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Hot deformation and static softening behavior of vanadium microalloyed high manganese austenitic steels

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The hot deformation and static softening behavior of various high Mn (20–30 wt%) austenitic steels microalloyed with different V (0.1, 0.2 wt%), C (0.2, 0.6, 1 wt%) and N (0.005–0.025 wt%) contents were investigated. Double-hit torsion tests at temperatures in the range 700–1100 °C were carried out and specimens quenched at selected conditions were examined using advanced microscopy techniques (EBSD-TEM) to characterize the recrystallization and strain-induced precipitation behavior. The results show that precipitation of vanadium at the hot working temperature range is sluggish. It mainly occurs for the combinations of 20%Mn–0.6%C–0.2%V and 30%Mn–1%C–0.1%V. When the carbon content is reduced to 0.2%C, strain-induced precipitation is suppressed at typical hot working temperatures, independently of the N level. The flow stress behavior was affected by the amount of C and by modifying the base composition from 30%Mn to 20%Mn–1.5%Al. However, the effect is complex and depends on deformation conditions. In the absence of strain-induced precipitation, the static softening kinetics was accelerated by increasing C content. However, no effect of Mn or V in solid solution was observed. In those cases where strain-induced precipitation took place, static recrystallization was severely delayed, leading to a major contribution of recovery to softening kinetics.

1. Introduction

In recent decades, the ongoing research of the steel industry has been focused on high strength steels with high toughness levels and excellent formability. This optimal combination is particularly attractive for automotive industry, since the increase in strength enables to reduce the mass of the car bodies and, therefore, CO_2 emissions, whereas the increase in ductility admits complex component design. Among others, TWinning Induced Plasticity (TWIP) steels have attracted special interest. Due to their low Stacking Fault Energy (SFE), the mechanical twinning assists in the implementation of larger plastic deformation [1,2]. As a consequence, these steel grades exhibit remarkable strain hardening properties, far in excess of other ferrous alloys [3–5]. Due to this effect, high-Mn steels are potentially attractive for automotive applications involving press-formed parts for energy absorption, or for structural reinforcement which assures vehicle structural stability in case of car crash. Although they are still in experimental stage, a wide range of mechanical properties and microstructural characteristics have been obtained by including different amounts of specific alloying elements (e.g., Mn, Si, Al) [6]. In addition, after hot rolling, cold rolling followed by annealing is known to be a powerful route for improving certain mechanical properties. Nevertheless, it is known that crash resistance is related to yield strength and even fine-grained (<3 μm) TWIP alloys have naturally low yield stress ($\sim$450 MPa). Accordingly, there is a clear benefit in increasing the yield strength of these materials without degrading the hot and cold rolling conditions. Microalloying with Nb, Ti or V has been revealed to be an ideal mechanism for increasing yield strength in TWIP steels [3,7]. These microalloying elements can precipitate in the form of nanometer-sized carbides and/or nitrides during the different stages of the steel production process, and depending on the size, volume fraction and coherence of the precipitate dispersion, this can lead to significant yield strength increase [3,7–9].

In the literature there are many works dealing with the mechanical property characterization and strengthening mechanisms of high Mn steels. However, in order to introduce TWIP steels with superior mechanical properties, the routes and parameters of Thermomechanical Processing (TMP) should be optimized. The main objective of any TMP route is controlling the kinetics of the restoration processes involved. This is highly dictated by TMP parameters such as strain, strain rate, temperature, initial grain size, etc. Several works have reported that the flow stress and the activation energies of hot deformation are higher for these steels

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[10–12], while their static softening kinetics is retarded compared to C–Mn steels [13,14]. However, at the moment the information is scarce. For instance, in the case of static softening behavior, the information available is limited to steel compositions of (0.1%C–25%Mn) with different Al levels [13,14]. However, the effect of varying C or Mn alloying levels within the range usually considered for TWIP steel compositions has not yet been investigated. The effect of microalloying elements on the static softening behavior of TWIP steels has not been analyzed either. It should be taken into account that, as in the case of low C–Mn steels [15], microalloying elements can lead to softening retardation and therefore to a further increase in the rolling loads. This effect is expected to be especially enhanced if strain-induced precipitation takes place during hot rolling [16,17]. In addition, strain-induced precipitation during hot deformation can lead to a decrease in the amount of microalloying element available for precipitation at later stages, which can result in a loss of microalloying efficiency [18]. Accordingly, there is a clear interest in investigating the effect of both microalloying and alloying levels on the hot deformation, static softening and strain-induced precipitation behavior of high Mn steels.

In this work, double hit torsion tests were performed with TWIP steels with different Mn (20–30 wt%), V (0.1, 0.2 wt%), C (0.2, 0.6, 1 wt%) and N (0.005–0.025 wt%) levels, in order to study their hot deformation and static softening behavior. In addition, specimens were characterized using advanced microscopy techniques to analyze the recrystallization and strain-induced precipitation kinetics of the vanadium TWIP microalloyed steels.

2. Experimental procedure

The compositions of the steels investigated are shown in Table 1. Two sets of laboratory casts with different base compositions (30%Mn and 20%Mn–1.5%Al) were analyzed. The 30%Mn steels present a constant microalloying addition of 0.1%V and different carbon and nitrogen levels. In the case of the 20%Mn steels, the carbon concentration is constant (0.6%) and the vanadium content is varied (0, 0.1, 0.2%).

