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Anisotropic strengthening of nanotwinned austenitic grains in a nanotwinned stainless steel

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ABSTRACT

Nanotwin lamellae orientation effect on tensile properties of nanotwinned grains strengthening stainless steels was investigated. It is found that the nanotwinned grains with twin lamellae oriented roughly parallel to loading direction exhibit the much higher strengthening effect (at least a GPa higher) than the nanotwinned grains containing twin lamellae inclined to loading direction, which is related to the anisotropic strengthening mechanisms of nanotwins.

Nanotwinned austenitic steels have attracted great interests in the past few years as a novel type of advanced high performance steels [1–8]. They are commonly based on a single-phase duplex-microstructure consisting of nanotwinned grains embedded in the statically recrystallized (SRX) grained matrix. These nanotwinned austenitic (referred to as *nt*-γ) grains act as "hard inclusions" to strengthen austenitic matrix due to containing a high density of nanoscale deformation twins [1–4]. Attractively, *nt*-γ grains not only possess the same elastic modulus as the matrix, but also co-deform homogeneously together with the surrounding soft matrix without generating notable strain localizations around the interfaces at low strains [9,10]. Hence, in comparison with conventional dual-phase (DP) steels, the nanotwinned austenitic steels exhibit an enhanced combination of strength and ductility [1,2,4].

Apparently, the mechanical properties of nanotwinned austenitic steels can be optimized by tailoring the structural parameters of *nt*-γ grains through various thermomechanical treatments. It has been reported that the higher volume fraction of *nt*-γ grains and the thinner of twin/matrix (T/M) lamellae in the samples are introduced, the higher strength can be achieved [2,3], like that in traditional DP steels [11,12]. But recent investigations [2,3,10] indicated that such relationship is not monotonously changing. For example, the strength did not rise but slightly reduced when the volume fraction of *nt*-γ grains increased from ~25 vol% to ~50 vol% in nanotwinned 316L samples with the fixed twin thickness [2,10]. So the strengthening of *nt*-γ grains is also influenced by other factors.

Investigations of electro-deposited nanotwinned Cu [13] indicated that the twin boundary (TB) orientation with respect to loading direction leads to an anisotropic nanotwin strengthening response. It implies twin lamellae orientation-dependent strengthening of *nt*-γ grains. Accordingly, in the present work, two types of nanotwinned 316L stainless steels were produced by means of dynamic plastic deformation (DPD) followed thermal annealing. One is strengthened by large *nt*-γ grains containing T/M lamellae which are roughly inclined to the tensile axis (TA) direction (referred to as *I-nt* 316L) and the other is strengthened by fine *nt*-γ grains containing T/M lamellae which are parallel to TA (referred to as *P-nt* 316L). The objective of this work is to investigate the relationship between the T/M lamellae orientation and the *nt*-γ grain strengthening effect.

The material studied in this work is a commercial AISI 316L stainless steel with a composition of Fe-16.42Cr-11.24Ni-2.12Mo-0.02C-0.37Si-1.42Mn-0.011S-0.040P (wt%). The original samples were solution-heat-treated at 1200 °C for 1 h and followed by water quenching to obtain a uniform austenitic structure with an average grain size of ~100 μm. The cylindrical samples were processed on a dynamic plastic deformation (DPD) facility with a strain rate of ~10^2–10^3 s^{-1} at room temperature to an accumulative strain of $\varepsilon = 0.8$. The DPD set up and processing parameters were described elsewhere in details [2]. Microstructures were characterized by using field emission gun scanning electron microscope (SEM) FEI Nova NanoSEM 430 with electron channeling contrast (ECC) imaging and a transmission electron microscope (TEM) JEOL 2010 operated at 200 kV.

The *I-nt* 316L stainless steels were produced by means of DPD to the coarse grained samples with random grain orientation followed by annealing at 770 °C for 90 min. As shown in Fig. 1a, the microstructure is mainly composed of *nt*-γ grains (outlined) embedded in SRX grains.

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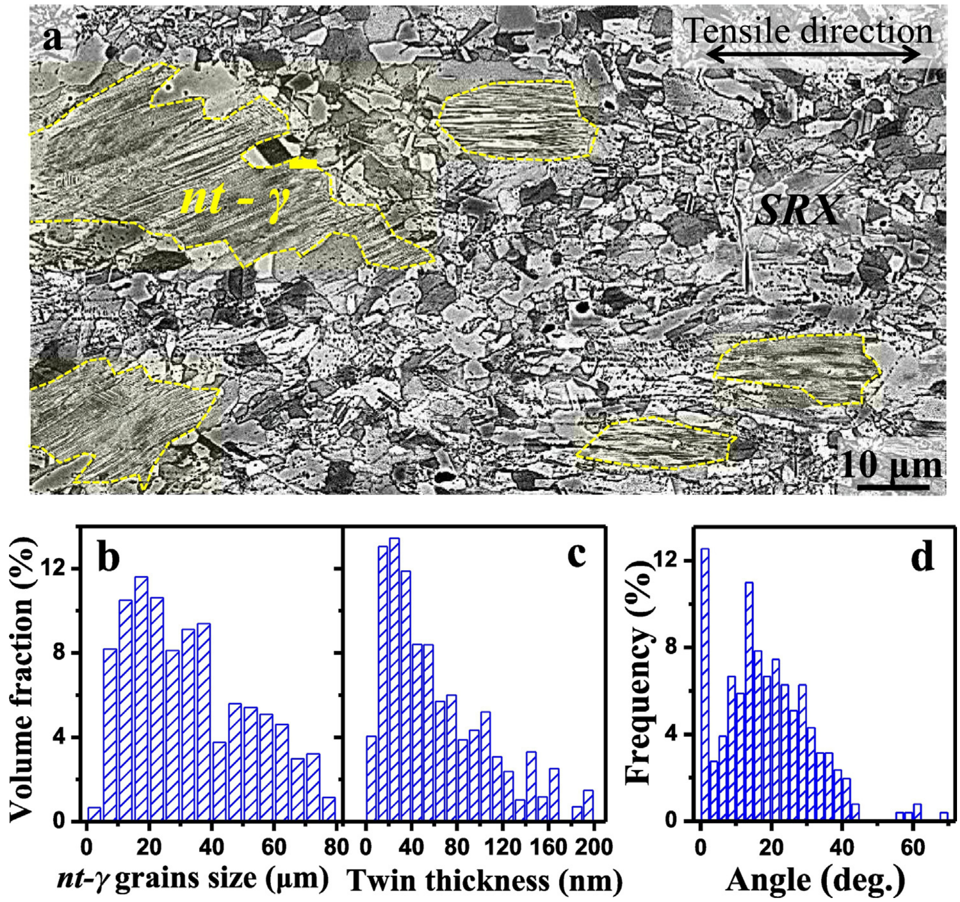

