Observation of $\gamma \rightarrow \alpha$ Transformation in Ultralow-carbon Steel under a High Temperature Optical Microscope

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The basic composition of IF steel is Fe-ultralow-carbon (<0.02 mass%). There is little research on the transformation behavior and microstructures of ultralow-carbon steel compared to low-carbon steels, and the details, for instance, transformation mechanism and the ratio of grain sizes of $\gamma$ and $\alpha$, remain unclear. Therefore, in situ observation of the $\gamma \rightarrow \alpha$ transformation in an Fe–0.004% C steel was performed at a cooling rate of 0.5–18°C/s under a high-temperature optical microscope. The main microstructure of ferrite was $\alpha_q$, and with an increase in the cooling rate, the fraction of Widmanstätten ferrite-like structure and ferrite having a severely ragged interface increased. The growth rate of $\alpha_q$ was $1 \times 10^{-4} - 9 \times 10^{-4}$ m/s which increased with the cooling rate. The growth rate of $\alpha_q$ decreased to about half when the amount of carbon increased to 0.01%. The ratio of $\gamma$ grain size to $\alpha$ grain size was about 1.2, and this value is considerably smaller than the values reported for the low-carbon steels. $\alpha_q$ crossed $\gamma$ grains frequently and some $\alpha_q$ grains were larger than $\gamma$ grains. Usually the curvature of the $\alpha_q/\gamma$ interface did not change at the intersection of the boundaries of $\alpha_q$ and $\gamma$. This shows that $\alpha_q/\gamma$ interfaces are usually incoherent in ultralow-carbon steels. Transformation temperature was in the single-phase region of $\alpha$. Therefore, the $\gamma \rightarrow \alpha_q$ transformation observed in the present research is thought to be massive transformation. The terminology of mechanisms of $\gamma \rightarrow \alpha$ transformation in ultralow-carbon steels and the microstructures of generated $\alpha$ was discussed.

KEY WORDS: ultralow-carbon steel; microstructure; in situ observation; growth rate; quasi-polygonal ferrite; equi-axed ferrite; massive transformation.

1. Introduction

The basic composition of IF steel is ultralow-carbon (<0.02 mass%). The number of research on the behavior of austenite ($\gamma \rightarrow$ ferrite ($\alpha$) transformation and the microstructures of $\alpha$ in an ultralow-carbon steel is small compared to research on steel containing a larger amount of carbon. Moreover, relationship between transformation behavior and the microstructures, and the mechanism of the transformation remain unclear.2,5 In addition, the terminology of the microstructures3,5 is confusing.

Shibata and Asakura5 observed $\gamma \rightarrow \alpha$ transformation in an ultralow-carbon steel in detail and found that the grain diameter of $\alpha$ did not decrease much even when the cooling rate was increased. Together with polygonal ferrite ($\alpha_p$),3,5 of which the grain boundaries are planar, and quasi-polygonal ferrite ($\alpha_q$),3,5 of which grain boundaries are curved or ragged, a variety of intermediate microstructures and martensite were observed and were found to be dependent of the cooling rate. In order to examine the relationship between the $\gamma$ grain boundary and the $\alpha$ grain of an ultralow-carbon steel, the present authors4,6,7 performed several experiments. For instance, they thermally etched $\gamma$ grain boundaries and attempted to observe the grain boundaries of both $\gamma$ and $\alpha$ grains simultaneously. However, an insufficient number of successful observations was obtained, because it was difficult to see grain boundaries of both phases over a wide area. Moreover, they attempted to observe $\gamma$ grain boundaries through the image of an alpha particle track23 showing boron segregation along $\gamma$ grain boundaries and to compare the track with the usual chemical etching microstructure of $\alpha$ in the same region. However, when the cooling rate was slower than 20°C/s, they only observed the track caused by the segregation of boron along $\alpha$ grain boundaries. When boron segregated along $\gamma$ grain boundaries, bainitic transformation occurred. Namely, they could not examine the relationship between $\gamma$ and $\alpha$ grain boundaries. Therefore, in the present research, in situ observation of the $\gamma \rightarrow \alpha$ transformation in an ultralow-carbon steel was performed using a high-temperature optical microscope (HTOM).

