Effect of Deformation Temperature on Mechanical Properties in 1-GPa-grade TRIP Steels with Different Retained Austenite Morphologies

Noriyuki TSUCHIDA,^{1)*} Takaaki TANAKA²⁾ and Yuki TOJI²⁾

1) Graduate School of Engineering, University of Hyogo, 2167 Shosha, Himeji, 671-2280 Japan.

2) Steel Research Laboratory, JFE Steel Co., 1 Kawasaki-cho, Chuo-ku, Chiba, 260-0835 Japan.

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The effect of deformation temperature on the mechanical properties of 1-GPa-grade TRIP steels with different retained austenite (γ_R) morphologies was studied. The temperature dependence on the deformation-induced martensitic transformation behavior was also investigated. The uniform elongation below room temperature was relatively large in the needle-like γ_R steel, whereas the tensile strength at each temperature was almost the same as it was independent of the γ_R morphology. The better tensile strength–uniform elongation balance was obtained at 373 K for the needle-like γ_R steel and at 473 K for blocky γ_R one. The mechanical stability of γ_R was higher in the needle-like γ_R steel, according to the x-ray diffraction experiments. γ_R was mechanically stable with increasing temperature, but its volume fraction decreased at temperatures above 473 K because of the deformation-induced bainitic transformation. In this paper, the quantitative conditions of deformation-induced transformation to obtain better uniform elongation in the 1-GPa-grade TRIP steels are summarized from the viewpoints of the volume fraction of deformation-induced martensite and the transformation rate.

KEY WORDS: TRIP; TRIP-aided multi-microstructure steel; temperature; deformation-induced martensitic transformation.

1. Introduction

Transformation-induced plasticity (TRIP) effect due to the deformation-induced martensitic transformation (DIMIT) of retained austenite (γ_R) can improve the ductility of steels and has been utilized in advanced high-strength steels (AHSS).^{1–6)} The tensile strength–uniform elongation balance of the third-generation AHSS is approximately 30 000–40 000 MPa%,¹⁾ and various studies on AHSS with a tensile strength of more than 1 GPa have been reported.^{1,2)} We have been investigating the tensile deformation behavior of 1-GPa-grade TRIP steels with different γ_R morphologies.^{7,8)} In our previous studies, the 1-GPa-grade TRIP steels with a needle-like γ_R morphology (needle-like γ_R steel) showed better mechanical properties at a wide range of strain rates, including high-speed deformation behavior.⁷⁾ It was found from the *in situ* neutron diffraction experiments⁸⁾ that not only the DIMIT behavior but also the stress partitioning between the austenite (γ) and the ferrite (α) phase plays an important role in the TRIP effect. The matrix microstructure of annealed martensite in the needle-like γ_R steel was also

found to be an important factor for the mechanical stability of γ_R .⁸⁾ In order to understand the TRIP effect of the 1-GPa-grade TRIP steel in more detail, it is important to drastically change the DIMIT behavior, which is an essential factor influencing the TRIP effect.^{3,5–7)} One of the conditions to make a significant difference in the DIMIT behavior is the deformation temperature.^{6,9–14)} Various data considerably changing the deformation temperature will lead to a deep understanding of the TRIP effect of high-strength TRIP steels.^{5,6)}

Therefore, in this study, the effect of the deformation temperature on the mechanical properties of the 1-GPa-grade TRIP steels with different γ_R morphologies was studied. The effect of deformation temperature on the DIMIT behavior in the 1-GPa-grade TRIP steels was also investigated in order to discuss the differences in their mechanical properties. Furthermore, the conditions to obtain better mechanical properties in high-strength TRIP steels were discussed by comparing with the mechanical properties of other TRIP steels.

