Mechanism of Lüders deformation of advanced ultra-high strength steel

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1. Introduction

In the history of high strength steels, TRansformation Induced Plasticity (TRIP) steel proposed by Zackay et al. [1] in 1960s exhibited very attractive properties showing high strength and ductility (2 GPa and 20%), but the steel was not used commercially because of high cost due to Ni alloying *etc.*. Their TRIP steel is strengthened by ausforming, where high ductility is realized by TRIP accompanying Lüders deformation (Type C flow curve in Fig. 1). After 1980s, the idea to obtain high ductility by TRIP effect has been extended to less expensive steels; so-called low-alloyed TRIP steel with lower than 1 GPa showing continuous yielding (Type A in Fig. 1) was developed by Matsumura et al [2]. Recently, various advanced steels like nano-bainite, Q&P, high or medium Mn steels, *etc.* have extensively been studied. The target of "1.5 GPa strength and 20 % elongation" for advanced steels has been set in the NEDO project by the Innovative Structural Materials research Association (ISMA: 2013-22). Ultra-fine grain refinement is effective to increase strength but loses ductility showing Type B flow curve in Fig. 1. The utilization of TRIP is effective and several promising steels have been realized. Such steels including gum-metal type steel [3], medium Mn steel [4], ultra-fine grained low alloy TRIP steel [5] and Cr steel of the present study processed by R&P [6] exhibit commonly Type C flow curve. The reason of the appearance and mechanism of Lüders deformation are not made clear. The purposes of this study include (1) to reveal the microstructural evolution in Type C steel and (2) to elucidate the Lüders deformation mechanism accompanying phase transformation and dislocation motion, referring to the case of conventional Type A steel.

[1] V.F. Zackay, *et al.*, *Trans. ASM*, 60 (1967). 252.
[2] O. Matsumura, *et al.*, *Trans. ISIJ*, 27 (1987), 570.
[3] I. Miyazaki *et al.*, *CAMP-ISIJ*, 30 (2017), 359.
[4] X.G. Wang, *et al*, *Mater. Sci. Eng. A*, 674 (2016), 59.
[5] K. Asoo, *et al.*, *ISIJ Int.*, 51(2011), 145.
[6] Y. Matsumura, *et al.*, *CAMP-ISIJ*, 30(2017), 951.

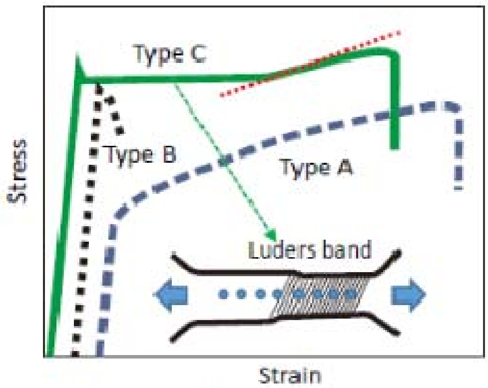

Fig. 1 Schematic drawings of continuous and discontinuous stress-strain curves.

2. Experiment

Sheets with 0.125 mm in thickness were prepared by cold rolling after austenitization treatment of a 16Cr-4Ni-3Mo-0.1C (mass%) steel (see ref. [6] for more details). *In situ* neutron diffraction measurements were performed during annealing and tensile testing using BL 19 at J-PARC MLF. First, 38 sheets of as-rolled (AR) specimen (5×20×0.125 mm) were stacked and annealed at 300, 400 and 500 ℃, respectively for 6 hs to monitor microstructural evolution. Next, tensile tests were carried out at RT for the AR and annealed specimens. Diffraction profiles along the tensile and transverse directions were simultaneously obtained using the two orthogonally installed detector banks, where 5×10 mm slit for the incident beam and 5 mm collimators for the diffracted beams were set. Tensile test specimens with a parallel portion of 6×55×0.125 mm were prepared in such a way that the tensile direction became 0, 45 and 90 degrees with respect to the rolling direction. The tensile tests were controlled by crosshead speed: 0.01 mm/min in the elastic region and then 0.05 mm/min (strain rate: 1.52×10^{-5}/s) in elasto-plastic region. In addition, pre-strained specimens with Lüders band were prepared and diffraction measurements were performed by travelling through the band front using 2×15 mm incident beam slit. The obtained results were analyzed using the Z-Rietveld software and the convolutional multiple whole profile (CMWP) fitting method.

