Phase Transformation from Fine-grained Austenite

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Microstructure formed by diffusional or martensitic transformation from fine-grained austenite of which grain size is smaller than 5 μm was studied. Grain refinement of austenite was established through two kinds of reversion processes; (1) cyclic transformation between martensite and austenite and (2) reverse transformation from tempered and cold-rolled lath martensite (or pearlite). In the process of (1), the fine austenite structures whose grain sizes of 5–10 μm are obtained. Refinement of austenite grain size results in the increase of hardness. In the process of (2), austenite grain size can be refined down to about 2 μm in low-carbon Mn steels by microalloying through pinning of austenite grain growth by alloy carbides. The ferrite grain size after continuous cooling transformation becomes finer as austenite grain size is refined. However, the grain size ratio of austenite and ferrite, d/doc, increases by refining austenite grain size. For the austenite of grain size smaller than 5 μm, the ferrite grain size becomes coarser than that of austenite for slow cooling. A similar trend in the change of ferrite grain size by refinement of austenite was recognized for isothermal pearlite transformation in eutectoid alloys. Thus, it is suggested that extensive accelerated cooling is important to obtain fine-grained ferrite by diffusional transformations from the fine-grained austenite. Packet and block sizes of lath martensite in low carbon steels are also refined by decreasing the austenite grain size. Several packets and blocks are formed even from the austenite matrix of 2 μm in grain size.

KEY WORDS: phase transformation; austenite; ferrite; pearlite; martensite; grain refinement.

1. Introduction

It is well known that grain refinement is the most effective way to improve strength and toughness. In the conventional processes such as controlled rolling, recrystallization during hot deformation is used to refine austenite grain size. The minimum grain size of austenite achieved by this process is about 20 μm in C–Mn steels.1) By further controlled rolling in unrecrystallized austenite region and accelerated cooling, fine ferrite structure of which grain size is about 5 μm has been produced in low carbon steels. On the other hand, it was recently shown that several thermo-mechanical processing routes successfully produce low-carbon steels consisting of ultrafine-grained ferrite structure whose grain size is finer than 1 μm in a bulk form Ref. 3). Those studies are mostly exploring the limit in conventional controlled rolling processes, i.e., lower deformation temperature and heavier deformation of metastable austenite before or during ferrite transformation. Since application of such concepts to the commercial steel production line still accompanies large difficulty, some modification such as short-interval multiple rolling has been attempted.4)

It is known that ferrite grain size decreases linearly with decrease in austenite grain size in continuously cooling of recrystallized hot rolled sheets for a wide range of austenite grain size between 20 to 250 μm.5) Therefore, another route proposed to produce ultrafine grained ferrite structure is transformation from ultrafine grained austenite structure. Although conventional route using static recrystallization of austenite does not normally produce austenite grain sizes under 10 μm, dynamic recrystallization could be effective to achieve further refinement. Samuel et al.6) studied dynamic recrystallization of austenite during strip rolling of Nb-contained low-carbon low-alloy steels and obtained austenite grain sizes just below 5 μm. During cooling after the rolling, a fine ferrite structure with a grain size of 3 μm was obtained. Later Salvatori et al.7) obtained ultrafine grained austenite of which grain size is 2–5 μm through dynamic recrystallization by applying large strain up to about 3 in an SUS304 austenitic stainless steel. It should be mentioned that even in steels processed by TMCP, refinement of austenite contributes to reduction of applied strain in controlled rolling in unrecrystallized regions.8)

Refrinement of austenite grain size smaller than 5 μm was previously studied for two kinds of reversion processes in low-carbon steels, i.e., (1) cyclic transformation between martensite and austenite, so called thermal cycling9) and (2) reversion from tempered and cold-rolled lath martens-
In both processes, fine-grained austenite structures with grain sizes of 1–2 μm were obtained. However, so far, phase transformations in fine-grained austenite of which grain size is in this range are not well understood. Thus, in the present paper, the formation of fine-grained austenite structures in the abovementioned two routes is described and subsequently, ferrite, pearlite and martensite structure formed from the fine-grained austenite are discussed.

