Materials Science & Engineering A 642 (2015) 71–83

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea

Temperature effect on deformation mechanisms and mechanical properties of a high manganese C+N alloyed austenitic stainless steel

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ARTICLE INFO

Article history:
Received 16 January 2015
Received in revised form
15 June 2015
Accepted 16 June 2015
Available online 20 June 2015

Keywords:
Austenitic stainless steels
Stacking fault energy
Twinning
Strain hardening
EBSD
Electron microscopy

ABSTRACT

Recently developed high-manganese stainless Fe–Cr–Mn–CN steels exhibit an exceptional combination of strength and ductility and show great promise for structural applications. Understanding the relationships between temperature, stacking fault energy (SFE) and strain-hardening behavior is critical for alloying, design, and further optimization of these steels. The present study investigates the influence of temperature and SFE on the microstructural evolution to explain the deformation behavior and mechanical properties of an austenitic Fe–14Cr–16Mn–0.3C–0.3N alloy. The flow behavior is homogenous and no serrations in the flow stress occur during tensile deformation in the temperature range from −150 to 250 °C. Mechanical twinning and the formation of (planar) dislocation substructures strongly influence the mechanical properties and work-hardening behavior in the intermediate temperature range from −40 to 45 °C (SFE range from 17 to 24 mJ m^{-2}). In the high temperature interval from 100 to 250 °C the SFE ranges from 29 to 44 mJ m^{-2} and the initiation of mechanical twinning is delayed leading to reduced work-hardening in the intermediate and final stages of strain-hardening. In the low temperature regime from −150 to 100 °C (SFE approximately 15 mJ m^{-2}), $\varepsilon_{h.c.p.}$-martensite is the dominant secondary deformation mechanism, contributing to the enhanced work-hardening in the early and intermediate stages of deformation and slightly lower total elongations. The yield strength of the studied alloy is significantly larger and exhibits greater sensitivity to temperature within the thermal and athermal ranges for dislocation motion compared to conventional Fe–Mn–(Al)–C TWIP or austenitic stainless steels, which may be attributed to phenomena such as short range ordering.

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1. Introduction

Stainless Fe–Cr–Mn–CN twinning induced plasticity (TWIP) steels alloyed with C+N exhibit increased strength and ductility compared to conventional stainless steels [1], along with high impact toughness [2,3], longer fatigue life [4] and improved wet corrosion resistance [5]. The synergistic effect of C+N alloying has been shown to improve the fatigue life of austenitic stainless steels [4]. With respect to conventional stainless steels, the substitution of nickel by manganese increases the interstitial solubility of C and N [6,7]. Several authors have investigated the mechanisms by which additions of C+N improve strength in austenitic stainless. According to Gavriljuk et al. [3,7], stabilizing the austenitic phase with C+N enhances the concentration of free electrons – more effectively than alloying only with N – promoting the metallic character of interatomic bonding and short-range ordering (SRO) rather than atomic-clustering. The increased concentration of conduction electrons strengthens the binding between immobile interstitial atoms and dislocations, enhancing the strength, ductility and impact toughness [2,7].

In comparison to conventional high-Mn TWIP steels [8–10] the homogenous flow and work hardening characteristics of Fe–Cr–Mn–CN steels indicate differences in the strain-induced hardening mechanisms. The occurrence of Cr–N SRO phenomena [11–13] and the resultant interactions with dislocations and stacking faults are believed to play a major role in the deformation behavior of these

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http://dx.doi.org/10.1016/j.msea.2015.06.047
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alloys. The strong affinity between Cr and N atoms generates Cr–N SRO clusters. Oda et al. [11] suggested that N in the austenitic Fe–1.5Mn–15Cr–15Ni alloy gathers around Cr atoms to form interstitial-substitutional complexes based on X-ray absorption fine structure analysis. More recently, Li et al. [14] reported Cr–N SRO in the Co–29Cr–6Mo–(0.1–0.16)N alloy based on results of atom probe tomography. Moreover, Cr–N SRO is assumed to influence the energy barrier for the γ- > ε phase transition [14], which along with the stacking fault energy (SFE), is relevant to the deformation mechanisms and mechanical behavior of Fe–Cr–Mn–CN steels.

