1. Introduction

Precipitation in steels is positively used for the control of strength, texture, grain size and over-aging, etc. of steels to obtain required properties thereof. Besides precipitation, the properties of steels are mainly controlled by means of transformation, recrystallization and grain growth. Because these metallurgical phenomena often interact with each other, it is indispensable to understand the influence of these interactions on each phenomenon so that the control by these phenomena may be optimized.

In this paper, the recent application of precipitation control to the development of steel sheets, including the interaction between precipitation and other metallurgical phenomena, is reviewed, special attention being paid to show how unique and sophisticated precipitation controls were applied to produce high quality steel sheets.

2. Recently Developed High Strength Steel Sheets Based on Precipitation Control

Precipitation hardening is an effective strengthening method. For example, in maraging steels, a strength increase of more than 1000 MPa can be achieved by proper precipitation hardening. The precipitation hardening is controlled by two different mechanisms, the Orowan mechanism and the cutting mechanism. Under the assumption of a fixed fraction precipitated, the precipitation hardening based on the Orowan mechanism decreases as the size of the precipitates increase while that based on the cutting mechanism increases as shown in Fig. 1. The most effective precipitation strength is realized at the transit point of the two mechanisms. The transit point depends on the strength of the precipitate, coherency of the precipitate in a matrix, etc.

In this section, unique precipitation hardened steel sheets recently developed are introduced.

2.1. High Strength Steel Sheets Hardened by Cu-precipitates

Hardening using Cu-precipitates was intensively studied in the 1960s and has been widely applied to strengthening steels. An advantageous application of Cu-precipitates is that they can be used in IF steels because Cu-precipitates don’t need C and N as precipitate forming elements. Using IF steel sheets strengthened by Cu-precipitates, deep drawable high strength steel sheets can be produced. For an example of a method for producing a deep drawable high strength steel sheet, Meyzaud et al. proposed that such a sheet could be produced through twofold batch annealing.
using a low carbon steel with C and Ti contents of 0.03% and 0.2%, respectively. Another example was given by Kishida et al.,\(^3\) who stated that a deep drawable cold rolled high strength steel sheet could be produced in a continuous annealing process using an extralow-carbon steel with C and Ti contents of 0.002% and 0.07%, respectively. These steel sheets have a microstructure consisting of fine precipitates dispersed in the ferrite matrix and show an excellent stretch flange formability.

**Figure 2** shows the change in tensile strength of a 1.6% Cu containing an extralow-carbon steel sheet during precipitation treatment at various temperatures.\(^3\) As can be recognized from this figure, the hardening can be realized within a relatively short time, and therefore it is possible that this kind of precipitation treatment could be carried out during the over-aging process in a continuous annealing line or in a short time heat treatment after forming.

As Cu-precipitates in steels, \(\varepsilon\)-Cu is well known. The smallest size of \(\varepsilon\)-Cu is several nanometers. In most cases, the increase in strength by Cu-precipitation cannot be fully explained by only the presence of \(\varepsilon\)-Cu precipitates observable by an electron microscope. Detailed TEM observations show a high density of contrast caused by strain induced by the coherency between fine precipitates and the matrix, which indicates the existence of very fine precipitates or clusters.

Goodman and his co-workers\(^6\) reported that \(10^{18}\) Cu-precipitates of a size of around 2.4 nm per unit cubic centimeter were observed in a 1.4% Cu bearing steel aged at 500°C by a field-ion microscope. Various researchers proved by means of micro-diffraction analysis and other methods that these very fine precipitates are \(\varepsilon\)-Cu.\(^5,7,8\) These very fine precipitates change their crystal structure from bcc to fcc over twinned 9R\(\rightarrow\)twinned 3R as their size increases during aging treatment.\(^4,7,8\) The bcc-Cu is formed before fcc-Cu, which is more stable than bcc-Cu according to the thermal equilibrium condition, because the interfacial energy between bcc-Cu and the ferrite matrix is lower than that between fcc-Cu and ferrite due to good coherency and because from the viewpoint of the total free energy change, it is preferable for the precipitation of bcc-Cu to take place at the initial stage of precipitation where the interfacial energy plays a significant role. If precipitates grow, the contribution of the interfacial energy to the overall free energy is reduced and as a consequence, the structure of the precipitates is changed to the stable fcc-Cu. Twinned 9R and twinned 3R, as the transit state of Cu-precipitates, can be considered crystal graphically as modified fcc-Cu with different amounts of stacking faults.\(^4\) The structural change from twinned 9R through twinned 3R to fcc Cu can be considered an apparent change through the annihilation of stacking faults. The change from bcc-Cu to the twinned 9R structure is thought to occur due to martensitic transformation driven by the strain energy which increases upon the growth of precipitates,\(^4,7\) but it has been also reported that a diffusional mechanism may partly contribute to this structural change.\(^10\)

This kind of structural change of precipitates has also been reported in other steels, for example, TiS\(\rightarrow\)Ti\(_2\)C\(_2\)S\(_2\)\(^11\) in Ti-bearing IF steel, M\(_6\)C\(_3\)\(\rightarrow\)M\(_7\)C\(_3\)\(\rightarrow\)M\(_{23}\)C\(_6\)\(^12\) in ferritic stainless steel, etc. At the present time, there are still many aspects regarding the mechanism and kinetics of the structural change of precipitates which are not understood, and thus further studies of this matter are required in the future.

**2.2. High Strength Steel Sheet with Ferrite Matrix Hardened by Precipitates Formed during Cooling After Hot Rolling**

The precipitation treatment of conventional precipitation hardened steel sheets is usually carried out in a coiling process. A new type of high strength steel sheet was developed in regard to which the precipitation treatment was carried out during cooling on the run-out table.

**Figure 3** shows the metallurgical principle for production of the steel sheet.\(^13\) The chemical composition of the steel is so controlled that the nose of TiC precipitation and that of \(\gamma/\alpha\) transformation almost overlap. To obtain a necessary strength increase, the content of C and Ti are adjusted to around 0.07% and 0.05–0.12%, respectively. C is fixed as TiC during the \(\gamma/\alpha\) transformation so that the en-
richment of C in austenite is avoided and consequently, the formation of pearlite, bainite and martensite is hindered and only precipitation hardened ferrite microstructure is formed. This kind of microstructure is good for stretch flange formability, as mentioned in Sec. 2.1. If the balance of Ti and C contents is properly controlled, an interstitial atom free state is realized in the steel, and thus a significant low yield ratio can be achieved.

