Martensitic transformation in SUS 316LN austenitic stainless steel at RT

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Abstract The results of high-accuracy magnetic measurements on SUS 316LN austenitic stainless steel compressively deformed at room temperature (RT) are reported here. Even after the mild deformation of $\sim$25% true strain (ε_t), the ferromagnetic phase (α'-martensite) could be clearly observed which increased sharply on further deformation. The amount of α' was very small (0.18 vol% at $\varepsilon_t \approx 60\%$) when compared to the reported data for other grades of austenitic steels such as 304, 304L, 316, and 316L. The strain-induced α'-martensite is further studied by magnetic hysteresis loops. The coercivity (H_C) and remanence (M_r) were analyzed by subtracting the paramagnetic contribution of the bulk austenite structure. While H_C was found to decrease with α', M_r remained the same ($\sim$67 emu/g) when normalized to the volume fraction of α'. The decreasing H_C with increasing α' and/or ε_t is presumed to be due to the domain wall pinning at the grain boundaries when the cluster size exceeds the domain wall width.

Introduction

Austenitic stainless steels are extensively used in widespread industrial and technological applications where the combination of high corrosion resistance with high strength, toughness, and stiffness is needed. The different grades of austenitic steels used in the hostile environments

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(like the one prevailing in nuclear reactors—coolant water of high temperature and pressure) are SUS 304, 304L, 316, 316L, and 316LN. Type 316LN is of superior quality among 300-grade stainless steels. It is a molybdenum-bearing nitrogen-alloyed low-carbon austenitic steel, which has high creep resistance, excellent formability, rupture and tensile strength at high temperature, corrosion, pitting resistance, etc. Nitrogen added to the alloy helps in strengthening the solid solution, while the low carbon content makes it resistant to sensitization in short-term exposures. Comparatively, it is an alloy of choice to minimize any occasional but catastrophic failures (e.g., environmentally assisted cracking). Hence SUS 316LN is considered as the major fabrication alloy, at least, in the crucial components of the upcoming reactors.

It is well known that due to strain-induced deformation, the austenite crystal structure (fcc, γ) of stainless steel, which is paramagnetic at room temperature (RT), is transformed to ferromagnetic martensite (bct, α') [1–7]. This transformation can be related to deterioration of the material property for the safe operation of steel-based plants, because it can lead to cracks due to the hard and brittle nature of martensite phase and associated microstructural changes. Assessing the degradation of stainless steel by nondestructive magnetic method is of current interest [3–14] because magnetic properties are very sensitive to the changes in microstructures.

The extent of $\gamma \rightarrow \alpha'$ transformation depends on the chemical composition of the alloy, deformation level, and temperature. Extensive studies on $\gamma \rightarrow \alpha'$ have been carried out using 304 austenitic steel by various deformations like compression [4], rolling reduction [5, 13, 14], and tensile testing [3, 7, 9, 15]. The $\gamma \rightarrow \alpha'$ was found to be accomplished relatively easily in 304 grade of stainless steel owing to its composition, especially when lower amounts

of Ni (<10 wt%) and/or higher amounts of Cr (>18 wt%) are present. This is because alloying elements like Ni, Mn, N, and Cu considerably stabilize the austenite structure while the Cr promotes α' structure. There are few studies on the deformation-induced α' in 316 stainless steel [15–20]. For instance, tensile testing [15] and selective generation of local ferromagnetism by nanoindentation [19] have been reported for the type 316 stainless steel. In the case of 316L stainless steel, severe deformations such as cold-drawing of wire [17] have produced a significant amount of α' as shown by X-ray diffraction (XRD) and transmission electron microscopy (TEM). Similarly, ball milling of 316L stainless steel powder also produced α' as determined by XRD and Mössbauer spectroscopy [20]. Miller et al. [12] have used magnetic force microscopy to image strain-induced α' resulting from fatigue crack growth in a 316L stainless steel. On the other hand, Botshekan et al. [18] have found no α' in 316LN deformed (tensile) up to rupture at RT, unlike when deformed in the low temperature of 77 K. Based on these studies, $\gamma \rightarrow \alpha'$ in different grades of austenitic stainless steels can be shown, broadly, in the decreasing order as $304 > 304L \gg 316 > 316L > 316LN$ when deformed to the same extent at a given temperature.