2. Grain Refinement of Austenite Matrix

2.1. Thermal Cycling

Grange\textsuperscript{8}) first applied thermal cycling of rapid heating and quenching to refine austenite grain size mainly in a medium carbon Ni–Cr–Mo steel (AISI8640). He successfully obtained ultrafine-grained austenite structure of which grain size was about 3 μm (13.5 in ASTM grain size number) and later he applied this technique to a variety of steels.\textsuperscript{13}) Porter and Darkowski\textsuperscript{14}) studied systematically effects of various parameters in thermal cycling, e.g. heating rate and number of cycles on austenite grain refinement.

In this section, we discuss microstructure change during thermal cycling. An alloy steel for machine structural use (SCM435; Fe–0.35C–1.05Cr–0.17Mo in mass%) was used to study the effect of thermal cycle conditions on austenite grain sizes. The specimens of lath martensite structure were obtained by quenching after austenitizing at 1 123 K for 1.8 ks. They were re-austenitized at 1 073 K or 1 123 K for various periods and quenched. This austenitizing and quenching cycle was repeated up to 6 times. Prior austenite (hereafter γ) grain size in the quenched specimen after each cycle was measured by an intersection analysis as a nominal grain size described as follows:

\[ d_\gamma = 1.128\bar{l} \]  

where \( \bar{l} \) is the average intercept length on the γ grain.

Figure 1 shows optical microstructures of the specimens processed by thermal cycling. The γ grain size obtained after the first austenitizing treatment is 24 μm (Fig. 1(a)). After each cycle, γ grain size becomes finer as seen in Figs. 1(b)–1(d). Figure 2(a) shows variation of the γ grain size with the number of thermal cycles. The γ grain size decreases drastically after the 1st cycle but remains nearly constant after the 3rd cycle. Such stabilization of γ grain size is similar to the behaviour previously reported.\textsuperscript{9,13,14) The shorter the austenitizing time at 1 123 K is, the smaller γ grain size obtained is. The γ grain size of the specimen after 6 cycles at 1 123 K for 300 s reaches to 4.0 μm. Vickers hardness of martensite increases by nearly 100HV with the decrease of γ grain size after repeated thermal cycles (Fig. 2(b)) presumably because the effective grain sizes (packet and block sizes) for lath martensite are refined as well as internal stress developed by thermal cycling. In Fig. 2(c), it is clear that a constant grain size is obtained in specimens of different initial grain sizes by employing the same thermal cycles.

It is found that new γ nucleates mostly at prior γ grain boundaries. There is also nucleation at martensite packet and block boundaries. The observed γ morphology was of a globular type as previously reported\textsuperscript{15,16) although acicular γ formed at low-angle boundaries between martensite laths was not seen. During those thermal cycles, cementite (hereafter θ) precipitates by tempering and remain undissolved in γ for some period. When martensite was tempered at 923 K for 1.8 ks after each cycle, γ grain size is slightly smaller than that without tempering because γ grain growth is retarded due to pinning by θ particles.
2.2. Reversion of Deformed Martensite

Tokizane et al.\(^2\) found that cold rolling of tempered martensite promotes intragranular nucleation of γ with recrystallization of lath martensite during reversion. Based upon this result, grain refinement of γ through reversion of tempered and cold-rolled martensite was attempted.\(^{10-12}\)

The minimum grain size obtained in this processing route was 5 μm in low carbon steels\(^10\) and 0.9 μm in medium carbon low alloy steel with modification in Mn and Mo content.\(^{11,12}\)

We applied this reversion process to low carbon low alloy steels as described in more detail elsewhere.\(^2\) Two Fe-0.2C-1.5Mn-0.2Si (mass%) steels (Steel A and B) of which chemical compositions were used for this processing. Steel A is a simple Fe-C-Mn-Si alloy whereas small amounts of vanadium (V: 0.15 mass%) and nitrogen (N: 0.02 mass%) were added in Steel B in order to obtain further grain refinement by pinning γ grain growth with V(C,N) precipitation. After austenitizing at 1473 K and quenching to obtain fully lath martensite structures, specimens were tempered at 923 K for 3.6 ks; (a) Steel A, reversion at 1103 K for 120 s→WQ, \(d_γ=10.3\) μm, (b) Steel A, 80% cold rolling→reversion at 1103 K for 60 s→WQ, \(d_γ=6.3\) μm, (c) Steel B, reversion at 1103 K for 120 s→WQ, \(d_γ=4.8\) μm, (b) Steel B, 80% cold rolling→reversion at 1103 K for 60 s→WQ, \(d_γ=2.3\) μm. Dashed lines outline prior γ grain boundaries.