The microstructure of the steel sheet is characterized by the ferrite matrix with linearly dispersed precipitates apparently induced by $\gamma/\alpha$ transformation. The size of these precipitates precipitated at $\gamma/\alpha$ phase boundaries is relatively large. The question as to whether these precipitates strengthen the steel enough or there exist smaller unobservables precipitates has not been answered yet. In addition, the modeling of the precipitation behavior at $\gamma/\alpha$ phase boundaries during transformation is a task which remains to be achieved in the future.

As a modification of the above described steel sheet, a DP steel sheet with a microstructure consisting of the ferrite matrix with fine precipitates embedded therein and dispersed martensite has also been developed. In this case, the Mn content is increased to achieve DP microstructure and in the same time, the Si content is increased to match the nose of TiC precipitation with that of the $\gamma/\alpha$ transformation. The precipitation hardened ferrite is formed at the former cooling stage on the run-out table, while martensite is formed by rapid cooling at the latter cooling stage.

2.3. Cold Rolled High Strength Steel Sheets with Low Yield Ratio Realized by PDZ (Precipitates Depleted Zone)

The feature of the chemical composition of this steel is that the contents of C and Nb are around 60 ppm and 0.1%, respectively, which is relatively high for an IF steel sheet. The solute drag effect of Nb and the pinning effect of NbC refine the ferrite microstructure. This refinement of the microstructure improves the secondary forming brittleness. The Nb segregated at grain boundaries may also contribute to the improvement of the secondary forming brittleness. Although the grain size after annealing is relatively small, this steel sheet shows a relatively high $r$-value and resultant good deep drawability, perhaps due to the formation of texture favorable to deep drawability. Solute Nb and finely precipitated NbC retard the recrystallization of austenite. The transformation texture of hot bands formed from the deformation texture of austenite leads to the formation of the recrystallization texture favorable to deep drawability. The fine microstructure of hot bands also contributes to forming the recrystallization texture with high intensity of ND//⟨111⟩ orientation resulting in high $r$-value.

Another remarkable feature of this steel sheet lies in its low yield strength and its resultant low yield ratio, although the grain size of the steel sheet is fine. The reason for this low yield strength was determined by observing the microstructure of the steel sheet. This observation revealed that, in the vicinity of grain boundaries, zones free from fine precipitates are formed which can yield easier than the matrix. Why such precipitate free zones are formed can be explained as follows: After the completion of recrystallization, precipitates are distributed relatively homogeneously all over the matrix. If the grain growth begins, fine precipitates contacting with a migrating grain boundary undergo the Ostwald growth at a higher rate than in the matrix and become coarsened. In the zones the grain boundary passed, only relatively coarse precipitates remain. These are the zones free from fine precipitates in the vicinity of grain boundaries. This result indicates that using this phenomenon, lowering of the YP and YR can be generally realized in precipitation hardened steel sheets if a proper heat treatment is carried out.

2.4. Nitrided High Strength Steel Sheets with High Dent Resistance

Nitriding is a well-known heat treatment for the surface region of steels which improves wear resistance, fatigue resistance, etc. This heat treatment is usually carried out at a temperature of around 550–600°C for a relatively long time of more than several hours. Steels used for the nitriding treatment are those containing nitride forming elements such as Cr, Al, Mo, V, Ti, etc. The hardening caused by nitriding is based on the formation of fine precipitates, clusters or G-P zones.

On the other hand, the nitrided steel sheet introduced here is heat-treated at around 700–800°C for quite a short time of less than a minute. The nitride forming element used in this case is Ti. The chemical free energy released from the formation of Ti–N complexes at these temperatures is extremely high, which enables this short time nitriding.

Increasing the thickness or YP of the steel sheet is an effective way to improve the dent resistance of an automotive outer panel. However, increasing the thickness of the steel sheet goes against the present trend in which automotive bodies are lightened. Further, as YP increases, the probability of the occurrence of a surface deflection in the vicinity of an emboss also increases.

As shown in Fig. 4, the nitrided steel sheet has a higher dent resistance than usual steel sheets with the same YP because the nitrided steel sheet has a kind of sandwich structure. The steel sheet developed here is a promising candidate for substitution in place of BH steel sheets if the bak-
ing temperature is lowered in the future and the attainment of an sufficient BH cannot be expected.

Figure 5 shows hardness distributions in the thickness direction of an Al-killed extralow-carbon steel sheet and Ti bearing extralow-carbon steel sheets nitrized at 800°C for 30 sec in an atmosphere consisting of 73.5% N\textsubscript{2}, 24.5% H\textsubscript{2} and 2% NH\textsubscript{3}. Their chemical composition of these steel sheets is given in Table 1. The Ti bearing extralow-carbon steel sheets show a higher increase in hardness than the Al-killed extralow-carbon steel sheet and despite the short nitrizing time, the increase in hardness in the vicinity of the surface is significant. This hardness distribution enables these Ti bearing extralow-carbon steel sheets to have a sandwich structure. The author and his collaborators\textsuperscript{16) have investigated the influence of NH\textsubscript{3} concentration, nitrizing temperature and time, the content of Ti, etc., on the hardness distribution and provided the conditions for producing a steel sheet good in both dent resistance and anti-surface deflection. They also developed a mathematical model for predicting the nitrizing behavior and the hardness distribution. Because the classical nucleation theory was assumed in the model, the formation of clusters or G-P zones cannot be considered which could be formed.\textsuperscript{17–19) Presently a detailed investigation is being carried out by means of a three-dimensional atom probe to clarify the precipitation behavior at the atomic level during nitrizing.

2.5. High Strength Steel Sheets with High Fatigue Resistance in Welded Areas\textsuperscript{20)}

The fatigue resistance of steel sheets themselves usually increases as TS increases. Heat affected zones (HAZ), however, often undergo softening by welding. The decrease in strength in HAZ is significant if the TS becomes higher than 500 N/mm\textsuperscript{2} and an expected improvement of fatigue resistance cannot be realized even though the strength of the bulk material is increased.

A new type of high strength steel sheet contains Nb and Mo and these elements form fine multiple precipitates such as NbC and Mo\textsubscript{23}C\textsubscript{6} in HAZ during cooling after welding. Mo hinders the annihilation of dislocations introduced by press forming and the dislocations remaining in HAZ act as preferential nucleation sites for NbC precipitation and strengthen the HAZ. Nb contributes to strengthening the HAZ not only by its precipitation but also by its solute drag effect which hinders the coarsening of the ferrite microstructure.