Fig. 1. (a) A typical SEM-ECC image showing the microstructure of the *I-nt* 316L samples mainly consisting of large *nt*-γ grains (outlined) embedded in the recrystallized (SRX) grain matrix. Statistical distribution of *nt*-γ grain size (b) and the twin/matrix (T/M) lamellar thickness (c) as well as angle distribution of twin boundary (TB) traces with respect to the tensile direction (d).

These *nt*-γ grains exhibit an average size of ~23 μm, much larger than SRX grains (averagely 2.7 μm) (Fig. 1b). They constitute ~20% in volume and contain a high density of T/M lamellae with the average thickness of ~31 nm by TEM statistics (Fig. 1c). The *nt*-γ grains distribute randomly in the matrix, but the orientation distribution of T/M lamellae inside *nt*-γ grains is not homogeneous. Most TBs are inclined to the tensile axis (TA) direction, as shown in Fig. 1d, ranging from 0° to ~45°, averagely ~19°. Rough estimation shows that the ~75% traces of TBs are inclined (>10°) to the TA direction while ~25% parallel to the TA direction (< 10°). It is noted that the statistical measurement of the angles between the TA and TB traces is more operable and efficient than that between the TB and the DPD direction (i.e. Compression axis, CA) by EBSD. In fact, the two measurements are self-consistent since the normal directions of most twin planes distribute randomly in the 3D-space [2,3, 14]. That is to say, most twin planes in *I-nt* 316L are also inclined to the CA direction.

The *P-nt* 316L samples were produced by DPD to the fully recrystallized samples (with average size of 4.6 μm and random orientation distribution) and then subsequently annealing at 730 °C for 30 min. As shown in Fig. 2a-b, the typical microstructure is also mainly consisted of the SRX grains matrix "dispersed" with *nt*-γ grains. Attractively, the *nt*-γ grain size is very small, ranging from several hundred nanometers to several micrometers (usually <4.5 μm) with the average transverse size of ~1.7 μm (Fig. 2c), which is nearly identical to the size of SRX grains (averagely 1.6 μm). Statistics show that the volume fraction of *nt*-γ grains and the T/M lamellar thickness is ~16% and ~29 nm (Fig.2d), respectively. These *nt*-γ grains "disperse" very homogeneously in the samples. But the angle distribution of TB traces with respect to TA is narrow, as shown in Fig. 2e, roughly ~70% angles below 10°, averagely 9°. Clearly, most T/M lamellae are roughly parallel to the TA direction, which means that they are almost perpendicular to the CA direction, analogous to the previous results [3,14].

Table 1 summarizes the microstructure characteristics of the two types of samples. Clearly, there are no significant differences in constitution and characteristic size except the size of *nt*-γ grains and T/M lamellar orientation distribution. The average size is ~23 μm for large *nt*-γ grains and ~1.7 μm for small ones. Meanwhile, most T/M lamellae are inclined to TA direction in large *nt*-γ grains of the *I-nt* samples but parallel to TA direction for small *nt*-γ grains of the *P-nt* samples. Such different twin lamellae orientation distribution may be due to the different degree of grain rotation at the same DPD strain of 0.8 [14,15]. The large grains containing multiple nanotwins may rotate slowly while the small nanotwinned grains rotate easily with increasing the strains. This microstructural distinction significantly influences the mechanical properties of nanotwinned 316L samples.

Uniaxial tensile tests with an initial strain rate of 5×10^{-3} s^{-1} at room temperature were performed in an Instron 5848 Micro-Tester system equipped with a contactless MTS LX300 laser extensometer to measure the tensile strain upon loading. As shown in Fig. 3, the *P-nt* 316L samples exhibit a high yield strength of ~760 MPa, which is 220 MPa higher than that of *I-nt* ones. Interestingly, the uniform elongation of the *P-nt* 316L is still as high as ~20%, ~6% lower than of the *I-nt* counterparts. It means that the yield strength is elevated by ~40% at the expense of a 20% loss in uniform elongation for the *P-nt* 316L samples, exhibiting an enhanced combination of strength and ductility.

To investigate and compare the strengthening mechanisms between the two samples, tensile deformation behaviors of SRX grains surrounding *nt*-γ grains in both samples are characterized at uniform strains of ε