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Modeling the hot flow behavior of a Fe–22Mn–0.41C–1.6Al–1.4Si TWIP steel microalloyed with Ti, V and Nb

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ABSTRACT

The present research work analyses the influence of Ti, V and Nb microalloying elements on the hot flow behavior of a high-Mn Twinning Induced Plasticity (TWIP) steel. For this purpose, flow curves were obtained by uniaxial hot compression tests performed at four strain rates (10^{-1}, 10^{-2}, 10^{-3} and $10^{-4}\,s^{-1}$) and three temperatures (900, 1000 and 1100 °C). The models of Estrin, Mecking and Bergstrom; Avrami and Tegart, and Sellars were applied to determine the hot working constants used to derive the constitutive equations describing the flow curves. The analysis of modeling parameters of the hot flow curves shows that Ti, V and Nb additions to TWIP steel generated slight increase in the peak stress (σ_p), retardation of the dynamic recrystallization (DRX) onset, particularly at low temperature, and decrease in the activation energy required to recrystallization (Q_t). Likewise, the softening effect promoted by DRV and DRX was more evident at high temperatures and low strain rates. On the other hand, the resulting deformed microstructures, analyzed by the SEM-EBSD technique, showed that the most important refining effect on recrystallized austenitic grain was in the presence of V and Ti. The good agreement between the experimental and predicted hot flow curves demonstrated that the developed constitutive equations predict with reasonable accuracy the hot flow behavior of the studied TWIP steels.

1. Introduction

High-Mn TWIP steels provide a great potential to the manufacturing, building and automotive industries [1–6]. There are several studies [1–5] about the chemical compositions of TWIP steels and their influence on the deformation mechanisms, the mechanical properties, the stacking fault energy (SFE) and the phase transformations during cold deformation processes. However, there are few researches about the hot flow behavior [6–14]. These studies show, as expected, the influence of temperature (T), plastic deformation or strain (ε) and strain rate ($\dot{\varepsilon}$) on the hot flow behavior. Nevertheless, less effort has been addressed to perform a good control of these parameters to obtain an optimal microstructural conditions and therefore to promote their mechanical properties.

Hot compression tests are usually carried out to simulate industrial processes such as forging, rolling, extrusion, etc. The obtained hot flow curves ($\sigma_t - \varepsilon_t$) show the typical plastic behavior of steels at high temperature, i.e., a single peak in the flow stress is noticed. The flow curves are furthermore characterized by three distinctive zones: (i) A stage I that is related to the balance between the generated and the stored dislocations by strain hardening, together with those rearranged and annihilated by dynamic recovery (DRV) [15,16]. (ii) A stage II extends from the peak stress (σ_p) to the steady state stress (σ_{ss}). This stage represents the progress of an additional softening mechanism called dynamic recrystallization (DRX), which can be modeled by the classical equations proposed by Kolmogorov–Johnson–Mehl–Avrami (KJMA) [17–21] assuming that the stress softening is proportional to the recrystallized volume fraction (X), and (iii) A stage III where the metal attains an ideal plastic flow behavior that can be modeled by using the Sellars and Tegart equations [22], modified by Cabrera et al. [23] by including a correction of the Young modulus dependence on temperature.

The appropriate design of the thermo-mechanical treatment is important in simulating hot forming processes, in order to control the grain size during the hot deformation. This is particularly true in TWIP steels [10], where no phase transformations occurs during cooling. Additionally, the austenitic grains can be further refined by a fine dispersion of particles randomly distributed within the