3. Grain Size Control

Grain size control is important for obtaining the properties steels require. The followings are some examples of why grain size control is important: The grain refinement of ferrite increases the toughness of steel. The coarsening of ferrite grains in hot bands deteriorates deep drawability of cold rolled steel sheets by decreasing mean r-value and increasing \( \Delta r \). The ridging problem of cold rolled stainless steel sheets, especially ferritic stainless steel sheets, is also caused by the large grain size of hot bands. Concerning cold rolled steel sheets, their formability usually deteriorates as grain size decreases but if the grain size becomes too large, a surface defect looking like orange peels appears after press forming. In the case of non-oriented electrical steel sheets, there exists an optimum grain size for core loss.

As shown above, to control the grain size is important for many products. The contribution of precipitates to the control of grain size is mainly based on the control of grain growth by their pinning effect. Further, precipitates are used as nucleation sites for solidification and transformation to refine the microstructure. This technology is called oxide metallurgy and has recently drawn much attention. In the following section, some of the recent developments in
3.1. Ultra-fine Grain Steels

Recently, the technology for obtaining ultra-fine ferrite microstructure has drawn worldwide attention. In Japan, big projects regarding such technology financed by governmental organizations are going on. Because grain refinement leads not only to an increase in strength but also an increase in the toughness of steel, there is keen expectation that steels with an ultra-fine grain microstructure will be utilisable as construction steels.

The methods for producing an ultra-fine ferrite microstructure have been studied and as a result, very high reduction rolling in the austenite region just above Ar3, in the γ/α dual phase region and the α region has been proposed for producing such a microstructure. In this connection, an effective use of precipitates promotes ultra-fine grain refinement.

Fujioka and his co-workers reported that they obtained an average ferrite grain size of 0.8 μm by a heavy working of 1.5 strain in the austenite region just above Ar3 using a steel containing 0.26% C and 0.3% Ti. In this case, TiC hindered the grain growth during heating before deformation and also suppressed the grain growth after deformation and transformation.

The ultra grain refinement achieved in the α-region by heavy deformation is realized by forming subgrains with high angle boundaries. To form grains of less than 1 μm with high angle boundaries, a strain of more than 2.3 is usually needed. In the case of a steel consisting of 0.16%C–0.015%Nb–0.013%Ti, an average grain size of 0.7 μm was achieved by a deformation of ε=2.3 due to a favorable effect of precipitates. In this case, fine cementite dispersed in grain boundaries is supposed to accelerate the formation of fine ferrite grains and to hinder their grain growth. In the case of a steel with 0.25% C and 0.5% Ti, the existence of abundant TiC helps to form the above mentioned microstructure, and an average grain size of less than 1 μm was achieved with a strain of 1.5. The detailed mechanism by which precipitates promote the formation of ultra-fine grain refinement has not been clarified yet, and thus its elucidation will have to be achieved in the future.

Concerning wire rods and rails using high carbon steels with a pearlitic microstructure, heavy deformation destroys the pearlite structure. That is, the cementite disappears and ferrite becomes ultra-fine and as a result, a significant increase in hardness and strength is achieved. In the case of cold drawing of wire rods, lamellar cementite is broken and dissolved due to heavy deformation and the C content in ferrite increases up to several percents. In order to explain the existence of the high amount of solute carbon in ferrite, the contribution of dislocations and grain boundaries to the equilibrium of free energy should be considered. Tarui and his co-workers reported that most carbon exists on dislocations. The segregation of carbon can decrease grain boundary energy and promote ultra-fine grain refinement.

Figure 6 shows the influence of the carbon content of pearlitic steels on the wear loss during wear tests. The initial hardness of each specimen is set to be an almost equal value. In this case, cementite does not completely disappear in the surface region which is heavily shear strained, but ultra-fine ferrite is observed. The different wear behavior of specimens was interpreted due to the different degree of the work hardening caused by wear tests. Upon the amount of carbon being increased, the grain refinement of ferrite is promoted and as a result, work hardening is accelerated.

The dissolution behavior of cementite due to heavy deformation has still not been quantitatively understood, and thus such understanding remains to be achieved in the future.

3.2. Oxide Metallurgy

So-called oxide metallurgy has drawn much attention for obtaining grain refinement in relatively light deformed or non-deformed parts, especially in the heat affected zone of welded parts where the microstructure tends to coarsen and grain refinement due to deformation cannot be expected. Oxide metallurgy is defined as a technology for refining a microstructure using oxides as nucleation sites for precipitation or transformation. The oxide metallurgy originally stemmed from the grain refinement of a welded part. Microstructural observation of a welded part resulted in the formation of acicular ferrite nucleated at oxides being recognized.

The grain refinement obtained by oxide metallurgy is divided into two cases: the refinement of a cast microstructure and that of a transformed microstructure. In the former case, oxides act as nucleation sites for solidification. This case is not concerned with the precipitation in the solid phase and is, therefore, out of the scope of this review. Thus, only some review papers which discuss this former case will be cited here.

In the above latter, ferrite seldom nucleates directly on oxides. Instead, ferrite nucleates on precipitates nucleated on oxides. TiN, BN and MnS precipitated on Ti2O3 or TiO are examples of favorable nucleation sites for ferrite. It was also reported that multiple precipitates such as VN or V2C3 precipitation on MnS precipitated on an oxide are effective nucleation sites for ferrite. There is another report that TiN itself without the help of oxides acts as an effective nucleation site for ferrite transformation.
The reasons why these precipitates are effective nucleation sites for ferrite transformation are reported to be as follows: 1. From the viewpoint of interfacial energy, a good coherency of the interface between the precipitate and matrix, say between VN and ferrite, helps ferrite to nucleate easily on the precipitate.\(^{31}\)

2. If an austenite forming element forms a precipitate, for example, Mn forms MnS, the austenite forming element is depleted in the vicinity of the precipitate and from the viewpoint of chemical free energy, a condition is created by which ferrite preferentially nucleates on the precipitate.\(^{34-36}\)

3. B is segregated in the interface between the oxide and ferrite matrix and is absorbed into the oxide to form a depleted zone similar to be one mentioned above.\(^{29}\)

4. The difference between the coefficient of the thermal expansion of the precipitate and that of matrix generates dislocations during cooling and the resultant strain energy provides ferrite with a preferential condition for nucleation.\(^{37}\)

Further, it is also well known that precipitates preferentially nucleate on certain oxides. The main reason for this is the interfacial energy between the precipitate and oxide based on the lattice coherency. The unique feature of the precipitation on oxides is that the behavior of ions influences the degree of preferentiality of the nucleation site.\(^{38}\)

It is reported that cation vacancy type oxides absorb Mn and let MnS precipitates easily on them.\(^{39}\)