It is generally believed that in 316LN stainless steels, such $\gamma \rightarrow \alpha'$ transformation is negligible when deformed at RT. However, it is of interest to see if there is any fractional amount of α' when the so-called stable steel, SUS 316LN, is deformed. The results of high-accuracy magnetic measurements on SUS 316LN compressively deformed at RT are reported here. It was found that the ferromagnetic fraction could be clearly observed even after the mild deformation of $\sim 25\%$ true strain. Nevertheless, the amount of α' was very low (<0.2 vol%), which is much less than that occurs in other grades of austenitic steels when deformed to same extent.

Experimental

The chemical composition of the SUS 316LN austenitic stainless steel used in this study is given in Table 1. As-received alloy (from The Japan Steel Works Ltd.) was in the form of a 1-cm thick disk sliced from a rod with a

Table 1 Chemical composition of SUS 316LN (wt%)

C	Si	Mn	P	S	Ni
0.015	0.58	1.39	0.022	<0.0005	12.71
Cr	Mo	N	Co	Cu	Fe
17.29	2.52	0.112	0.04	0.08	Bal.

diameter of ~ 8 cm. Then, it was cut to $3 \times 3 \times 2$ mm size using an electric discharge wire-cutting machine. The samples were solution annealed at 1,050 °C for 80 min in vacuum and then water quenched. Before compression, the samples were electro-polished using a mixture of ethanol and perchloric acid in 9:1 volume ratio and then cleaned with ethanol in an ultrasonic bath. Samples were compressed at RT (at a strain rate of 0.5 mm/min) using a Shimadzu Autograph Type SFL-250 kN AG Instron testing machine. The different levels of compression up to 61.4% of true-strain (ε_t) were obtained. Magnetization (M) was measured at 300 K using a superconducting quantum interference device magnetometer, Quantum Design MPMS-XL. The applied field (H) was perpendicular to the axis of compression.

Results and discussion

Saturation magnetization or α'-martensite versus ε_t

As can be seen in Fig. 1, the initial magnetization curve of an undeformed ($\varepsilon_t = 0$) sample is linear in the entire field range showing the typical paramagnetic behavior due to pure austenite phase. Such a linear behavior was seen for 7.5% and 15% compressed samples also (not plotted here). However, after compressing to $\varepsilon_t > 25\%$, magnetization increased sharply in the lower field region, and thereafter gradually increased with applied field. The first part of magnetization curve can be attributed to the strain-induced ferromagnetic phase (α'). This ferromagnetic component becomes magnetically saturated soon after the field was applied, and the saturation field varied from 0.5 to 5 kOe depending on its volume fraction.

Figure 2 shows the derivative plots for the deformed samples. In corroboration with the data in Fig. 1, the increasing derivative maxima can be clearly seen after $\varepsilon_t \approx 25\%$.

It is well known that only α' structure contributes to the ferromagnetic part, due to strain-induced $\gamma \rightarrow \alpha'$ transformation of austenitic steels at RT and above [3–18]. To evaluate the amount of α', the saturation magnetization (M_s) was obtained from the plots in Fig. 1 corresponding to the intercept of the linear extrapolation from the high-field region to ordinate, and plotted in Fig. 3. The magnetization of the deformed samples after subtracting the paramagnetic contribution (equivalent to magnetization of undeformed sample) is shown in the inset of Fig. 3, and the M_s obtained from these plots is same as that from Fig. 1. M_s of 100% martensite-transformed 304-grade steel is reported to be ~ 153 emu/g [21], while it is ~ 143 emu/g for 316LN-grade steel [18]. Accordingly, Fig. 3 also shows the amount of α' which increased sharply when $\varepsilon_t > 25\%$.

