Effects of Heat Treatment and Alloying Elements on the Microstructures and Mechanical Properties of 0.15 Wt Pct C Transformation-Induced Plasticity–Aided Cold-Rolled Steel Sheets

SUNG-JOON KIM, CHANG GIL LEE, ILDONG CHOI, and SUNGHAK LEE

The main emphasis of this study has been placed on understanding the effects of manganese and silicon additions and of heat-treatment (intercritical annealing and isothermal treatment) conditions on the microstructures and mechanical properties of 0.15 wt pct C transformation-induced plasticity (TRIP)–aided cold-rolled steel sheets. The steel sheets were intercritically annealed and isothermally treated at the bainitic region. Microstructural observation and tensile tests were conducted, and volume fractions of retained austenite were measured. Steels having a high manganese content had higher retained austenite fractions than the steels having a low manganese content, but showed characteristics of a dual-phase steel such as continuous yielding behavior, high tensile strength over 1000 MPa, and a low elongation of about 20 pct. The retained austenite fractions and mechanical properties varied with the heat-treatment conditions. In particular, the retained austenite fractions increased with decreasing intercritical annealing and isothermal treatment temperatures, thereby resulting in the improvement of the elongation and strength-ductility balance without a serious decrease in the yield or tensile strength. These findings suggested that the intercritical annealing and isothermal treatment conditions should be established in consideration of the stability of austenite and the solubility of alloying elements in the austenite formed during the intercritical annealing.

I. INTRODUCTION

THE demand for high-strength steel sheets having excellent ductility has been increasing in various industries and in the automotive industry in particular, in order to improve productivity and safety levels and to reduce auto body weight. Dual-phase steels having a high tensile strength above 900 MPa have been widely used as high-strength steel sheets. However, their elongation is below 20 pct because of the high volume fraction of hard martensite, and, thus, they are inappropriate for undergoing various deformation processes. Recently, high-strength transformation-induced plasticity (TRIP)–aided cold-rolled steel sheets have been receiving increasing attention, as the steel sheets have both high strength and ductility due to the martensitic transformation of retained austenite during plastic deformation.[1–5] Previous studies on these steel sheets have mainly focused on cases of a 0.2 to 0.4 wt pct carbon content, because higher volume fractions of retained austenite can be easily obtained at those carbon contents.[6,7,8] However, weldability, which is an important property required for many applications and especially for productivity during cold rolling, can deteriorate due to the higher carbon content. To overcome these shortcomings, more studies are required to lower the carbon content.

In the meantime, when the carbon content is lowered so as to improve the weldability and rolling capability, it becomes harder to obtain a large amount of stabilized retained austenite. In order to achieve the elongation improvement by an appropriate level of stabilized retained austenite in low-carbon C-Si-Mn TRIP steels, it is necessary to add alloying elements such as Si and Mn in optimum ratios and to establish heat-treatment conditions under which the effect of alloying elements can be utilized to the maximum.

The present study aims to establish optimum heat-treatment conditions in order to achieve the elongation improvement by retained austenite in low-carbon TRIP-aided cold-rolled steel sheets, in which the carbon content is lowered to 0.15 wt pct in order to enhance the weldability and rolling capability. To these purposes, the effects of heat-treatment conditions on mechanical properties, mircrostructures, and retained austenite volume fractions were investigated by varying the intercritical annealing and the subsequent isothermal bainite treatment conditions.

II. EXPERIMENTAL

Three kinds of steel ingots, termed ECO-1, ECO-2, and ECO-3, were fabricated by vacuum induction melting and aluminum killing. The carbon content was fixed at 0.15 wt pct, and the silicon and manganese contents were varied from 1 to 1.5 wt pct and 1.5 to 2.5 wt pct, respectively. Table I lists their chemical compositions, together with the A_{C1}, A_{C3}, and martensite start (M_s) temperatures measured using a dilatometer. The A_{C1} and A_{C3} temperatures were measured at a heating rate of 20 °C/min up to 1000 °C, and the M_s temperature was measured, after maintaining the sample for 5 minutes at 1000 °C, at a cooling rate of 50 °C/s

SUNG-JOON KIM, Principal Researcher, and CHANG GIL LEE, Senior Researcher, are with the Materials Processing Department, Korea Institute of Machinery and Materials, Changwon 641-010, Korea. ILDONG CHOI, Associate Professor, is with the Department of Materials Engineering, Korea Maritime University, Busan, 606-791, Korea. SUNGHAK LEE, Professor, is with the Center for Advanced Aerospace Materials, Pohang University of Science and Technology, Pohang, 790-784, Korea.

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Table I. Chemical Compositions (Weight Percent) and A_{C1}, A_{C3}, and M_s Temperatures (°C) of the Steels Used in This Study

Steel	C	Si	Mn	P	S	Al	A_{C1}	A_{C3}	M_s
ECO-1	0.16	1.42	1.47	0.0016	0.0036	0.046	728	922	425
ECO-2	0.16	1.41	2.52	0.0017	0.0039	0.055	716	891	409
ECO-3	0.15	1.00	2.55	0.0016	0.0039	0.047	709	867	400

using cold-rolled specimens. Steel ingots were rough rolled to slabs of 25 mm in thickness. The slabs were homogenized at 1250 °C for 2 hours, hot-rolled to sheets of 3 mm in thickness, and then compulsorily air cooled. The finish rolling temperature was 900 °C. The hot-rolled steel sheets were pickled by a 10 pct HCl solution at 80 °C and were cold rolled to sheets of 1 mm in thickness.

The conditions of intercritical annealing and isothermal bainitic transformation significantly affect the formation of retained austenite and the subsequent mechanical properties in TRIP-aided cold-rolled steels. According to recent research articles,[9,10,11] when intercritically annealed at an $(A_{C1} + A_{C3})/2$ temperature, the highest fraction of retained austenite and the most excellent mechanical properties were achieved, and the volume-fraction ratio of ferrite to austenite was about 40:60. Other research, however, has shown that, when intercritically annealed at a temperature at which the volume-fraction ratio of ferrite to austenite is 50:50, excellent TRIP effects can also be obtained.[12,13] In this study, the intercritical annealing was conducted for 5 minutes at temperatures at which the fraction ratios of ferrite to austenite were 40:60 and 50:50. The isothermal treatment was carried out in the temperature range of M_s to (M_s + 50 °C) for 1 to 20 minutes, followed by air cooling. Two separate salt baths were used for both the intercritical annealing and isothermal treatment.

Tensile specimens (longitudinal direction) with a gage length of 25.4 mm and a width of 6.3 mm were prepared from the cold-rolled steel sheets. The specimens were tensioned at room temperature at a crosshead speed of 2 mm/min using a tensile tester, and the yield strength, tensile strength, and elongation were measured. A lower yield point was adopted as the yield strength in the case of the yield-point phenomenon, whereas a 0.2 pct offset value was adopted in the case of the continuous yielding.

Since C-Mn-Si TRIP-aided steels have very complex microstructures composed of ferrite, bainite (or martensite), and retained austenite, identification of each phase is unclear when the sample has been etched in a nital solution. Thus, a sodium metabisulfite solution (10 g $Na_2S_2O_3 \cdot H_2O$ + 100 mL H_2O) was used in conjunction with the nital etching. When the specimens etched by the sodium metabisulfite solution are observed by an optical microscope, it is easier to define the phases, since ferrite is displayed as gray, bainite or martensite as black, and retained austenite as white.