Concerning MnS precipitated on oxides, they were more observed if the deoxidation is carried out using Mn, Si and Zr than using Al. The following mechanism for this fact was proposed: If the Mn containing oxide is a liquid oxide with high solubility of Mn at a high temperature, MnS is crystallized on the surface of oxide during cooling. This crystallized MnS is a preferential nucleation site for MnS.\(^{39}\)

As shown above, there are many studies on the preferential nucleation phenomena for the precipitation on oxides and for the ferrite transformation on oxides and precipitates. Most of these studies are phenomenological analyses. That is, there are few quantitative evaluations. Even some controversial results were reported. For example, Ishikawa and his coworkers\(^{31}\) reported that the crystal orientation relationship between V(CN) and the ferrite nucleated on the precipitate is the Baker–Nutting one, which proves that the interface has a good coherency, while Yamaguchi and his coworkers showed many ferrite grains which nucleated on V(CN) do not possess any special crystal orientation relationship with the precipitates such as Baker–Nutting one.\(^{40}\)

Anyway, there is still much work remaining to do in this area, including enriching the data base on interfacial energy and modeling the nucleation process in the oxide metallurgy.

### 3.3 Extralow-carbon Ferritic Stainless Steel Sheets

Hot rolled ferritic stainless steel sheets often reveal a coarse ferrite microstructure because transformation as a means for grain refinement cannot be used. The reduction of carbon enhances the grain coarsening. If the grain size of hot bands is large, the surface quality of cold rolled products deteriorates due to roping or ridging.

To refine the microstructure of hot bands, the refinement of the cast microstructure and the promotion of recrystallization in hot rolling processes are of great importance. To refine the cast microstructure, the crystallization of TiN in liquid iron acting as a solidification site is effective.\(^{41}\) The oxide metallurgy mentioned in the previous section is also applied for refining the cast microstructure of extralow-carbon ferritic stainless steels.\(^{42}\)

In extralow-carbon ferritic stainless steels, Ti is added in an weight amount at least 10 times as much as the sum of C and N to maintain the corrosion resistance at welded parts because a rapid precipitation of titanium carbonitrides during cooling after welding avoids the formation of chromium carbonitrides and, as a consequence, the formation of a Cr depleted zone is suppressed. If titanium carbonitrides finely precipitate in hot rolling processes, they retard the recrystallization of hot bands. The fine precipitates are mainly TiC. If C is precipitated in the form of Ti\(_4\)C\(_2\)S\(_2\), whose solubility in ferrite is lower than that of TiC,\(^{43}\) the amount of fine precipitates can be apparently reduced. This precipitation control promoting recrystallization resulted in a significant improvement of ridging.\(^{44,45}\) On the other hand, Kimura and his co-workers investigated the influence of P on the recrystallization behavior of the hot band and showed that the recrystallization was markedly retarded if FeTiP finely precipitated.\(^{36}\)

### 4. Texture Control

#### 4.1 Deep Drawable Al-killed Steel Sheets

Reviewing the history of deep drawable steel sheets, the first breakthrough technique for improving deep drawability of steel sheets was the box annealing of Al-killed steel sheets where a sophisticated precipitation control was conducted.\(^{47}\) Controlling the heating rate in the box annealing according to an amount of Al and N, AlN can be precipitated just after recrystallization of ND/⟨111⟩ grains to suppress the recrystallization of grains having other orientations. ND/⟨111⟩ grains have a higher amount of stored energy than most of the grains with other orientations and, therefore, they are recrystallized earlier. The recrystallization proceeds by the migration of recrystallized ND/⟨111⟩ grains into deformed grains with the other orientations, and finally a texture with a high intensity of ND/⟨111⟩ favorable for deep drawability is achieved. Because in this case the precipitation should occur in the box annealing process, in the hot coil Al and N must be in a solution. Therefore, a low temperature cooling is required.

The annealing process has been changed from box annealing to continuous annealing to improve productivity. Since the holding time in a continuous annealing process is short and, as a consequence, the grain growth is restricted, the grain size of cold rolled steel sheets is relatively small after continuous annealing. It is known that the deep drawable steel sheet shows better deep drawability if the grain size becomes larger. Therefore, to obtain a good deep drawability in a continuous annealing process, an effort was made to improve the grain growth rate. The coarsening of precipitates was one of the effective measures for improving the grain growth rate. This coarsening was realized for Al-killed steel sheets by a high temperature cooling which coarsened AlN. At the same time, the cementite was coarsely precipitated at grain boundaries, which helped to form a texture good for deep drawability.
High temperature coiling, however, has some problems, such as a deterioration of pickling efficiency or a decrease in formability in leading end and tail end parts of the coil. Therefore, high temperature coiling should be avoided. An effort was made to obtain mechanical properties by low temperature coiling which are as good as those obtained by high temperature coiling. Ushioda and his co-workers\(^{49}\) obtained coarse cementite and AlN by low temperature coiling through a low temperature slab heating below 1150°C using steels containing less than 0.02% C, less than 0.15% Mn and more than 0.05% Al. The main points of this technique are as follows: The precipitation rate of cementite is increased by reducing the amount of Mn which can be included in cementite and retard the precipitation, and a high amount of Al promotes a precipitation of coarse AlN. The low temperature slab heating promotes the formation of MnS which avoids the occurrence of the hot shortness caused by segregated S at edge parts of hot bands. This development is a typical example of how a new production technique can be developed by combining the various precipitation controls.

4.2. Deep Drawable IF Steel Sheets

To improve the formability of steel sheets, their C content has been gradually reduced during the last 50 years. Reducing the carbon content certainly improves ductility, but it does not always improve deep drawability. The reasons for this are that even a small amount of C in solution can detrimentally influence the formation of texture favorable for deep drawability, and the grain size of hot bands tends to become large as C decreases, which also detrimentally affects the formation of the final texture of products. That is, the smaller the grain size of hot bands, the better the deep drawability of cold rolled steel sheets. To assure the improvement of deep drawability, not only the reducing of C but also the scavenging of C and N in solution as a form of precipitates are important.\(^{49,50}\) In view of these requirements, IF steel sheets have been developed. An interstitial atom free state was realized by adding Ti and/or Nb, which form carbides or carbo-nitrides. As IF steels, Ti, Nb and both Ti and Nb bearing extralow-carbon steels were developed and their optimum chemical composition and processing were investigated.\(^{51-53}\) From the viewpoint of precipitation control, C and N in solution are scavenged by forming precipitates and these carbides or carbo-nitrides are coarsened so they do not hinder the grain growth of ferrite in the continuous annealing. The coarsening of precipitates can be realized at a high temperature cooling. But the high temperature cooling has some drawbacks mentioned in the previous section, such as the deterioration of pickling efficiency and the increase in fluctuation of mechanical properties.

