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Thermally activated dislocation dynamics in austenitic FeMnC steels at low homologous temperature

S. Allain,[a,*] O. Bouaziz[a] and J.P. Chateau[b]

[a]*Arcelormittal Maizières Research, Voie Romaine, F-57283 Maizières les Metz, France*

[b]*Institut Jean Lamour, Ecole des Mines de Nancy, Parc de Saurupt, F-54042 Nancy Cedex, France*

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Thermally activated dislocation dynamics in austenitic FeMnC steels is investigated based on a collection of tensile yield stresses at different temperatures and strain rates of numerous steels available in the literature. A classical viscoplastic potential is suitable to describe the phenomena in binary FeMn alloys. The description of ternary FeMnC alloys is more complex as two basic mechanisms should be involved.

Keywords: Dislocation dynamics; Austenitic steels; TWIP; Thermally activated processes

During the last decade, high-manganese austenitic FeMnC steels have been the object of intense worldwide scientific work [1–4], due not only to the renewal of interest of steelmakers in finding breakthrough solutions compared to traditional ferritic steels, but also for the complex and multiple origins of their work-hardening [5]. The main factors accounting for the excellent balance between flow stresses and ductility of these alloys reported in the literature are an atypical dynamic strain ageing (DSA) mechanism and the occurrence of deformation mechanisms subsidiary to the dislocation gliding [7–12]. Depending on the stacking fault energy of the alloy, these mechanisms are either an ε-martensitic transformation or mechanical twinning. Both resulting microstructures lead to a so-called "dynamic Hall–Petch" effect, more widely known as transformation-induced plasticity (TRIP) effect or twinning-induced plasticity (TWIP), respectively. These mechanisms are activated after a critical amount of deformation, higher than the plastic yield onset in the steels studied above, and coexist with dislocation gliding.

The respective contributions to work-hardening are still a matter of debate, even though recent published works on the measurements of the kinematic hardening of these alloys give a pre-eminent role to the TRIP/TWIP effects [6,13].

Nevertheless, whatever the mechanisms involved, the huge elongation reported for this family of steels can only be explained by an intense dislocation gliding, as the final volume fractions of martensite or twins remain below 0.1. The main objective of this paper is to focus on the thermally activated nature of dislocation gliding in these austenitic alloys. The paper will highlight the fundamental role played by the carbon content and give new ways for a better understanding of the work-hardening behaviour of these alloys. This analysis relies on numerous experimental data available in the literature.

The characteristics of dislocation gliding are first investigated in terms of the evolution of the yield stresses (YS) of different FeMnC grades as a function of temperature at low strain rate. The strain-rate effect will be discussed hereafter. This stress is thought to be the most representative parameter for dislocation motion, in so far as the work-hardening rate may also involve the DSA mechanism, leading to negative strain-rate sensitivity, and the TRIP or TWIP effects (not yet activated at this low plastic strain level).

A large collection of results from different authors have been selected, with various manganese and carbon contents presented in Table 1.

All the considered steels are fully austenitic at room temperature (RT) and at low temperature, considering the Schumann's map [11], except steel #3, which is the less stable. The behaviour of steel #3 will be analyzed only at temperatures higher than RT. According to the estimated Néel temperatures, steels #9 and #10 are

* Corresponding author. Tel.: +33 3 87 70 47 47; fax: +33 3 87 70 47 12; e-mail addresses: perso.allain_s@hotmail.fr; sebastien.allain@arcelormittal.com

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Table 1. Chemical composition, grain size, Néel temperature estimated from Ref. [11] and tensile procedure (strain rate $\dot{E}$ and ε_p plastic strain onset for measuring YS) of the 10 reference steels for this study.

Steel No. and Ref.	Studied steels					Tensile procedure	
	wt.% Mn	wt.% C	Others	Grain size (μm)	$T_{Néel}$ (K)	$\dot{E}$	ε_p
#1 – Tomota 86 [14]	31	0.001		>20	432	0.00330	0.2%
#2 – Tomota 86 [14]	36	0.004		>20	470	0.00330	0.2%
#3 – Choi 99 [15]	24	0.120		>50	361	0.00100	1.0%
#4 – Rémy 75 [12]	26	0.200		>50	375	NS	0.2%
#5 – Kim 86 [16]	30	0.300	0.1 Nb	80	403	NS	NS
#6 – Kim 86 [16]	30	0.300	5 Al 0.1 Nb	80	340	NS	NS
#7 – Allain 04 [11]	22	0.600		2.3	313	0.00070	1.0%
#8 – Kuntz 08 [17]	22	0.600		>10	292	0.00380	0.2%
#9 – Allain 04 [11]	22	1.000		2.3	292	0.00070	1.0%
#10 – Adler 86 [18]	13	1.200		220	173	0.00083	0.2%

NS = not specified.

paramagnetic at room temperature and the others antiferromagnetic. The magnetic state of each steel is able to change with the tensile temperature, depending of its Néel transition.