2. Experimental Procedures

Two steels were prepared by melting in a vacuum induction furnace for this research. Chemical compositions of these steels are shown in Table 1. Steel B (0.01% C) was used only to examine the effect of carbon on the growth rate of $\alpha$ grains. The ingots were forged and then hot-rolled. The specimens for high-temperature optical mi-
crosscopy were prepared by the following procedures. Blocks were cut from the hot-rolled plate perpendicularly to the rolling direction, and were machined into rods of 3 mm in thickness with a precision cutter. The discs were then polished mechanically. Heating, cooling and observation of the $\gamma \rightarrow \alpha$ transformation under HTOP were carried out in a vacuum of about $10^{-3}$ Torr. A disc specimen was set on a tantalum heater. The specimen temperature was measured using an R-type thermocouple of 0.2 mm which was spot-welded to the specimen. The specimen was heated to 1000°C at the rate of 1.7°C/s and held at this temperature for 3 min. The cooling rate was controlled to be constant. The cooling rate was varied in the range from 0.5 to 18°C/s.

**Figure 1** shows the main components of the system for high temperature optical microscopy. They are an optical microscope, a heating stage, a charge-coupled device (CCD) camera, a video tape recorder (VTR) and a temperature controller. The display rate of temperature is about 30 times/s and the capture rate of the image is 30 frames/s. Hereafter, microstructures observed using such a procedure will be called ISSs (in situ structures). The growth rate of \( \alpha \) grains was measured as the mean value of the velocity of the \( \alpha/\gamma \) interfaces of \( \alpha \) grains observed on the VTR screen. Growth rate was not measured for \( \gamma \) phase at 1000°C. Figures 2(b) and 2(c) show ISS during the \( \gamma \rightarrow \alpha \) transformation, while (d) shows ISS after the transformation. The arrows in (b) and (c) indicate the position of the \( \alpha/\gamma \) interface and the growth direction of \( \alpha \) grains. The \( \alpha \) grain grows chiefly as a result of movement of the curved \( \alpha/\gamma \) interface. The growth rate was about $1 \times 10^{-4}$ m/s on average. **Figure 3** shows the observed result for the cooling rate of 0.8°C/s. Figure 3(a) shows ISS of the \( \gamma \) phase at 1000°C. Figure 3(b) shows ISS after the transformation, while (c) shows CES. Figure 3(d) shows a trace of the grain boundaries of \( \gamma \) and \( \alpha \) observed in (b). It can be seen that an \( \alpha \) grain crosses \( \gamma \) grains. As shown in the circles in (d), the change in the curvature of the \( \alpha/\gamma \) interface could not be seen in many cases at the intersection of grain boundaries of \( \alpha \) and \( \gamma \). When the cooling rate was 1.6°C/s, \( \gamma \rightarrow \alpha \) transformation behavior was similar to that in the cases of 0.5 and 0.8°C/s cooling.

At cooling rates from 2.7 to 3.3°C/s, the main microstructure of ferrite was also \( \alpha_{\gamma} \). In addition to \( \alpha_{\gamma} \), \( \alpha_{q} \) and Widmanstätten ferrite-like structures were observed. The Widmanstätten ferrite-like structure appeared to nucleate at the \( \gamma \) grain boundaries. The growth of an \( \alpha_{g} \) grain with crossing a \( \gamma \) grain was more marked when the cooling rate was slower. Hardness was HV86–90. **Figure 4** shows the

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**Table 1.** Chemical composition of steels (mass%)

| steel | C  | Si | Mn | P  | S  | Al | Ti | B   | N  | O  |
|-------|----|----|----|----|----|----|----|-----|----|----|
| A     | 0.004 | <0.01 | 0.01 | 0.004 | 0.016 | 0.008 | 0.004 | 0.0004 | 0.0040 | 0.0090 |
| B     | 0.010 | 0.05 | 0.06 | 0.008 | 0.005 | 0.0057 | 0.001 | <0.0001 | 0.0040 | 0.0017 |

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**Fig. 1.** Diagram of the system for high temperature optical microscopy.