* Corresponding author: E-mail: tsuchida@eng.u-hyogo.ac.jp

2. Experimental Procedures

In this study, two types of 1-GPa-grade TRIP steels with different γ_R morphologies obtained by using a 0.3C–1.5Si–2Mn steel were used. One is the needle-like γ_R steel, and the other is the blocky γ_R one.^{7,8)} The details of the 1-GPa-grade TRIP steels have been reported elsewhere.^{7,8)} Microstructures were observed using scanning electron microscope (SEM) with electron backscatter diffraction (EBSD).⁷⁾ Tensile test specimens with a gage length of 25 mm and gage width of 5 mm were machined from the sheets (thickness = 1.4 mm). Static tensile tests were conducted with an initial strain rate of $3.3 \times 10^{-4} \text{ s}^{-1}$ at various test temperatures between 77 K and 573 K by using a gear-driven-type Instron machine.⁷⁾ Here, in order to control the test temperatures, liquid nitrogen (77 K), environmental chamber (123–373 K), and heating furnace (473–573 K) were used.^{12–14)} In order to investigate the effect of temperature on the DIMT behavior, test samples deformed by various true strains (ϵ) were prepared for the x-ray diffraction analysis. The quantitative estimation of γ and deformation-induced martensite (α') by x-ray diffraction was based on the principle that the total integrated intensity of all the diffraction peaks for each phase in a mixture is proportional to the volume fraction of the phase.^{12–14)} In this study, the increase in the volume fraction of α' was considered to be the decrease in the volume fraction of γ from the analyses of the γ and α phases.⁸⁾

On the other hand, the M_s^σ temperature was experimentally determined by the single-specimen temperature-variable tensile test (SS-TV-TT).^{5,15,16)} In the SS-TV-TT, the test specimen was given a pre-strain of approximately 0.5% and was then unloaded. The temperature was lowered by 10 K, and the specimen was reloaded to the new 0.5% strain. This procedure was repeated at temperatures between 323 K and 123 K.

3. Results

3.1. Effect of Temperature on Mechanical Properties of 1-GPa-Grade TRIP Steels

Figure 1 shows the SEM–EBSD inverse pole figure and phase maps of the needle-like γ_R steel and the blocky γ_R one. In the phase maps, green and red indicate γ and α , respectively. As reported in the previous paper,^{7,8)} the volume fraction of γ_R before deformation ($V\gamma_0$) estimated by x-ray diffraction experiments is 24.6% for the needle-like γ_R steel and 22.9% for the blocky γ_R one. The matrix of the needle-like γ_R steel is tempered martensite and partially bainite and that of the blocky γ_R one is ferrite and bainite.^{7,8)} The carbon contents of γ_R in the needle-like and blocky γ_R steels are 1.19 mass% and 1.11 mass%, respectively. **Figure 2** shows the nominal stress–nominal strain curves for the needle-like γ_R steel (a) and the blocky γ_R one (b), obtained by tensile tests at various deformation temperatures. In **Fig. 3**, the nominal stress–nominal strain curves at the early stage of deformation are shown. All of nominal stress–nominal strain curves were continuous yielding type without yield drop. The nominal stress increased with a decrease in the temperature at temperatures below 373 K, and both the TRIP steels were fractured without necking at 77 K. On the other hand, the nominal stress increased with an increase

in the deformation temperature at temperatures above 473 K.^{6,9,10)} **Figure 4** shows the 0.2% offset proof stress (0.2% PS), tensile strength (TS), and uniform elongation (U.El) as functions of the deformation temperature. The 0.2% PS at a given temperature was larger in the needle-like γ_R steel. The 0.2% PS of the blocky γ_R one decreased below approximately 300 K and increased again below 173 K. This seemed to be associated with the transformation strain due to DIMT before the yielding of γ .^{5,12,15,16)} On the other hand, the effect of deformation temperature on TS was almost the same for both the TRIP steels. The TS decreased with increasing temperature but increased at temperatures of above 523 K. The U.El indicated the maximum value at 373 K (needle-like γ_R) and 473 K (blocky γ_R). **Figure 5** shows U.El as a function of TS obtained by tensile tests at various temperatures in the present TRIP steels and other TRIP steels.^{3,6)} The dashed lines in Fig. 5 are contour lines for the products of TS and U.El. **Figure 6** shows the product of TS and U.El as a function of temperature. Here, $V\gamma_0$ in the 0.22C and the 0.1C TRIP steels were 14.4% and 10%, respectively.^{3,6)} In Fig. 5, the TS–U.El balance decreased with an increase in TS. The decrease in the TS–U.El balance with an increase in TS was larger in the present TRIP steels than in the 0.1C and 0.22C TRIP steels.^{3,6)} In the comparisons of the TS–U.El balance between the needle-like γ_R steel and the blocky γ_R one, the TS–U.El balances at TS > 1 GPa were larger in the needle-like γ_R steel. Figures 4 and 6 show that the temperature dependence of the TS–U.El balance is affected by the U.El. A better TS–U.El balance of the TRIP steels was obtained at the maximum U.El.

