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Progress of Continuous Annealing Technology for Cold-Rolled Sheet Steels and Associated Product Development

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1 Introduction

Continuous casting (CC) is one of the most innovative processes in the modern steel industry. Continuous casting has made it possible to produce a slab directly from molten steel. It is very effective for saving energy and omitting manufacturing processes as compared with the conventional ingot making and slabing process. Steel products manufactured from CC slabs exhibit homogeneous properties with small differences in chemical composition and microstructure.

Continuous annealing (CA), also recently developed technology, is used in combination with continuous casting technology. Ordinary sheet steel products of approximately 1-mm thickness are finally cold-rolled from hot bands which are themselves hot-rolled from slabs with about 200-mm thickness. Cold-rolled products (cold-rolled sheet steels) are extensively applied, for example, in automotive use, due to their excellent surface and high dimensional accuracy. As-cold-rolled steels, including high strain induced by rolling, are too hard to be press-formed. Therefore, cold-rolled steels are usually annealed (recrystallized) following cold-rolling.

Box annealing (BA) has been used to recrystallize ordinary cold-rolled steels. In the box annealing process using a bell type furnace as described in **Fig. 1**, it takes approximately one week to complete annealing. The box annealing process requires batch-type cleaning for degreasing prior to the annealing process. Also, after annealing, it is necessary to perform temper-rolling and finishing under the off-line process. On the other hand, in the continuous annealing process, it is possible to link the above-mentioned additional processes, since the annealing time is only about 10 min. Therefore, the continuous annealing line includes not only annealing but also all processes from cleaning to finishing.

As well as the continuation of processes before and after annealing, the continuous annealing process includes the following additional advantages:

(1) Homogeneity of Products
Continuous annealed products exhibit more homogeneous mechanical properties in coil length and width than box annealed products annealed as coil.

(2) Product Development
The continuous annealing process has a high poten-

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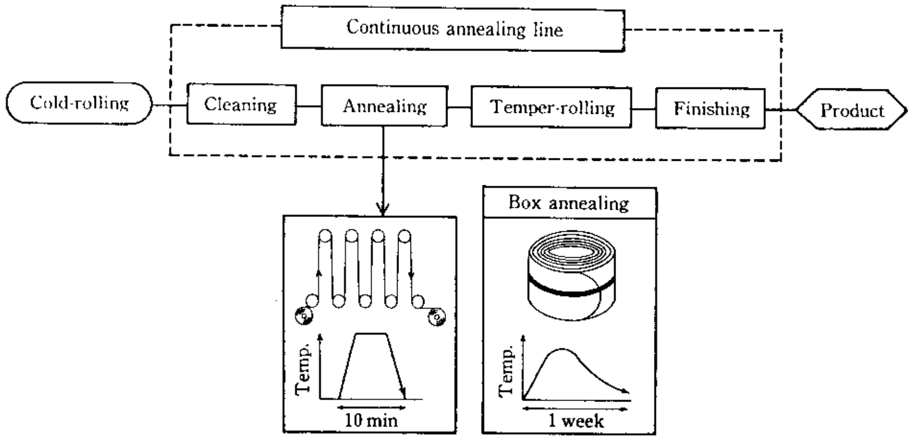

Fig. 1 Process flow of continuous annealing compared with conventional box annealing process

tial for product development since continuous annealing can be applied to higher soaking temperatures and higher cooling rates than box annealing.

In Kawasaki Steel (KSC), KM-CAL operations for cold-rolled sheet steels were first introduced to Chiba Works in 1980[1]. Continuous annealing lines with high capacity were successively installed at Mizushima Works in 1984 and at Chiba Works in 1988. Over 60% of KSC's cold-rolled sheet steels are currently produced by continuous annealing. This CA ratio is the highest in the world. In this paper, the history of continuous annealing technology will be reviewed mainly from a metallurgical view point. Associated product development using the continuous annealing process will be also described, with the focus on processing principles.

2 History of Continuous Annealing Technology

As-cold-rolled steel is very hard and thus difficult to press-form since many dislocations are contained in the steel. Recrystallization through heating is effective to obtain a soft microstructure. Heating and cooling rates of continuous annealing are generally higher than those of box annealing. Since the recrystallization temperature of an ordinary low C steel is 600 to 700°C, with sufficient soaking temperature and time, it is possible to use continuous annealing to complete recrystallization. However, CA-processed low C steels contain a high quantity of solute C due to higher cooling rate than box annealing process. The ductility of CA-processed low C steels, with a high content of solute C, is inferior to BA-processed steels. Furthermore, a high content of solute C easily causes aging, resulting in a deterioration of the steels' mechanical properties over time.