**Fig. 2.** In situ micrographs showing growth of ferrite observed in Steel A cooled at 0.5°C/s. (a) 1000°C, (b) 832.2°C, (c) 831.8°C, (d) 830.8°C.

**Fig. 4.** α growth direction
The $\gamma \rightarrow \alpha$ transformation behavior when the cooling rate was 2.7°C/s. **Figure 5** shows the $\gamma \rightarrow \alpha$ transformation behavior when the cooling rate is 3.3°C/s. The arrows in (4) and (5) indicate the position of the $\alpha/\gamma$ interface and the growth direction of the $\alpha$ grain. **Figures 6(a) and 6(b) show traces corresponding to Figs. 5 (4) and (5), respectively.** $\alpha$ grains which cross $\gamma$ grains can be seen. $\alpha$ grains grow mainly as a result of the movement of the curved $\alpha/\gamma$ interface, as in the cases of lower cooling rates. However, planar interfaces can be observed locally, as shown by the letter (A) in Fig. 5 (5).

When the specimens were cooled at 4–10°C/s, the main microstructure of ferrite was again $\alpha$. $\alpha$ and Widmanstätten ferrite-like structures coexisted. However, the formation frequency of the Widmanstätten ferrite-like structure was higher than that for cases cooled at rates slower than 3.3°C/s. The average diameter of $\alpha$ was about 140 µm and the maximum diameter was about 170 µm. The growth of $\alpha$ grains with crossing $\gamma$ grains was also observed. Hardness was HV91–98. **Figure 7** shows the $\gamma \rightarrow \alpha$ transformation behavior when the specimen was cooled at 4°C/s. The $\alpha/\gamma$ interface which began to be observed in (2) crosses a $\gamma$ grain. The Widmanstätten ferrite-like structure was observed in the area enclosed by a circle in (4). In addition,
it was observed that a planar interface, as shown in the square in (5), appeared in a short period such as 0.1 s. In Fig. 8, the relationship between the Widmanstätten ferrite-like structure and γ grain boundary is shown in detail. Figure 8(a) shows ISS of γ at 1000°C, (b) shows CES and (c) shows the trace of γ and α grain boundaries. It can be seen that the Widmanstätten ferrite-like structure nucleates at the γ grain boundary and grows. However, the ragged structure that can be seen at the bottom of Figs. 8(b) and 8(c) does not nucleate at the γ grain boundary. Therefore, this ragged microstructure can be distinguished from above the Widmanstätten ferrite-like structure. Such a ragged microstructure is thought to be a microstructure wherein the degree of curving of the α boundary increases markedly. The frequency of such a microstructure increased with an increase in the cooling rate. The growth rate of α was about $3 \times 10^{-4}$ m/s at the cooling rate of 4°C/s. The transformation behavior and the transformation microstructure observed at the cooling rate of 10°C/s were similar to those in the case of the cooling rate of 4°C/s.

When the cooling rate was 15–18°C/s, the main ferrite structure was again α. Ferrite with planar boundaries and the Widmanstätten ferrite-like structure were observed in addition to α. The growth rate of α was $8 \times 10^{-4}$–9×$10^{-4}$ m/s. Hardness was HV103–108. Figure 9 shows the transformation behavior when the cooling rate was 15°C/s. In Fig. 9, α which has planar grain boundaries (□) and α which seems to grow from the γ grain boundary (□) are observed. Figure 9(c) shows the trace of the grain boundaries of α and γ. Also at this cooling rate, α grains crossed γ grains. However, the frequency is less than in the cases of cooling rates of 10°C/s or lower. The transformation behavior and the transformation structure at the cooling rate of 18°C/s were similar to those in the case of the cooling rate of 15°C/s.