Figure 7 shows the work-hardening rate ($d\sigma/d\epsilon$) and true stress (σ) as a function of ϵ at various deformation temperatures for the needle-like γ_R steel (a) and the blocky γ_R one (b). The $d\sigma/d\epsilon$ at a given ϵ generally increased with a decrease in temperature.^{11,12)} At temperatures above 373 K, the $d\sigma/d\epsilon$ stopped decreasing and began to increase again.^{12–14)} Such $d\sigma/d\epsilon$ behavior affects the U.El of the TRIP steels.^{7,12,13)} The changes in $d\sigma/d\epsilon$ were undulated at temperatures above 373 K. Because the σ – ϵ curves are not

Fig. 1. Orientation color map and EBSD phase mapping image in the 1-GPa-grade TRIP steels with different γ_R morphologies of needle-like ((a), (b)) and blocky ((c), (d)). (Online version in color.)

Fig. 2. Nominal stress–nominal strain curves of the 1-GPa-grade TRIP steels with different γ_R morphologies of needle-like (a) and blocky (b) obtained by tensile tests at various temperatures between 77 K and 573 K. (Online version in color.)

Fig. 3. Nominal stress–nominal strain curves at the early stage of deformation in Fig. 2 to show the temperature dependence of 0.2% offset proof stress of the 1-GPa-grade TRIP steels with different γ_R morphologies of needle-like (a) and blocky (b). (Online version in color.)

Fig. 4. 0.2% offset proof stress, tensile strength and uniform elongation as functions of temperature in the 1-GPa-grade TRIP steels with different γ_R morphologies. (Online version in color.)

Fig. 5. Uniform elongation as a function of tensile strength in the 1-GPa-grade TRIP steels and various steels (0.1C–5Mn³) and 0.22C–1.64Mn–1.51Al–0.05Si⁶) obtained by tensile tests at various temperatures. (Online version in color.)

Fig. 6. Product of tensile strength and uniform elongation as a function of temperature in the 1-GPa-grade TRIP steels and various steels.^{3,6)} (Online version in color.)

the so-called saw-toothed waves,^{9–11)} these $d\sigma/d\varepsilon$ behaviors are different from dynamic strain aging, judging from other studies on TRIP steels.

3.2. Effect of Deformation Temperature on Deformation-Induced Martensitic Transformation Behavior

Figure 8 shows the volume fraction of γ_R (V_γ) as a function of ε at various deformation temperatures for the needle-like γ_R steel (a) and the blocky γ_R one (b). The lines in Fig. 8 are the calculated results using the following equation formulated by Matsumura *et al.*:¹⁷⁾

$$V_\gamma = \frac{V_{\gamma 0}}{1 + (k/q)V_{\gamma 0}\varepsilon^q} \quad \dots\dots\dots (1)$$

where k is a constant relating to mechanical stability of γ_R , and q is the strain exponent concerning autocatalytic effect.¹⁷⁾ The values of k and q are summarized in Table 1. In the blocky γ_R steel, γ_R was transformed to α' before the tensile tests at 77 K and 173 K. The V_γ values at 77 K and 173 K were 14.8% and 17.7%, respectively. In terms of the

Fig. 7. True stress and work-hardening rate as functions of true strain at the temperatures between 77 K and 573 K in the needle-like γ_R steel (a) and the blocky γ_R one (b). (Online version in color.)

Fig. 8. Volume fraction of retained austenite as a function of true strain at the temperatures between 77 K and 573 K in the needle-like γ_R steel (a) and the blocky γ_R one (b). (Online version in color.)

Table 1. Values of k and q in Eq. (1) in the 1-GPa-grade TRIP steels with different γ_R shapes at various deformation temperatures.

Temperature (K)	(a) Needle-like γ_R		(b) Blocky γ_R	
	k	q	k	q
77	5 609.1	1.764	2 361.5	1.226
173	140.2	0.995	128.0	0.876
243	207.0	1.425	199.1	1.298
296	54.6	1.264	33.0	1.037
373	78.8	1.871	60.6	1.764
473	14.8	1.228	190.1	2.425
573	59.3	1.251	42.2	1.417

Fig. 9. Volume fraction of deformation-induced martensite as a function of true strain at 243, 296, 373 and 473 K in the needle-like γ_R steel and the blocky γ_R one. (Online version in color.)