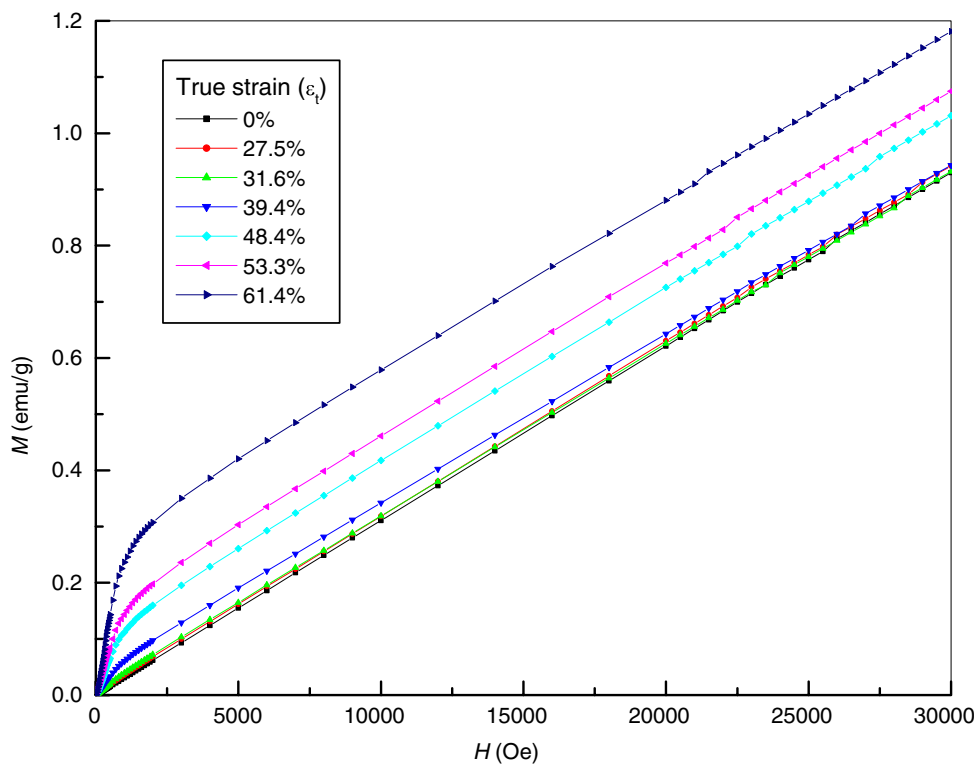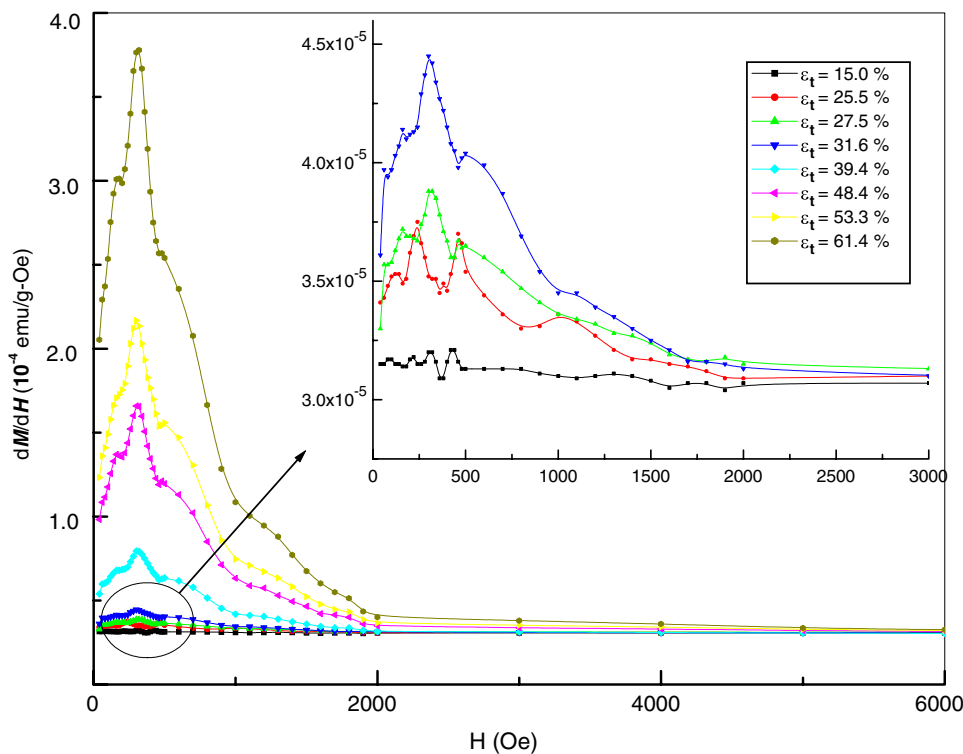**Fig. 1** Initial magnetization curves for different levels of compression measured at 300 K