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**Fig. 7.** *In situ* micrographs showing growth of ferrite observed in Steel A cooled at 4°C/s.

**Fig. 8.** *In situ* observation of austenite and ferrite grain boundaries observed in Steel A cooled at 4°C/s; corresponding to Fig. 7. (a) ISS (*in situ* structure) of austenite, (b) CES (chemically etched structure) of ferrite and (c) trace of austenite and ferrite grain boundaries.

**Fig. 9.** *In situ* observation of austenite and ferrite grain boundaries observed in Steel A cooled at 15°C/s. (a) ISS (*in situ* structure) of austenite, (b) CES (chemically etched structure) of ferrite and (c) trace of austenite and ferrite grain boundaries.
4. Discussion

4.1. Variation of the Velocity of α/γ Interface with Cooling Rate and Carbon Content

Effects of cooling rate on the velocity of the α/γ interface (growth rate of the α phase) and transformation time are shown in Fig. 10. The velocity of the α/γ interface is 1–9×10⁻⁴ m/s which increases with an increase in the cooling rate. Table 2 shows the velocities of the α/γ interface previously reported. The values 1×10⁻³ m/s reported by Leslie,⁹) 1×10⁻² m/s by Wilson¹) and 4.4×10⁻² m/s by Rasanen¹) are similar to the values obtained in the present research. Leslie⁹) reported that the value 1×10⁻³ m/s corresponds to the velocity of massive transformation. Wilson¹) mentioned that the values 1×10⁻² m/s and 4.4×10⁻² m/s correspond to the velocity of equi-axed ferrite transformation. This equi-axed ferrite transformation is thought to be the same as massive transformation and the details will be discussed later.

The transformation time is fairly long, about 2.5 s, for the cooling rate of 0.5°C/s. However, the transformation time decreases as the cooling rate increases and is very short, about 0.5 s, at the cooling of 18°C/s. Temperature difference between Fs and Ff is 1.3°C for 0.5°C/s cooling and 9°C for 18°C/s cooling. Namely, in the case of 0.5°C/s cooling, γ→α transformation is completed with a small decrease in temperature.

Figure 11 shows the growth rates of α in Steels A and B. The growth rate of α in Steel B (0.010 mass% C) is generally slower than that in Steel A (0.004 mass% C). In both steels, the velocity of α/γ interfaces increases with an increase in the cooling rate. Namely, the velocity of α/γ interfaces of Steel A is about two times higher than that of Steel B.

4.2. Growth of α Crossing γ Grain

It was found that α grains grow across prior γ grains during γ→α transformation at the range of the cooling rates in the present research (0.5–18°C/s). Such growth behavior was more marked at lower cooling rates. Goodenow et al.,¹¹) Wilson,¹) Maki²) and Araki et al.³) reported on the growth behavior of α, but did not mention the above growth behavior.

Figure 12 schematically shows two kinds of growth behavior of α in γ→α transformation of (a) low-carbon and (b) ultralow-carbon steels. In low-carbon steels, α grains which nucleated mainly on γ grain boundaries and inclusions impinge on each other before crossing the prior γ grains. Therefore, the final grain size of α is usually smaller than that of the prior γ. In ultralow-carbon steels, however, α grains can cross the prior γ grains frequently and therefore the size of some α grains is larger than that of the prior γ grains.