DIMT behavior at various temperatures, $V\gamma'$ at a given ϵ usually decreased with a decrease in temperature. However, $V\gamma'$ at a given ϵ decreased when the temperature increased from 473 K to 573 K in both the TRIP steels. This is different from the common temperature dependence on DIMT behavior.^{6,9,10} Similar deformation-induced transformation behaviors were also reported by Sugimoto *et al.*^{9–11} Sugimoto *et al.* clarified that the mechanical stability of γ_R above 523 K in the 0.4C steel was mainly controlled by the deformation-induced bainitic transformation, as revealed by TEM observations.⁹ During this time, TS and elongation increased with an increase in deformation temperature and indicated the maximum value near the austempering temperature.^{9–11} The increase in TS at temperatures above 523 K, as shown in Fig. 4, also seemed to be associated with the deformation-induced bainitic transformation.

Figure 9 shows the volume fraction of α' ($V\alpha'$) as a function of ϵ at 243, 296, 373, and 473 K in order to compare the DIMT behavior between the needle-like γ_R steel and the blocky γ_R one. At 243, 296, and 373 K, $V\alpha'$ was larger in the needle-like γ_R steel at $\epsilon > 0.15$. At 473 K, $V\alpha'$ at a given ϵ was smaller at $\epsilon < 0.25$ but became larger at $\epsilon >$

0.25 in the blocky γ_R steel. Our previous studies at room temperature^{7,8)} indicated that the mechanical stability of γ_R was higher in the needle-like γ_R steel. Thus, the mechanical stability of γ_R was higher in the needle-like γ_R steel at temperatures below 373 K. As seen in Table 1, both of k and q in Eq. (1) were larger in the needle-like γ_R steel at temperatures below 373 K.

4. Discussions

4.1. 0.2% Offset Proof Stress

The 0.2% PS of the blocky γ_R steel decreased below 296 K and increased again at 173 K, as shown in Fig. 4. Because σ usually increases with a decrease in temperature, the temperature dependence of 0.2% PS in the blocky γ_R steel seems to be related to the M_s^σ point.^{5,12,15,16} **Figure 10** shows the nominal stress–nominal strain curves obtained by using the SS-TV-TT technique for the needle-like γ_R steel (a) and the blocky γ_R one (b). In the blocky γ_R steel (Fig. 10 (b)), the yield point was observed at 303 K, and the lower yield stress was almost the same between 303 and 293 K. The yield point was also observed below 283 K, and the lower yield stress increased a little with a decrease in the temperature. At temperatures below 183 K, the stress–strain curves became the continuous yielding type. In the SS-TV-TT technique, the M_s^σ temperature is designated in the case that the flow stress immediate after yielding decreases irrespective of the decreasing temperature or the yield point appears.^{5,15,16} Thus, the M_s^σ of the blocky γ_R steel can be determined as 303 K from Fig. 10(b), which is coincident with the temperature dependence of 0.2% PS shown in Fig. 4. The M_s^σ of the needle-like γ_R steel was determined as 273 K from Fig. 10(a), which was lower than that of the blocky γ_R one. This also means that the mechanical stability of γ_R is higher in the needle-like γ_R steel.

We also investigated the temperature dependence on elastic limit because the 0.2% PS contains an influence of work hardening after the plastic deformation on stress. **Figure 11** shows elastic limit as a function of temperature in the needle-like γ_R and blocky γ_R steels. The temperature at which the elastic limit was decreased was almost coincident with the M_s^σ temperature of each TRIP steel. As described before, in the needle-like γ_R steel, γ_R was not transformed to α' before the tensile tests in the present deformation temperature range. Therefore, the decrease in elastic limit of the needle-like γ_R steel near the M_s^σ temperature is mainly associated with the DIMT before yielding of γ_R . On the other hand, the amount of decrease in the elastic limit near M_s^σ was larger in the blocky γ_R steel. This seems to be correlated with the matrix microstructure. It is difficult to understand the decrease in the elastic limit from only the DIMT because the increase in $V\alpha'$ during elastic deformation is a little. Considering that the elastic limit is affected largely by the deformation of matrix microstructure with the volume fraction of about 80%, the difference of strength (or hardness) of matrix microstructure is associated with the amount of decrease in the elastic limit in Fig. 11. From Figs. 10 and 11, we think that the inverse temperature dependency of 0.2% PS is associated with the DIMT before yielding of γ_R . The difference of temperature dependence on 0.2% PS between the needle-like γ_R steel and the blocky γ_R one

Fig. 10. Nominal stress–nominal plastic strain curves of the needle-like γ_R steel (a) and the blocky γ_R one (b) obtained by the SS-TV-TT technique. (Online version in color.)

