was chosen, in agreement with Turkdogan [39].

Solubility temperatures (T_s) were calculated for the aforementioned precipitates and the values of nose temperature (T_N) and minimum incubation time (t_N) corresponding to the curve nose were measured from the PSTT diagrams and are set out in Table 3.

4.2. Influence of strain

P_s values corresponding to the nose of the curves were determined from the SPTT diagrams, and both were seen to decrease as the microalloying element content or the strain increased. According to expression (3), the time ($t_{0.05}$) is related with the strain (ε) in accordance with the following expression:

$$\ln t_{0.05} \propto -\beta \ln \varepsilon \quad (5)$$

According to expression (5), and accepting that $t_{0.05}$ and P_s may be assumed to be approximately equal, the values of β were determined. The value taken for $t_{0.05}$ corresponding to each strain has been the minimum nucleation time (t_N) corresponding to the nose of the P_s curve (Table 3).

The graphic representation of β in equation (4) and the values obtained for the three steels used (Table 3) versus the microalloying element content clearly shows that this parameter depends on the microalloying content and the influence of its nature can be practically disregarded (Figure 5).

Steel	ε	T_s (K)	T_N (K)	t_N (s)	β
N1	0.20	1438.4	1252	24	0.72
	0.35			16	
N8	0.20	1309.9	1184	58	1.65
	0.35			23	
V8	0.20	1399.5	1151.5	19	0.54
	0.35			14	

Table 3. Solubility temperature (T_s) according to Turkdogan [12], experimental and predicted nose temperature (T_N) and minimum incubation time t_N . $\dot{\varepsilon} = 3.63 \text{ s}^{-1}$.

The value of β will indicate what type of nucleation will be preponderant. For high β values it is obvious that the preponderant nucleation will be heterogeneous nucleation on dislocations produced by the deformation. At low β values the nucleation should be homogeneous due to the relatively high percentage of microalloying.

5. Calculation of dislocations density and number of nodes

In equation (1), $N_0 = 0.5\Delta\rho^{1.5}$ is the number of available sites (nodes) for heterogeneous nucleation in the dislocation network, being $\Delta\rho = (\Delta\sigma/0.2\mu b)^2$ the variation in the density of dislocations associated with the recrystallisation front movement in the deformed zone at the start of precipitation [40], $\Delta\sigma$ is the difference between the flow stress and yield stress at the deformation temperature, b is the Burgers vector ($2.59 \times 10^{-10} \text{ m}$) and μ is the shear modulus ($4 \times 10^4 \text{ MN/m}^2$).

The flow stress increment ($\Delta\sigma$) has been calculated using the model reported by Medina and Hernández [41] which facilitates the calculation of flow stress. The dislocation density has been calculated at the nose temperature of the P_s curves corresponding to strains of 0.20 and 0.35, respectively. The calculated values of $\Delta\sigma$, $\Delta\rho$ and N_0 are shown in Table 4. The $\Delta\rho$ values are of the same order for the three steels and increase when the strain goes

from 0.20 to 0.35 as a consequence of the increase in flow stress. When the austenite is not deformed the dislocation density is approximately 10^{12} m^{-2} [42], which indicates that deformation has led to an increase in $\Delta\rho$ of approximately three orders of magnitude. Bearing in mind that the nose of the P_s curve corresponds approximately to a 50% recrystallised fraction [9], the dislocation density corresponding to the curve nose will be given by $10^{12+0.5\Delta\rho}$ (Table 3).

Steel	ε	T_N (K)	$\Delta\sigma$ (MPa)	$\Delta\rho$ (m^{-2})	$0.5 \times \Delta\rho$ (m^{-2})	N_0 (m^{-3})
N1	0.20	1252	143.0	8.81×10^{14}	4.42×10^{14}	4.64×10^{21}
	0.35		159.4	1.09×10^{15}	5.49×10^{14}	6.42×10^{21}
N8	0.20	1184	154.1	1.02×10^{15}	5.13×10^{14}	5.81×10^{21}
	0.35		173.7	1.30×10^{15}	6.51×10^{14}	8.30×10^{21}
V8	0.20	1151.5	158.6	1.08×10^{15}	5.43×10^{14}	6.33×10^{21}
	0.35		180.3	1.40×10^{15}	7.01×10^{14}	9.29×10^{21}

Table 4. Calculated values of nodes (N_0), flow stress ($\Delta\sigma$) and dislocation density ($\Delta\rho$).