The composition and temperature dependent SFE strongly influences the plasticity mechanisms and work-hardening behavior of austenitic high-Mn steels [8–10,15]. Increasing the SFE causes the active deformation mechanisms to change from strain-induced martensite formation and dislocation glide, to mechanical twinning and dislocation glide, and to pure dislocation glide [16]. The effect of N additions on the SFE in Fe–Cr–Mn steels is reported to be non-monotonic, exhibiting a minimum SFE at ~0.4 wt% N in Fe–15Cr–17Mn [17,18] and Fe–20Cr–17Mn [19] alloys, from quantitative SFE measurements on extended three-fold dislocation nodes using transmission electron microscopy (TEM). Including these results in a thermodynamics based model, along with a composition dependent interfacial energy term, allows for calculation of the SFE [20]. The decrease in SFE at low N contents was explained due to the segregation of interstitial N atoms to stacking faults [21]. While at higher N contents the SFE increases as the bulk effect becomes more pronounced [18]. Limited available data on the influence of C on SFE in the Fe–Cr–Mn system exists since C contents are generally kept low (< 0.1 wt%) in austenitic stainless steels to avoid intergranular corrosion during welding. Remy [22] measured a room temperature (RT) SFE of 11 mJ m^{-2} for a Fe–4Cr–20Mn–0.48C alloy. C acts as a strong austenite stabilizer that lowers the $\varepsilon_{h.c.p.}$-martensite-start and the $\gamma_{f.c.c.} \rightarrow \varepsilon_{h.c.p.}$ transition temperatures [23,24] in agreement with thermodynamic models that report increases in the SFE with C in the Fe–Mn–C system [8,25,26]. Experimental SFE values reported by Petrov [17,27] on Fe–22Mn–(0.1–0.7)C exhibit a linear increase in SFE with C content, while at dilute concentrations the segregation effects of C leads to a decrease in SFE. There are conflicting reports about the influence of C+N alloying on SFE in Fe–Mn–Cr steels. According to Lee et al. [28,29] the SFE increases linearly with increasing C+N content for a Fe–18Cr–10Mn alloy. However, experimental investigations by Roncery et al. [1,30] in the Fe–14Cr–(21–30)Mn–(0.57–0.77)C+N system imply a drop in SFE with increase in N content, which is explained by a decrease in the density of states at the Fermi level which is inversely proportional to the SFE [18]. Furthermore, segregation effects of N to dislocations and stacking faults [17,21], Cr–N SRO [31] and the distribution of alloying elements within the *f.c.c.* lattice are also reported to affect the SFE [13]. In addition, the experimental method of SFE measurement may also affect the SFE values. For instance, the method of measuring SFE by analyzing the geometry of extended three-fold nodes using TEM, which is utilized in several of the aforementioned studies, overestimated the SFE in Fe–Cr–Ni alloys [32].

The goal of this research is to provide a detailed description of the influence of temperature and SFE on the deformation mechanisms and strain-hardening behavior of a newly developed Fe–14Cr–16Mn–0.3C–0.3N alloy. Analysis of partial dislocation spacings using weak-beam dark-field (WBDF) TEM, comparisons of the active deformation mechanisms with testing temperature, and the results of previous SFE investigations on similar materials yielded the temperature dependence of the SFE. The microstructure was characterized by electron back-scatter diffraction and TEM at different levels of strain. The strain-hardening behavior was assessed from uniaxial tensile tests conducted at temperatures ranging from −150 to 250 °C. The intrinsic properties are correlated to the temperature dependent deformation mechanisms, with respect to the twin and dislocation substructure evolution, to explain the flow and work-hardening behavior.

2. Experimental procedure

2.1. Material

The investigated steel with composition of 14.6 wt% Cr, 15.9 wt% Mn, 0.31 wt% C and 0.29 wt% N was strip cast (thickness of 2.2 mm), cold rolled to a final thickness of 1.1 mm and recrystallized at 1150 °C for 4 min. The samples were fully austenitic in the as-received state and after deformation to fracture at RT as determined with X-ray diffraction, with an average grain size of ~20 μm approximating a log-normal distribution.

2.2. Ultrasonic pulse velocity measurements

The ultrasonic pulse velocities (longitudinal and shear) of the Fe–14Cr–16Mn–0.3C–0.3N wt% steel and type 304 stainless steel for comparison were measured. A Tektronix TDS 2001C oscilloscope, Olympus 5072PR pulser/receiver, 20 MHz normal incident longitudinal transducer and a 10 MHz normal incident shear transducer were used to measure the transit time of longitudinal and shear pulses through the specimens. Soundsafe longitudinal and shear wave ultrasound couplants were used to couple specimens to the longitudinal and shear wave transducers, respectively. The transducers were used in a pitch/catch method. Transit time measurements were made by measuring the time between the crest of leading cycles of consecutive echoes [33]. Sheets with area between 1–2 cm^2 were gently grinded, using a tool to ensure each surface is nearly parallel (standard deviation in thickness across the samples did not exceed 0.009 mm), with successively finer SiC paper up to 1200 grit. A transit time error associated with electronics and transducer specimen bond was determined using different thicknesses of stainless steel type 304 [33,34]. The longitudinal and transverse pulse velocities in the stainless steel type 304 were 5739 ± 15 μm μs^{-1} and 3116 ± 8 μm μs^{-1}, respectively, agreeing well with that of 5759 ± 15 μm μs^{-1} and 3134 ± 18 μm μs^{-1} reported by Ledbetter et al. [33].

2.3. Stacking fault energy measurements

Three millimeters diameter disks were cut from the gage length of samples deformed to 0.015 plastic tensile strain using electro discharge machining. The disks were ground to ~100 μm thickness and electro-polished to electron transparency with a TenuPol-5 using a solution of 95% ethanol and 5% perchloric acid at −20 °C. Partial dislocations were analyzed with a Philips CM20T TEM operating at 200 kV. Measurements of Shockley partial-dislocation separation distances were made with a beam direction near [111] on defects in the (111) habit plane using < −220 > type g-vectors. Weak-beam dark-field (WBDF) imaging was employed using g(3 g) diffracting conditions. Multiple measurements of the separation distances were made on each dislocation pair to achieve an average separation distance for each dislocation. A correction was applied to account for the slightly smaller separation distance of the dislocation cores in relation to the intensity peaks, which arises due to asymmetries in the strain fields outside and between partial dislocations [35]. An average d_{actual} and standard deviation of the measurements were obtained for each partial-dislocation pair. The perfect dislocation character angle, β, was determined from Burgers vector analysis on the partial dislocations in WBDF imaging mode. For Shockley partial dislocations in the [111]/(111) zone/habit plane configuration, $|g \bullet b_p|$ (where b_p

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