Fig. 11. Elastic limit as a function of temperature in the needle-like γ_R and the blocky γ_R steels. (Online version in color.)

is correlated with the mechanical stability of γ_R from their M_s^σ temperatures and the strength of matrix microstructure.

4.2. Tensile Strength

Figure 12 shows the TS (a) and the true stress (σ) at the maximum load point (b) as functions of $V\alpha'$ at the maximum load. In Fig. 12, the experimental results of the present TRIP steels obtained by using tensile tests at various strain rates⁷⁾ and of the other TRIP steels^{6,18,19)} are also shown. The TS in the TRIP steels was roughly dependent on $V\alpha'$. Harjo *et al.*²⁰⁾ reported that the σ of α' in the 0.2C TRIP steel estimated by the neutron diffraction experiments was approximately 2–2.5 GPa. In our previous study,⁸⁾ the phase strains of α' in the present TRIP steels, which was associated with the σ of α' , were almost the same as those in the 0.2C TRIP steel²⁰⁾ and were almost independent of the γ_R morphologies. These results also indicate that the σ of α' was considerably larger than those of other microstructures in TRIP steels.^{8,20,21)} Because $V\alpha'$ at the maximum load at

a given temperature was also almost the same, as it was independent of the γ_R morphology, the TS of the present TRIP steels could be roughly determined using $V\alpha'$.^{5,6,22)} As shown in Fig. 12(b), the σ at the maximum load instead of TS might be a better parameter to consider because the σ at the maximum load is different when U.El is different even though TS is the same. Thus, the relationship between $V\alpha'$ and the σ at the maximum load is reasonable.

4.3. Uniform Elongation

As shown in Fig. 4, the temperature at the maximum U.El ($T_{\max}$) was different between the needle-like γ_R steel and the blocky γ_R one. With respect to the temperature dependence on U.El in TRIP steels, the $T_{\max}$ value depended on the mechanical stability of γ_R .^{6,10,12–14)} The higher the mechanical stability of γ_R was, the lower was the $T_{\max}$ value. The carbon content of γ_R , which was closely related to the mechanical stability of γ_R , was larger in the needle-like γ_R steel.^{7,8)} Our previous study⁸⁾ reported that the stress partitioning between γ and α also affected the mechanical stability of γ_R including the DIMT behavior. In the needle-like γ_R steel, the stress partitioning was small at the early stage of deformation and increased with an increase in ϵ .⁸⁾ In particular, the stress partitioning at the early stage of deformation was closely related to the σ value at which the DIMT started (σ_{D-s}).⁸⁾ The smaller the stress partitioning was, the larger was the σ_{D-s} value. As shown in Fig. 9, $V\alpha'$ at the early stage of deformation was smaller and the values at $\epsilon > \text{approximately } 0.15$ became larger in the needle-like γ_R steel at temperatures below 373 K. As described above, $T_{\max}$ was lower in the needle-like γ_R steel with the higher mechanical stability of γ_R .

U.El also affected the TS–U.El balance, and a better TS–U.El balance was obtained at the maximum U.El. As shown in Fig. 5, the better TS–U.El balances of the needle-like γ_R steel at the TS > 1 GPa were associated with the temperature dependence on U.El. Considering the improvements of the mechanical properties of the 1-GPa-grade TRIP steels at room temperature, a higher mechanical stability of γ_R is required. In contrast, in Fig. 6, the TS–U.El balance at 296 K was almost the same between the blocky γ_R steel and

Fig. 12. Tensile strength (a) and true stress at the maximum load (b) as functions of the volume fraction of deformation-induced martensite (α') at the maximum load in the 1-GPa-grade TRIP steels and other TRIP steels.^{6,18,19)} (Online version in color.)