Sekine and Maruyama¹²) measured the ratio of grain size d(γ)/d(α) in hot-rolled Fe–0.1C–0.25Si–1.4Mn and Fe–
to the large grain size of commercial low-carbon steels because the number of pre-
helpful for elucidating the reason for the small
rate contributes to such a small
d\(m\) and \(18°C/s\) is about 140
cooling for Steel A in the present research. Some
thought that the nucleation rate of
carbon steels. In addition, in ultralow-carbon steel, it is
Average grain size of
ing rate, although the amount of the decrease is small.
Grain size of \(\alpha\) decreases with an increase in cool-
ing rate, although the amount of the decrease is small.
Average grain size of \(\alpha\) in the cooling range between 0.5
and \(18°C/s\) is about 140 \(\mu m\). Using the values of 168 \(\mu m\) and
\(140\mu m\), the value 1.2 is obtained for the ratio 
d(\(\gamma\)/d(\(\alpha\)). Namely, the ratio is markedly small compared to
the ratio for the above two low-carbon steels. High growth
rate contributes to such a small d(\(\gamma\)/d(\(\alpha\)) ratio in ultralow-
carbon steels. In addition, in ultralow-carbon steel, it is
thought that the nucleation rate of \(\alpha\) is lower than that in
commercial low-carbon steels because the number of precipitates and inclusions is smaller compared to commercial
low-carbon steels. Such a low nucleation rate also contributes to the large grain size of \(\alpha\) and a small d(\(\gamma\)/d(\(\alpha\)) ratio. Furthermore, it is expected that \(\alpha\) grains can more easily cross the prior \(\gamma\) grain boundaries in ultralow-carbon steels and contributes to a small d(\(\gamma\)/d(\(\alpha\)) ratio, because impingement effects of precipitates and inclusions are thought to be small in ultralow-carbon steel, although these effects have not yet been clarified by experiments.
Computer simulation of the \(\gamma\rightarrow\alpha\) transformation may be helpful for elucidating the reason for the small d(\(\gamma\)/d(\(\alpha\)) ratio of ultralow-carbon steel.

0.05C–0.25Si–1.2Mn commercial steels. In their results, the ratios were 5 for the former steel and 2.5 for the latter steel when the diameter of \(\gamma\) was 170 \(\mu m\). Figure 13 shows \(\alpha\) grain sizes at various cooling rates and \(\gamma\) grain size before cooling for Steel A in the present research. Some \(\alpha\) grains are larger than the average grain size of \(\gamma\), namely 168 \(\mu m\). Grain size of \(\alpha\) decreases with an increase in cooling rate, although the amount of the decrease is small.

Namely, the cooling rates in the present research are much
higher than the cooling rates previously reported for mas-
sive ferrite formation. In addition, \(\gamma\) temperatures observed
in the present research are higher than those reported in the
above-mentioned papers, except for the high plateau temperature of 833°C measured by Mirzaev et al.\(^{21}\).

4.3. Effect of Cooling Rate on \(F_s\) Temperature and Transformation Mechanism

Figure 14 shows the relationship between cooling rate and \(\gamma\rightarrow\alpha\) transformation start temperature (\(F_s\)) of Steel A cooled at 0.5–18°C/s. The \(F_s\) temperature falls with an increase in cooling rate and is in the range from 845 to 813°C. The relationship between cooling rate and \(\gamma\rightarrow\alpha\) transformation temperature has so far been examined by many researchers: Gilbert and Owen,\(^{14}\) Bibby and Parr,\(^{15}\) Izumiya, et al.\(^{16}\) Ackert and Parr,\(^{17}\) Morozov et al.,\(^{18}\) Leslie,\(^{9}\) Wilson\(^{19,20}\) and Mirzayev et al.\(^{21}\). According to their results, the transformation temperature decreases as the cooling rate increases in the range of relatively slow cooling. Beyond the critical cooling rate, the transformation temperature becomes constant and shows some plateaus on the \(F_s\) versus cooling rate curve. They reported that the temperature of the plateau and the critical cooling rate depend on the content of carbon and that the plateaus correspond to different transformation mechanisms. Wilson\(^{9}\) noted that the first plateau was for equi-axed ferrite formation and the second plateau for massive ferrite formation. Other researchers, however, did not observe two plateaus at those cooling rates, but only one plateau corresponding to massive ferrite formation.