In 1936, Hague and Brace[2] proposed the rapid cooling and over-aging process after heating and soaking in continuous annealing line in order to reduce solute C in low C steels. **Table 1** summarizes the history of CA metallurgy, CA installation, and product development using CA. As for manufacturing formable cold-rolled steels with low C steels, Blickwede suggested that high temperature coiling at hot-rolling was effective in softening CA-processed steel due to the coarse carbides in hot bands[3]. He also pointed out that rapid cooling and over-aging after soaking was remarkably effective in reducing solute C[4].

In the beginning of 1970, two Japanese steel producers first started operation of the world's first continuous annealing lines for cold-rolled sheet steels. Both lines included the over-aging heat cycle following rapid cooling proposed by Hague et al. However, they applied different rapid cooling methods, namely gas-jet and water-quenching. The main products manufactured by both lines were the CQ grade (formable or flat use) and dual phase high strength steels.

Kawasaki Steel started operation of its continuous annealing line with accelerated gas-jet cooling in 1980. The line is characterized as a multipurpose facility which manufactures many kinds of products including dual phase high strength steels. The CA-processed products included an extra-deep drawing quality, EDDQ steel, when made from an extra-low C steel ($C < 30$ ppm)[5]. This was a different type of interstitial-free (IF) steel from conventional IF steels which contained a high amount of excessive carbide-forming elements such as Ti and Nb.

Since then, the continuous annealing line has been widely installed in many steel mills throughout the world. As for cooling techniques, mist-cooling and roll-cooling methods have also been adopted. The initial gas-jet and water-quenching cooling methods achieved a cooling rate of 10 to 30°C/s and approximately 1000°C/s, respectively. Accelerated gas-jet and roll-cooling techniques having an intermediate cooling rate are currently being used by many steel producers. This may have resulted because this cooling rate range is suitable for obtaining desired mechanical properties, surface quality and sheet shape.

Table 1 History of continuous annealing technology for cold-rolled sheet steels

Year	Metallurgy	Line installation: Company name	Line installation: Cooling type	Product
	Low C steel			
1936	Proposal for sheet CAL (by Hague & Brace)			
1959	High CT in hot-rolling (by Blickwede)			
1969	Rapid cooling & overaging (by Blickwede)			
1972		NSC (Kimitsu) NKK (Fukuyama)	Gas jet Water	Mild steel (CQ grade)
1975	Dual phase steel			High strength steels
1976		NKK (Fukuyama)	Water, Roll	
1979		NSC (Yawata)	Gas jet	
1980	New IF steel (Extra-low C steel)	KSC (Chiba)	A. gas jet	EDDQ mild & high strength steels
1981		SMC (Kashima) Novlipetsk Sidmar	Gas jet Water Gas jet	
1982		NSC (Yawata) NSC (Nagoya) NSC (Hirohata) Kobe (Kakogawa) SSAB	Mist Mist Mist Roll Water	
1983		Inland Bethlehem	Water Water	
1984	Bake-hardenable (BH) steel	KSC (Mizushima) Rasselstein CSN	Roll A. gas jet Gas jet	EDDQ-BH steels
1985		Cockerill	Hot water	
1986		Usinor Hösch Posco Nisshin (Sakai)	Roll Roll Roll Roll	
1987		NKK (Fukuyama) ISCOR Sollac CSC	Roll Roll A. gas jet Roll	
1988		KSC (Chiba)	Roll	

* A. gas jet : Accelerated gas jet cooling

In 1984, a high capacity continuous annealing line (CAL) was opened at Mizushima Works. Its main products are drawable steels (DQ to EDDQ) for automotive use, whereas previously installed continuous annealing lines have produced mainly CQ grade steel. Having the capability to maintain high annealing temperatures around 900°C has made it possible to develop a bake-hardenable EDDQ steel[6]. On the other hand, since no extended soaking periods are required for new type interstitial-free steels, the compact CAL, with short heating and soaking sections, was installed in Chiba Works in 1988.

The Continuous annealing process has spread more rapidly among domestic and foreign steel suppliers than the CC process. In the beginning of 1988, the total number of sheet CAL reached approximately 30. This number is expected to increase steadily in the future.

3 Product Development with CA Process

3.1 Dual Phase High Strength Steel

Dual phase (DP) high strength steel developed under the continuous annealing process is a typical example of steel which has been difficult to process under the box annealing method. The transformation temperature of ferrite to austenite (Ac_1) in low C steels ($C \cong 0.05$ wt.%) is approximately 700°C. Heating temperatures above Ac_1 creates a dual phase structure of ferrite and austenite. The austenite phase transforms to a hard martensite phase (α') when the heated steel is rapidly cooled above the critical rate. It is possible to obtain a duplex microstructure with $\alpha + \alpha'$ by controlling the steel's chemistry and the annealing heat cycle.