Fig. 13. Transformation rate calculated by Eq. (1) as a function of true strain at various temperatures in the needle-like γ_R steel (a) and the blocky γ_R one (b). (Online version in color.)

the 0.22C steel⁶⁾ despite an approximately 10% difference in $V\gamma_0$. This implied that a better TS–U.EI balance could be obtained by the control of DMT in spite of a smaller $V\gamma_0$. As a way to obtain the higher mechanical stability of γ_R other than the chemical compositions, a relatively small stress partitioning between γ and α is effective.⁸⁾

Next, the conditions of the DMT behavior at $T_{\max}$ in the two TRIP steels, *i.e.*, 373 K for the needle-like γ_R steel and 473 K for the blocky γ_R one, are discussed. As shown in Fig. 9, the γ_R was mechanically stable in the early stage of deformation and $V\alpha'$ increased gradually in the latter part of the deformation at $T_{\max}$. This was coincident with the qualitative condition of DMT to obtain better U.EI because of the TRIP effect, as reported in the previous studies.^{12,22,23)} We now summarize the quantitative conditions for the deformation-induced transformation behavior in the present TRIP steels. Here, $V\alpha'$ and the transformation rate are focused upon as the quantitative indexes.^{12,13,23)} Figures 8 and 9 show that at 373 K for the needle-like γ_R steel and at 473 K for the blocky γ_R one, $V\alpha'$ at ϵ of 0.1 was less than 5% and approximately 50% of γ_R was transformed to

α' at ϵ of 0.3. In the 0.22C TRIP steel,⁶⁾ as shown in Fig. 6, the maximum U.EI of approximately 40% was obtained at 333 K. From the DMT behavior at various temperatures,⁶⁾ the two conditions that $V\alpha'$ at ϵ of 0.1 is less than 5% and approximately 50% of γ_R is transformed to α' at ϵ of 0.3 were applied in the case of the 0.22C steel. We also verified the two conditions for the 0.4C TRIP steel reported by Mukherjee *et al.*¹⁰⁾ As a result, the DMT behavior at the better TS–U.EI balance was almost in agreement with the two conditions. **Figure 13** shows the transformation rate ($dV\alpha'/d\epsilon$) estimated by using Eq. (1)^{12,13)} as a function of ϵ in the case of the needle-like γ_R steel (a) and the blocky γ_R one (b). The estimated $dV\alpha'/d\epsilon$ was shown up to the U.EI at each temperature, and the arrows in Fig. 13 show the maximum $dV\alpha'/d\epsilon$ at $T_{\max}$ in each TRIP steel. The maximum $dV\alpha'/d\epsilon$ value at $\epsilon < 0.1$ was observed at most temperatures. At $T_{\max}$, $dV\alpha'/d\epsilon$ gradually increased with an increase in ϵ and indicated the maximum value at ϵ from 0.15 to 0.2, as denoted by the arrows in Fig. 13. The maximum $dV\alpha'/d\epsilon$ value was approximately 0.5, as this parameter was independent of the γ_R morphology. From the above discussions,

in the present 1-GPa-grade TRIP steels, the conditions of DIMT to obtain better U.El have something in common, and the following three quantitative conditions are summarized:

- (i) $V\alpha'$ at ε of 0.1 is less than 5%.
- (ii) Approximately 50% of γ_R is transformed to α' at ε of 0.3.
- (iii) The maximum $dV\alpha'/d\varepsilon$ value of approximately 0.5 is indicated at larger ε near uniform elongation.

5. Summary

In this study, the effect of deformation temperature on the mechanical properties of 1-GPa-grade TRIP steels with different γ_R morphologies was clarified by tensile tests at temperatures between 77 K and 523 K. The temperature dependence of the deformation-induced martensitic transformation behavior was also investigated. The main conclusions are as follows:

(1) The mechanical properties at a given temperature were greater in the needle-like γ_R steel at temperatures below 373 K. The U.El below room temperature was larger in the needle-like γ_R steel, whereas the TS at each temperature was almost the same, as it was independent of the γ_R morphology.

(2) The mechanical stability of γ_R was higher in the needle-like γ_R steel, according to the x-ray diffraction experiments. However, at temperatures more than 473 K, the $V\alpha'$ at a given ε was larger in the needle-like γ_R steel. γ_R was usually mechanically stable with increasing temperature, but its volume fraction decreased at temperatures above 473 K. It seemed to be associated with the deformation-induced bainitic transformation.