The volume fraction of retained austenite was measured using an X-ray diffractometer. Specimens were prepared by both mechanically and chemically polishing the sheets to half of their thickness. A Mo K_α characteristic X-ray was used, and the volume fraction of retained austenite (V_γ) was calculated from the integrated intensity of ferrite and austenite peaks using the equation $V_\gamma = 1.4I_\gamma/(I_\alpha + 1.4I_\gamma)$.[14,15] Here, I_γ is the average integrated intensity

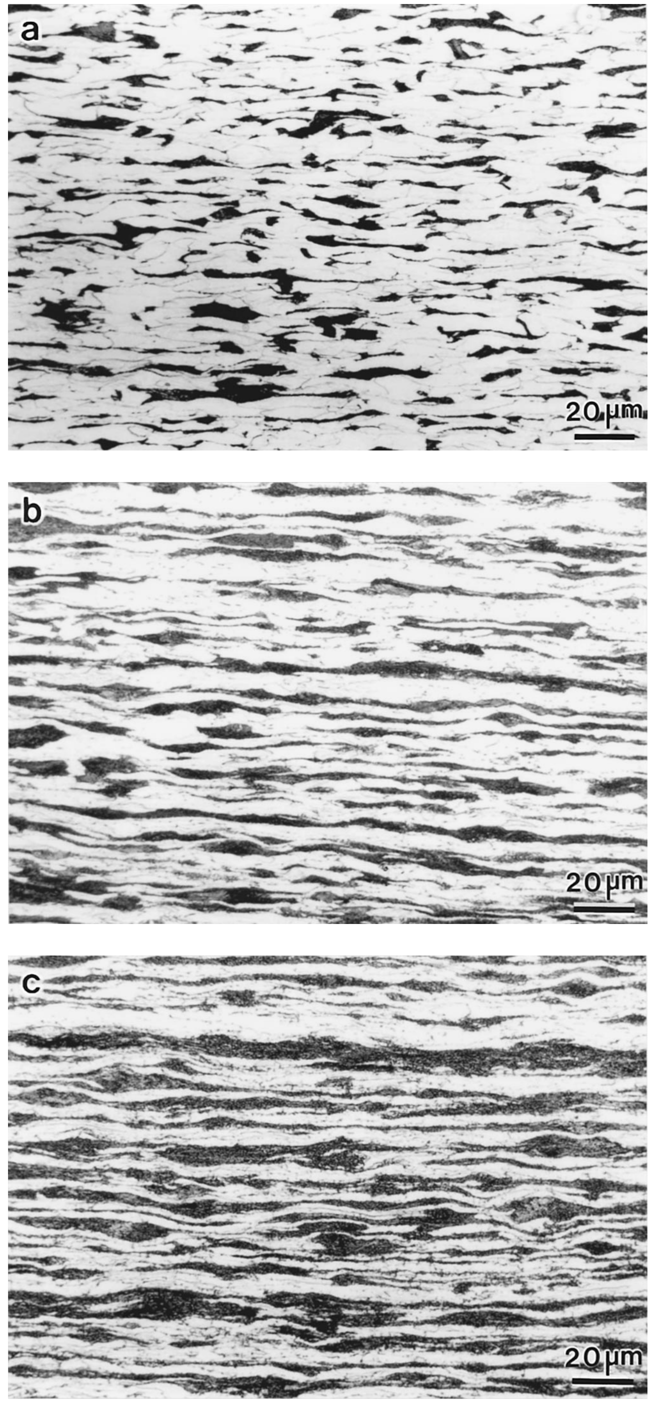


Fig. 1—Optical micrographs of the cold-rolled (*a*) ECO-1, (*b*) ECO-2, and (*c*) ECO-3 steels. Nital etched.

obtained from $\{220\}_\gamma$ and $\{311\}_\gamma$ peaks, and I_α is that obtained from the $\{211\}_\alpha$ peak.

III. RESULTS

A. *Microstructure of Cold-Rolled Steel Sheets*

Figures 1(a) through (c) are optical micrographs of the cold-rolled ECO-1, ECO-2, and ECO-3 steels. All of these three steels have a mixed structure of ferrite and pearlite elongated along the rolling direction, showing a band structure. Particularly as the manganese content increases or as the silicon content decreases, the formation of the band structure is accelerated; thus, the ECO-3 steel, with a high manganese content, is completely composed of the band structure (Figure 1(c)). It is generally known that the development of the band structure becomes stronger with higher manganese contents. This is because the areas where manganese and carbon are simultaneously segregated during casting are not completely homogenized, due to the reduced diffusion rate of carbon by the presence of manganese as well as the slow diffusion of manganese during homogenization;[16] thus, they are elongated in a band-type manner wherein pearlite grains are intensively formed. Since this band structure works as a factor to deteriorate ductility, it can adversely affect the elongation enhancement when pearlite grains are not completely decomposed during intercritical annealing.

B. *Microstructural Variation* vs *Alloying Elements and Heat Treatment*

Figures 2(a) through (c) are optical micrographs of the ECO-1, ECO-2, and ECO-3 steels, which were intercritically annealed and then isothermally treated. They were intercritically annealed for 5 minutes at temperatures where the volume fractions of ferrite and austenite were 40 and 60 pct, respectively (*i.e.*, 840 °C for the ECO-1 steel, 770 °C for the ECO-2 steel, and 740 °C for the ECO-3 steel). After the intercritical annealing, they were subjected to the isothermal treatment for 3 minutes at M_s + 20 °C (*i.e.*, 450 °C, 430 °C, and 420 °C for the ECO-1, ECO-2, ECO-3 steels, respectively). The ECO-1 steel shows a homogenous microstructure overall, because of a relatively even distribution of secondary phases including ferrite and bainite (Figure 2(a)). In the ECO-2 and ECO-3 steels, the band structure still exists, and it is more visible in the ECO-3 steel than in the ECO-2 steel (Figures 2(b) and (c)). This indicates that bainite and retained austenite are mainly formed in this band region. Because manganese is segregated in the band structure, a considerable amount of austenite can transform to martensite during the cooling following the isothermal treatment, due to the increased hardenability of austenite.[17,18] Such a microstructure is more brittle than the one wherein hard phases are evenly distributed overall, and, thus, the formation of the band structure does not help to enhance the elongation. Consequently, the manganese content should be controlled at a lower amount in order to restrict the formation of the band structure.

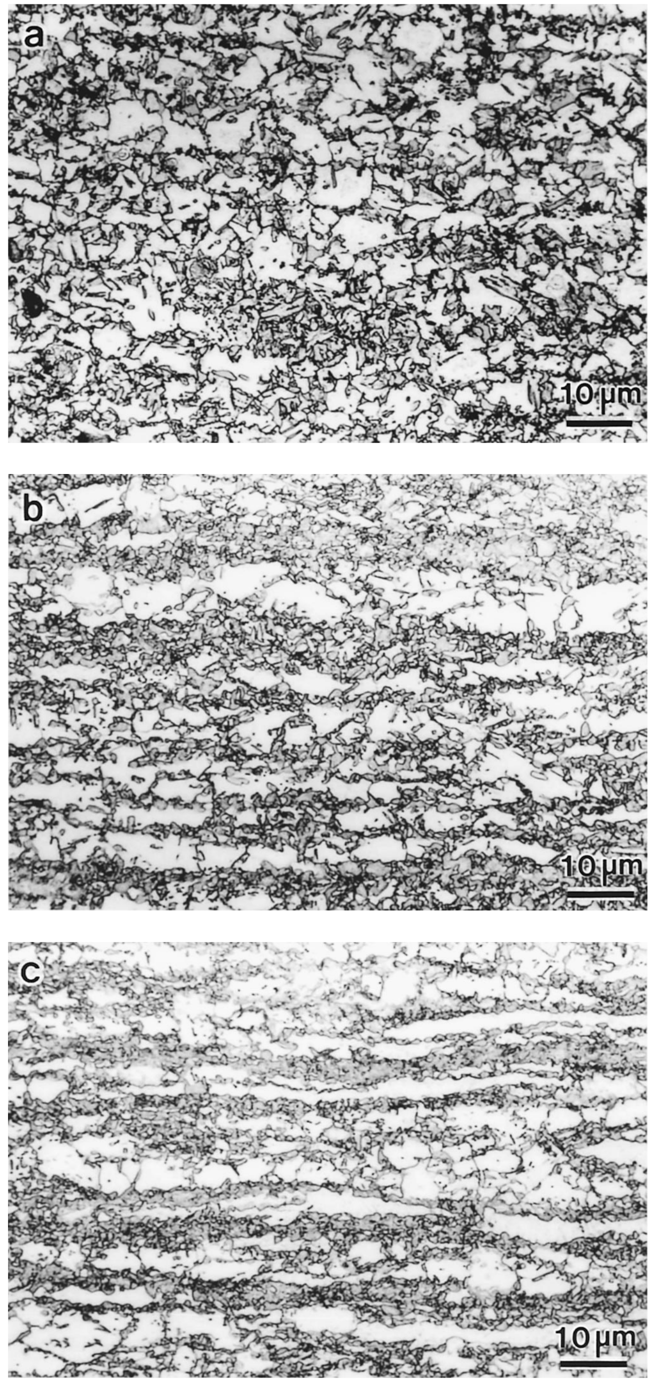


Fig. 2—Optical micrographs of the (*a*) ECO-1, (*b*) ECO-2, and (*c*) ECO-3 steels intercritically annealed at 840 °C, 770 °C and 740 °C for 5 min and isothermally treated at 450 °C, 430 °C, and 400 °C for 3 min, respectively. Nital etched.