Figure 2 shows a schematic illustration of the dual phase structure and its mechanical properties as a function of the CA cooling rate[7]. The dual phase structure is obtained above the critical cooling rate, for example, at about 10°C/s in 0.05%C-1.2%Mn-0.5%Cr steel, resulting in a rapid decrease of yield stress (YS). Since tensile strength (TS) increases with an increase in the cooling rate, the decrease of YS provides a low yield ratio (YR = YS/TS). Lower YR steel, which has better deformation properties at a lower stress, is more suitable for pressing than sheet steels having the same TS grade. By controlling the volume fraction of α', it is possible to manufacture an extremely high TS steel. This is the important feature of dual phase steel. **Table 2** demonstrates the typical mechanical properties of dual phase steels with 40 to 100 kgf/mm^2 TS. Dual phase steels have been widely applied in automotive use, for example, in bumpers and doorguard bars.

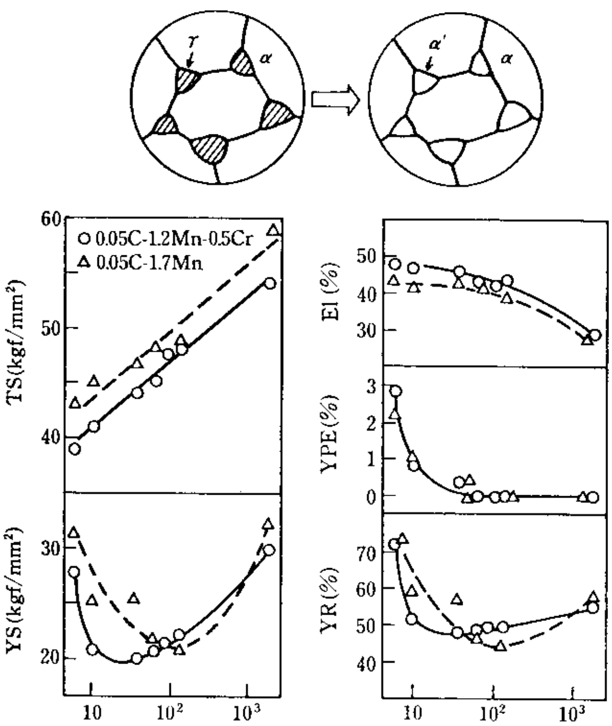

Fig. 2 Effect of cooling rate after intercritical annealing (soaking temp. 770°C) on mechanical properties of 0.05 wt.%C-1.7 wt.%Mn and 0.05 wt.%C-1.2 wt.%Mn-0.5 wt.%Cr steels

Table 2 Typical mechanical properties of dual-phase high-strength sheet steels

Grade	YS (kgf/mm²)	TS (kgf/mm²)	YR* (%)	El (%)
CHLY40	22	43	51	40
CHLY60	33	62	53	31
CHLY80	46	82	56	25
CHLY100	68	103	66	17

* YR = (YS/TS) × 100

3.2 Low C Drawing Quality Steel ($C = 0.02 \sim 0.05$wt.%)

Drawing property is required for automotive applications, for example, in white body parts. Drawing property is governed by the crystal orientation of sheet steels. The stronger the {111} orientation normal to the sheet plane is developed, the better is the drawability that results. Drawability is evaluated by *r*-value (Lankford-value). The *r*-values of DQ (drawing quality; SPCD in JIS) and EDDQ (extra-deep drawing quality) are approximately 1.7 and 2.0, respectively.

Figure 3 shows the manufacturing principle of low C DQ steel in the continuous annealing process as compared to the conventional box annealing process. In the box annealing process, DQ steel has been produced with low C Al-killed steels. AlN precipitation during BA heating develops a strong {111} recrystallization texture favorable for drawability. However, the effect of AlN precipitation is not available in the rapid heating of continuous annealing. High temperature coiling at hot-rolling improves the drawability of CA-processed low C Al-killed steels[8]. This may result because the coarse carbides in hot bands coiled at high temperature are effective in reducing the amount of solute C during recrystallization.

