# Matrix switch related to microstructural connectivity and its effect on strength in metals with duplex microstructure

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## ARTICLE INFO

## ABSTRACT

To understand the microstructural factors dominantly affecting the mechanical properties of metals with duplex microstructures on a mesoscale, the change in the 0.2% proof stress as a function of the hard austenite fraction of an Fe–Ni austenitic alloy having a duplex microstructure composed of soft and hard austenite was investigated in terms of the microstructural connectivity. The 0.2% proof stress continuously increased with increasing fraction of the hard austenite. However, the strengthening behavior changed when the hard austenite fraction exceeded a critical value; the 0.2% proof stress shifted from the lower limit stress to the upper limit one calculated by secant method at the critical fraction. This demonstrates that the matrix of duplex microstructure was switched from soft austenite to hard austenite at the critical fraction. On the other hand, it was also confirmed that the non-monotonic strengthening behavior has a good correlation with the microstructural connectivity of the hard austenite, that is, the percolation phenomenon. From these result, it was concluded that the strength of metals with duplex microstructure can be wholly predicted based on secant method while taking into account the matrix switching between soft and hard phases, and the percolation index has a possibility to be a useful parameter to distinguish between soft and hard phases which is the matrix of the duplex microstructure, although further investigation is needed on a degree of the hard phase connection.

## 1. Introduction

Metallic materials are often used as duplex microstructures[2] rather than as single phases. Appropriate control of the duplex microstructure provides compatibility between contrary properties. For instance, low-carbon dual-phase (DP) steels with a duplex microstructure composed of soft ferrite and hard martensite have a good strength–ductility balance for use as high-strength steel because hard martensite enhances the strength while the soft ferrite matrix maintains sufficient ductility. The variability in the strength of the DP steels is explained on the basis of stress/strain partitioning between ferrite and martensite [1–5]. Given the fact that the strength of martensite is much higher than that of ferrite, the martensite acts as a strengthening component in DP steels. Therefore, the flow stress of DP steels increases as the fraction and the strength of martensite increases according to the simple mixture rule. On the other hand, it is suggested that the flow stress is strongly affected by the connectivity of martensite grains as well. When the martensite grains surround the ferrite grains and form a chain-like networked structure, the strain hardenability of DP steels is greatly improved in comparison to the case where they are isolated within the ferrite matrix [6–8]. Moreover, the authors recently investigated the development of local stain distribution in a DP steel using a new strain measurement technique, and it was proved that when the martensite distribution was changed from an isolated type to a chain type, the average strain in the martensite grains was markedly increased [9,10]. In other words, the connection between hard martensite grains reduced the degree of strain partitioning between the soft ferrite and hard martensite, leading to near iso-strain condi-

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**Table 1**
Chemical composition of the austenitic steel used in this study (mass%).

| C | Si | Mn | P | S | Cr | Ni | Ti | Fe |
|------|------|------|-------|-------|-------|------|------|------|
| 0.002 | 0.05 | 0.91 | 0.002 | 0.001 | 0.024 | 27.7 | 0.25 | Bal. |

tions in a duplex microstructure. These previous studies suggest that the strength of DP steels should markedly change in response to the connectivity of hard martensite grains. Although there are some kinds of steel with duplex microstructures, such as ferrite–martensite DP steels, ferrite–pearlite steels, and duplex stainless steel, in this study, a unique austenitic steel with a duplex microstructure composed of soft and hard austenite was used as a model alloy material to investigate the effect of microstructural connectivity on the strength of metals with duplex microstructures because the fraction of hard austenite can be optionally and solely controlled by simple heat treatment. The change in the strength of the austenitic steel as a function of the fraction of hard austenite was investigated. The strengthening behavior was then discussed in terms of the connectivity of hard austenite.

