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Effect of stacking fault energy on work hardening behaviors in Fe–Mn–Si–C high manganese steels by varying silicon and carbon contents

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ABSTRACT

To improve the low work hardening capacity of Hadfield steel at low stress, the effect of stacking fault energy (SFE) on the microstructures and the work hardening behaviors of the Fe–Mn–Si–C high manganese steels were investigated by varying the silicon and carbon contents. The work hardening rates of the Fe–17Mn–Si–C steels with lower SFE were higher than that of the Hadfield steel at the strain below 0.28. The reason was that the amount of deformation-induced ε-martensite or mechanical twins was higher in the Fe–17Mn–Si–C steels than in the Hadfield steel due to their earlier onset. The work hardening rate of the Fe–17Mn–Si–C steels increased with decreasing the SFE because the rate of the formation and the amounts of martensite and twins increased with lowering the SFE.

1. Introduction

Conventional Hadfield steels (1.0%–1.4%C, 11%–14%Mn, wt.% here and throughout) are widely used in the engineering fields, i.e. metallurgy, mining and railway due to the excellent properties such as the high toughness, good wear resistance, especially the high work hardening capacity [1–3]. However, their applications are limited under low stress as their high work hardening capacity can only be obtained under heavy stress or high load impact. Therefore, it is still of significance to improve the work hardening capacity of Hadfield steels under low stress. On the other hand, although Hadfield steels have been invented for over 120 years, the real origin of their high work hardening rate is still under intense discussion [2,4–10]. Obviously, to clarify the real origin should be first made to find some methods to improve the work hardening capacity of Hadfield steels under low stress.

So far, there exist two dominating opinions responsible for the exceptionally high work hardening rate in the Hadfield steels, that is, dynamic strain aging [2,5,6] and mechanical twinning [4,10]. Dastur and Leslie proposed that the primary cause of high work hardening rate of Hadfield steel was the dynamic strain aging, instead of mechanical twinning. The reason was that the stress–strain curves of the Hadfield steel exhibited a serrated flow, a negative strain rate dependence of flow stress and high work hardening, which are characteristic of dynamic strain aging. Furthermore, they thought that the strong interactions between Mn and C atoms led to the dynamic strain aging [2]. However, although Owen and Grujicic further confirmed the existence of the strong interactions between Mn and C atoms through thermodynamic calculations [5], the dynamic strain aging cannot explain why the work hardening rate is higher at 173 K, at which the dynamic strain aging cannot take place [4]. The recent studies by Koyama et al. also showed the dynamic strain aging made a minor contribution to the high work hardening rate as compared to the mechanical twinning in an Fe–17Mn–0.8C steel [8].

On the other hand, Adler et al. proposed that the mechanical twinning was responsible for the high work hardening rate of Hadfield steels based on the results that the work hardening rate increased with lowering the deformation temperature and can be rationalized with the change in the volume fraction of mechanical twins. However, only the mechanical twinning is not enough to explain the anomalous hardening of Hadfield steel. A Co–33Ni alloy showed a lower work hardening rate than the Hadfield steel although it exhibited the same twinning kinetics as Hadfield manganese steel did [4]. The recent studies by Idrissi et al. also showed that the work hardening rate was lower in the Fe–28Mn–3.5Si–2.8Al steel than in the Fe–20Mn–1.2C steel although they had the same twinning kinetics [7].

Recently, we argued that the formation of mechanical twins and their high hardness induced by carbon atoms may be responsible for the anomalous hardening rate in the Hadfield steels [10]. After the formation of mechanical twins, large octahedral interstitial sites originally

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occupied by carbon atoms are converted to small tetrahedral ones. Consequently, serious distortions exist inside the mechanical twins in Hadfield steels unless carbon atoms undergo diffusive motions during the lattice shear process. Accordingly, the mechanical twins in Hadfield steels with much more carbon atoms have a higher hardness than those in the Co–33Ni alloy with very few interstitial atoms [10]. This can explains why the Co–33Ni alloy shows a lower work hardening rate than the Hadfield steel although they exhibit the same twinning kinetics.

Crystallographically, the occurrence of deformation-induced ε-martensite (DIEM) transformation also converts the octahedral interstitial sites to the tetrahedral ones. If enough carbon atoms exist, the concomitant carbon-induced distortions will also be produced in the ε-martensite. Based on this fact, a novel high manganese austenitic steel Fe–18Mn–5Si–0.35C with lower stacking fault energy (SFE) was designed with the aim of promoting the DIEM transformation under low stress by remarkably lowering the carbon content and increasing the silicon content [10]. The results confirmed our argument. That is, the Fe–18Mn–5Si–0.35C steel showed much higher work hardening rate at low stress than Hadfield steel under tensile deformation because of easy occurrence and rapid increase of the amount in the DIEM at low strain. In the present paper, the effects of SFE variations on the deformation mechanism and work hardening behavior in Fe–Mn–Si–C steels were further investigated by varying the silicon and carbon contents.