Resistance to aging is also important in DQ steel. A large amount of solute C in annealed sheet steels causes strain aging by C diffusion around dislocations even at ambient temperatures, resulting in the deterioration of mechanical properties such as ductility. Rapidly cooled low C steel in the continuous annealing process contains a high quantity of solute C, whereas most C is stabilized as Fe_3C during the cooling stage in the case of BA-processed low C steel. In order to reduce solute C in the continuous annealing process, supersaturation

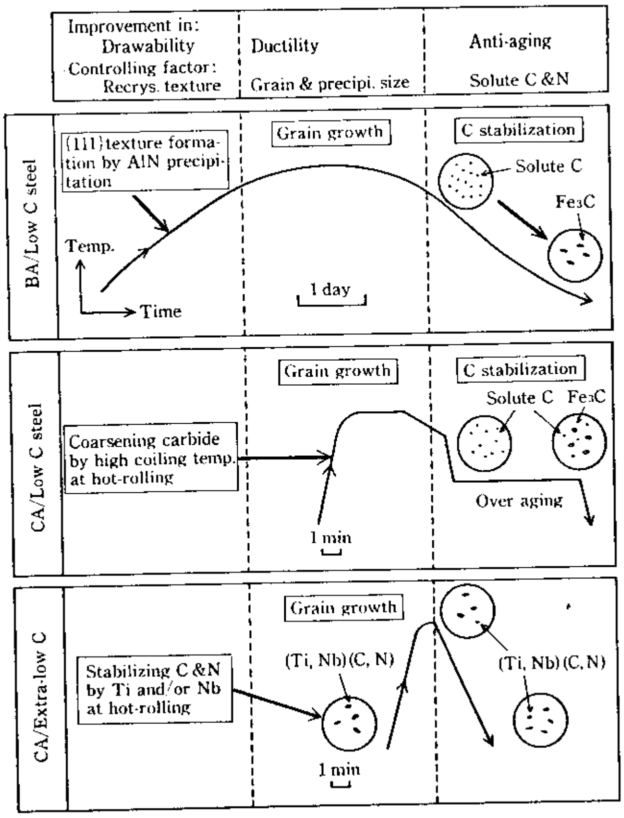

Fig. 3 Metallurgical principle of annealing process using low C and extra-low C steels for drawing quality sheet steels

by rapid cooling to intermediate temperatures (about 400°C) and subsequent over-aging for precipitating C as Fe_3C are efficient. The over-aged low C steel contains approximately 10 ppm solute C, while under the box annealing process it is possible to reduce solute C to less than 1 ppm. However, no distinct deterioration of mechanical properties within 3 months is recognized in overaged steel with approximately 10 ppm solute C or lower. On the other hand, N is also an interstitial atom. In the case of Al-killed steels, aging caused by solute N seldom occurs due to the stabilization of N as AlN.

The effect of the cooling rate, during continuous type annealing, on the mechanical properties of a low C steel (0.035%C-0.02%Al, coiling temp. 700°C) is represented in **Fig. 4**.[9] The soaking temperature is 800°C and slow cooling rate to 650°C is 2.5°C/s. Aging index (AI) of 4 kgf/mm^2 corresponds to solute C of approximately 10 ppm. The AI linearly decreases with an increase of the cooling rate, independent of over-aging temperature, T_0. However, ductility, characterized by elongation (El), has a maximum value at the intermediate cooling rate. At a too rapid cooling rate, ductility deteriorates due to the dense dispersion of carbides as shown in **Photo 1**. Therefore, in CA-processed low C Al-killed steels, an intermediate cooling rate is suitable for obtaining high ductility and a low aging index. For obtaining this range

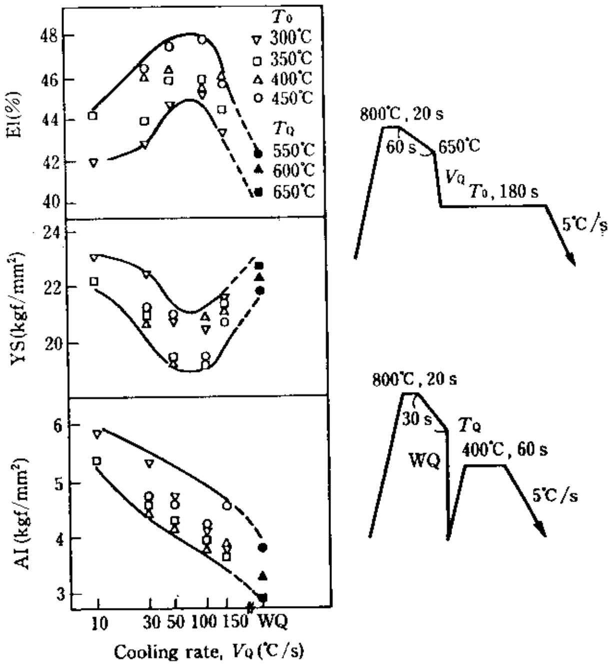

Fig. 4 Effect of cooling rate before over-aging on mechanical properties of 0.035 wt.%C-0.020 wt.%Al steel