## 2. Experimental procedure

The material used in this study was an Fe–28%Ni alloy with the chemical composition shown in Table 1 (mass%). A tiny amount of carbon was fixed as carbide by the addition of Ti to avoid solute carbon partitioning between the fcc and bcc phases. This material had a fully austenitic structure with an average grain size of 84 μm after solid-solution treatment at 1273 K for 1.8 ks. To achieve the duplex microstructure, a sequence of heat treatments stimulating the fcc–bcc martensitic transformation and subsequent bcc–fcc martensitic reversion was carried out, as illustrated in Fig. 1; the solution-treated material was subjected to subzero treatment for 1.8 ks at $T_I$ K below martensitic transformation start temperature, $M_s$, to partially form lenticular martensite within the austenite matrix. The material was then annealed for 60 s at 773 K above martensitic reversion finish temperature, $A_f$, to completely reverse the lenticular martensite to austenite, followed by water cooling. It was confirmed by dilatation tests that $M_s$, martensitic reversion start temperature, $A_s$, and $A_f$ were 203, 623, and 723 K, respectively. Because the forward and reverse martensitic transformations were accompanied by lattice invariant deformation, thus introducing high-density dislocations into the austenite, the reversed austenite was much harder than the untransformed initial austenite, as explained later [11–17]. Therefore, the heat-treated material had an austenitic single phase but it exhibited a duplex microstructure consisting of the untransformed soft austenite and the reversed hard austenite. The reversed austenite was directly transformed from the lenticular martensite, and thus its fraction could be
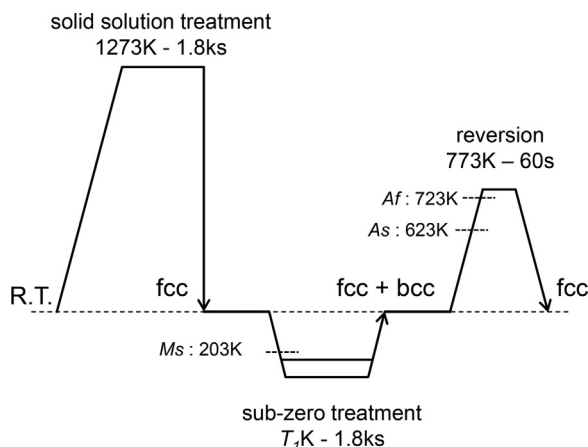
controlled by the subzero temperature $T_I$. The microstructure was observed using an optical microscope and a transmission electron microscope, JEOL-2010F. The average fraction of lenticular martensite after subzero treatment was measured by X-ray diffractometry in order to evaluate the average fraction of reversed austenite. Also, the local fraction of reversed austenite was quantified by measuring the area fraction in optical micrographs. Hardness was evaluated by Vickers hardness, and tensile testing was conducted at an initial strain rate of $1.67 \times 10^{-3}$ s$^{-1}$ using plate test pieces with a gauge size of $6^l \times 3^w \times 1^t$ mm$^3$.

## 3. Results and discussion

### 3.1. Duplex microstructure of Fe–28%Ni austenitic alloy

Fig. 2 shows optical images of the microstructure of the Fe–28%Ni alloy subzero treated at 193, 159, 130, and 77 K. Lenticular martensite with a typical morphology formed in the austenite matrix with a relatively uniform distribution, and its fraction increased with decreasing subzero temperature. The transformation behavior as a function of subzero temperature is displayed in Fig. 3. The martensitic transformation from fcc to bcc started suddenly below $M_s$ with a burst phenomenon, and then the fraction gradually increased with decreasing subzero temperature. As a result, the martensite fraction reached 0.34, 0.50, 0.70, and 0.86 by subzero treatment at 193, 159, 130, and 77 K, respectively. As an example, a TEM image showing the interface between the untransformed austenite and reversed austenite in the material annealed at 773 K for 60 s after the subzero treatment at 77 K is shown in Fig. 4 [17]. The reversed austenite, which seems to have been martensitically reversed from some of the secondary lenticular martensite, appears on the right in this image. Selected area diffraction patterns obtained from both sides of the austenite (insets) look very similar, which demonstrates that the reversed austenite had the same orientation as the untransformed austenite because of crystallographic reversibility. However, a zigzag interface with a low misorientation angle is clearly observed, because the reversed austenite contains higher density of dislocations. Since the untransformed austenite has the same orientation as the reversed austenite, slip systems are continued through the interface. Therefore, it is thought that moving dislocations in untransformed austenite are able to pass through the interface, but interact with high density dislocations in reversed austenite. In response to the difference in dislocation density, the reversed austenite had much higher hardness (190 ± 4 HV) than the untransformed austenite (128 ± 3 HV). Furthermore, it was confirmed that the hardness of each type of austenite changed little, even when the fraction of reversed austenite changed. These results indicate that this austenitic alloy with the unique duplex microstructure can be optionally and solely controlled while retaining the hardness of both austenite phases by simple heat treatment.

### 3.2. Tensile behavior of Fe–28%Ni austenitic alloy with duplex microstructure

Fig. 5 shows the nominal stress–strain curves of the Fe–28%Ni austenitic alloy with average fractions of reversed hard austenite, $F_H$, of 0, 0.34, 0.50, 0.70, and 0.86. It was confirmed that a deformation-induced martensitic transformation did not occur, even at the tensile fracture surface, because of the sufficient mechanical stability of the austenite. The solution-treated material with $F_H$=0 has a stress–strain curve characterized by high strain hardenability and a resulting large uniform elongation. An increase in $F_H$ significantly enhances the yield stress and reduces uniform elongation. Although the reason why a material mostly comprising hard austenite has low uniform elongation is unclear, it is obvious that the yield stress of this kind of austenitic steel is very sensitive to $F_H$. The change in the 0.2% proof stress, $\sigma_{0.2}$, as a function of $F_H$ is summarized in Fig. 6. Additionally, the 0.2%



**Fig. 1.** Heat treatment route to make duplex microstructure in Fe–Ni austenitic alloy.