The values of $\Delta\rho$ and N_0 are of the same order for the three steels, although a slight increase is seen from steel N1 followed by steels N8 and V8. This slight increase is due to the fact that the T_N temperature decreases from one steel to another, reaching higher $\Delta\sigma$ values from one steel to another. Thus there is no relationship between the values of β and those of $\Delta\rho$ and N_0 . The lowest value of β was 0.54 (steel V8) and corresponds precisely to the highest values of $\Delta\rho$. These results indicate that the nucleation of precipitates does not depend only on the strain applied, but that there are other conditions which modify the influence of the strain.

6. Driving forces for precipitation

The driving force for precipitation is defined as the free energy change that accompanies the formation of a unit volume of precipitate from the supersaturated matrix and is given by [43]:

$$\Delta G_v (Jm^{-3}) = -\frac{R_g T}{V_m} \ln \left(\frac{C_x C_y}{C_x^e C_y^e} \right) \quad (6)$$

where C_x and C_y are the instantaneous concentrations of V or Nb and N or/and C, respectively, C_x^e and C_y^e are the equilibrium concentrations at the deformation temperature, V_m is the molar volume of the precipitate species, R_g is the universal gas constant and T is the deformation absolute temperature.

According to Turkdogan [39], the supersaturation ratio defined by $K_s = \frac{C_x C_y}{C_x^e C_y^e}$ will be:

Nb-Steels:

$$K_s = \frac{[Nb][C]^{0.7}[N]^{0.2}}{\left[10^{4.12 - \frac{9450}{T}} \right]} \quad (7)$$

V-Steels:

$$K_s = \frac{[V][N]}{\left[10^{2.86 - \frac{7700}{T}} \right]} \quad (8)$$

It has been considered that Nb forms carbonitrides and V forms nitrides with the stoichiometry used by Turkdogan. In order to calculate expression (6), the values used for V_m were 1.305×10^{-5} and 1.052×10^{-5} (m^3/mol) for Nb-carbonitride and V-nitrides precipitates, respectively [40]. In this way, ΔG_v was calculated at the nose temperature (T_N) of the P_s curve. The values calculated of ΔG_v are displayed in Table 5.

Steel	T_N (K)	ΔG_v (J/m ³)
N1	1252	-1.80×10^9
N8	1184	-1.33×10^9
V8	1151.5	-1.48×10^9

Table 5. Values of driving force for precipitation (ΔG_v) for steels used.

Graphic representation of parameter β versus ΔG_v clearly shows that the influence of the strain depends on the driving force for precipitation (Fig. 6). In other words, the increase in the dislocation density does not alone guarantee that nucleation will take place on the dislocation nodes and this will depend on the value of ΔG_v at the deformation temperature. If the ΔG_v value is high, but small in absolute terms as it is negative, nucleation will be preponderantly heterogeneous on the dislocation nodes. In contrast, if the ΔG_v value is small, but large in absolute terms, nucleation will become preponderantly homogeneous. Given that this occurs whatever the austenite grain size [32], the loss of influence of heterogeneous nucleation on the dislocation nodes is due to a greater occurrence of homogeneous nucleation. According to equation (3), the grain boundaries continue to act as heterogeneous nucleation sites, and it has not yet been demonstrated whether their potential to promote heterogeneous nucleation will be reduced by the increase in driving forces for precipitation.

7. Conclusions

When the driving force for precipitation is high, or low in absolute terms, the strain as a promoter of the increase in the dislocation density exerts an important influence on the nucleation time ($t_{0.05}$), whose β exponent in equation (3) reaches values of close to 2. In contrast, when the driving force is low, or high in absolute terms, the strain loses its

influence on the nucleation time and homogenous nucleation is preponderant over heterogeneous nucleation. In this case the value of the β exponent decreases considerably as the microalloying content (Nb, V) increases.