Figure 3 shows schematically the microstructure change during thermal cycles. New γ is formed at high-angle boundaries in the lath martensite structure. When the γ grain size is large, packet and block boundaries within a prior γ grain act as nucleation sites of new γ precipitates. As the γ grain size decreases, γ tends to nucleate only at prior γ grain boundaries and eventually γ grain size saturates by achieving the balance between reversion by nucleation and grain growth after completion of reversion as already pointed out before.\(^13\) The saturated γ grain size becomes coarser as the austenitizing is performed at a higher temperature or for a longer period. Finer γ grain size achieved for faster heating rate previously reported\(^14\) can be explained in terms of suppression of grain growth during heating as well as more frequent nucleation.

Utilization of thermal cycling was made by several researchers to improve strength–toughness balance of steels. Grange\(^9\) studied various transformed microstructures from fine-grained γ matrices produced by thermal cycling. He clearly demonstrated that both of ferrite+pearlite and tempered martensite structures were refined by γ grain refinement which led to monotonous increase in yield strength and more complex changes in tensile strength and elongation.\(^9\) Although he did not report resultant ferrite/pearlite or martensite grain sizes quantitatively, later Karlsson\(^6\) showed that thermal cycling produces fine ferrite/pearlite grains down to about 5 μm in size in Fe-C hypo-eutectoid alloys. For tempered martensite of Ni-Cr-Mo steels, the balance of absorption energy and yield strength by thermal cycling was reported.\(^14,18\) Although its application to low-carbon steels results in γ grain size about 10 μm,\(^19\) combination with intercritical annealing resulted in formation of (ferrite+martensite) microduplex structures with better tensile properties in low-carbon steels.\(^20\)
whereas the grain size is nearly 10 μm without tempering. The γ grain sizes of both alloys can be refined further by employing cold-rolling before reversion. Figure 5(b) describes relationship between γ grain size and the reduction of cold rolling. As rolling reduction increases up to 50%, γ grain size decreases. TEM observation revealed fine V(C, N) precipitate of which diameter is about 20–30 nm is present at prior γ grain boundary as well as in the matrix.22) Those V(C, N) precipitates are considered to act as pinning particle, resulting in the fine-grained γ structure obtained in Steel B. Addition of V is effective to obtain ultrafine γ structure in eutectoid alloys as briefly reported previously.23) To study isothermal pearlite transformation from fine-grained γ matrix as described in the next section, two eutectoid carbon steels (Fe–0.8C–0.5Mn–0.2Si (mass%) with or without 0.3 mass% V addition) were used. Hereafter, we denote those alloys as V-free and V-added eutectoid alloys. Refining of γ grain size was attempted by 70% cold rolling of martensite (tempered at 923 K) or pearlite (transformed at 873 K) followed by reversion to austenite at 1 043 K. Figure 6 shows the relationship between γ grain size of eutectoid steels with and austenitizing time during reversion. For the V-free eutectoid alloy, γ grain size is reduced to about 13 μm by cold rolling of pearlite and 8 μm by cold rolling of martensite, respectively. For the V-added eutectoid alloy, the resultant γ grain size is much finer than that of the V-free alloy. It is seen that γ grain growth is significantly suppressed by the V addition. The finest γ grain size obtained is 2–3 μm for the initial structure of either pearlite or martensite before reversion, which also contain undissolved θ particles.

3. Phase Transformations from Fine-grained Austenite Matrices

3.1. Proeutectoid Ferrite Transformation

We examined transformed α grain sizes in the Fe–0.2C–1.5Mn–0.2Si hypoeutectoid steels (Steels A and B described in the previous section) austenitized after tempering of martensite as described above and cooled at various cooling rate between 1–20 K/s.22) Figure 6 shows microstructure transformed by continuous cooling from γ matrices of different grain sizes. α grain sizes were measured in the same manner as the γ grain size by using Eq. (1) by means of optical and scanning electron microscopy.