Fig. 2 Derivative of magnetization to magnetic field as a function of applied magnetic field for different levels of compression

However, the amount of α' obtained here was very small, 0.18 vol% at $\varepsilon_t \approx 61.4\%$. This suggests that the $\gamma \rightarrow \alpha'$ transformation is highly hindered in SUS 316LN steels unlike in the other grades such as 304, 304L, 316, and 316L austenitic steels [3–18].

During plastic deformation, the martensitic transformation takes place by shearing of the lattice like generation of twins or stacking faults. Due to plastic deformation, the dislocation density increased and, in consequence, also the stacking fault density. In fcc lattices, stacking faults (SF) are

Fig. 3 Saturation magnetization and volume percent of α'-martensite as a function of true strain. Inset: magnetization plots after subtracting paramagnetic contribution
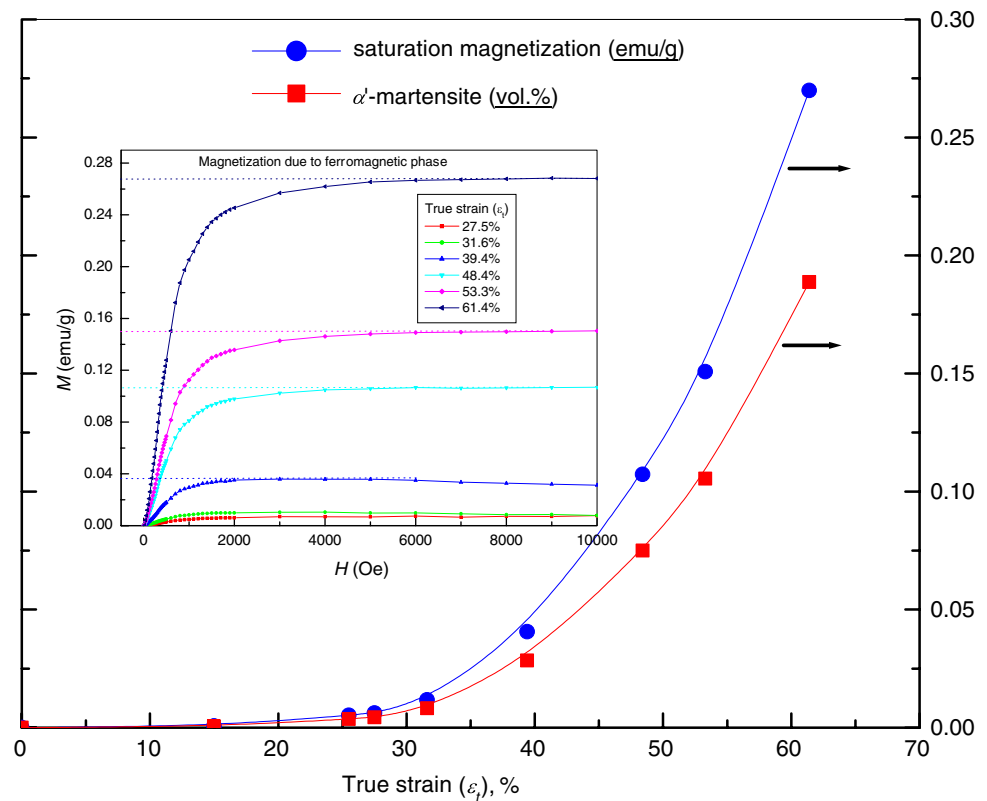
of hexagonal symmetry, like ε-martensite (hcp); n stacking faults can form a hexagonal ε-martensite crystal. The transformation from austenitic to α'-martensite was thus believed to follow [22, 23], $\gamma \rightarrow \text{SF} \rightarrow n\text{SF} \rightarrow \varepsilon \rightarrow \alpha'$.