Figure 3 shows optical micrographs of the ECO-1 steel intercritically annealed for 5 minutes at 840 °C, at which the fraction ratio of ferrite to austenite is about 40:60, and then isothermally treated at 465 °C, 450 °C, and 430 °C for 1 and 10 minutes, respectively. When isothermally treated at 465 °C, the case of the 10-minute treatment shows a slightly higher ferrite fraction than the case of the 1-minute treatment. This tendency is also observed in the cases of isothermal treatment at both 450 °C and 430 °C, but is reduced as the isothermal treatment temperature decreases. The fraction of secondary phases including bainite increases with decreasing isothermal treatment temperatures. Such a microstructural variation *vs* the isothermal treatment temperature and time is also observed in the ECO-2 and ECO-3 steels. This variation tendency is thought to arise (1) because the bainitic formation is delayed to a later stage of the isothermal treatment as the ferritic transformation proceeds at the initial stage in the higher isothermal temperature and (2) because bainite is more actively formed as the incubation time up to the start of the bainitic transformation is shortened by decreasing the isothermal treatment temperature. It is, thus, found that the isothermal treatment temperature should be lowered to promote a more active bainitic transformation

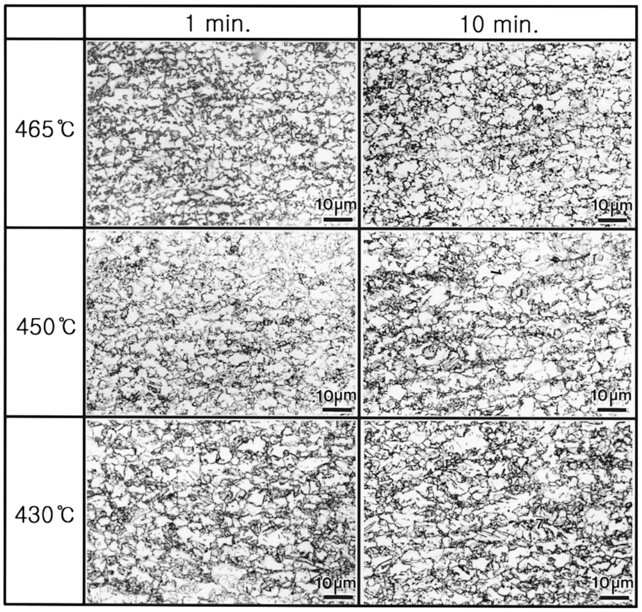


Fig. 3—Optical micrographs of the ECO-1 steel intercritically annealed at 840 °C and isothermally treated at 430 °C to 465 °C for 1 and 10 min. Nital etched.

and, subsequently, a higher fraction of retrained austenite during the isothermal treatment.

Optical micrographs of the ECO-1 steel, etched by a sodium metabisulfite solution after it was intercritically annealed at 840 °C and 800 °C and then isothermally treated for 5 minutes at 430 °C, are shown in Figures 4(a) and (b). As mentioned in Figure 3, 840 °C is the temperature at which the fraction ratio of ferrite to austenite is 40:60, whereas 800 °C is the one where it is 50:50. The areas displayed as light or dark gray are ferrites, the black ones are bainites or martensites, and the white ones are retained austenites. Both specimens are observed to have homogeneously distributed retained austenites, most of which are coexisting in association with nearby ferrites, bainites, or martensites, whereas some are isolated in ferrite grains. The fraction of retained austenite in the specimen intercritically annealed at 840 °C is about 10 pct (Figure 4(a)), the same as in the specimen intercritically annealed at 800 °C (Figure 4(b)). The particle-shaped retained austenite is fine in general, and a considerable amount of film-shaped retained austenite is observed to locate at bainite or martensite lath boundaries. It is also confirmed that such a microstructural variation tendency *vs* the intercritical annealing temperature occurs in the ECO-2 and ECO-3 steels.

As observed in Figures 3 and 4, the microstructures and the fraction and shape of retained austenite are varied, depending on the temperature and time of intercritical annealing and isothermal treatment. Besides the fraction of retained austenite, its shape and size and its microstructural relationship with other phases such as ferrite and bainite are also very crucial to the improvement of strength and ductility by the TRIP effect. Chung[10] reported from *in-situ* transmission electron microscope tensile experiments that retained austenite associated with nearby ferrite and bainite is transformed to martensite during plastic deformation, thereby working favorably for the ductility improvement. However, film-shaped retained austenite grains located at bainite or martensite lath boundaries and those isolated in ferrite grains are not transformed to martensite even when a considerable amount of plastic deformation is applied, thus not contributing to the ductility improvement. Chung interpreted the latter case by the size stability effect[19,20] due to the extremely small size of retained austenite, and many other researchers have reported the same.[21,22] The former case is associated with the deformation constraint effect,[23,24] as retained austenites are surrounded by bainites, which are not deformed easily. In view of this, when intercritically annealed at 840 °C, a lower ductility is expected in comparison with the intercritical annealing at 800 °C, because of the considerable amount of film-shaped retained austenites located along bainite or martensite lath boundaries as well as those isolated in the ferrite grains.

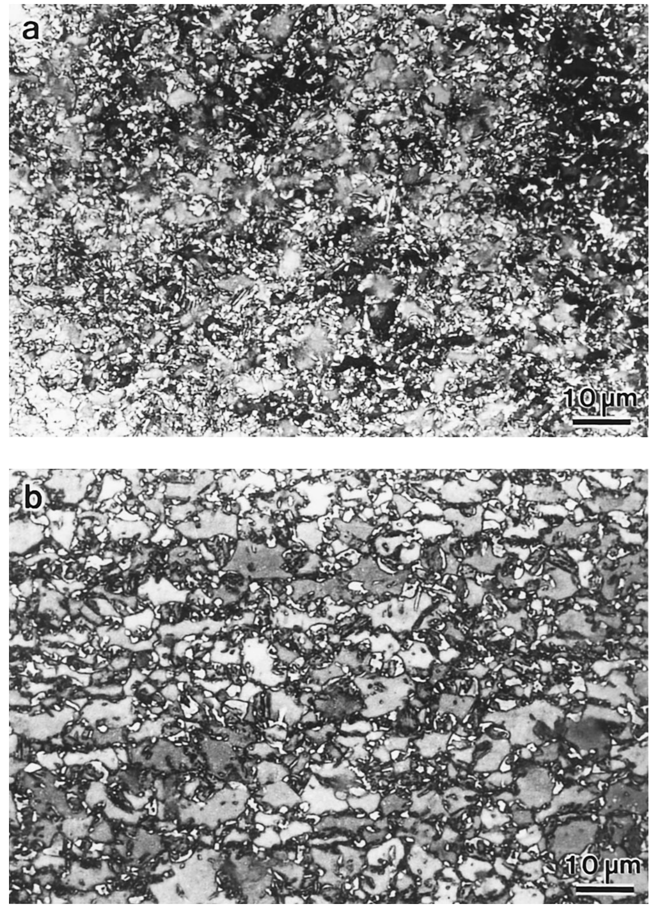
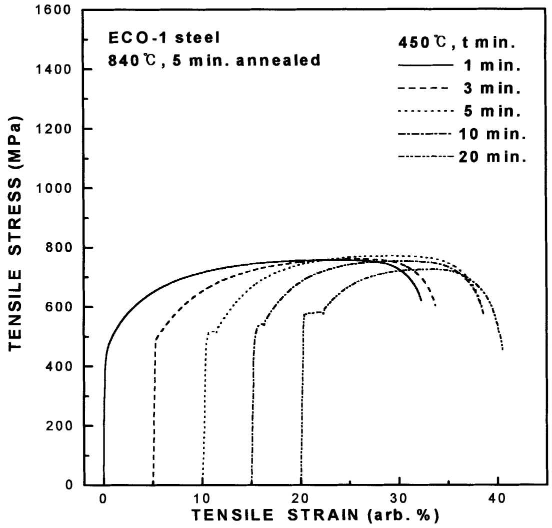


