Microstructure evolution of 316L stainless steel during simple shearing

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1. Introduction

Austenite in steel exhibits deformation induced transformation into martensite following the sequence of γ austenite → ε martensite → α´ martensite transformation [1, 2]. However, some researchers suggested the formation of α´ transformation has been shown conclusively to be independent of ε martensite [3, 4]. It is suggested that α´ martensite is formed at highly stress-concentrated regions such as the impact point of mechanical twins and grain boundaries [5, 6]. Thus, fine α´ martensite is expected to be formed preferentially at quite local regions such as intersections of mechanical twins with a thickness of a few nanometers. The amount of α´ martensite is known to be dependent on the deformation methods, amount of plastic strain.

In this study, forward or reverse shearing was applied to 316L stainless steel samples with different strain amounts, and the microhardness variations were investigated. It was observed that the evolution of α´ martensites was more active while the sample was subjected to a reverse shearing after a forward one. Moreover, the amount of α´ martensites increased as the shear strain amount increased. The hardness value became also higher when the sample was subjected to the reverse shearing after the forward one.

2. Methods and Results

The chemical composition of the 316L stainless steel sample was as follows (in wt%) : 17% Cr. 12% Ni, 2% Mn, 0.03% C, 1% Si, 0.03% S, and 2.5% Mo; with the balance Fe. Simple shear strained cylindrical 316L stainless steels are shown in figure 1. The diameter of clamping area of the sample was three times thicker than the central cylindrical area which had 5mm in diameter. The straining were controlled by torsion. A black line was drawn on each sample surface before straining, and the total amount of shear strain could be measured after straining. In figure 1(b), 1(c), 1(d), and 1(e), it is observed that the shear strain was controlled up to 0, 0.4, 0.8, and 1.6, respectively.

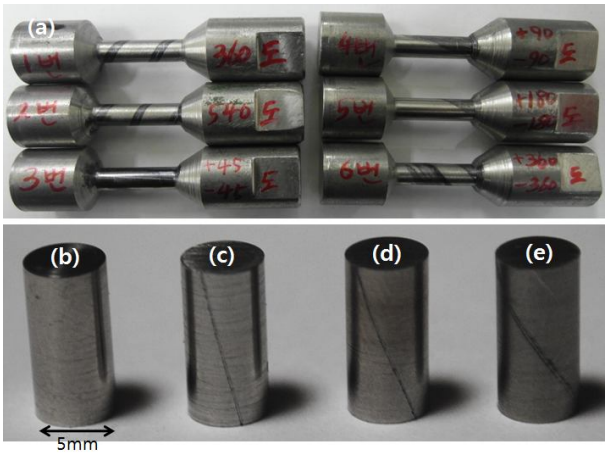

Figure 1. Simple shear strained cylindrical 316L stainless steels : The shearing amounts were controlled up to 0, 0.4, 0.8, and 1.6 in figure 1(b), 1(c), 1(d), and 1(e), respectively

The microstructure and microhardness were evaluated in the perpendicular plane to the radius. In order to ensure that all the samples have the same surface condition, samples were finished by an electrolytic polishing with the same voltage and etchant flow condition. Electro polishing was effective to relieving the local defects on the surface. An SEM/EBSD system (JSM 7000F / Oxford INCA) was utilized to observe the microstructural evolutions. α´ martensites with a few nanometers in size were found at the intersections of mechanical twins by high-resolution TEM (JEM 2100F). By using automated diffraction pattern analysis system(ASTAR/REDS), the γ and α´ phases were identified in the micrographs, and the twin boundaries were distinguished from the conventional high angle boundaries.

In figure 2, the microhardness of 316L stainless steel samples are plotted versus the variables of the shearing methods and the amounts. It was observed that the hardness value generally increased as the shear strain amount increased. Moreover, the hardness value became higher when the sample was subjected to the reverse shearing after forward one. It suggests that hardening of 316L stainless steel is dependent on both shear straining mode and the amounts.