(3) From the present study, the conditions of DIMT in order to obtain better U.El in the present 1-GPa-grade TRIP steels are as follows: The maximum transformation rate of approximately 0.5 is observed at larger ε and $V\alpha'$ at ε of 0.1 is less than 5%. Furthermore, approximately 50% of γ_R is transformed to α' at ε of 0.3. These three conditions need to be studied further using TRIP steels with a wide range of $V\gamma_0$.

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REFERENCES

- 1) W. Bleck, X. Guo and Y. Ma: *Steel Res. Int.*, **88** (2017), 1700218. <https://doi.org/10.1002/srin.201700218>
- 2) K. Sugimoto, J. Kobayashi and T. Hojo: *Tetsu-to-Hagané*, **103** (2017), 1 (in Japanese).
- 3) J. Hu, W. Cao, C. Wang, H. Dong and J. Li: *ISIJ Int.*, **54** (2014), 1952.
- 4) M. K. Hatami, T. Pardoen, G. Lacroix, P. Berke, P. J. Jacques and T. J. Massart: *J. Mech. Phys. Solids*, **98** (2017), 201.
- 5) B. C. De Cooman: *Curr. Opin. Solid State Mater. Sci.*, **8** (2004), 285.
- 6) J. A. Jimenez, M. Carsi, O. A. Ruano and G. Frommeyer: *Mater. Sci. Eng. A*, **508** (2009), 195.
- 7) N. Tsuchida, S. Okura, T. Tanaka and Y. Toji: *ISIJ Int.*, **58** (2018), 978. <https://doi.org/10.2355/isijinternational.ISIJINT-2017-635>
- 8) N. Tsuchida, T. Tanaka and Y. Toji: *ISIJ Int.*, **60** (2020), 1349. <https://doi.org/10.2355/isijinternational.ISIJINT-2019-751>
- 9) K. Sugimoto, M. Kobayashi and S. Hashimoto: *Metall. Trans. A*, **23** (1992), 3085.
- 10) M. Mukherjee, O. N. Mohanty, S. Hashimoto, T. Hojo and K. Sugimoto: *ISIJ Int.*, **46** (2006), 316.
- 11) K. Sugimoto, M. Tsunazawa, T. Hojo and S. Ikeda: *ISIJ Int.*, **44** (2004), 1608.
- 12) N. Tsuchida, Y. Morimoto, T. Tonan, Y. Shibata, K. Fukaura and R. Ueji: *ISIJ Int.*, **51** (2011), 124.
- 13) N. Tsuchida, Y. Yamaguchi, Y. Morimoto, T. Tonan, Y. Takagi and R. Ueji: *ISIJ Int.*, **53** (2013), 1881.
- 14) N. Tsuchida, T. Kawahata, E. Ishimaru and A. Takahashi: *ISIJ Int.*, **54** (2014), 1971.
- 15) G. F. Bolling and R. H. Richman: *Acta Metall.*, **18** (1970), 673.
- 16) A. N. Vasilakos, K. Papamantellos, G. N. Haidemenopoulos and W. Bleck: *Steel Res.*, **70** (1999), 466.
- 17) O. Matsumura, Y. Sakuma and H. Takechi: *Scr. Metall.*, **21** (1987), 1301.
- 18) N. Tsuchida and K. Osaki: *Tetsu-to-Hagané*, **99** (2013), 524 (in Japanese).
- 19) N. Tsuchida, T. Araki, Y. Yamaguchi and K. Fukaura: *Tetsu-to-Hagané*, **98** (2012), 558 (in Japanese).
- 20) S. Harjo, N. Tsuchida, J. Abe and W. Gong: *Sci. Rep.*, **7** (2017), Article No. 15149. <https://doi.org/10.1038/s41598-017-15252-5>
- 21) Y. Tomota, H. Tokuda, Y. Adachi, M. Wakita, N. Minakawa, A. Moriai and Y. Morii: *Acta Mater.*, **52** (2004), 5737.
- 22) P. J. Jacques, E. Girault, P. Harlet and F. Delannay: *ISIJ Int.*, **41** (2001), 1061.
- 23) I. Tamura: *Tetsu-to-Hagané*, **56** (1970), 429 (in Japanese).