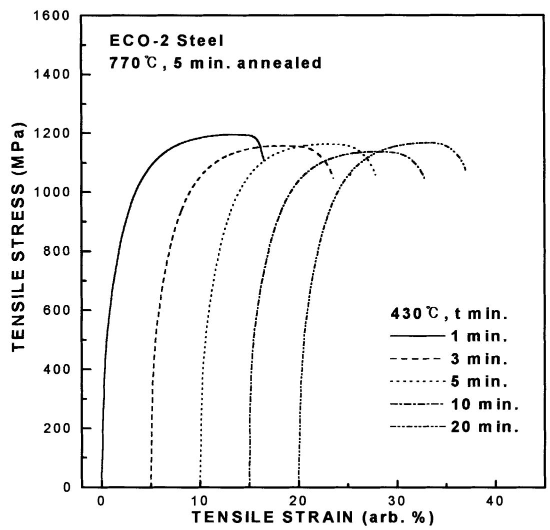
Fig. 4—Optical micrographs of the ECO-1 steel intercritically annealed at (*a*) 840 °C and (*b*) 800 °C for 5 min and isothermally treated at 430 °C for 5 min. Etched by a 10 pct sodium metabisulfite solution.

C. *Mechanical Properties and Volume Fraction of Retained Austenite*

Figures 5(a) through (c) provide tensile stress-strain curves of the steel specimens after the isothermal treatment. The ECO-1 steel shows a very peculiar tensile behavior (Figure 5(a)) in comparison with other steels. The specimen isothermally treated for 1 minute shows a continuous yielding behavior, but those treated for longer times show a discontinuous yielding behavior. It is well known that the discontinuous yielding behavior in steels arises, in general,



(*a*)



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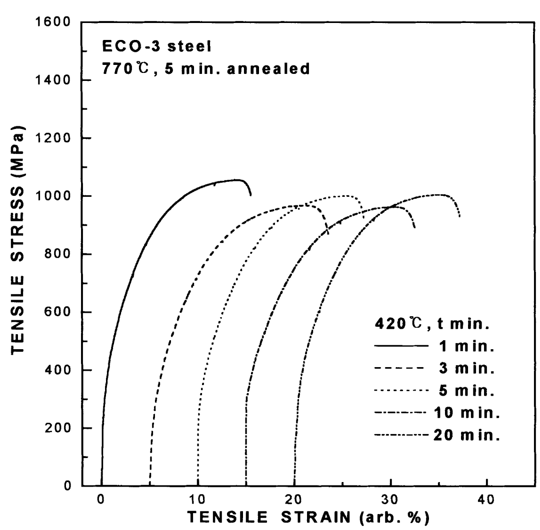

(*c*)

Fig. 5—Tensile stress-strain curves of the (*a*) ECO-1, (*b*) ECO-2, and (*c*) ECO-3 steels intercritically annealed and isothermally treated. Heat treating conditions are described in each diagram.

from the interaction between dislocations and interstitial solute atoms.[25] This phenomenon is frequently observed in low-carbon steels and is accompanied by the reduction of discontinuous stress, which is related to the formation of Luders bands, and by upper and lower yield points.[25] In the tensile curves of Figure 5(a), a rather constant stress is maintained throughout the yield elongation, and the yield elongation increases with increasing isothermal treatment time, showing a different pattern from typical cases. It is, thus, found that some other factors different from the discontinuous yielding behavior occurring in typical low-carbon steels are working here. According to Sakuma *et al.*,[26] the discontinuous yielding behavior occurs when the dislocation density in ferrite grains is low after the isothermal treatment. Nevertheless, there remain many uncertainties that cannot be explained by the dislocation density alone, but are supposedly related to the martensitic transformation of unstable retained austenite at the initial stage of tensile deformation. Therefore, further in-depth studies on the discontinuous yielding behavior of TRIP-aided steels are required.

Unlike the case of the ECO-1 steel, the tensile stress-strain curves of the ECO-2 steel show a continuous yielding behavior, together with a high tensile strength of about 1200 MPa, irrespective of the time (Figure 5(b)). Such a continuous yielding behavior is also observed in the ECO-3 steel (Figure 5(c)). This tensile deformation behavior of the ECO-2 and ECO-3 steels is very similar to that of typical low-carbon dual-phase steels, which are characterized by a continuous yielding, a low yield-to-tensile ratio, a high strain-hardening rate at the initial stage of homogeneous deformation, and a tensile elongation of about 20 pct. Since the ECO-2 and ECO-3 steels have a higher manganese content, the hardenability of austenite increases, thereby forming lots of martensites in the final microstructure. Consequently, the strengthening effect of the dual phases overrides the TRIP effect of the retained austenite.

Figure 6 lists the mechanical properties and fractions of retained austenite of the three tensile specimens intercritically annealed and isothermally treated under the conditions specified in Figures 5(a) through (c), as a function of the isothermal treatment time. In the ECO-1 steel, the yield strength significantly increases to the level of 550 MPa from 400 MPa as the time increases, whereas its elongation is radically reduced to 20 from 32 pct and the tensile strength is slightly reduced at the level of 780 MPa. In the ECO-2 and ECO-3 steels, the yield strength drops when isothermally treated for 1 to 3 minutes, but increases slightly as the time increases. The tensile strength is maintained at about 1200 and 1000 MPa, respectively, regardless of the time, and the elongation is relatively low at about 20 pct. Comparing the mechanical properties of each steel, the variations in the strength-ductility balance in particular, with the variations in fraction of retained austenite, it is found that they are closely related. In the ECO-1 steel, when isothermally treated for 1 minute, the retained austenite fraction reaches the maximum of 12 pct but decreases with time. Both the elongation and strength-ductility balance are also reduced with increasing isothermal treatment time to about 32 pct and 24×10^3 MPa·pct, respectively. The ECO-2 and ECO-3 steels reach the maximum fraction of retained austenite of 12 pct under the isothermal treatment time of 3 to 5 minutes, together with the maximum values of the elongation