For IF steel sheets, the same effort as that made for Al-killed steel sheets was made. Namely, an effort was made to have low temperature cooling achieve the same mechanical properties as those obtained by high temperature cooling. In the case of IF steel sheets, carbo-sulfides are used to obtain coarse precipitates in hot bands coiled at low temperatures. Figure 7 shows the mole fraction of precipitates at various temperatures in Ti and S bearing IF steel sheets calculated using Thermo-calc.\(^{54}\) If the S content is increased, C can be fixed in the form of coarse Ti\(_4\)C\(_2\)S\(_2\) during slab heating and the precipitation of fine TiC during hot rolling, cooling and coiling can be avoided and highly formable steel sheets produced even though a low temperature coiling is employed.\(^{54}\) For Ti and Nb bearing steel sheets, the similar precipitation control performed for Ti bearing steel sheets was carried out using (Ti, Nb)\(_4\)C\(_2\)S\(_2\).\(^{55}\)

Not directly related to texture control, a modification of the precipitation control mentioned above was applied to produce a highly formable IF-BH steel sheet. In this case, the precipitation of Ti\(_4\)C\(_2\)S\(_2\) is suppressed in hot bands by reducing S or increasing the Mn content or applying high temperature slab heating, and TiC is precipitated. Using lower solubility of TiC than Ti\(_4\)C\(_2\)S\(_2\), TiC can be dissolved at a high temperature annealing. A subsequent rapid cooling in a continuous annealing line enables C to remain in solution and a BH of more than 30 MPa to be achieved in IF steel sheets.\(^{54}\)

4.3. Grain Oriented Electrical Steel Sheets (GO)

The existence of precipitates in the final product of electrical steel sheets is harmful because they hinder the movement of magnetic walls and increase the core loss. Therefore, a so-called purification treatment was carried out at high temperature batch annealing to dissolve precipitates and get rid of impurity elements from the steel. However, precipitates are indispensable in the preceding processes for producing the final product to realize a required control of texture.

Precipitates mainly used for the texture control of GO are MnS, MnSe, AlN.\(^{56,57}\) Besides, CuS and BN can also be used. The role of precipitates in GO is twofold, firstly, to control the grain size and texture of the steel sheet after the primary recrystallization and secondarily, to enable a stable
occurrence of secondary recrystallisation and a resultant texture formation with high intensity of Goss texture favorable for the electro-magnetic properties. The grain size control in the primary recrystallization means the control of the driving force for the secondary recrystallization. The texture control in the primary recrystallization means letting Goss grains remain in a sufficient amount and intensifying and sharpening the coincidence relationship between a Goss orientation and main orientations. A secondary recrystallization microstructure having a sharp Goss orientation can be obtained if the grain boundary between a Goss grain and the matrix moves at a significantly higher rate than the other grain boundaries. Such a boundary was proved to be a coincidence boundary with Σ9 relationship. On the other hand, it was reported that such a boundary need not be a coincidence boundary, but it should have a misorientation angle of 20 to 45°.

To produce grain oriented electrical steel sheets with a sharp Goss orientation, the grain size of primary recrystallization, the amount, size and kind of inhibitors, annealing conditions etc. have to be optimized. Because of numerous influential factors, this optimization has been carried out based on stored knowledge. This means the know-how plays a major role in the production of GO. In the recent years, however, some simulation models for predicting the secondary recrystallization behavior have been developed and it is foreseen that a new approach based on the model calculations will set up the optimum conditions for producing high quality GO.

In the conventional production of GO, precipitates formed in hot rolling processes have to play two roles simultaneously, to control microstructure and texture of primary recrystallization and those of the secondary recrystallization, respectively. In recent years, inhibitors controlling the secondary recrystallization have been additionally introduced by nitrizing after primary recrystallization in the continuous annealing process. This additional process made possible the independent optimization of precipitation control for performing both of the above roles and lead to a significant improvement of the production technology.

The precipitation behavior in GO has not been clarified thoroughly. For example, the modeling of the precipitation behavior in the γ/α dual phase in slab heating and in hot rolling process and that during and after nitrizing remain to be achieved in the future.

4.4. Non-oriented Electrical Steel Sheets (NO)

Non-oriented electrical steel sheets are used for rotated electrical parts like motor cores and their required properties are high magnetic flux density, low core loss and low planar isotropy of magnetic properties. To obtain these properties, the texture is controlled to have {100}uvw as main orientations. From the viewpoint of core loss, there is an optimum grain size, which is around 150 μm. This grain size is usually achieved after the stress release annealing. To obtain this relatively coarse microstructure, the precipitation of fine precipitates should be avoided. For this purpose, the elements forming carbides or carbonitrides such as Ti, Nb, Zr, V etc. are reduced as much as possible. Concerning AlN, there are two measures for avoiding the precipitation of fine precipitates; the first one is to coarsen AlN by increasing Al content and the second one is to reduce the amount of AlN by reducing Al content. In the latter case, N harmful to magnetic properties should be scavenged by forming BN which precipitate relatively coarsely. Concerning MnS, there are three ways to avoid hindering the grain growth. The first one is to reduce S content as low as possible. The second one is to form large MnS by increasing Mn content or by lowering the slab heating temperature. The third one is to add elements such as Ca, REM etc. forming sulfides at a higher temperature than that of MnS. If the precipitation of fine precipitates is suppressed by these measures, the grain size of hot bands is generally increased. This is favorable for NO because the larger grain size of hot bands provides a lower intensity of ND//(111) and a higher intensity of ND//(100) in the recrystallization texture of the product.

5. Over-aging Treatment

If C and/or N in solution exceed several ppm in the steel sheet, a surface defect called stretcher strain (SS) occurs after press forming. To avoid the occurrence of SS, C and/or N in solution are reduced by forming carbides, nitrides or carbo-nitrides. In the box annealing, the cooling rate is low enough for C and/or N to complete the necessary precipitation. To complete this precipitation in a continuous annealing process, some measures must be taken. First of all, the over-aging process is conducted in which the steel sheet is held at a temperature of around 400°C for several minutes after recrystallization treatment. If the C content is decreased to obtain softer steel sheets, the precipitation rate is reduced and as a consequence, the necessary over-aging time is increased and productivity decreases.

To increase the precipitation rate, the nucleation of cementite is promoted by a proper control of precipitation. Figure 8 shows the efficient nucleation sites of cementite. AlN has not proved to be an efficient nucleation site and MnS whose size is less than 0.05 μm does not seem to be a promising nucleation site. These facts indicate that MnS in a proper size promotes the ingrain precipitation of cementite and accelerates the over-aging treatment. It was also reported that in a Ti and B bearing steel sheet, multiple precipitates of TiN, MnS, BN, Fe23(C, B)6 can act as effective nucleation sites of cementite.