Proeutectoid ferrite transformation starts from γ grain boundaries. When coarse-grained γ was transformed by cooling at 1 K/s, grain boundary α was initially formed. However, Widmanstätten α plate developed during further cooling and then pearlite transformation proceeded until completion of transformation. In fine-grained γ specimens, transformed microstructure only consists of polygonal ferrite and pearlite as seen in the SEM micrographs of Figs. 7(a) to 7(d). In the specimen transformed from the γ matrix of which grain size is 19.2 μm at cooling rate of 1 K/s, the resultant α grain size is 10.0 μm which is finer than the γ matrix (Fig. 7(a)). On the other hand, α grain size transformed from γ of which grain size is 2.3 μm at cooling rate of 1 K/s is 4.2 μm (Fig. 7(c)). This is actually coarser than the grain size of its γ matrix. This kind of a overgrowth occurs because hardenability is decreased by γ grain refinement and thus, at a given cooling rate, ferrite transformation takes place at a higher temperature as Grange pointed out.3) So we applied accelerated cooling which is known to effectively lower transformation temperature and refine α grain sizes in TMCP. Figure 7(d) shows SEM microstructure of the specimen with γ grain size of 2.3 μm cooled by 20 K/s. α grain size is refined to 2.2 μm which is nearly the same as the γ matrix by increasing the cooling rate.

We examined theoretically the observed change in α
where following equations considering nucleation and diffusional growth of $\alpha$ on $\gamma$ grain boundary faces:

$$d_a = \left( \frac{2}{3n_sS_{gb}} \right)^{1/3} \quad \text{(2)}$$

where $d_a$ is $\alpha$ grain size, $n_s$ is number of $\alpha$ grain formed on $\gamma$ grain boundary per unit area and $S_{gb}$ is area of $\gamma$ grain boundary per unit volume. $S_{gb}$ and $n_s$ are expressed by the following equations

$$S_{gb} = \frac{4}{\sqrt{\pi d_{\gamma}}} \quad \text{(3)}$$

where $d_{\gamma}$ is $\gamma$ grain size.

$$n_s = \sqrt{\frac{T}{\tau}} \left[ \frac{I_s(T')}{Q(T')} \right] \times \exp \left[ -9\pi \int_{T_{\alpha}}^{T} \frac{L^2(T''')}{Q(T''')}dT'' \right] \frac{dT}{T} \quad \text{(4)}$$

where $I_s$ is nucleation rate of $\alpha$ phase per unit area of $\gamma$ grain boundary, $Q$ is cooling rate, $L$ is parabolic rate constant calculated using a constant carbon concentration gradient model and $T_{\alpha}$ is $A_{c1}$ temperature in paraequilibrium. Nucleation rate of $\alpha$ at a $\gamma$ grain boundary was calculated by classical nucleation theory which assumes pearlite transformation and double spherical cap model as a critical nucleus. We fix the aspect ratio of the nucleus, which is equal to the maximum thickness/diameter of nucleus on $\gamma$ grain boundary plane, to be 1/3 for simplicity. Then $\alpha/\gamma$ interphase boundary energy and nucleation site density on $\gamma$ grain boundary were taken to be 0.07 J/m$^2$ and 7.8 x 10$^{17}$ m$^{-2}$, respectively, as fitting parameters which give reasonable matching between experimental and calculated values.