The approximate amount of α' in 300 grades of austenitic stainless steels when deformed to the same extent, $\varepsilon_t \approx 65\%$ at 300 K, could be shown in the order: 304 (90%) > 304L (30%) >> 316 (3%) > 316L (2%) >> 316LN (0.2%)—value in parentheses is the amount of α' in vol%. The large difference in $\gamma \rightarrow \alpha'$ transformation especially between 316L and 316LN is usually attributed to the alloying element nitrogen. Nitrogen in austenitic solid solution is an enormously useful element with respect to austenite stability, corrosion resistance, and mechanical properties. The superior properties of 316LN, such as better resistance to fatigue crack growth, irradiation embrittlement, and high-temperature oxidation have been ascribed to the presence of nitrogen. During the strain-induced deformation, nitrogen is known to make an obstacle for the movement of dislocations. The role of nitrogen seems to be more complex, decrease in stacking fault energy (SFE) and other mechanisms (short-range order effects), which can favor planar slip of dislocations or, in other words, to reduce the tendency for cross slip and avoid a tri-dimensional cellular structure [23, 24]. Vogt et al. have reported [24] the TEM observation of 316L and 316LN after a low-cycle fatigue test at 300 K. Accordingly, in 316L, the dislocations are entangled and arranged in uncondensed and

poorly defined walls. Increase in strain amplitude ($\Delta \varepsilon_t$) results in a more homogeneous structure having well-formed misoriented cells. On the other hand, in 316LN, the dislocations are quite homogeneously distributed, often confined in their [111] slip planes, and sometimes they tend to form planar bands. At higher $\Delta \varepsilon_t$, the dislocation walls are formed but are not very condensed and are less often misoriented than with 316L. Further, for comparison, the α' formed (due to low-cycle fatigue test, $\Delta \varepsilon_t = 2.5\%$) at 77 K was ~ 66 vol% in 316L while it is only ~ 4 vol% in 316LN. The occurrence of a difference in the mechanism of $\gamma \rightarrow \alpha'$ transformation at low temperature presumably originated by the effect of nitrogen at low temperature through a decrease in SFE and a reduced mobility of the dislocations, martensite transforms directly from austenite.

Hysteresis loops—coercivity (H_C) and remanence (M_r)

To demonstrate the ferromagnetic behavior further, the hysteresis loops were measured in the field range of ± 30 kOe. Figure 4 shows the enlarged view of the as-measured loops of deformed samples in comparison with undeformed sample. It is evident that the loops are bulged as the α'-martensite was increased with ε_t. Inset in Fig. 4 shows the full view of loops of undeformed and maximum deformed sample, as a typical case.