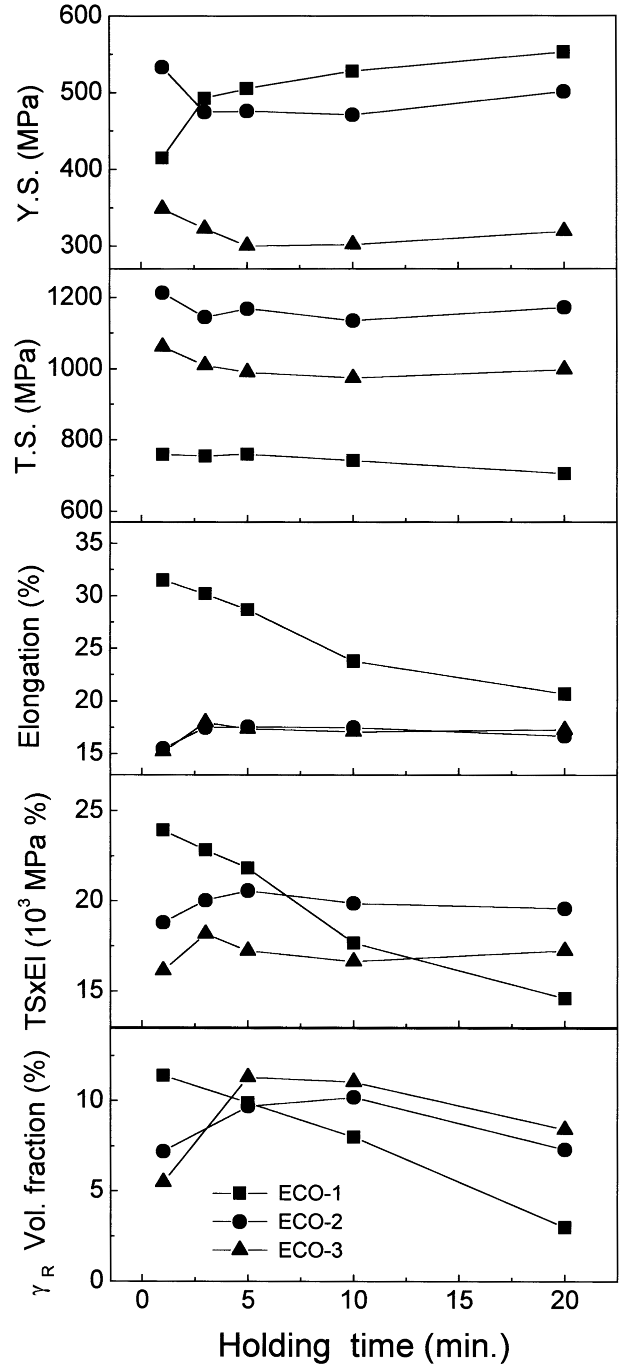

Fig. 6—Tensile properties and retained austenite volume fractions of the three steels as a function of isothermal treatment time. Intercritical annealing conditions and isothermal treatment temperatures are the same as specified in Figs. 5(a) through (c).

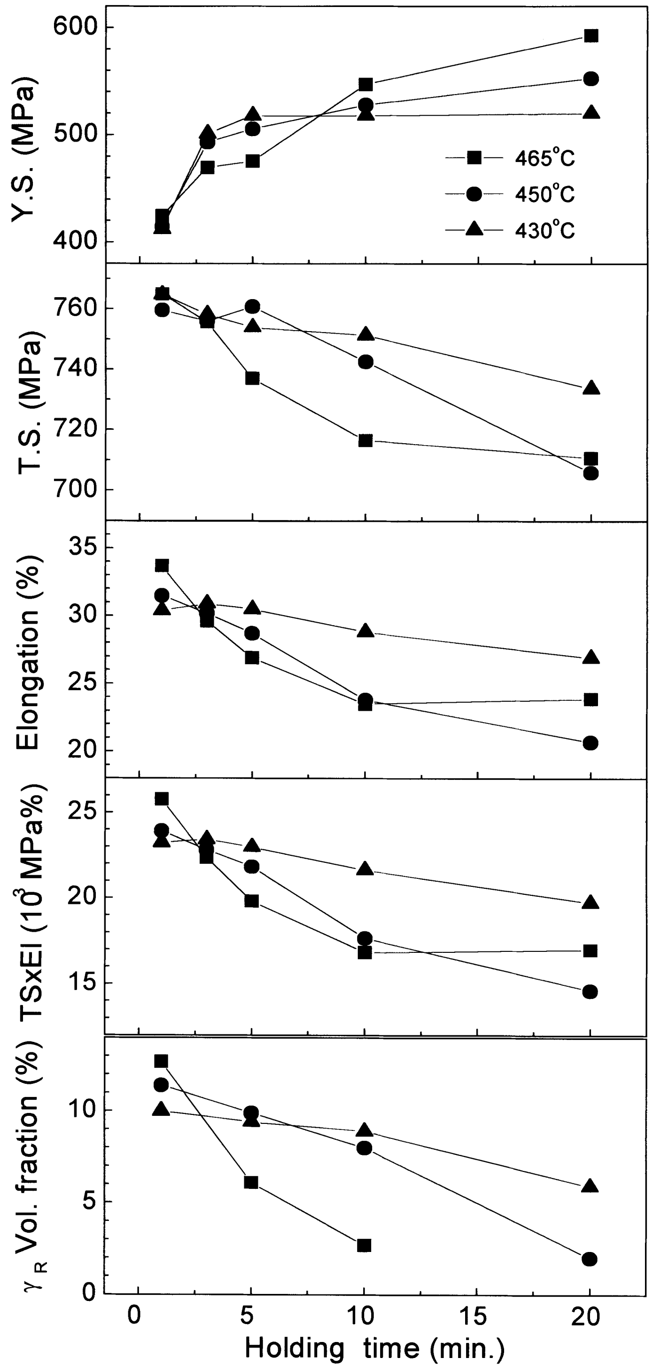

Fig. 7—Tensile properties and retained austenite volume fractions of the ECO-1 steel intercritically annealed at 840 °C for 5 min and isothermally treated at 430 °C to 465 °C for 1 to 20 min.

and strength-ductility balance. In comparison with the ECO-1 steel, they have a lower elongation and strength-ductility balance despite the high fraction of retained austenite, which indicates that the contribution of retained austenite to the ductility improvement is low.

Figure 7 summarizes the mechanical properties and fractions of retained austenite *vs* isothermal treatment temperature of the ECO-1 steel, which had the superior ductility among the three steels after it was intercritically annealed at 840 °C. With an increasing isothermal treatment time, the tensile strength decreases by 30 to 50 MPa from the level of 760 MPa, but the yield strength increases considerably. As the isothermal treatment temperature is lowered, the increase in yield strength becomes smaller. Particularly at 430 °C, the yield strength stays almost constant when the isothermal treatment time is over 5 minutes. The elongation is abruptly reduced from about 30 to 35 pct when the time increases, irrespective of the temperature. However, such a reduction tends to be less with a decrease in the temperature. The fraction of retained austenite is also radically reduced from 12 to 13 pct to below 3 pct with increasing treatment