On the other hand, it was often observed in the present research that the curvature of the \(\alpha/\gamma\) interface did not change markedly at the intersection of the grain boundaries of \(\alpha\) and the prior \(\gamma\) grains. This observation shows that most \(\alpha/\gamma\) interfaces are incoherent which corresponds to the mentions by Wilson\(^{19}\) and Massalski\(^{13}\) that the growth of new phases in massive transformation is caused by movement of incoherent interfaces.
inferred, as mentioned in Sec. 4.2, that almost all moving
effects of other elements are considered. In addition, it was
authors consider that
Christian\textsuperscript{22)} mentioned that allotropic transformation occurs
has the following characteristics: (1) the nucleation site is
sive ferrite transformation. Equi-axed ferrite transformation
rate. However, he mentioned that the former was not mas-
temperature shows the plateau in the range of high cooling
showing the plateau in the transformation temperature
versus cooling rate curve.

Here, two problems arise: (1) the difference in the mech-
isms of the two kinds of massive transformation and (2)
the difference in the microstructures generated by these two
types of transformation. The former problem will be dis-
cussed in this section and the latter will be discussed in the
following section.

Hillert\textsuperscript{23)} analyzed the relationship between the driving
force of the transformation (difference in Gibbs free ener-
gies of $\gamma$ and $\alpha$) and the mobility of the $\alpha/\gamma$ interface by
considering that carbon atoms accumulate on the $\alpha/\gamma$ inter-
facing during $\gamma$ to $\alpha$ transformation. His calculated results for
steels with very low contents of carbon indicated that the
transformation temperature decreases with an increase in
cooling rate in the range of slow cooling rate and that the
temperature shows the plateau in the range of high cooling
rate. However, he mentioned that the former was not mas-
sive transformation but diffusional transformation. Jönsson
and Ågren\textsuperscript{24)} performed a detailed analysis similar to that
conducted by Hillert and showed that there are two kinds of
massive transformation in the ($\alpha + \gamma$) two-phases region:
slow-growth-rate transformation with carbon dragging and
high-growth-rate transformation without carbon dragging.
Regarding the transformation in an $\alpha$ single-phase region,
they concluded that there is only one high-growth-rate mas-
sive transformation.

On the other hand, Wilson\textsuperscript{1,20)} has proposed two new trans-
formations: equi-axed ferrite transformation and mas-
sive ferrite transformation. Equi-axed ferrite transformation
has the following characteristics: (1) the nucleation site is
the corner of the $\gamma$ grains, (2) an $\alpha$ grain is bounded by one
or more coherent grain boundaries, (3) volume change
upon transformation can be easily relaxed and the disloca-
tion density is relatively low, (4) growth of $\alpha$ is caused by
the movement of high-energy incoherent boundaries with
rapid atom transfer across the boundaries and (5) solute
drag does not occur. As for the latter massive ferrite trans-
formation, (1) $\alpha$ nucleates as a semispherical cap on the $\gamma$
grain boundaries, (2) the planar interface of the semispheri-
cal cap is coherent and the curved interface is incoherent,
(3) volume change upon the transformation cannot easily
be relaxed and the dislocation density is relatively high, (4)
growth of $\alpha$ is caused by incoherent boundaries, (5) move-
ment of these incoherent boundaries is not as high as in the
equi-axed ferrite transformation due to solute drag, (6)
solute drag makes the interface ragged. Wilson mentioned
that the rapid atom transfer across the boundaries is caused
by grain boundary diffusion. Therefore, the mechanism of
atom transfer at the interface in the equi-axed ferrite trans-
formation seems to be the same as that in allotropic trans-
formation and recrystallization. Wilson\textsuperscript{1} calculated the
growth rate of $\alpha$ of zone-refined iron growing at 800°C by
the equi-axed ferrite transformation and obtained $1 \times 10^{-4}$
m/s. This value corresponds to the growth rate, about
$1 \times 10^{-4}$ m/s, which was obtained in the range of the cooling
rate from 0.5 to 3.3°C/s in the present research. In addition, in
addition, the equi-axed ferrite transformation occurs at lower
temperature with an increase in cooling rate. Therefore, it is
thought that $\alpha_{q}$ is transformed from $\gamma$ through the same
mechanism as in Wilson's equi-axed ferrite transformation,
although the morphology of $\alpha_{q}$ is not always equi-axed and it
is not certain that $\alpha_{q}$ nucleates at the corner of a $\gamma$ grain.