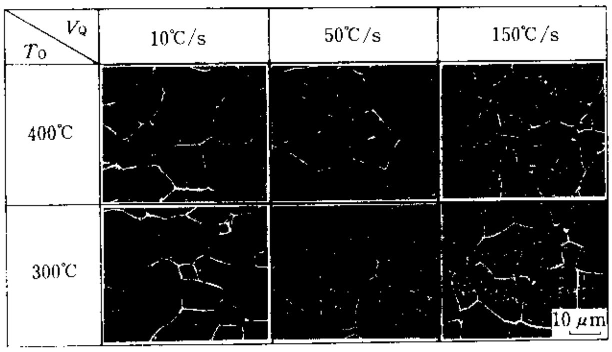

Photo 1 Scanning electron micrographs showing the effects of cooling rate and overaging temperature on carbide microstructures etched in nital

of cooling rate, accelerated gas-jet cooling, roll cooling, and a combination of both methods are widely used. Slow cooling treatment, before rapid cooling to the over-aging temperature, is also effective for obtaining the desired over-aging effect and improved drawability[10].

3.3 Extra-low C Mild Steel

Excellent drawability is obtained through the use of extra-low C cold-rolled steels containing C of less than 100 ppm (0.01wt.%) and sufficient carbide-forming elements such as Ti and Nb for stabilizing C[11,12]. This kind of steel is called an interstitial-free steel. C in the interstitial-free steel is stabilized as carbides at the hot band. Since the carbides are stable up to approximately 800°C during post-cold-rolling annealing, there is only a little solute C at recrystallization around 700°C, resulting in a strong {111} texture favoring drawability. The stable carbides also provide a non-aging property independent of the cooling conditions after soaking.

Figure 5 describes the potential of IF type extra-low C steels for manufacturing formable cold-rolled mild sheet steels and compares such potential with that of low C steels. CA-processed low C steel has a production limit of up to DQ grade. IF extra-low C type steel makes it possible to produce CQ to EDDQ grades under the CA process.

The serious disadvantage of using extra-low C steel was its high cost at steel making. However, recent advances in steel making technology have made it possible to reduce C content to less than 30 ppm within a reasonable cost.[13] There have been many advances, particularly in the RH-degassing process. The RH-refractory consumption, which is the main reason for increased costs, has been lowered remarkably, as shown in **Fig. 6**.[14] In order to improve the mechanical properties of cold-rolled sheet steels, lowering C content is efficient, however, it is difficult to obtain grades higher than DDQ (see Fig. 5) merely by reducing C to approximately 20 ppm as demonstrated in **Fig. 7**. Therefore, addition of strong carbide-forming elements such as Ti and Nb is necessary to produce the grades higher than DDQ even when using extra-low C steels with a C content of 20 to 30 ppm. The effect of Nb addition on mechanical properties is demonstrated in Fig. 7. In the case of 20 to 30 ppm C steels, the required amount of Nb or Ti is nearly equal to the stoichiometrically equivalent amount to the C content. This kind of steel might be classified under the new interstitial-free steel since the conventional interstitial-free steel contains

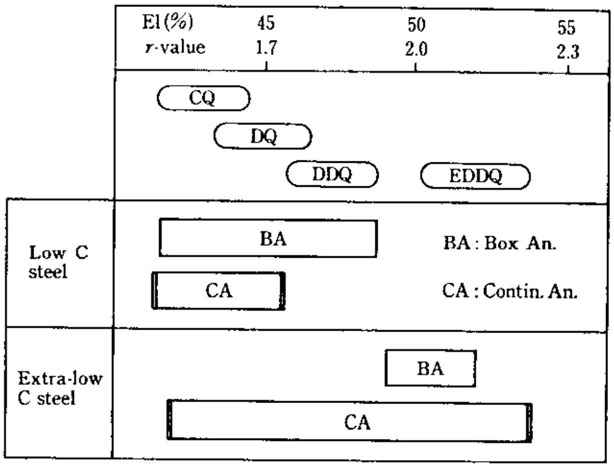

Fig. 5 Potential of extra-low C steels for manufacturing formable cold-rolled sheet steels with continuous annealing process