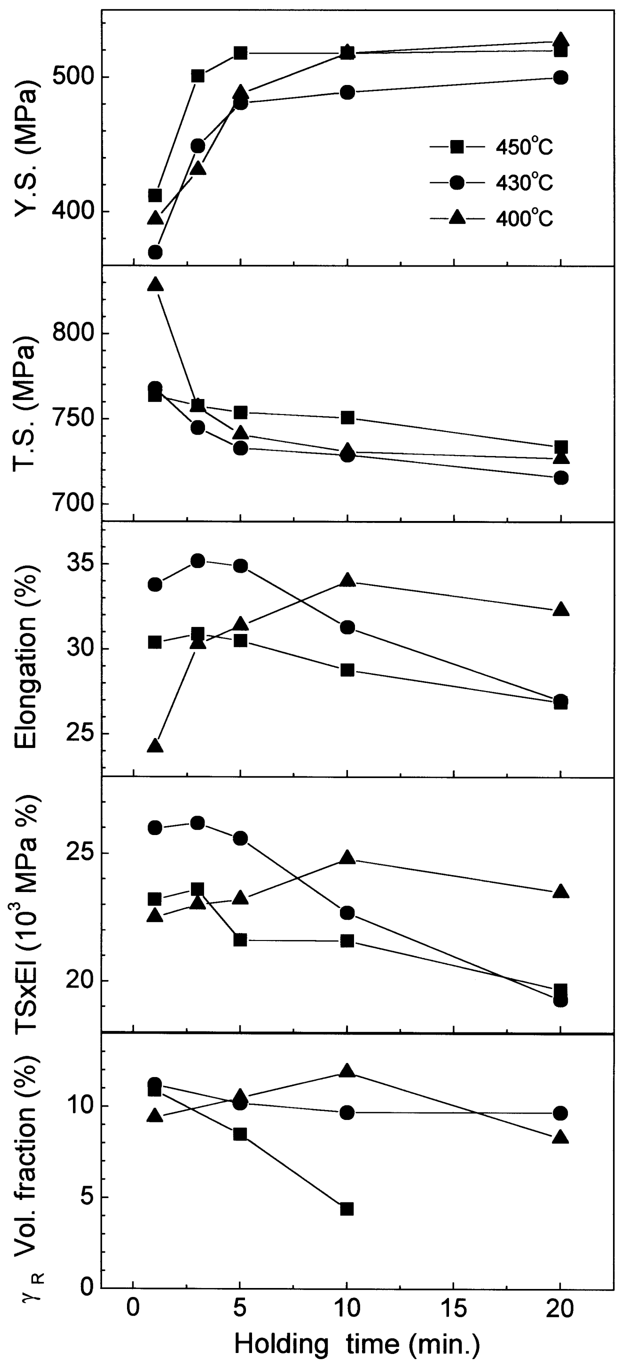

Fig. 8—Tensile properties and retained austenite volume fractions of the ECO-1 steel intercritically annealed at 800 °C for 5 min and isothermally treated at 400 °C to 450 °C for 1 to 20 min.

time when treated at 465 °C and 450 °C, whereas at 430 °C, the reduction in the retained austenite fraction is very small at about 3 pct.

Figure 8 also summarizes the mechanical properties and fractions of retained austenite of the ECO-1 steel *vs* the isothermal treatment temperature and time after it was intercritically annealed at 800 °C for 5 minutes. The variations in yield strength and tensile strength *vs* isothermal treatment conditions tend to be similar to the case of intercritical annealing at 840 °C, and the yield strength and tensile strength are kept almost constant when isothermally treated for 5 minutes or over 10 minutes. When isothermally treated for 3 minutes at 450 °C and 430 °C, the elongation reaches the maximum of 31 and 35 pct, respectively, and then drops. In the case of isothermal treatment at 400 °C, it reaches the maximum of 33 pct when treated for 10 minutes. Compared with the case of the intercritical annealing at 840 °C (Figure 7), it is expected to have superior formability because the value of the strength-ductility balance is higher due to its higher elongation in general. In the case of the isothermal treatment at 450 °C, the retained austenite fraction abruptly drops from about 12 pct as the treatment time increases, but it stays at about 10 pct in the case of treatment at 430 °C, regardless of the treatment time. When isothermally treated at 400 °C, it reaches the maximum of 12 pct with the 10-minute treatment. Figures 7 and 8 confirm that the changes in the mechanical properties, especially the elongation and strength-ductility balance, correlate well with those in the retained austenite fraction. These figures also show that the higher the retained austenite fraction is, the higher the elongation and strength-ductility balance get, whereas the elongation and strength-ductility balance decrease with a decreasing retained austenite fraction. It is found that under identical intercritical annealing conditions, the retained austenite fraction increases with a decreasing isothermal treatment temperature and the elongation and strength-ductility balance also improve. This is because the stability of retained austenite improves when the bainitic transformation proceeds more actively, while the ferritic transformation is prevented at lower isothermal treatment temperatures. Thus, in order to improve the elongation and strength-ductility balance due to the increased retained austenite fraction, it is more favorable to conduct the isothermal treatment at a temperature as close to the M_s point as possible.

Figure 9 summarizes the effects of the intercritical annealing temperature on the mechanical properties and retained austenite fractions of the ECO-1 steel isothermally treated at 430 °C, the same as in Figures 7 and 8. The yield and tensile strengths are higher in the intercritical annealing at 840 °C, but the elongation is higher in the case of annealing at 800 °C, particularly showing a high elongation of about 35 pct when isothermally treated for 3 to 5 minutes. Consequently, it is expected that the ECO-1 steel intercritically annealed at 800 °C has better formability due to its higher strength-ductility balance than does the sample annealed at 840 °C. The retained austenite fraction also is higher at 800 °C, and its variance *vs* the isothermal treatment time is small. The higher fraction of retained austenite in the intercritical annealing at 800 °C than that present at 840 °C can be explained by the improved austenite stability due to the higher carbon solubility in austenite.

IV. DISCUSSION

A. *Effects of Manganese and Silicon*

The mechanical properties of TRIP-aided high-strength cold-rolled steel sheets having a mixed microstructure of ferrite, bainite, and retained austenite are closely related to the retained austenite fraction. Particularly, it was confirmed by experiments that the elongation and strength-ductility balance vary with the retained austenite fraction. Thus, important considerations in developing TRIP-aided high-strength cold-rolled steel sheets are to promote the formation

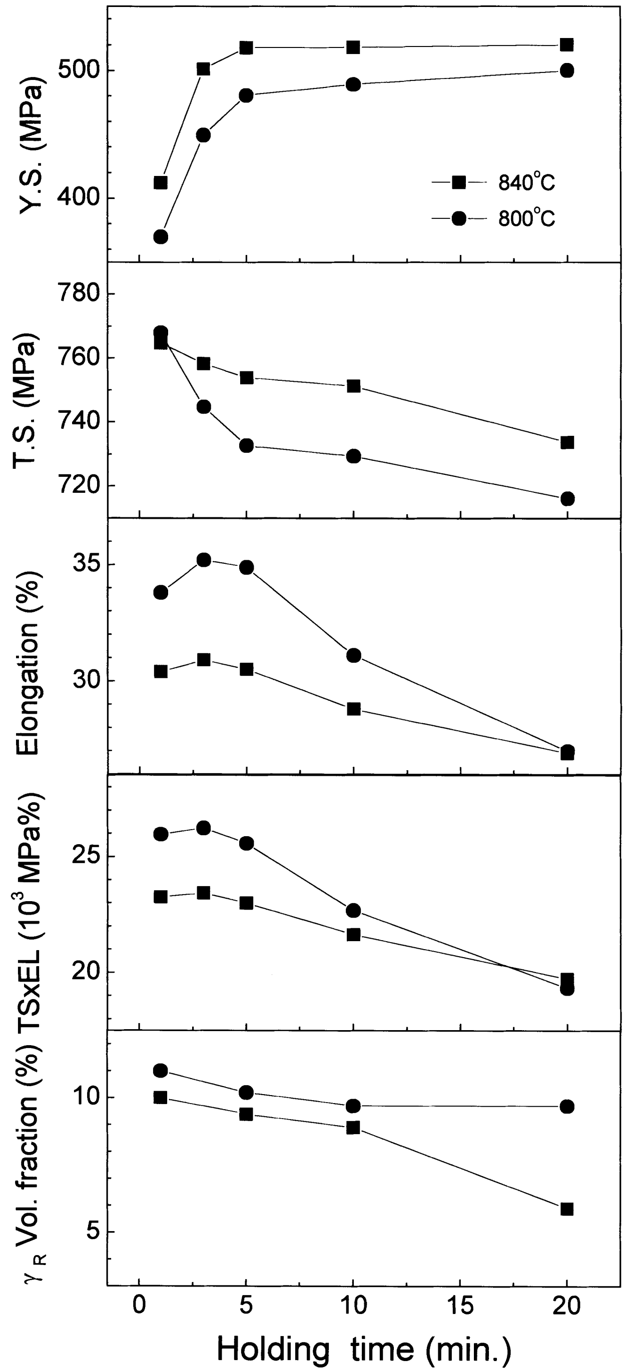

Fig. 9—Tensile properties and retained austenite volume fractions of the ECO-1 steel intercritically annealed at 800 °C and 840 °C for 5 min and isothermally treated at 430 °C for 1 to 20 min.

of stabilized retained austenite as much as possible and to design adequate alloy compositions and appropriate heat-treatment conditions.