The grain sizes of the studied steels are also indicated in Table 1, as reported in the original papers or estimated when micrographs were available, as well as the parameters of the tensile procedure when available. For all the steels, tensile tests have been performed at low strain rate and the onsets for yield strength can be considered as comparable, as no continuous yielding occurs in these steels, compared to some ferritic steels. The differences will be compensated thanks to the thermal analysis procedure described below.

Figure 1 shows the evolution of the measured YS of each reference steel, as reported in the respective publications. All the curves present the same tendency: a large decrease in the YS as a function of the temperature below RT and a small variation above. The sensitivities above RT are similar and confirm the consistency of the database.

The results proposed by Shun et al. [19] have not been retained in the database due to the lack of consistency between Figure 6 of their article and the proposed evolution of the 0.2% YS. The study of Grässel et al. on a Fe30Mn3-Si3Al steel [2] is, on the other hand, discussed below.

The YS in austenitic steels is the sum of a number of major contributions: solid solution, grain size effect and thermally activated contributions. To compare the latter contribution in the different steels, all the curves presented in Figure 1 are normalized by subtraction of the value of the yield stress at room temperature,

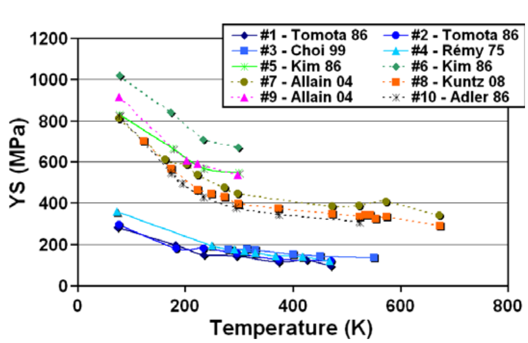

Figure 1. Evolution of the tensile yield stresses of the steels described in Table 1 as a function of the temperature.

YS(RT), as shown in Figure 2a. YS(RT) reflects the contributions of solid solution and grain size.

Figure 2a shows that the YS present very similar evolutions as a function of the temperature for all these alloys. Above RT, all the curves are perfectly superimposed with a weak linear decrease about -0.25 MPa K^{-1}. This evolution can be explained in terms of a non-thermally activated phenomenon due to the decrease in the solid solution and grain size contributions, which scale like the elastic modulus. As analyzed and modelled in Ref. [20], the elastic bulk modulus in a paramagnetic FeMnC austenite in the considered temperature range depends linearly on the temperature.

Below RT, the evolution of the YS with temperature depends on the steel: the higher the carbon content, the higher the sensitivity of the YS to the temperature. This typical two-range behaviour was also observed by Shun et al. [19] and in an austenitic Fe22Ni0.6C alloy [9].

An initial simple thermal analysis has been performed on the data to characterize the contribution of the thermal activation in the YS. According to Ref. [11], this contribution can be described thanks to a classical visco-plastic potential [21]:

$$YS^{therm}(T) = \frac{1}{M} sinh^{-1}\left(\frac{\dot{E}}{2M\rho_m b_{110}^2 \nu_D} exp\left(\frac{\Delta G_0}{k_B T}\right)\right)\frac{k_B T}{V^*} \tag{1}$$

where $M \approx 3$ is the Taylor factor, b_{110} is the Burgers' vector of perfect dislocations, $\nu_D = 10^{13}$ s^{-1} is the Debye frequency, ρ_m is the density of mobile dislocations (taken equal to 10^{12} m^{-2}), k_B is the Boltzmann constant, T and $\dot{E}$ are the temperature and the initial strain rate related to the tensile conditions. ΔG_0 and V^* are the activation energy and volume, respectively, characterizing the physical mechanism responsible for the thermal activation. These two physical values are unknown and must be identified with the help of experimental results.

The prediction of Eq. (1) is compared to the experimental results from which the athermal contributions have been removed:

$$YS^{therm}(T) = YS^{exp}(T) - YS^{exp}(298\text{ K}) - [60 - 0.25T] \tag{2}$$