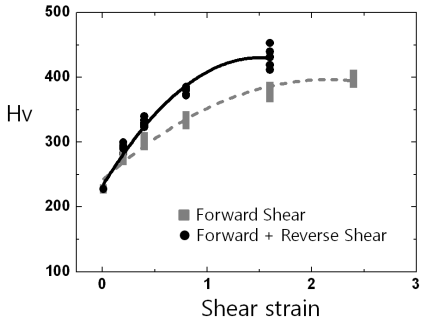

Figure 2. Microhardness values of 316L stainless steel samples versus the shear straining methods and amounts

In figure 3, the deformed microstructures were observed by EBSD band contrast map in the perpendicular plane to the radius. The microstructure of unstrained 316L stainless steel is shown in figure 3(a) as reference. The average grain size was estimated as 25 ㎛ in diameter, and the planar twin boundaries were observed inside the grains. In figure 3(b), 3(c), and 3(d), the shear deformed microstrucutures were observed; the grains became more aligned to shear band direction as the shearing amounts increased. The grain size was observed several hundreds nanometers. It is also observed that tens of nanometer sized twin boundaries are developed inside the grains.

In figure 3(e), 3(f), and 3(g), forward and then reverse shear deformed microstrucuture are demonstrated. The grains were refined to hundreds of nanometer in size. However, it was observed that the grain shapes were not aligned to the shear band direction. It suggests that the strain hardening behavior in forward and then reverse shear would be different from the forward shearing condition. Refering to the higher hardness values in figure 2, fine α´ martensite is expected to be formed in the forward and then reverse shear deformed sample.

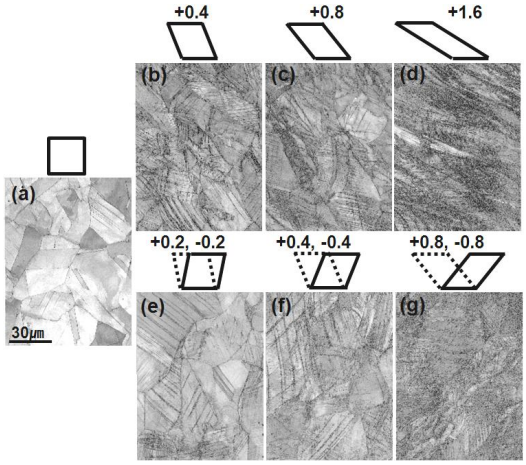

Figure 3. The forward and then shear deformed microstructures were observed by EBSD band contrast map. Schematic diagrams in each figures showing the different deformation mode and strains amount

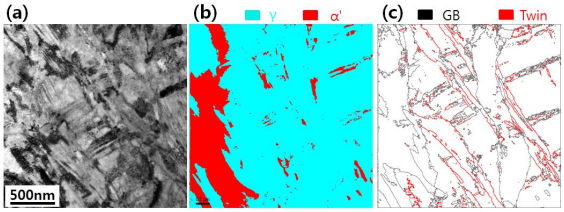

Fig. 4. TEM analysis of forward and then reverse sheared sample : (a) TEM bright field image, (b) γ and α´ phase map, (c) grain and twin boundary distributions

In figure 4, the nucleation of fine α´ martensite is shown. A TEM bright field image of forward and then reverse sheared sample is shown in figure 4(a). By using automated diffraction pattern analysis system (ASTAR/REDS), the γ and α´ phases were identified in figure 4(b) and the twin boundaries were distinguished from the conventional high angle boundaries in figure 4(c). It was observed that fine α´ martensite is formed quite local intersections of mechanical twins with a thickness of a few nanometers.

3. Conclusions

In this study, forward (and reverse) shear strains were applied to 316L stainless steel samples with different strain amounts, and the microstructural change and the microhardness are investigated. The α´ martensites was formed preferentially at local intersections of mechanical twins with a thickness of a few nanometers. The same tendency was also found in the microhardness variations.

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