The relationship between $\gamma$ matrix grain size and transformed $\alpha$ grain size obtained by continuous cooling in experiment and calculation is shown in Fig. 8. It is seen that the transformed $\alpha$ grain size is finer than the $\gamma$ matrix grain size when the $\gamma$ grain size is greater than 5 $\mu$m both in the experimental and the calculation. However, the grain size ratio, i.e. $d_a/d_{\gamma}$, increases by decrease in $\gamma$ grain size and exceeds unity in calculation for an ultrafine $\gamma$ grains smaller than 5 $\mu$m. Indeed, as observed in Fig. 7(b), $\alpha$ grain size is larger than $\gamma$ grain size when the cooling rate is 1 K/s. It is because that as cooling rate gets lower, supercooling and thus $\alpha$ nucleation rate becomes smaller. In such cases, transformation proceeds by the growth of smaller number of $\alpha$ nucleus resulting in a coarser $\alpha$ structure than the $\gamma$ matrix. However, when cooling rate is increased to 10 or 20 K/s for example, $\alpha$ grain size becomes finer than $\gamma$ grain size. This result clearly indicates that further accelerated cooling is needed to refine $\alpha$ structure effectively by transformation from fine-grained $\gamma$ structure.

### 3.2. Pearlite Transformation

Pearlite is a eutectoid transformation product containing lamellar $\theta$ in $\alpha$. This structure is formed by highly cooperative growth of the two constitutive phases. The lamellar pearlite structure formed in a $\gamma$ grain can be divided into several units of substructures. Takahashi et al. reported that the pearlite structure consists of ‘block’ (a region in which $\alpha$ orientations are nearly the same) and ‘colony’ (a region containing $\theta$ lamellas of nearly the same orientation) as its substructures. According to them, the pearlite block act as a unit of ductile fracture. Thus, it is important to refine the $\alpha$ component in pearlite by $\gamma$ grain refinement.

Figure 9 shows SEM microstructure and $\alpha$ orientation map of lamellar pearlite transformed from $\gamma$ with different grain sizes. The size of pearlite block is finer ($d_a=7.5 \mu$m) with $d_{\gamma}=7.7 \mu$m (Fig. 9(c)) than the value ($d_a=16.5 \mu$m) with $d_{\gamma}=77 \mu$m (Fig. 9(a)). However, the size of pearlite colony seems nearly the same at 6–7 $\mu$m in the measurement on the SEM micrographs in Figs. 9(b) and 9(d).

Figure 10(a) shows the relationship between block/
colony sizes and γ grain size. A decrease in the block size by refining γ matrix is more significant than that of the colony size. Figure 10(b) shows that the ratio between \( d_\alpha/d_\gamma \) for pearlite increases and approaches to unity as in the case of proeutectoid α transformation when γ grain size becomes smaller than 10 µm.

According to Umemoto et al.,\(^{25}\) when grain boundary face nucleation is assumed for constant nucleation and growth rate of spherical transformed phase (\( I_\gamma \) and \( G \), respectively), transformed grain size (\( d_\gamma \)) is related to matrix grain size (\( d_\gamma \)) as follows based upon the kinetic theory proposed by Cahn\(^{27}\):

\[
\frac{d_\alpha}{d_\gamma} = 0.695 \left( \frac{I_\gamma}{G} \right)^{-2/9} \frac{d_\alpha^{-2/3}}{G} \text{ (5)}
\]

This assumption is appropriate for pearlite transformation because the driving force remains constant during this transformation. In Fig. 10(b), it is seen that the exponent of \( d_\gamma \) with respect to the grain size ratio (\( d_\alpha/d_\gamma \)) is about \(-2/3\) for pearlite block. This corresponds well to the case of grain boundary face nucleation described above. On the other hand, the exponent for pearlite colony is nearly equal to unity which corresponds to homogeneous nucleation with no dependence of α grain size on the γ grain size. It was seen that neighbouring pearlite colonies are often misoriented by small angles. It is noticed that there are smaller regions with different contrast within a single pearlite block in the orientation maps of Figs. 9(a) and 9(c). Those regions are bounded by low angle boundaries in TEM observation. This implies that a new colony nucleates at the boundary between matrix and pre-existing colony as suggested previously in degenerate pearlite.\(^{28}\) The weak dependence of colony size against matrix grain size seen in Fig. 10(a) can be attributed to such intragranular nucleation of pearlite colonies. However, further study such as in-situ SEM observation is necessary for clarifying the formation process of pearlite colony in more detail.