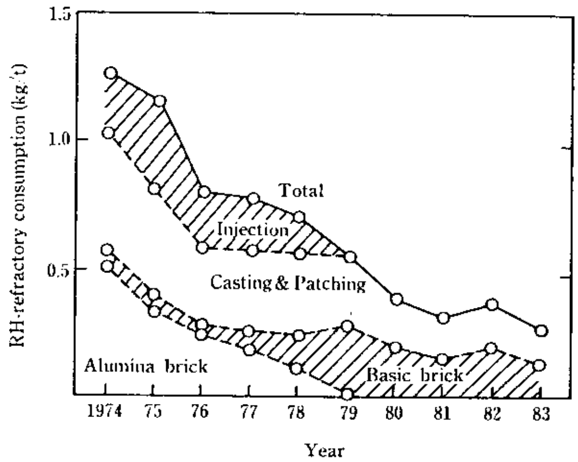

Fig. 6 Change of RH-refractory consumption at Chiba Works

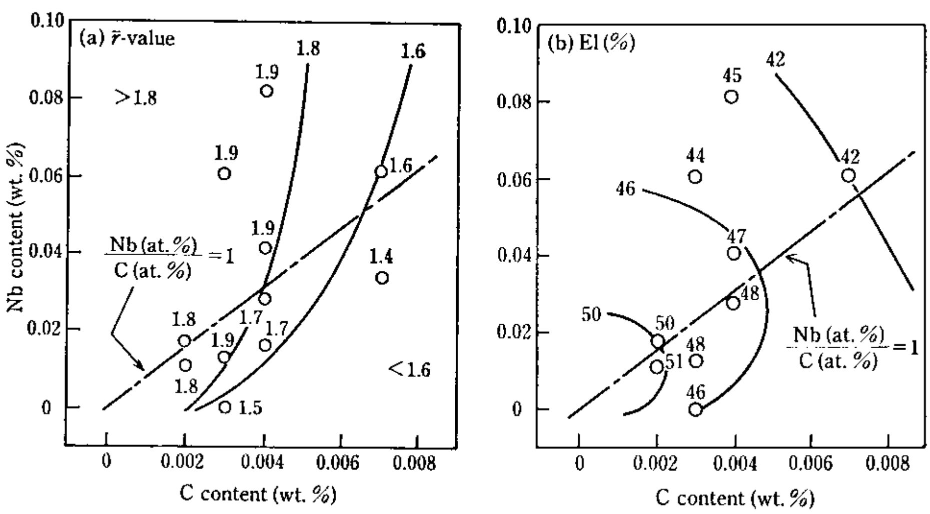

Fig. 7 Mechanical Properties as functions of C and Nb contents

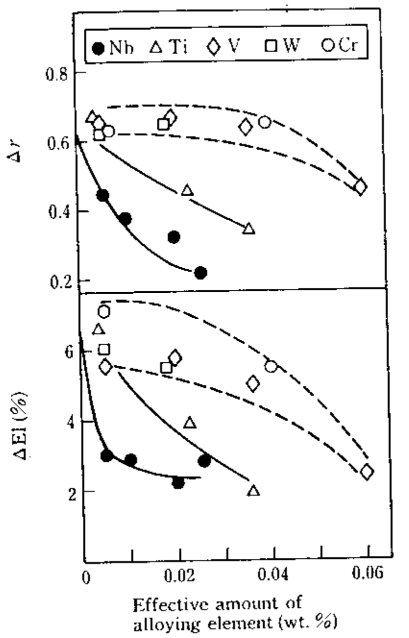

Fig. 8 Planar anisotropy of elongation and r-value in extra-low C (0.002 wt.%C) steels plotted against effective amount of carbide forming elements

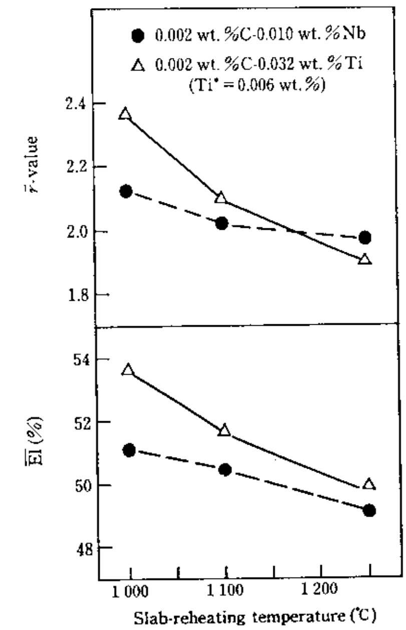

Fig. 9 Effect of reheating temperature before hot-rolling on mechanical properties of extra-low C cold-rolled sheet steels with Nb or Ti

excessive amounts of alloying elements relative to a C content.

Nb and Ti have a similar affinity to C in steel,[15,16] however, they exhibit some different effects on the mechanical properties of cold-rolled sheet steels. For example, the effect of adding alloying elements, on the planar anisotropy of r-value and El (Δr, ΔEl), is represented in **Fig. 8**. Δr and ΔEl were calculated by measuring r-value and El in the three directions (longitudinal, diagonal, and transverse) relative to the rolling direction. Ti and Nb are effective in lowering their planar anisotropy. The effect is particularly distinct in Nb-addition. This is mainly due to the grain refinement in hot bands[17].

