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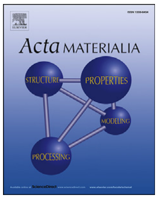

Full length article

Self-stabilization of untransformed austenite by hydrostatic pressure via martensitic transformation

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ABSTRACT

For improving the understanding of austenite stability in steel, hydrostatic pressure in untransformed austenite that is generated via martensitic transformation was evaluated from macro- and micro-viewpoints, and its effect on austenite stability was investigated in a Fe-27%Ni austenitic alloy. X-ray diffractometry revealed that the lattice parameter of untransformed austenite is continuously decreased via martensitic transformation only when martensite becomes the dominant phase in the microstructure. This suggests that the untransformed austenite is isotropically compressed by the surrounding martensite grains, i.e., hydrostatic pressure is generated in untransformed austenite dynamically at a later stage of martensitic transformation. On the other hand, microscopic strain mapping using the electron backscatter diffraction technique indicated that a finer untransformed austenite grain has a higher hydrostatic pressure, while a high density of dislocations is also introduced in untransformed austenite near the austenite/martensite interface because of lattice-invariant shear characterized by non-thermoelastic martensitic transformation. Furthermore, it was experimentally demonstrated that the hydrostatic pressure stabilizes the untransformed austenite; however, the austenite stabilization effect alone is not large enough to fully explain a large gap between martensite start and finish temperatures in steel.

1. Introduction

Various kinds of high-strength steel containing the austenite phase have been developed recently, e.g. low-alloyed transformation-induced plasticity (TRIP)-assisted steel and quenching and partitioning (Q&P) steel, where the austenite contributes to an enhancement in the mechanical properties because of its high strain hardenability and deformation-induced transformation. In order to develop and optimize the mechanical properties of this material class, the stability of austenite must be controlled as accurately as possible [1]. It is well known that the austenite stability is affected by factors such as chemical composition [2], austenite grain size [3,4], and stacking fault energy [5]. However, it is unclear why the gap between the martensite start and finish temperatures (*Ms* and *Mf*, respectively) is so large in steel, i.e., the austenite transformed at *Mf* is much more stable than that transformed at *Ms*. This implies that the austenite stability is originally variable or dynamically improved through martensitic transformation. Although no definite reason is established, this issue is usually discussed based on the following phenomena: (1) a variation in the non-uniform nucleation ability [6], (2) a dynamic austenite grain refinement [7,8], and (3) a rapid carbon diffusion leading to an atmosphere formation and a partitioning between austenite and martensite [2,9–11]. However, the effect of carbon is not obvious, because the temperature gap, *Ms*-*Mf*, is large enough even at an ultralow carbon level (more than 100 K) [12]. On the other hand, it is often reported that X-ray diffractometry (XRD) reveals that when the martensitic transformation occurs, the austenite lattice parameter is decreased [13–17]. These results enable us to understand that untransformed austenite is isotropically compressed because of the hydrostatic pressure caused by the martensitic transformation accompanied with a large volume

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expansion. Considering that an external hydrostatic pressure suppresses martensitic transformation [18,19], the untransformed austenite may be dynamically stabilized by the hydrostatic pressure upon quenching. In this study, the characteristics of hydrostatic pressure generated via the martensitic transformation were investigated via microscopic strain mapping using the electron backscatter diffraction (EBSD) technique, the so-called EBSD/Wilkinson method [20], as well as macroscopic XRD to understand the dynamic austenite stabilization phenomenon. The stabilization effect on untransformed austenite was subsequently investigated.