Fig. 4 Enlarged view of the as-measured M-H loops for different levels of compression. Inset is the full view of undeformed and the maximum deformed sample
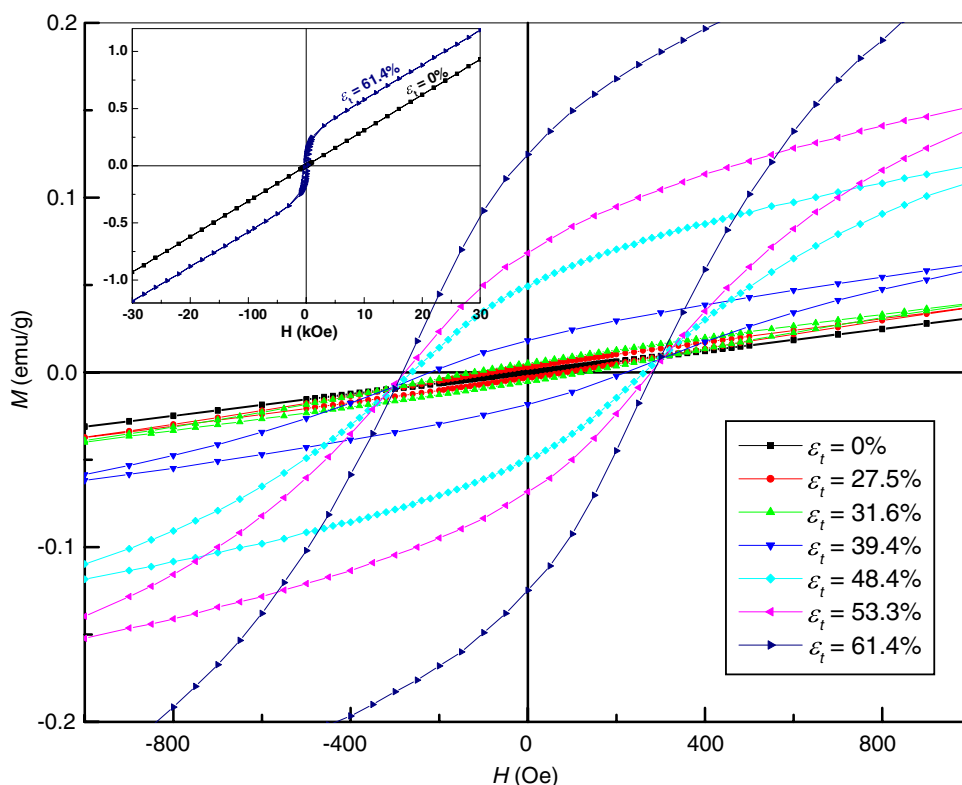

To estimate the actual behavior of H_C and M_r versus α' fraction and/or ε_t, it is essential to subtract the paramagnetic contribution especially when a very small amount of α' was formed (as in the present study). Accordingly, Fig. 5 shows the select view of the loops after subtracting the paramagnetic contribution (equivalent to the undeformed sample), and normalizing to the corresponding amount of α'. As shown in the inset of Fig. 5, H_C was found to decrease with increasing α'. On the other hand, M_r remained more or less the same at ~ 67 emu/g suggesting that M_r is independent of the ferromagnetic volume fraction.

The variation of H_C as a function of the volume fraction of α' and/or strain rate has been reported in many of the studies [7, 13, 14, 25], but the decreasing or increasing trend of H_C is not consistent. For instance, H_C was reported to decrease continuously with volume fraction of α' in SUS 304 after tensile testing [7], cold-rolled AISI 304L steel [13], and in the mechanically milled 304 stainless steel powder [25]. On the other hand, H_C increased with α' in AISI 304 wherein the intermediate amount of martensite was formed [14]. Further, it is not clear in those studies whether the paramagnetic contribution is subtracted or not. Hence, the results obtained from Fig. 5 here could not be compared directly with the previous reports. Nevertheless, the decreasing or increasing trend of H_C may be attributed

to the different microstructural interactions of α', based on its volume fraction [14, 26].

As revealed in the study of Mumtaz et al. [4], at about 0.12 vol% of α', the size was in the order of nanometer scale (<100 nm). Thus, it is presumed here that the isolate α' particles (nanoclusters or grains equivalent to the thickness of domain wall) are distributed wide apart and that the intercluster interaction is negligible while the intracluster interaction is dominant. On increasing the amount of strain, the cluster size and also number density are expected to increase. When the cluster size exceeds the domain wall width, the magnetization process is determined by domain wall pinning at the grain boundaries. Under such situation, H_C is inversely proportional to the grain size [4, 26, 27]. Our experimental data of H_C versus martensite (Fig. 5 inset) indeed show this behavior. On the other hand, random anisotropy model [14, 26, 27] explains the decreasing H_C based on intercluster exchange interactions (when a significant amount of α' was formed) and increasing H_C based on intracluster exchange interactions (when an intermediate amount of α' was formed). However, in the present study, total amount of α' is too small to invoke such exchange interaction. Nevertheless, more detailed studies are needed to support the microstructural information of α' and will be considered in our future studies.