As shown in the previous section, the γ grain size finer than 5 µm can be established by reversion process in the eutectoid alloys. However, coarse q remains undissolved in those fine-grained γ matrices even after α is fully converted to γ in such cases. When γ containing θ particles is transformed at 873 K, lamellar pearlite is not formed and, instead, spheroidized θ + α aggregate (divorced pearlite) becomes dominant in the transformed microstructure. Pearlite

![Fig. 9. α orientation map and SEM micrograph of pearlite transformed at 873 K for 0.6 ks in eutectoid alloys; (a), (b) V-free alloy, \( d_\gamma = 77 \text{ mm} \); (c), (d) V-added alloy, \( d_\gamma = 7.7 \text{ µm} \); high angle α boundaries of which misorientation is larger than 15 degrees are drawn as solid lines in the maps.](image)

![Fig. 10. (a) Variation of pearlite block or colony size and (b) grain size ratio between α and γ with γ grain size, respectively.](image)

![Fig. 11. SEM micrographs before and after pearlite transformation in the V-added eutectoid alloy. (a) Without undissolved θ, austenitized at 1 123 K for 0.6 ks (\( d_\gamma = 7.7 \text{ µm} \)) and (b) subsequently transformed at 873 K for 0.6 ks. (c) With undissolved θ, austenitized at 1 043 K for 0.6 ks (\( d_\gamma = 6.1 \text{ µm} \)) and (b) subsequently transformed at 873 K for 0.6 ks.](image)
structures are compared for \( \gamma \) matrices of which grain sizes are nearly the same at about 7 \( \mu \)m in Fig. 11. When the specimen without retained \( \theta \) (Fig. 11(a)) was transformed, lamellar pearlite is fully obtained (Fig. 11(b)). On the other hand, from the austenite containing retained \( \theta \) (see arrows in Fig. 11(c)), a considerable amount of divorced pearlite is obtained (Fig. 11(d)). Thus, it is suggested that \( \theta \) needs to be dissolved completely in order to refine pearlite grain sizes by \( \gamma \) grain refinement.

### 3.3. Martensitic Transformation

There is a variety in crystalline size in lath martensite structure. A prior \( \gamma \) grain can be divided into “packets” each of which consists of a group of laths holding the same parallel close-packed plane relationship in the Kurdjumov–Sachs orientation relationship.\(^{29,30}\) In general, a packet is partitioned into several blocks each of which contains laths of a single variant of the K–S relationship. Blocks and packets are mostly surrounded by high-angle grain boundaries whereas lath boundaries in a block are of low-angle type. Packet and block sizes are important factors for both strengthening and toughening of martensitic steels. It is well known that \( \gamma \) grain refinement is most effective for refinement of packet and block sizes.

In general, \( \gamma \) grain refinement suppresses martensitic transformation.\(^{31}\) There are several reasons suggested on the mechanism. (1) Small elastic stress concentration at a martensite tip which is more difficult to initiate burst transformation,\(^{32}\) (2) strengthening of \( \gamma \) by a Hall–Petch effect,\(^{33}\) (3) segregation of nucleation into a few grains.\(^{34}\) Mechanical properties of lath martensite were examined for various packet sizes by changing \( \gamma \) grain size between a few to few hundreds \( \mu \)m.\(^{35,36}\) It was found that both of the 0.2% proof strength increases and ductile-brittle transition temperature is lowered with refinement of packet size obeying Hall–Petch type relationships. However, there are few studies on effect of block size on mechanical properties of martensitic steel.

We examined the variation of packet and block sizes\(^{37}\) and strength\(^{38}\) by changing the \( \gamma \) grain size in a wide range by the same procedure described in the case of proeutectoid transformation. Figure 12 shows the lath martensite structures formed from different \( \gamma \) grain sizes. As \( \gamma \) grain size decreases, packet and block sizes are both refined (Figs. 12(a) to 12(c)). We note that a prior \( \gamma \) grain is subdivided into a few packets even in the case of ultrafine grained \( \gamma \) of \( d_\gamma=2.3 \mu \)m in the TEM micrograph of Fig. 12(c). Figure 13 shows the relationship between prior \( \gamma \) grain size and packet or block size measured including the previous works.\(^{35,39}\) As \( \gamma \) grain size decreases, packet and block sizes decrease. And those grain sizes approach toward the line of \( d_\gamma/d_p=1 \) when \( d_\gamma \) is smaller than 50 \( \mu \)m. However, it should be noted that unlike diffusional transformations, martensite does not grow crossing \( \gamma \) grain boundaries and thus, \( d_\gamma/d_p \) does not exceed unity.