Table 1
Chemical composition of the steels investigated (wt%). The soaking temperatures used in the torsion tests and the initial austenite grain sizes are also included in the table.

Steel	Mn	Al	C	V	N	T_{Soak} (°C)	D_0 (μm)
30Mn–0.2C-low N	29	–	0.19	0.10	0.005	1150	23 ± 1
30Mn–0.2C-high N	29	–	0.19	0.10	0.022		28 ± 1
30Mn–0.6C-low N	29	–	0.56	0.10	0.005		30 ± 2
30Mn–0.6C-high N	29	–	0.58	0.09	0.025		25 ± 1
30Mn–1C	31	–	1.06	0.12	0.009		34 ± 2
20Mn–0V	21	1.5	0.61	–	~0.005	1200	44 ± 6
20Mn–0.1V	21	1.5	0.63	0.10	~0.005	1250	38 ± 3
20Mn–0.2V	20	1.5	0.58	0.19	~0.005		43 ± 6

Double-hit torsion tests were performed to study the static softening kinetics of the steels. The geometry of the torsion specimens was a reduced central gauge section 16.5 mm in length and 7.5 mm in diameter. The thermomechanical schedule employed in the tests is shown schematically in Fig. 1(a). First, the specimens were soaked at temperatures ranging from 1150 to 1250 °C (see Table 1) for 10 min in order to dissolve the microalloying elements. Next, a roughing pass of $\varepsilon = 0.3$ was applied at 1150 °C to obtain a refined recrystallized microstructure. Finally, the specimens were cooled down to the deformation test temperature. In all cases a fast cooling rate of 10 °C/s was applied to prevent precipitation of the microalloying elements during this cooling step. Once the deformation temperature was reached, the specimens were deformed up to $\varepsilon = 0.36$, unloaded and held for increasing times before reloading again up to $\varepsilon = 0.1$. The strain-rate was kept constant ($\dot{\varepsilon} = 1\ s^{-1}$) and deformation temperatures between 700 and 1100 °C were considered. From the stress–strain curves, the Fractional Softening (FS) was determined using the 2% offset method [19]:

$$FS = \frac{\sigma_m - \sigma_r}{\sigma_m - \sigma_0} \quad (1)$$

where σ_m is the value of the stress before reloading, and σ_0 and σ_r are the stresses corresponding to 2% strain value of the first and second curves, respectively (see Fig. 1(b)).

In addition, specimens were water quenched after the roughing treatment to analyze the initial microstructure or after deformation and selected holding times to study the evolution of the austenite microstructure. Analysis of the quenched samples was carried out on a section corresponding to 0.9 of the outer radius of the torsion specimen, *R*, also known as the sub-surface section [19]. The quenched specimens were examined via Electron Back-Scatter Diffraction (EBSD) to determine the initial austenite grain size and the recrystallized fraction and by Transmission Electron Microscopy (TEM) to study the precipitation state.

Orientation imaging was carried out on a Philips XL30cp scanning electron microscope, equipped with a W-filament, using the TSL OIM™ Analysis 4.6 software. The scans were made over a regular hexagonal grid, using step sizes of 0.5–1.5 μm and an analysis area of $600 \times 600\ \mu m^2$ for grain size determination, and a step size of 0.5 μm and an analysis area of $250 \times 250\ \mu m^2$ for recrystallized fraction calculation. Between 4 and 9 scans were recorded for each condition. Grains have been defined as regions surrounded by 5° misorientation boundaries with a minimum grain size of 5 μm for the initial microstructure characterization, and of 1.2 μm for recrystallized fraction measurements. In order to

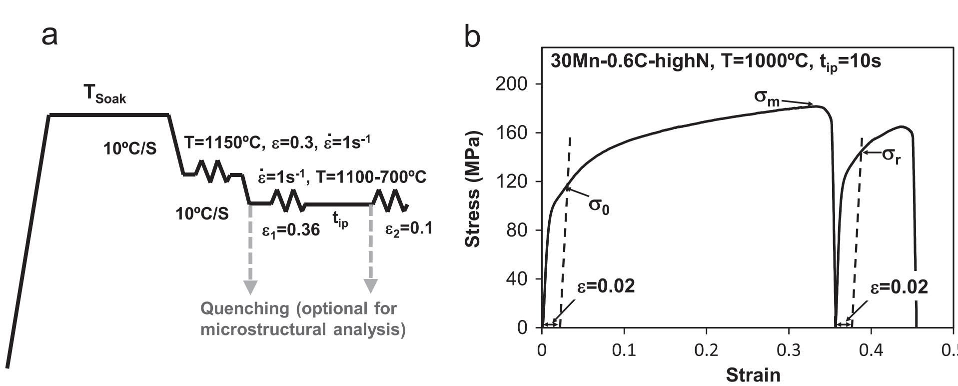

Fig. 1. (a) Thermomechanical treatment used in the double-hit torsion tests, and (b) definition of the stresses used in the 2% offset method (Eq. (1)).