Fig. 5 M-H loops after
subtracting the paramagnetic
contribution and normalized to
the corresponding weight
fraction of α'-martensite. Inset
(middle) is the coercivity (H_C)
as a function of α'-martensite
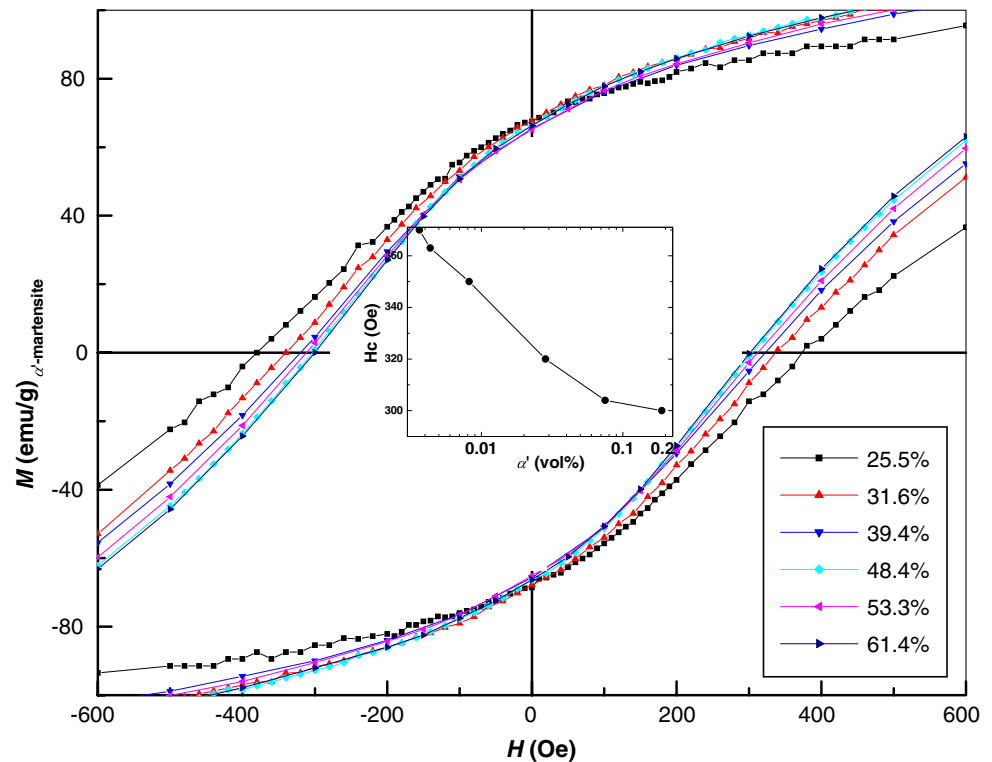
Conclusion

The ferromagnetic fraction in SUS 316LN austenitic stainless steel was clearly seen when deformed (compressively) at RT, due to $\gamma \rightarrow \alpha'$ transformation. However, such a transformation was considerably hindered, and the α' was <0.2 vol% when deformed to ~60% true strain. H_C and M_r were analyzed by subtracting the paramagnetic contribution of the bulk austenite structure. While H_C was found to decrease with α', M_r remained the same (~67 emu/g) when normalized to the volume fraction of α'. The decreasing H_C with increasing α' and/or true strain is presumed to be due to the domain wall pinning at the grain boundaries when the cluster size exceeds the domain wall width. We believe that these results can serve as the experimental evidence for the small fraction of α' formation in SUS 316LN, which is a less common alloy to study by the magnetic method when deformed at RT, unlike low-temperature deformations.

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