Further, the precipitation of cementite can be promoted if the over-aging process is optimized. Figure 9 shows a new type of over-aging process called the reheated over-aging (R-OA). This over-aging process consists of a rapid cooling down to around 250–300°C after a recrystallization treatment and a subsequent rapid reheating up to around 350°C and then a gradual cooling to around 300°C. The initial rapid cooling enhances the driving force for precipitation and promotes the nucleation of cementite and the increase in temperature promotes the diffusion of C and accelerates the growth of cementite. The gradual cooling decreases the equilibrium C content in solution and contributes to the reduction of solute carbon. The optimum R-OA pattern was investigated by means of a mathematical model for predicting the cementite precipitation. Koyama and his co-workers developed a model for predicting the cementite precipitation at grain boundaries and in inside grain
simultaneously, and thus contributed to the development of a new continuous annealing line.69,70)

6. Precipitation Models

Numerous precipitation models have already been developed in the steel industry, most of these models being based on the classical nucleation theory.71) The growth of precipitates is usually calculated by means of diffusion equations under the Zener's assumption.71) Akamatsu and his co-workers introduced a model for predicting the precipitation behavior of NbC in austenite, in which model the para-equilibrium condition for the calculation of the concentration of the precipitate forming elements at the interface between the precipitate and matrix is assumed.72) The feature of this model is that the promotion of precipitation due to deformation or the influence of recrystallization on the precipitation behavior can be determined by assuming that the nucleation site is a function of dislocations caused by deformation and annihilated by recovery and recrystallization. Akamatsu and his co-workers improved their model by introducing the partition local equilibrium condition for the calculation of the concentration of the precipitate forming elements at the interface between the precipitate and matrix.73) Figure 10 shows the concentration of Nb (M) and C at the interface between the precipitate and matrix. To determine the Nb concentration at the interface in accordance with the assumption that the partition local equilibrium condition had been introduced, the concentration of C was assumed to follow its isoactivity line.

Okamoto and Suehiro74) further improved the Akamatsu's model in the way that the free energy increase due to the curvature of the precipitate radius was considered in thermodynamic calculations and the interfacial energy used in the model was assumed to be described as a function of the radius of precipitates from the viewpoint of coherency. Because the interfacial energy is influenced by the size of precipitates and the amount of segregated elements, this energy is difficult to determine theoretically and is thus usually used as a fitting parameter of the model. Okamoto and Suehiro's model is connected with the Thermocalc75) and is applicable for the calculation of the precipitation behavior of various precipitates beside NbC. Using this model, the size, number and amount of precipitates were calculated from the nucleation through the Ostwald ripening. A similar model was independently developed by Fujita and Bhadeshia and was used to discuss the precipitation behavior of Mo2C.76) A simulation on the mathematical model is also useful for predicting creep behavior, but in many cases several precipitates nucleate and grow simultaneously and some precipitates transform during creep processes, thus the further development of more sophisticated models is required.

Most models have been developed for predicting the precipitation behavior in austenite but there are also some models which have already been developed for predicting the precipitation behavior in ferrite77) and in the γ/α dual phase region.78)
Besides the models for predicting the precipitation behavior in slab heating and hot rolling processes, mathematical models have been developed for simulating the precipitation behavior in continuous annealing processes. The model for predicting the precipitation behavior of cementite has already been introduced in Sec. 5. A model for predicting the dissolution behavior of cementite during heating in a continuous annealing process has been developed. The influence of the size of cementite and the heating rate on the dissolution of cementite was quantitatively discussed based on the model. The prediction of the amount of carbon in solution during recrystallization in the continuous annealing process is important because C in solution forming C–Mn dipoles may influence the formation of annealing texture, and as a consequence, the deep drawability of the product.

Concerning AlN, a model for predicting the dissolution and precipitation in austenite and the ferrite region was developed. There is also another model for predicting the dissolution and precipitation of TiC and NbC in continuous annealing processes with the aim of controlling BH.

The mathematical models simulating the grain growth and the secondary recrystallization closely related with precipitates are not introduced here because of space limitations. Instead, reference should be made to the review papers.

In all precipitation models introduced here, the nucleation of precipitates is assumed to follow the classical nucleation theory. However, showed in the observation of the initial stage of precipitation in an Al-alloy on a three-dimensional atom probe that the nucleation process is quite different from that described in the classical nucleation theory. At the initial stage of nucleation, precipitate forming elements form clusters containing Al with a non-equilibrium composition. The clusters change their composition, while growing with time, to an equilibrium condition. It is not probable that the same nucleation process occurs in steels but it is required to clarify the real nucleation phenomena by an atom-level analysis, for example, using a three-dimensional atom probe. The result of these nanometer-level observations can lead to a development of new models.

7. Interaction with Recrystallization and Transformation

The precipitates can both suppress and accelerate recrystallization depending on each case. In an example of such suppression, precipitates pin the dislocations and grain or subgrain boundaries and retard recovery, recrystallization and transformation. Figure 11 shows an example of this kind of suppression where the recrystallization is significantly retarded if precipitation begins to occur.

In an example of said above acceleration, precipitates provide preferential nucleation sites for recrystallization and transformation. As shown in Sec. 3.2, certain precipitates act as preferential nucleation sites for transformation. Concerning recrystallization, a large precipitate can be a potential nucleation site. If the solute dragging effect overwhelms the pinning effect, the reduction of the amount of the solute atoms by forming precipitates also accelerates recrystallization and transformation. Figure 12 shows an example where transformation is promoted by the precipita-
further development of precipitation control. \(^9\) On the other hand, recrystallization influences the precipitation behavior. If recrystallization occurs, the number of dislocations often acting as nucleation sites for precipitation decrease and thus precipitation is retarded. A quantitative analysis on the influence of deformation on the precipitation behavior was carried out by Akamatsu and his co-workers by means of a mathematical model. \(^7\)

To make a model for predicting the precipitation, recrystallization and transformation behavior in consideration of their interactions with each other, the Johnson–Mehl–Avrami type equations being often used because of their simplicity are not suited. Instead, equations which can describe the effect of solute dragging and pinning on the nucleation and growth process mathematically are required. Although it will certainly be difficult to develop such equations, it will be worth the effort.

8. Closing Remarks

In this paper, the recent industrial research on the precipitation control in steel sheets was reviewed. This field is too broad to allow a review to cover the whole field in limited pages. Therefore, this review has mainly covered the field related to the author's work. Nevertheless, the author hopes that the reader of this review has understood the present status of and future prospects for the precipitation research in the industry in some measure.