References

- [1] Sun L, Muszka K, Wynne BP, Palmiere EJ. *Acta Mater* 2014; 66: 132.
- [2] Andrade HL, Akben MG, Jonas JJ. *Metall Trans A* 1983; 14: 1967.
- [3] Kwon O. *ISIJ Int* 1992; 32: 350.
- [4] Medina SF, Mancilla JE. *ISIJ Int* 1996; 36: 1063.
- [5] Luton MJ, Dorvel R, Petkovic RA. *Metall Trans A* 1980; 11: 411.
- [6] Gómez M, Rancel L, Medina SF. *Mater Sci Eng A* 2009; 506: 165.
- [7] Vervynckt S, Verbeken K, Lopez B, Jonas JJ. *Int Mater Rev* 2012; 57:187.
- [8] Hu J, Du LX, Wang JJ, Xie H, Gao CR, Misra RDK. *Mater Sci Eng A* 2013; 585:197.
- [9] Medina SF, Quispe A. *ISIJ Int* 2001; 41: 774.
- [10] Medina SF, Quispe A, Gómez M. *Mater Sci Technol* 2003; 19: 99.
- [11] Pereloma EV, Kostyrychev AG, AlShahrani A, Zhu C, Cairney JM, Killmore CR, Ringer SP. *Scripta Mater* 2014; 75: 74.
- [12] Hu J, Du LX, Wang JJ. *Scripta Mater* 2013; 68: 953.
- [13] Medina SF, Gómez M, Rancel L. *Scripta Mater* 2008; 58: 1110.
- [14] Ishikawa F, Takahashi T. *ISIJ Int* 1995; 35: 1128.
- [15] Zajac S, Siwecki T, Hutchinson WB, Lagneborg R. *ISIJ Int* 1998; 38: 1130.
- [16] Quispe A, Medina SF, Valles P. *ISIJ Int* 1997; 37: 783.
- [17] Dutta B, Sellars CM. *Mater Sci Technol* 1987; 3: 197.
- [18] DeArdo AJ. *Int Mater Rev* 2003; 48: 371.
- [19] Ardell J. *Acta Metall* 1972; 20: 61.
- [20] Bhadeshia H K D H, Honeycombe RWK. *Steels Microstructure and Properties*. London: Elsevier; 1981.
- [21] Park SH, Yue S, Jonas JJ. *Metall Trans A* 1992; 23A: 1641.
- [22] Liu WJ. *Metall Trans A* 1995; 26A: 1641.

- [23] Faessel A. Rev Metall Cah Inf Tech 1976; 33: 875.
- [24] Medina SF, Hernández CA. Acta Mater 1996; 44:149.
- [25] Andrade HL, Akben MG, Jonas JJ. Metall Trans A, 1983; 14: 1967.
- [26] Pertula JS, Karjalainen LP. Mater Sci Technol 1998; 14: 626.
- [27] Kampmann R, Wagner R. A comprehensive treatment. Weinheim: VCH Verlagsgesellschaft mbH; 1991.
- [28] Perez M, Dumont M, Acevedo-Reyes D. Acta Mater 2008; 56: 2119.
- [29] Mukherjee M, Prahl U, Bleck W. Acta Mater 2014; 71: 234.
- [30] Dutta B, Sellars CM. Mater Sci Technol 1987; 3: 197.
- [31] Sakui S, Sakai T, Takeishi K. Trans ISIJ 1977;17: 718.
- [32] Medina SF, Quispe A, Gomez M. Metall Mater Trans A, 2014; 45A: 1524.
- [33] Vignes A, Philebert J, Badía J, Lavaşeur L. The Use of the Electron Microprobe for the Determination of Impurity Diffusion Coefficients, in Proc 2nd National Conference on Microprobe Analysis. Boston; 1967. Paper 20.
- [34] Medina SF, Mancilla J. ISIJ Int 1996; 36: 1063.
- [35] Kwon O, DeArdo A. Acta Metall Mater 1990; 39: 529.
- [36] Vervynckt S, Verbeken K, Thibaux P, Houbaert Y. Mater Sci Eng A 2011; 528:5519.
- [37] Vervynckt S, Verbeken K, Lopez B, Jonas JJ. Int Mater Rev 2012; 57: 187.
- [38] Quispe A, Medina SF, Gómez M, Chaves JI. Mater Sci Eng A 447 2007; 447: 11.
- [39] Turkdogan ET. Iron Steelmaker 1989; 16: 61..
- [40] Dutta B, Valdes E, Sellars CM. Acta Metall Mater 1992; 40: 653.
- [41] Medina SF, Hernández CA. Acta Mater 1996; 44: 155.
- [42] Gladman T. The physical metallurgy of microalloyed steels. London: The Institute of Materials; 1997.
- [43] Dutta B, Palmiere EJ, Sellars CM. Acta Mater 2001; 49: 785.

CAPTIONS OF FIGURES

Fig. 1. Variation of the recrystallised fraction (X_a) with the time for steel N1. Austenite grain size (D)=122 μ m;

Fig. 2. P_s curves for steel N1 at strain of 0.20 and 0.35, respectively.

Fig. 3. P_s curves for steel N8 at strain of 0.20 and 0.35, respectively.

Fig. 4. P_s curves for steel V8 at strain of 0.20 and 0.35, respectively.

Fig. 5. Plot of β against the total microalloying element content (wt %).

Fig. 6. Parameter β against driving force for precipitation for the steels used.

Fig. 1.

Fig. 2.

Fig. 3.

Fig. 4.

Fig. 5.

Fig. 6.