2. Experimental procedure

Fe-27%Ni austenitic alloy with chemical composition (wt.%) shown in Table 1 was used in this study. Ti was added to fix the tiny amount of solute C to prevent stabilization of the untransformed austenite by a rapid carbon diffusion. After multipass hot rolling and the subsequent sufficiently long homogenization, this material was austenitized at 1273 K for 1.8 ks, followed by water cooling to obtain a full austenitic structure. *Ms* of this material was measured to be 205 K by thermal dilatometry. This initial solution-treated material was subjected to subzero treatment at three different temperatures, 158, 143, and 77 K for 1.8 ks to stimulate athermal martensitic transformation in stages. Furthermore, the specimens subjected to subzero treatment at 158 and 77 K were tempered at 523 K for 3.6 ks, and then subzero-treated again at the same cryogenic temperatures. Since the tempering temperature is lower than the start temperature of the martensitic reversion of this material (approximately 573 K), not only diffusional reversion but also martensitic reversion from martensite to austenite never takes place through the heat treatment. The above heat treatment process is illustrated in Fig. 1. The microstructure was observed by optical and scanning microscopies. Crystallographic orientation was mapped via EBSD using a field emission scanning electron microscope (SEM) S-4300 developed by Hitachi High-Tech Ltd. The captured EBSD patterns were analyzed by a software, OIM analysis ver. 7.01, developed by TSL solutions. In addition to the volume fraction, the lattice parameters of austenite and martensite were measured at ambient temperature by XRD to macroscopically evaluate the hydrostatic pressure. The microscopic distribution of hydrostatic pressure was mapped by the EBSD/Wilkinson method using a dedicated software Cross Court 3.0 developed by TSL solutions. In this method, the shift between similar features in two EBSD patterns can be measured using two-dimensional cross-correlation functions and then elastic strain (e_{ij}) and crystal rotation (w_{ij}) components (i,j = 1,2,3) are separately determined on the basis of plane stress boundary condition using elastic constants compiled in the software [20–23]. Local hardness was measured at 2.0 mN constant load using a nano-indentation hardness tester with a diamond Berkovich indenter ENT-1100a developed by ELIONIX Inc., and then the nano-hardness was calculated from the obtained load–displacement data and converted to conventional Vickers hardness using an empirical formula [24]. Since the above experiments were conducted on a cross-sectional area of each specimen, the effect of residual stress on the specimen surface was ignored in this study.

Table 1
Chemical compositions of the alloy used in this study (mass%).

C	Si	Mn	P	S	Ni	Ti	Fe
0.003	0.036	0.836	0.002	0.003	26.75	0.15	bal.

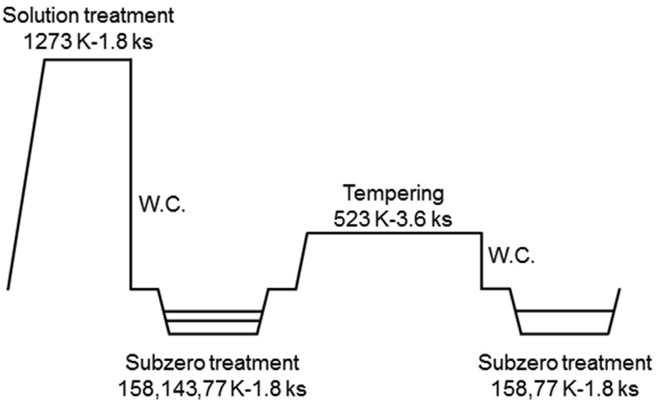

Fig. 1. Heat treatment route.

3. Results and discussions

3.1. Macroscopic characteristics of hydrostatic pressure in untransformed austenite

Fig. 2 shows the optical micrographs of (a) the solution-treated material and materials subjected to subzero treatment at (b) 158, (c) 143, (d) 77 K. The initial solution-treated material has a typical austenitic single-phase structure with an average grain size of 56 μm. It was confirmed by electron probe micro analysis that the austenitic structure has little Ni segregation. In contrast, lenticular shaped martensite is partially formed in subzero-treated materials and the fraction increases with decreasing subzero temperature (b–d). The athermal martensitic transformation behavior is represented in Fig. 3. The martensite fraction rapidly rises with a burst phenomenon and eventually exceeds 90% at 77 K. Fig. 4 exhibits a change in the lattice parameter of both (a) untransformed austenite and (b) the newly formed martensite as a function of the subzero temperature. Since XRD was conducted at ambient temperature after subzero treatment, the thermal expansion/contraction effect is not visible in this figure. It is interesting that the austenite lattice parameter continuously decreases below *Ms*, while the martensite lattice parameter has a constant value. This result clearly proves that hydrostatic pressure is generated via martensitic transformation dynamically and it intensively compresses the untransformed austenite, as reported in the previous studies [13–17]. Combining information from Figs. 3 and 4, a change in the austenite lattice parameter as a function of the martensite fraction is shown in Fig. 5. The austenite lattice parameter drops at an accelerated rate when the martensite fraction is higher than 60%. This experimental result strongly suggests that hydrostatic pressure is generated in untransformed austenite only when it is surrounded by martensite grains. In other words, the untransformed austenite experiences no hydrostatic pressure before a later stage of martensitic transformation where martensite becomes the dominant phase in the microstructure [25]. This is in agreement with a sense of micromechanics [26], and the absence of lattice parameter change in the martensite (Fig. 4) can be explained from this point of view.

Fig. 6 shows the optical micrographs of materials tempered at 523 K that were first subzero-treated at (a) 158 and (b) 77 K. It was found that martensite maintains a lenticular shape in both samples and no microstructural evolution occurs. In addition, XRD clarified that the martensite fraction never changes through this tempering process. The change in lattice parameter of the untransformed austenite upon tempering and the subsequent subzero treatment at the same temperature is displayed in Fig. 7. In this figure, the