Figure 9 shows the effect of heating temperature, during hot-rolling, on the mechanical properties of Nb- or Ti-added sheet steels (C = 20 ppm).[18] The practical slab-reheating temperature is approximately 1250°C. The El of Ti-added steel is higher than that of Nb-added steel. This is clear at the lower heating temperatures. The r-value of Ti-added steel is also higher than that of Nb-added steel, however, the difference is small at higher heating temperatures. The difference in mechanical properties between Ti- and Nb-added steels is mostly caused by the difference in dispersion of precipitates. The main precipitates in Ti-added steel are TiN, TiS, and TiC. Nb-added steel contains AlN, MnS, and NbC. TiN and TiS are coarser than AlN and MnS[18] since TiN and TiS begin to precipitate at higher temperatures than AlN and MnS, respectively. Generally speaking, low slab-reheating temperature treatment provides coarse precipitates in hot bands. Coarse dispersion of precipitates is more advantageous in grain growth during annealing than fine and dense dispersion, resulting in improved mechanical properties, especially in regards to ductility. The material above is useful in understanding the difference in ductility between Ti- and Nb-added steels.

Ti and Nb have a dissimilar effect on the mechanical properties of extra-low C steels, yet both elements are added mainly to stabilize C in steel. Thus addition of both Ti and Nb effectively ensures both of their advantages[19,20].

3.4 Extra-low C High Strength Sheet Steel

High strength sheet steels for automotive panels require excellent drawability. For this kind of application, EDDQ high strength sheet steels have been developed with solution (P, Mn etc.)-hardened extra-low C steels[21]. Surface distortion is the main cause of trouble in pressing automotive outer panels when an ordinary high strength steel, that is, a high yield strength steel, is applied. To solve this problem, a bake-hardenable (BH) steel has been developed. The bake-hardenable steel contains a small amount of solute C thereby guaranteeing resistance to aging. The yield strength after pressing is raised by stabilizing dislocations during bake-painting (approximately 170°C). Therefore, the bake-hardenable steel has a low yield strength at pressing but a high yield strength after bake-painting.

As mentioned before, a little amount of solute C at recrystallization retards the development of {111} texture. Therefore, it was difficult to obtain both bake-hardenability and extra-deep drawability (EDDQ-BH).

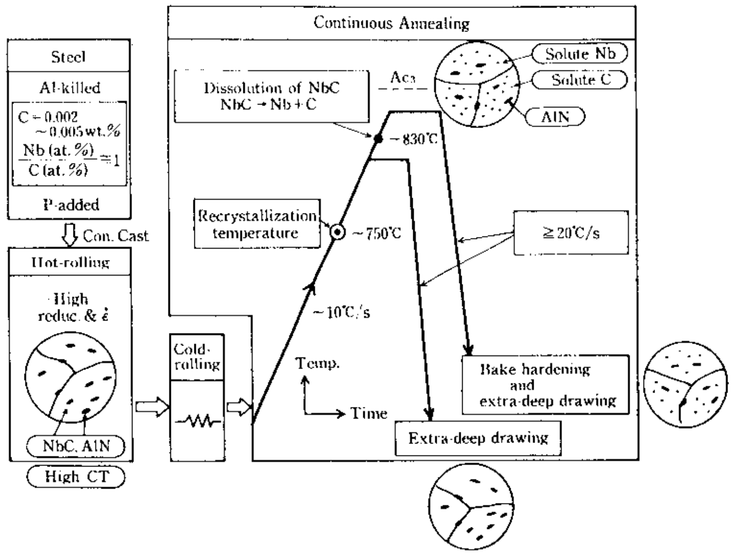

Fig. 10 Processing principle of bake-hardenable and extra-deep drawing sheet steel

Figure 10 illustrates the processing principle of EDDQ-BH steel when using extra-low C steel in the continuous annealing process. C is stabilized as stable carbides before completing recrystallization, resulting in a strong {111} recrystallization texture. Further heating produces some additional solute C due to the dissolution of carbides when the atomic ratio of alloying element to C is approximately unity. At this stage, the influence of solute C on texture development might be negligible since this stage corresponds to the growth of recrystallized grains. It is possible to retain a small amount of solute C by subsequent rapid cooling so as not to precipitate again.

According to this principle, the dissolution of carbides above the recrystallization temperature is necessary. Ti- and Nb-carbides are possible under practical conditions. Since Ti preferentially combines with N and subsequently stabilizes S, it is difficult to adjust practically an adequate amount of Ti to the C content. Thus Nb has been used for commercial production.