Martensite nucleation starts at a \( \gamma \) grain boundary and a block is formed by repeated nucleation of the same variant of lathes adjacent to each other. Normally, development of a block leads to accumulation of elastic strain energy and thus, self accommodation comes to play a part to form different martensite variants. This results in formation of different blocks in a packet or formation of different packets. However, if a \( \gamma \) grain size gets smaller than a few \( \mu \)m, the first nucleated variant dominates in a single \( \gamma \) grain and strain would be accumulated at \( \gamma \) grain boundary. In such cases, because of small martensite sizes, stress concentration is small enough to accommodate elastically as suggested by Unemoto and Owen.\(^{32}\) Furthermore, strain accommodation might occur across the grain boundary by selecting some preferred variant in an adjacent \( \gamma \) grain which accommodates strain accumulated at grain boundary. In our study,\(^{37}\) one packet is larger than the others in fine grained \( \gamma \) matrix as seen in Fig. 12(d). Furthermore, Takaki et al.\(^{40}\) reported that an ultrafine grained \( \gamma \) structure of which grain size is smaller than 1 \( \mu \)m transforms to almost a single variant of martensite in a Fe–16Cr–10Ni austenitic stainless steel of which composition is in the range of \( \gamma \rightarrow e \rightarrow \alpha' \) transformation sequence.\(^{31}\) They also suggested that a large strain field in the single variant transformation suppresses transformation kinetics.

It was found that yield strength is proportional to square root of the width of block, which is mostly surrounded by high-angle boundaries, as same as the packet size. Since the coefficient in this Hall–Petch equation with respect to the block is more close to that of equi-axed grained ferrite structure than the case of packet, it was proposed that the
Hall–Petch dependence for packet boundaries previously reported\(^{(35)}\) could be apparent and the block size controls the component of grain boundary strengthening in the high strength of lath martensite. However, due to the specific (here K–S) orientation relationship between martensite and its austenite matrix, the role of grain boundaries as barriers could be different between block boundaries and packet boundaries in slip and fracture. Morris et al.\(^{(32-44)}\) proposed that a packet boundary would be more effective as a barrier of slip on the \(\{110\}_a\) planes whereas a block boundary inhibits cleavage fracture along \(\{100\}_a\) planes in terms of boundary misorientation. Therefore, the deformation behavior of lath martensite needs to be clarified in more detail by further studies.

4. Summary

The formation of fine-grained austenite structures in the abovementioned two processing routes with reversion was described and subsequently, ferrite, pearlite and martensite structures formed from the fine-grained austenite were discussed.

(1) The ultrafine-grained \(\gamma\) structure of which grain size is a few \(\mu\)m can be achieved for various kinds of steels through cyclic transformation or cold deformation of carbonitride is effective to suppress the coarsening of \(\gamma\) grains. It was pointed out that extensive accelerated cooling should be applied to refine effective grain sizes (packet and block sizes) in lath martensite formed.

(2) Both for ferrite and pearlite transformations, transformed \(\alpha\) grain size becomes finer by refining \(\gamma\) grain size down to about \(5 \mu\)m. However, the grain size ratio of \(\gamma\) and \(\alpha\), \(d_\gamma/d_\alpha\), increases as \(\gamma\) grain size is refined. For \(d_\gamma<5 \mu\)m, the ratio \(d_\gamma/d_\alpha\) exceeds unity for a slow cooling rate (1 K/s). It was pointed out that extensive accelerated cooling should be applied to refine \(\alpha\) grain size by diffusional transformations from the fine-grained \(\gamma\) matrix of \(d_\gamma<5 \mu\)m.

(3) Refinement of the \(\gamma\) grain size is useful to refine effective grain sizes (packet and block sizes) in lath martensite structure after quenching. As decreasing the \(\gamma\) grain size smaller than \(5 \mu\)m, the number of martensite variant formed seems decreased.

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