In the heat treatment of TRIP-aided steels, the first consideration is to decide on the manganese content. Since manganese is an austenite stabilizer, it works favorably to form retained austenite. The ECO-2 and ECO-3 steels with a higher manganese content show higher retained austenite fractions than in the ECO-1 steel. With too a high manganese content, excessively stabilized retained austenites can be formed. Since they are not transformed to martensites even

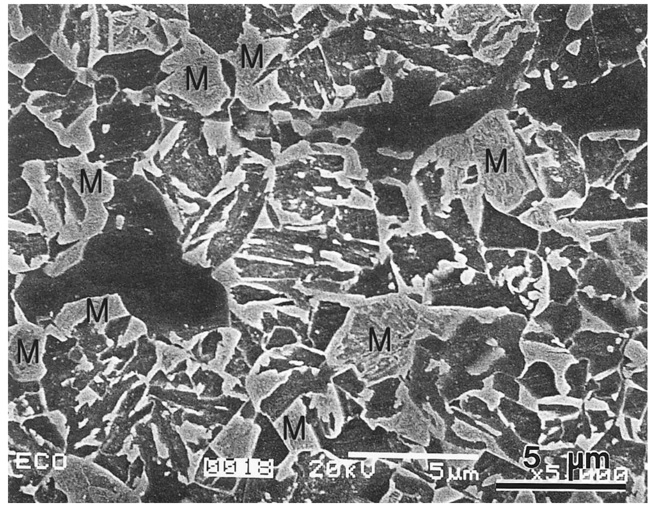

Fig. 10—SEM micrograph of the ECO-2 steel intercritically annealed at 770 °C for 5 min and isothermally treated at 430 °C for 10 min. Martensites are marked by "M."

during plastic deformation, they do not contribute to ductility enhancement. As they can also increase the hardenability of cold-rolled steels,[17,18] it is likely that the microstructure turns into a structure mainly consisting of ferrite and martensite due to the martensitic transformation of austenite during cooling after the isothermal treatment. In this case, the eventual fraction of retained austenite is reduced, together with the occurrence of the deformation behavior of a dual-phase structure; thus, a higher strength can be expected but not an improved elongation. Figures 5(b) and (c) of the ECO-2 and ECO-3 steels show a typical tensile deformation behavior of a dual-phase steel, because a considerable amount of martensite is formed. For the ECO-2 and ECO-3 steels intercritically annealed at 770 °C followed by the isothermal treatment at 430 °C for 10 minutes, the same conditions as specified in Figures 5(b) and (c) were observed by a scanning electron microscope (SEM). A representative SEM micrograph of the ECO-2 steel is shown in Figure 10. Here, lots of martensites are observed, as marked by the letter M. Distinguishing martensite from retained austenite in the SEM micrograph is possible by the surface appearance of both phases. That is, the martensite consists of many laths and the retained austenite looks rather smooth. This figure confirms that, with too much manganese above an appropriate level, the strengthening effect of dual phases overrides the TRIP effect, thereby adversely affecting the ductility enhancement.

Another point to consider when adding manganese is the formation of the band structure. This can be confirmed by the optical micrographs of cold-rolled steels, ECO-3 steel in particular, as shown in Figures 1(a) through (c). This band structure can remain even after the intercritical annealing (Figures 2(b) and (c)), wherein retained austenite and martensite are intensively formed, since manganese can be segregated in the band structure. The final microstructure would have the band structure, in which hard phases such as bainite and martensite are locally concentrated. Since such a microstructure is brittle, the band structure does not work favorably for the ductility improvement. When the microstructures of Figures 1(a) through (c) are compared with the chemical compositions of Table I, it is interesting to note that the ratio of silicon to manganese is related to the formation of the band structure. In the ECO-1 steel, where the Si:Mn ratio is close to 1, the pearlitic band structure is formed weakly, whereas in the ECO-2 and ECO-3 steels having Si:Mn ratios of 0.6 and 0.4, respectively, the band structure is developed completely, especially in the ECO-3 steel. These observations indicate that the Si:Mn ratio should be made high in order to prevent the formation of the band structure and to promote a homogeneous pearlite-ferrite structure. On this point, Thomas and Koo[27] and Chung[10] reported the disappearance of the band structure in steels having an increased silicon content. Sawai *et al.*[28] and other researchers[29,30,31] investigated the effect of various kinds of alloying elements on the formation of retained austenite and reported that the amount of retained austenite increases by increasing the addition of alloying elements, with silicon working better than manganese. To improve ductility due to the increased amount of retained austenite under a given carbon content, thus, it is desirable to set the Si:Mn ratio as high as possible.

B. *Effects of Heat-Treatment Conditions*

Since the main emphasis in developing high-strength TRIP-aided cold-rolled steel sheets is placed on promoting stabilized retained austenite as much as possible, it is imperative to select appropriate alloy compositions and to establish heat-treatment conditions under which alloying elements can be utilized to the maximum. As the mechanical properties *vs* heat-treatment conditions vary similarly in the three steels, the effects of heat treatment on mechanical properties are discussed here, based on the results of the ECO-1 steel.

In this study, the isothermal treatment temperature was set on the basis of the M_s point cold-rolled steels. In the ECO-1 steel specimens which were under the isothermal treatment for a short time at 430 °C (just above the M_s point) after being intercritically annealed at 800 °C, where the austenite fraction was about 50 pct, a relatively high retained austenite fraction of over 10 pct was obtained, together with a considerable improvement in ductility. Examining the microstructural variations during the isothermal treatment, the phase that is transformed to bainite in the isothermal treatment is the austenite formed during the intercritical annealing. This indicates that the M_s point is not appropriate as the standard to be applied to determine the isothermal treatment condition. Because the chemical composition of austenite formed during the intercritical annealing is different from the overall composition of the cold-rolled steel and since austenite stabilizers such as carbon and manganese are mostly solved in austenite, the M_s point of the austenite is supposed to be lower than that of the steel itself. For instance, when the ECO-1 steel is intercritically annealed at 800 °C, the fraction ratio of ferrite to austenite is 50:50, according to the Fe-Mn binary-phase diagram.[32] Thus, the manganese content in the austenite becomes about 2 pct as the remaining manganese, except about 0.5 pct solved into ferrite out of the 1.47 pct manganese (Table I), is solved into the austenite. Assuming that all the carbon is solved into the austenite, disregarding the carbon solubility in the ferrite, the carbon concentration in the austenite becomes 0.3 pct. Consequently, the austenite having 0.3 pct carbon and 2 pct manganese can be assumed to transform to ferrite or bainite during the isothermal treatment. The M_s point of the austenite having this chemical composition is calculated to be about 370 °C, according to Andrew's equation.[33] The isothermal treatment temperature of 400 °C used in Figure 8 was, thus, set in reference to research results[10] that reported that the optimum isothermal temperature is M_s + (20 to 30 °C), based on the M_s point as calculated previously. When isothermally treated under such conditions, comparatively higher retained austenite fractions, together with equal or better mechanical properties, were obtained (Figure 8). Therefore, the chemical composition of the austenite formed during the intercritical annealing should be predicted first, and the M_s point should be calculated on the basis of this prediction to arrive at the appropriate isothermal treatment temperature.