A research workshop on precipitation control organized by ISIJ published a proceedings of a seminar reviewing fundamental research made on precipitation control. \(^5\) The proceedings showed that there is still quite a lack of fundamental knowledge on precipitation phenomena. Using recently developed sophisticated analytical devices such as a three-dimensional atom probe, the fundamental knowledge on precipitation phenomena should be deepened. Also, close cooperative among universities, national research laboratories and the steel industry will surely contribute to the further development of precipitation control.

REFERENCES

1) E. Hornbogen and R. C. Glenn: Trans. Metall. Soc. AIME, 218 (1966), 1064.
2) Y. Meyzaud and P. Parniere: Mater. Sci. Eng., 29 (1977), 41.
3) K. Kishida, M. Oda, N. Ikenaga, M. Takita and O. Akisue: SAE Technical Paper Series 900735, Detroit, Michigan, USA, (1990).
4) N. Maruyama, M. Sugiyama, T. Harai and H. Tamehiro: Mater. Trans. JIM, 40 (1999), 268.
5) Y. Okitu, T. Ukena and M. Oda: Tetsu-to-Hagané, 81 (1995), 739.
6) S. R. Goodman, S. S. Brenner and J. R. Low, Jr.: Metall. Trans., 4 (1973), 2363.
7) P. J. Othen, M. J. Jenkins and G. D. W. Smith: Philos. Mag. A, 70 (1994), 1.
8) H. A. Hardouin Duparc, R. C. Doole, M. L. Jenkins and A. Barbu: Philos. Mag. Lett., 71 (1995), 325.
9) G. M. Worrall, J. T. Buswell, C. A. English, M. G. Hetherington and G. D. W. Smith: J. Nucl. Mater., 148 (1987), 107.
10) Y. N. Ossetsky and A. Serra: Philos. Mag. A, 75 (1997), 1097.
11) M. Hua, C. I. Garcia and A. J. DeAredo: Metall. Trans. A: 28A (1997), 1769.
12) Y. Yazawa, T. Furuhara and M. Maki: CAMP-ISIJ, 11 (1998), 1127.
13) M. Morita, N. Nakaosawa, S. Masui, T. Kato, T. Higashino and N. Aoyagi: CAMP-ISIJ, 5 (1992), 1863.
14) M. Morita, T. Shizumi, O. Furukimi, N. Aoyagi and T. Kato: Mater. Jpn., 37 (1998), 513.
15) T. Urabe, T. Fujita, K. Nakajima, F. Kitano and Y. Hosoya: Int. Conf. on the Processing, Microstructure and Properties of IF Steels, ISS, Warrendale, PA, (2000), 215.
16) K. Kusumi, T. Senuma, M. Suehiro, M. Sugiyama and M. Matsuo: Tetsu-to-Hagané, 86 (2000), 682.
17) E. Yajima and K. Furusawa: Trans. JIM, 26 (1962), 141.
18) X. F. Hu, Q. L. Ge and Z. L. Wu: Acta Metall. Mater., 41 (1993), 1625.
19) S. Suzuki and K. Naito: Tetsu-to-Hagané, 81 (1995), 655.
20) H. Tanabe, K. Anai, M. Yamazaki and T. Kiyotomo: Mater. Jpn., 38 (1999), 242.
21) M. Fujisaka, Y. Abe and Y. Hagiwara: CAMP-ISIJ, 13 (2000), 1136.
22) M. H. Hong, W. T. Reynolds, T. Tarui and K. Hono: Metall. Mater. Trans. A, 30A (1999), 717.
23) T. Tarui, N. Maruyama and H. Tashiro: CAMP-ISIJ, 12 (1999), 460.
24) M. Ueda, K. Uchino and T. Senuma: Tetsu-to-Hagané, 87 (2001), 190.
25) J. Takamura and S. Mizoguchi: Proc. 6th. Int. Iron and Steel Cong., Vol. 1, ISIJ, Tokyo, (1999), 591.
26) M. Wako: Materia Jpn., 35 (1996), 1311.
27) T. Ohashi, H. Tamehiro and M. Takahashi: Materia Jpn., 36 (1997), 159.
28) K. Funakoshi, T. Tanaka, S. Ueda, M. Ishikawa, N. Koshizuka and K. Kobayashi: Tetsu-to-Hagané, 63 (1977), 303.
29) N. Mori, H. Homma, S. Ohkita and M. Wakabayashi: IIW Doc. IX–6 (1987), 1196–881, (1981).
30) K. Yamamoto, T. Hasegawa and J. Takamura: Tetsu-to-Hagané, 79 (1993), 1169.
31) F. Ishikawa, T. Takahashi and T. Ochi: Metall. Trans. A, 25A (1994), 929.
32) K. Yamamoto, T. Hasegawa and J. Takamura: ISIJ Int., 36 (1996), 80.
33) Y. Morikage, K. Ohi, F. Kawabata and K. Amano: Tetsu-to-Hagané, 84 (1998), 510.
34) S. Akamatsu, T. Senuma and H. Yada: CAMP-ISIJ, 2 (1989), 1933.
35) Y. Teshima, H. Yuyama, S. Mizoguchi and H. Kajoka: Metall. Trans. B, 17B (1986), 845.
36) G. Shigesato, M. Sugiyama, S. Aihara, R. Uemori and Y. Tomita: Tetsu-to-Hagané, 87 (2001), 93.
37) M. Ferrante and K. Akune: Conf. Solid-Solid Phase Transformations, TMS-AIME, Warrendale, PA, (1988), 193.
38) J. M. Gregg and H. K. D. H. Bhadeshia: Acta Metall. Mater., 42 (1994), 3321.
39) M. Wako, T. Sawai and S. Mizoguchi: Tetsu-to-Hagané, 78 (1992), 1697.
40) J. Yamaguchi, N. Takemura, T. Furuhara, T. Maki and R. Uemori: CAMP-ISIJ, 11 (1998), 1128.
41) Y. Ishii, S. Takeo, C. Okajama and K. Tashiro: Tetsu-to-Hagané, 66 (1980), 710.
42) H. Fujimura, S. Tsuge, Y. Komizo and T. Nishizawa: CAMP-ISIJ, 14 (2001), 416.
43) N. Yoshinaga, K. Ushioda, S. Akamatsu and O. Akisue: ISIJ Int., 34 (1994), 24.