2. Materials and methods

2.1. Materials

The experimental steels were prepared by induction melting in the air, using the high purity iron, manganese, silicon and graphite. After homogenization at 1373 K for 15 h, the ingots were hot-forged to the round rods of 15 mm in diameter. The forged rods were solution treated at 1373 K for 40 min, followed by a water quenching. Table 1 lists the chemical compositions of the experimental steels and their SFE at 300 K based on thermodynamic calculation [11].

2.2. Tensile and impact test

The solution-treated bars were machined into two kinds of specimens to determine tensile properties and impact ductility, respectively. The specimens for the tensile test were cylindrical and button headed, whose gauge diameter and length were 10 mm and 50 mm, respectively. The specimens for the impact ductility were Charpy-U shape, whose dimensions were $10 \times 10 \times 55$ mm^3. The tensile tests were conducted according to ASTM: E-8M and performed using a tensile test machine (WAW-300B) at a cross-head speed of 1 mm per minute at room temperature. The mechanical properties of the experimental steels are also listed in Table 1.

2.3. Microstructure observations

The metallographic observations were performed using an optical microscope (OM, OLYMPUS, CK40M) and an electron back-scattered diffraction system (EBSD, TSL, OIM) attached to a field-emission scanning electron microscope (FE-SEM, JEOL, 6500F). The specimens for OM and EBSD observations were first mechanically ground and then electropolished in a solution comprising 10% $HClO_4$ and 90% C_2H_5OH (volume fraction). An optical color etching method was used in a solution comprising 1.2 g $K_2S_2O_5$ and 0.5 g NH_4HF_2 in water. The volume fraction of mechanical twins was measured based on the point-counting method using the OM images [12,13]. From stereology principles, the estimation of the volume fraction can be reached by sampling a random surface of a material, for example by estimating the proportion of a set of points superimposed to the object of interest [13]. For each specimen, an average value based on 10 random optical images was given. An area of 260 μm × 260 μm (a grid of 40 points × 40 points) was set on each image.

The constituent phases and their volume fractions were determined by X-ray diffraction apparatus (XRD, Philip, X'Pert Pro) with a speed of 0.04 degree per second. The X-ray source was Cu-K_α. The austenite (γ) reflections (200), (311), (222) and the ε-martensite reflections $(10\bar{1}0)$, $(10\bar{1}1)$ peaks were used to calculate their volume fractions, respectively.

Both the γ and the ε-martensite are paramagnetic in Fe–Mn steels, while the α′-martensite is ferromagnetic. Thus, a magnetic method was used to determine the volume fraction of α′-martensite. When the saturation magnetization per unit volume of a fully 100 pct α′-martensite is taken as 650 emu/cm^3 [14], the volume fraction of α′-martensite can be calculated using the following formula: $\alpha'(\mathrm{vol\%}) = (Ms/650) \times 100\%$, where Ms is the saturation magnetization (emu/cm^3). For the present experimental Fe–Mn–Si–C steels, the magnetization Ms at 15,000 Oe determined by a Superconducting Quantum Interference Device (SQUID, MPMS-7T, Quantum Design) was taken as the saturation magnetization, and their density was taken as 7.5 g/cm^3.

3. Results

3.1. Tensile properties

Fig. 1a shows the engineering stress versus engineering strain curves of the solution-treated Fe–17Mn–Si–C and Hadfield steels. Their true stress versus true strain curves are shown in Fig. 1b, in which both the elastic section and the section close to rupture were removed. Fig. 1c gives their corresponding work hardening rate versus true strain curves, which were determined from the data in Fig. 1b. The work hardening rate of Fe–17Mn–6Si–0.3C steel was the highest when the strain (true strain, here and after) was between 0.01 and 0.18. The work hardening rate of Hadfield steel was the lowest when the strain was below about 0.28, over which it was higher than that of Fe–17Mn–3Si–0.6C steel. The work hardening rate of Fe–17Mn–5Si–0.5C steel was higher than that of Fe–17Mn–3Si–0.6C steel at the same strain.

3.2. Microstructures

3.2.1. Microstructures of solution-treated samples before deformation

Fig. 2 shows the microstructures of the solution-treated samples before deformation, which were reported in our previous studies [11]. Only peaks of the γ appeared in all steels except the Fe–17Mn–6Si–0.3C steel, in which one obvious peak (10.0) of ε-martensite presented

Table 1
Chemical compositions, mechanical properties and stacking fault energy (SFE) at 300 K of experimental steels.

Steel	Element (wt.%)				Mechanical properties				SFE (mJ/m^2)
	Mn	Si	C	Fe	YS (MPa)	UTS (MPa)	El_t (%)	α_k (J/cm^2)	
Fe–17Mn–6Si–0.3C	17.10	5.71	0.32	Bal.	226	816	17.3	170	2.9
Fe–17Mn–5Si–0.5C	17.69	4.76	0.46	Bal.	306	927	26.0	261	8.1
Fe–17Mn–3Si–0.6C	16.42	3.0	0.59	Bal.	390	963	40.6	254	12.5
Hadfield	13.08	1.23	0.96	Bal.	370	904	40.9	>300	23.4

Note: YS, UTS, El_t and α_k represent the yield strength, the ultimate strength, the total elongation and the impact ductility, respectively.