The intercritical annealing is as important a heat treatment as is the isothermal treatment. Figure 9 illustrates that the mechanical properties and retained austenite fractions vary with the intercritical annealing conditions. This is because the fraction and chemical composition of austenite also vary with the intercritical annealing temperature. In the ECO-1 steel, the fraction of retained austenite formed at 840 °C, which is the $(A_{C1} + A_{C2})/2$ temperature, is about 60 pct, while it is about 50 pct at 800 °C. According to Figure 9, the ECO-1 steel specimens intercritically annealed at 800 °C show slightly lower yield and tensile strengths than those annealed at 840 °C, but show a superior elongation and strength-ductility balance, together with a higher retained austenite fraction. These results indicate that the intercritical annealing temperature should be set in such a way as to increase the stability of austenite by keeping the fraction of austenite formed during the intercritical annealing as low as possible, to promote the higher solubility of alloying elements in the austenite. With higher intercritical annealing temperatures, the austenite fraction increases, but a higher retained austenite fraction cannot be obtained because the solubility of alloying elements in the austenite and the stability of the austenite decrease. At too-low an intercritical annealing temperature, the solubility of alloying elements in the austenite can increase, but the obtainable fraction of retained austenite is reduced because of the lower austenite fraction and the excessively high austenite stability. On this point, many researchers[9,10,11] have reported that the appropriate intercritical annealing temperature to obtain the highest fraction of retained austenite is $(A_{C1} + A_{C3})/2$. Since the cold-rolled steels used in the present study contain a lower carbon content than in conventional TRIP-aided cold-rolled steels, the fraction of austenite formed when intercritically annealed at 840 °C, which is the $(A_{C1} + A_{C3})/2$ temperature, is rather high at about 60 pct, but the contents of alloying elements in the austenite, carbon in particular, are considerably reduced. Thus, the ferrite fraction increases with increasing isothermal treatment time, because the ferritic transformation occurs prior to the bainitic transformation at higher isothermal treatment temperatures (Figure 3), but the retained austenite fraction drops radically (Figure 7). When intercritically annealed at 800 °C, a higher fraction of retained austenite can be obtained (Figures 8 and 9) than at 840 °C, because the solubility of carbon and manganese in the austenite increases as the austenite fraction decreases by about 10 pct. Therefore, in establishing the intercritical annealing temperature, the concentration of alloying elements such as carbon and manganese solved in the austenite during the intercritical annealing and the austenite stability should be taken into consideration.

V. CONCLUSIONS

In the present study, three kinds of low-carbon TRIP-aided cold-rolled steel sheets were designed to contain 0.15 pct carbon, in order to improve their weldability. They were intercritically annealed and isothermally treated, and their microstructures, mechanical properties, and retained austenite fractions were investigated and analyzed to reach the following conclusions.

1. With a higher manganese content, the retained austenite fraction increases, but the martensitic transformation of austenite occurs during cooling because of the improved austenite hardenability. Thus, characteristics of a dual-phase steel occur and the elongation deteriorates, whereas the strength improves.
2. The smaller the Si:Mn ratio is, the stronger the band structure that is formed. The formation of the band structure diminishes the effect of the ductility improvement, and, thus, the Si:Mn ratios should be set as high as possible under a given carbon content in order to not promote the band structure but to increase the retained austenite fraction and ductility.
3. The isothermal treatment temperature should be set on the basis of the M_s temperature, in consideration of the fraction of austenite formed during the intercritical annealing and the solubility of alloying elements in the austenite, instead of the M_s temperature of the steel itself. In establishing the intercritical annealing temperature, the alloying elements and the fraction of austenite formed during the intercritical annealing, as well as its stability, should be taken into consideration. When intercritically annealed at a temperature at which the fraction ratio of ferrite to austenite is about 50:50, a higher fraction of retained austenite is obtainable, accompanied by a more effective ductility improvement without forsaking the deterioration of strength.
4. Higher ductility and strength can also be achieved in low-carbon cold-rolled steel sheets having a lower carbon content than conventional TRIP-aided steel sheets by designing the alloy composition and by establishing the intercritical annealing and isothermal treatment conditions appropriately.

REFERENCES

1. K.I. Sugimoto, N. Ushi, M. Kobayashi, and S.I. Hashimoto: *Iron Steel Inst. Jpn.*, 1992, vol. 32, p. 1311.
2. K.I. Sugimoto, N. Ushi, M. Kobayashi, and S.I. Hashimoto: *Metall. Trans. A*, 1992, vol. 23A, pp. 3085-91.
3. H.C. Chen, H. Era, and M. Shimizu: *Metall. Trans. A*, 1989, vol. 20A, pp. 4371-45.
4. S.K. Kim, H.C. Shin, J.H. Chung, and Y.W. Chang: *J. Kor. Inst. Metall. Mater.*, 1998, vol. 36, p. 151.
5. C.G. Lee and S.J. Kim: *J. Kor. Inst. Metall. Mater.*, 1998, vol. 36, p. 1024.
6. B.V.N. Rao and A.K. Sachdev: *Metall. Trans. A*, 1982, vol. 13A, pp. 1899-1906.
7. B.V.N. Rao and M.S. Rashid: *Metallography*, 1983, vol. 31, p. 19.
8. N.C. Goel, J.P. Chakravarty, and K. Tangri: *Metall. Trans. A*, 1987, vol. 18A, pp. 5-9.
9. W.C. Leslie and G.C. Rauch: *Metall. Trans. A*, 1978, vol. 9A, pp. 343-49.
10. J.H. Chung: Ph.D. Thesis, Pohang University of Science Technology, Pohang, 1994, p. 64.
11. O. Matsumura, Y. Sakuma, and H. Takechi: *Iron Steel Inst. Jpn.*, 1987, vol. 27, p. 570.
12. C.G. Lee, S.J. Kim, S.G. Park, and I. Choi: *J. Kor. Inst. Metall. Mater.*, 1998, vol. 36 p. 1382.
13. S.J. Kim and C.G. Lee: *J. Kor. Inst. Metall. Mater.*, 1999, vol. 37, p. 774.
14. R.L. Miller: *Trans. ASM*, 1964, vol. 57, p. 892.
15. R.L. Miller: *Trans. ASM*, 1968, vol. 61, p. 592.
16. W.C. Leslie: *The Physical Metallurgy of Steels*, McGraw-Hill Press, New York, NY, 1981, p. 172.
17. A.D. Romig, Jr. and R. Salzbrenner: *Scripta Metall.*, 1982, vol. 16, p. 33.
18. N. Pussegoda, W.R. Tyson, P. Wycliffe, and G.R. Purdy: *Metall. Trans. A*, 1984, vol. 15A, pp. 1499-1502.
19. B.V.N. Rao and M.S. Rashid: *Metallography*, 1983, vol. 13, p. 19.
20. J.M. Rigsbee and P.J. VanderArend: in *Formable HSLA and Dual-Phase Steels*, A.T. Davenport, ed., TMS-AIME, Warrendale, PA, 1979, p. 56.
21. H.K.D.H. Bhadeshia and D.V. Edmonds: *Met. Sci.*, 1983, vol. 17, p. 411.
22. G.S. Ansell, P.J. Brofman, and G. Judd: *Proc. Int. Conf. on Martensitic Transformations ICOMAT 1979'*, MIT Press, Boston, MA, 1980, p. 350.
23. S.J. Kim, C.G. Lee, and S. Lee: *J. Kor. Inst. Metall. Mater.*, 1997, vol. 35, p. 1278.
24. N.J. Kim and G. Thomas: *Metall. Trans. A*, 1981, vol. 12A, pp. 483-89.
25. G.E. Dieter: in *Mechanical Metallurgy*, S.I. Metric, ed., McGraw-Hill Press, New York, NY, 1988, p. 197.
26. Y. Sakuma, D.K. Matlock, and G. Krauss: *Metall. Trans. A*, 1992, vol. 23A, pp. 1233-41.
27. G. Thomas and J.Y. Koo: *Structure and Properties of Dual-Phase Steels*, R.A. Kot and J.W. Morris, eds., TMS-AIME, Warrendale, PA, 1979, p. 183.
28. I. Sawai, S. Uchida, and E. Kamisaka: *Tetsu-to-Hagané*, 1985, vol. 71, p. 1292.
29. H.K.D.H. Bhadeshia and D.V. Edmonds: *Metall. Sci.*, 1983, vol. 17, p. 411.
30. B.P.J. Sandvik and H.P. Novalanien: *Met. Technol.*, 1981, vol. 8, p. 213.
31. V.T.T. Mijhkinen and D.V. Edmonds: *Mater. Sci. Technol.*, 1987, vol. 3, p. 422.
32. *Binary Alloy Phase Diagrams*, 2nd ed., T.B. Massalski, ed., ASM, Metals Park, OH, 1990, p. 1725.
33. K.W. Andrew: *J. Iron Steel Inst.*, 1965, vol. 203, p. 721.