At steelmaking, Nb-added extra-low C steel with an approximate unity of Nb/C (atomic ratio) must be processed. In the case of Nb/C < 0.5 steel, solute C tends to retard {111} texture development. On the other hand, the dissolution of NbC in steels, having Nb/C > 1.5, shifts to higher temperatures close to the Ac_3 transformation point (about 900°C).[22] Annealing above Ac_3 drastically decreases {111} texture, resulting in the deterioration of drawability. Thus severe control of Nb/C by advanced steel-making technology is significant to manufacture EDDQ-BH steel.

High temperature annealing at around 850°C is required for dissolution of NbC. Since ordinary automotive sheet steels are very thin (about 0.7 mm) and wide (about 1500 mm), high temperature annealing causes problems such as heat-buckling. To prevent these problems, new techniques, for example, systems for tension control and optimizing hearth roll profile, have been developed[23,24].

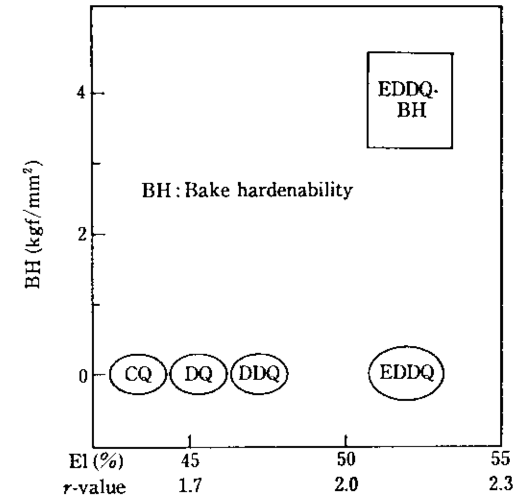

Fig. 11 Situation of bake-hardenable and extra-deep drawing sheet steel (EDDQ-BH) compared with conventional mild sheet steels

Figure 11 shows the effect of using Nb-added extra-low C steel for producing EDDQ-BH mild steel and compares it with non-BH mild steels. EDDQ-BH steel has been used in the production of automotive outer parts which require excellent formability and high dent-resistance.

4 Future Trends

4.1 Continuous Annealing Technology

Requirements for shape, homogeneity, and mechanical properties in cold-rolled steels have become more severe. The continuous annealing process is expected to spread more rapidly and widely since it is superior to the box annealing process in meeting the above mentioned requirements. In the continuous annealing process, further linking with pre- and post-processes will be anticipated. When controllability of heating and cooling in continuous annealing line is improved more, it might be possible to reduce differences in steel properties due to the change in steel chemistry and hot-rolling conditions. To achieve this improvement, on-line measuring and inspecting techniques of mechanical properties and surface quality should be improved.

Another significant technique is pre- and post-treatment in the continuous annealing process, so as to develop high value-added products, since customers require not only a large amount of stable property steels but also a small amount of unique products.

4.2 Steels Development

Since low C steel has limits in obtaining deep-drawing quality or the higher grades of steel for automotive use, extra-low C steel will become important. Further progress in steel making, which makes it possible to reduce the production cost of extra-low C steels, will increase the use of extra-low carbon steels. An overaging process for a low C steel is not necessary when extra low-C steels are used. Therefore, it is possible to shorten the line length when a continuous annealing line processes mainly extra-low C steels.

The continuous annealing process has a high potential for developing new products. We expect that the control of heating and cooling conditions would make it possible to develop new products which are difficult to manufacture under the BA process. Applying it to develop super high strength and super ductile steels is anticipated in the near future. Furthermore, metallurgical background of CA-processed products is almost at the stage where it can be applied to a continuous hot-dip galvanizing process with in-line annealing. Thus continuous annealing technology and associated product development will contribute to progress in surface-coated steel products.

5 Conclusions

The history of continuous annealing (CA) technology for cold-rolled sheet steels was reviewed from a metallurgical viewpoint, and processing principles for some new products developed under the continuous annealing process were described.

(1) Continuous annealing process for cold-rolled sheet steels proposed by Hague et al. in 1936 was first installed at Japanese steel mills in the early 1970s and quickly spread throughout the world. From the viewpoints of product quality and productivity, the continuous annealing process is expected to evolve globally in the future.

(2) Dual phase high strength steel and bake-hardenable extra-deep drawing quality steel have been developed by using CA's potential for rapid cooling and high temperature soaking.

(3) For automotive drawing quality steels, extra-low C steel is superior to low C steel because of its high potential for producing a wide grade of steels from CQ to EDDQ with no over-aging required in the continuous annealing process.

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