3. Results

The AR specimen shows very strong texture and the microstructural change during annealing is very small as is observed in Fig. 1 for the 400 ℃ holding case. The change in volume fraction of the retained austenite was hardly observed. The Full-Width at Half Maxim of austenite phase was decreased by 500 ℃ annealing, suggesting the decrease of dislocation density which was confirmed by CMWP analysis.

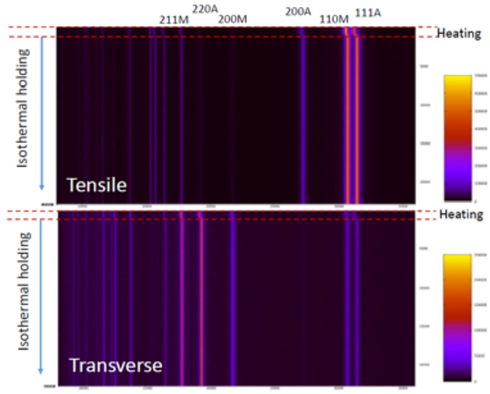

Fig. 1 Changes in diffraction profiles during annealing at 400 ℃.

Stress-strain curves of the AR and annealed specimens were presented in Fig. 2. The AR specimen exhibits continuous yielding whereas annealed specimens discontinuous yielding accompanied by Lüders band. The influence of tensile direction was also examined for the 400 ℃ annealed specimens and it was found the Lüders strain parallel to the rolling direction (0 degree, see Fig. 1) became smaller in 45 and 90 degree, indicating the influence of variant selection in deformation-induced martensitic transformation. Figure 3 shows the diffraction profiles obtained in the tensile direction after 13% (after Lüders deformation) for the annealed specimens and 10% for AR. As seen, the austenite peaks were hardly observed in the AR case indicating that the retained austenite was unstable. On the other hand, the austenite peak intensity decreases slightly with increasing of annealing temperature in Fig. 3. The lattice constant of austenite before deformation increased slightly with increasing of annealing temperature, suggesting the enrichment of carbon and/or nitrogen in the retained austenite. Hence, it is suspected the decrease in dislocation density for 500℃ annealing would make easier to transform to martensite during tensile deformation. The locking of dislocations and austenite stabilization by annealing is speculated to cause discontinuous yielding. The microstructure was found drastically to change at the Lüders band front; martensite volume fraction was large within the band.

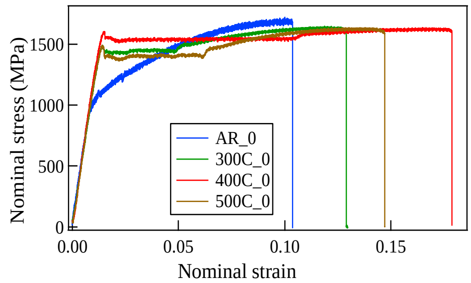

Fig. 2 Stress-strain curves of the four kinds of heat-treated specimens.

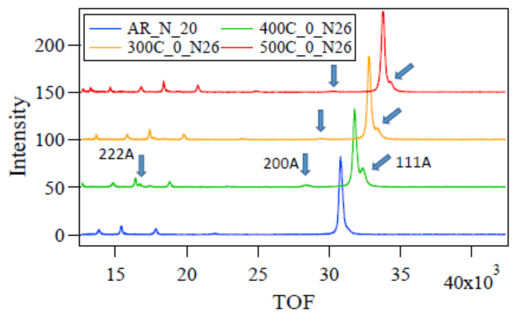

Fig. 3 Comparison of diffraction profiles obtained along the tensile direction for the four kinds of heat-treated specimens.

4. Conclusions

The ultimate tensile strength is determined mainly by austenite stability, *i.e.*, martensite volume fraction, whereas the Lüders elongation and uniform elongation are not only by austenite stability but also transformation shear strain with respect to variant selection and applied stress direction. The transition of yielding manner is likely caused by the locking of dislocations with annealing.