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Nomenclature

Q_{rec}	activation energy for DRX
Q_{HW}	activation energy for hot working
Q	activation energy required into the deformation process
Q_t	activation energy required to recrystallization
a, b	Avrami constants
n_g	Avrami exponent calculated for every test condition
n_A	Avrami exponent
b	Burgers vector
n	Creep exponent
K_D	Derby's coefficient
n_D	Derby's exponent
ρ	dislocations density
X	dynamically recrystallized volume fraction
ρ_0	initial dislocations density
d_0	initial grain size
α	inverse of the stress associated with power-law breakdown
α_a	material constant due to DRV
A	material constant due to steady state stage
A_p, α_p, A_{ss}, α_{ss}	material constants correspondent to the peak stress and steady state, respectively
$k_{t50\%}$, $m_{t50\%}$	material constants related to 50% of recrystallization
k_t, m_t, n_t	material constants related to recrystallization
K_ε, m_ε, n_ε, $m_{\sigma p}$	material constants calculated during the strain hardening and DRV
T_m	melting temperature
K_Ω, m_Ω, K_U, m_U	modeling constants calculated during the softening and hardening stages, respectively
ε_p	peak strain
σ_p	peak stress
ε	plastic strain
d_{rec}	recrystallized grain size
σ_s	saturation stress
μ_o	shear modulus
Ω	softening due to DRV
σ_{ss}	steady state stress
$\dot{\varepsilon}$	strain rate
$(\sigma_p(T)/E(T))$	temperature dependent stress normalized by the temperature dependent Young modulus
$D^{\gamma}_{Fe}(T)$	temperature dependent γ-Fe lattice self-diffusion coefficient
T	test temperature
$t_{50\%}$	time necessary for 50% of recrystallization
t	time
ε_t	true strain
σ_t	true stress
R	universal gas constant
U	work hardening due to DRV
Z	Zener–Hollomon parameter
A-UHSS	advanced ultra-high strength steels
BC	band contrast
CSL	coincidence site lattice
DRV	dynamic recovery
DRX	dynamic recrystallization
GAM	grain average misorientation
GB	grain boundaries
KJMA	Kolmogorov–Johnson–Mehl–Avrami model
RF	recrystallized fraction
SEM-EBSD	scanning electron microscopy-electron back-scattering diffraction
SFE	stacking fault energy

matrix. In this case, particles must be sufficiently stable to not dissolve or to not grow during the reheating and hot forming operations [24]. The addition of microalloying elements such as Nb and Ti can promote an extra strengthening via precipitation hardening or via grain refinement in high-Mn steels, and the imminent production of finer grain sizes at the end of the steel processing as previously reported by Dobrzanski et al. [10]. Additionally, they reported [11] that the relative low strain values (ε_p) to initiate DRX produce a more refined austenitic microstructure after multiple deformation passes because the earlier activation of DRX. On the other hand, Hamada et al. [6] reported that increasing the Al content in TWIP steels is possible to change the ε_p values from 0.17 to 0.28, with a consequent retardation on the DRX onset.

The present work analyzes from experimental true stress–strain data of uniaxial hot compression tests performed at three different temperatures and four strain rates, the separate effect of microalloying elements (Nb, V and Ti) on the hot flow behavior of a high-Mn TWIP steel. Additionally, mathematical modeling of the hot flow curves has been carried out by using previous models and equations obtained from scientific literature [23,25–41], which are summarized as follows.

2. Constitutive models to the hot flow behavior

As already pointed out, previous research works [23,25–41] have established constitutive models from experimental data to describe and predict the hot flow behavior during the thermo-mechanical processing. The models analyze separately the three stages above described of the hot flow curves. In the stage I, strain hardening and DRV are the dominant mechanisms, and the so-called one variable approach model can be used [15,16]. It states that during hot deformation, the dependence of the dislocation density (ρ) on plastic strain (ε) is generally considered to be a balance between dislocations generated by work hardening and dislocations annihilated by DRV, resulting in the following equation:

$$\frac{\partial \rho}{\partial \varepsilon} = U - \Omega\rho \tag{1}$$

where U represents the work hardening and Ω is the softening due to DRV. This expression has already been used in a wide range of materials [16,23,25,28,29] and its integration is relatively simple assuming that both U and Ω are independent of strain (ε). Assuming the classical relationship between the stress and dislocation density (ρ):

$$\sigma = \alpha_a \cdot \mu_0 \cdot b \cdot \sqrt{\rho} \tag{2}$$

where α_a is a material constant, μ_0 is the shear modulus and b is Burger's vector, then, Eq. (1) becomes in the following constitutive expression:

$$\sigma^2 = \sigma_0^2 e^{-\Omega\varepsilon} + \sigma_s(1 - e^{-\Omega\varepsilon}) \tag{3}$$

being $\sigma_0 = \alpha_a \mu b \sqrt{\rho_0}$ and $\sigma_s = \alpha_a \mu b \sqrt{U/\Omega}$, where ρ_0 is the initial dislocation density. Here σ_s is the saturation stress, i.e., the stress in the absence of DRX, equal to σ_{ss} in the case of softening solely by