Fig. 12. Effect of precipitation treatment on the ferrite transformation behavior of 0.03% Nb containing steel (A, B, C and D) and C–Mn steel (E). \(^{50}\)
44) K. Kimura, T. Senama, A. Yamamoto, M. Abe and T. Takeshita: CAMP-ISIJ 10 (1997), 1204.
45) S. Tsuge, H. Fujimura and A. Yamagishi: Sumitomo Met., 49 (1997), 65.
46) K. Kimura, M. Tendou, A. Yamamoto, M. Abe and T. Senama: CAMP-ISIJ 11 (1998), 1144.
47) N. Takahashi: Proc. of JIM in Fall Meeting, (1965), 51.
48) K. Ushioda, K. Koyama and M. Takahashi: Tetsu-to-Hagané, 76 (1990), 1536.
49) A. Okamoto: Tetsu-to-Hagané, 70 (1984), 1906.
50) H. Abe: Proc. ICOTOM-8, ed. by I. S. Kallend and G. Gottstein, TMS, Warrendale, PA, (1988), 661.
51) N. Fukuda and M. Shimizu: Tetsu-to-Hagané, 61 (1975), 817.
52) S. Satoh, T. Obara, M. Nishida and T. Irie: Trans. Iron Steel Inst. Jpn., 24 (1984), 838.
53) Y. Tokunaga, M. Yamada and K. Itoh: Tetsu-to-Hagané, 73 (1987), 341.
54) K. Kawasaki, S. Sanagi, T. Senuma, S. Akamatsu, N. Yoshinaga and O. Aksue: Tetsu-to-Hagané, 79 (1993), 76.
55) N. Hashimoto, N. Yoshinaga and M. Suehiro: CAMP-ISIJ, 9 (1996), 1346.
56) S. Taguchi, T. Yamamoto, A. Sakakura, T. Wada, F. Matsumoto and K. Ueno: Seitetsu Kenkyu, 276 (1972), 135.
57) M. Koizumi, T. Kikuchi and S. Bandou: Tetsu-to-Hagané, 66 (1980), 1351.
58) T. Nakayama and Y. Ushigami: Proc. 7th RISO Int. Symp. Met. Mat. Sci., Roskilde, (1986), 463.
59) Y. Hayakawa and J. A. Szpunar: Acta Mater., 46 (1998), 1063.
60) G. Abbruzzese and K. Luecke: Proc. 7th RISO Int. Symp. Met. Mat. Sci., Roskilde, (1986), 1.
61) D. J. Srolovitz, M. Anderson, P. S. Sahnu and G. S. Grest: Acta Metall., 35 (1987), 2233.
62) N. Takahashi and H. Harase: Mater. Sci. Forum, 204–206 (1995), 143.
63) H. Shimanaka, Y. Ito, T. Irie, K. Matsumura, H. Nakamura and Y. Shono: Energy Efficient Electrical Steel, TMS-AIME, Warrendale, PA, (1980), 193.
64) Y. Iida, T. Nozaki and F. Sudo: Scaninject 3 Part 2, (1983), 49.
65) K. Koyama: Dissertation given at Kyoto University, (1987).
66) K. Kawasaki and H. Katoh: Trans. JIM, 49 (1985), 928.
67) K. Osawa, C. Maeda, T. Tsunashika, T. Sakata, O. Furukami and M. Morita: CAMP-ISIJ, 12 (1999), 431.
68) T. Hayashida, M. Oda, T. Yamada, Y. Matsukawa and J. Tanaka: Proc. Symp. On High-Strength Sheet Steels for the Automotive Industry, ed. By R. Pradhan, ISS, Warrendale, PA, (1994), 135.
69) K. Koyama, Y. Kuroda, H. Kato and M. Nagumo: Tetsu-to-Hagané, 71 (1985), 1497.
70) K. Koyama, H. Kato and M. Nagumo: Tetsu-to-Hagané, 72 (1986), 823.
71) R. Becker and W. Doering: Ann. Phys., 24 (1935), 719.
72) S. Akamatsu, Y. Matsumura, T. Senuma, H. Yada and S. Ishikawa: Tetsu-to-Hagané, 75 (1989), 933.
73) S. Akamatsu, T. Senuma and M. Hasebe: Tetsu-to-Hagané, 78 (1992), 102.
74) C. Okamoto and M. Suehiro: Tetsu-to-Hagané, 84 (1998), 650.
75) B. Sundman, B. Jansson and J.-O. Andersson: Calphad, 9 (1985), 153.
76) N. Fujita and H. K. Bhadeshia: Mater. Sci. Technol., 15 (1999), 627.
77) C. Okamoto, M. Suehiro and K. Kishida: CAMP-ISIJ, 10 (1997), 453.
78) K. Nakamura and H. Okaguchi: CAMP-ISIJ, 11 (1998), 577.
79) J. Agren and G. P. Vassilev: Mater. Sci. Eng., 64 (1984), 95.
80) K. Ushioda, W. B. Hutchinson, J. Agren and U. von Svhlippenbach: Mater. Sci. Technol., 2 (1986), 807.
81) H. Abe, T. Suzuki and S. Okada: Trans. JIM, 25 (1984), 215.
82) L. M. Cheng, E. B. Hawbolt and T. R. Meadowcroft: Metall. Mater. Trans. A, 31A (2000), 1907.
83) M. Kameda and T. Senuma: CAMP-ISIJ, 8 (1995), 777.
84) G. Abbruzzese: Acta Metall., 33 (1986), 905.
85) Y. Ushigami, K. Murakami and T. Kubota: Proc. 3rd Int. Conf. on Grain Growth, Pittsburgh, (1998), 491.
86) Y. Hayakawa and J. A. Szpunar: Acta Mater., 45 (1997), 4713.
87) K. Hono: Acta Mater., 47 (1999), 3127.
88) A. Lévêque, V. Rofes-Vernis and C. Rossard: Met. Sci., 9 (1975), 36.
89) I. Weiss and J. J. Jonas: Metall. Mater. Trans. A, 10A (1979), 831.
90) S. S. Hansen, J. B. Vander Sande and M. Cohen: Metall. Trans. A, 11A (1980), 389.
91) O. Kwon and A. J. DeArdo: Acta Metall. Mater., 39 (1991), 529.
92) W. P. Sun, M. Müllitzer, D. Q. Bai and J. J. Jonas: Acta Metall., 41 (1993), 3595.
93) F. J. Humphreys: Acta Metall., 25 (1977), 1323.
94) A. Yoshie, M. Fujioka, H. Morikawa and Y. Onoe: Proc. Int. Conf. On Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals, ISIJ, Tokyo, (1988), 799.
95) Proc. of Seminar on Precipitation Control: ISIJ, Tokyo, (2001).