As mentioned above, Christian\textsuperscript{22)} noted that allotropic
transformation occurs through massive transformation, and
atom transfer in Wilson's equi-axed ferrite transformation
seems to be the same as that in allotropic transformation.
Therefore, it is possible to regard Wilson's equi-axed ferrite
transformation massive transformation. In Wilson's nomen-
clature of transformation, however, the relationship among
massive transformation, equi-axed ferrite transformation
and massive ferrite transformation is not clear. In addition,
the present authors do not agree with his nomenclature in
which a term indicating morphology is used in the name of
transformation mechanism, because the mechanism which
is thought to be the same as Wilson's equi-axed ferrite
transformation may not always generate equi-axed mor-
phology, such as $\alpha_{q}$ observed in the present research.

4.4. Terminology of $\alpha$ Observed in Ultralow-carbon
Steel

Figure 16 shows representative optical micrographs
shown in previous papers as massive structure or massive
ferrite. Micrograph (a) shows the microstructure of Cu–
9.3at%Al, called a massive structure for the first time by
Greninger.\textsuperscript{25)} Micrographs (b) and (c) show microstructures
of Cu–37.8at%Zn\textsuperscript{13)} and pure iron,\textsuperscript{26)} respectively. They
were shown by Massalski\textsuperscript{13)} to be representative massive
structures. Micrograph (d) shows the microstructure of Fe–
4at%Ni, which was presented as the representative massive
ferrite by Owen and Wilson.\textsuperscript{26)} and Maki.\textsuperscript{27)} They noted that
this microstructure transformed from $\gamma$ by massive transfor-
information. Wilson, however, stated in his recent paper that it transformed by the equi-axed ferrite transformation. Micrographs (e) and (f) show microstructures of Fe–2%Mn steel and Fe–4%Cu steel, respectively. Wilson showed them to be massive ferrite.

In these micrographs, with the exception of (c), zigzagging lines surround almost all ferrite grains. Grains in the micrograph (c) are surrounded by curved boundaries and resemble α in the present research. Compared to the micrograph (a), it seems unreasonable to call microstructures in (c) and α in the present research massive structures or massive ferrite, although microstructures in (c) and α may transform from γ by massive transformation. Araki and coworkers have proposed calling ferrite surrounded by curved or ragged boundaries quasi-polygonal ferrite, irrespective of the transformation mechanism. Following their nomenclature, all of the microstructures, revealed in Fig. 16 are classified as αq.

On the other hand, in the present research, it was observed that the Widmanstätten ferrite-like structure formed from prior-austenitic grain boundaries, as shown in Figs. 5, 8 and 9. This microstructure resembles that of Widmanstätten ferrite. Therefore, it seems possible to call such structure Widmanstätten ferrite and to use the symbol αw for this microstructure following the nomenclature by Araki and coworkers. However, at the present stage it is thought to be reasonable to call such a structure as the Widmanstätten ferrite-like structure, because details of this microstructure, for instance, orientation relationship between γ and this kind of α, have not yet been performed.

In addition, a microstructure having severely ragged α/γ interfaces was also observed, as shown in Fig. 8. This microstructure could be distinguished from the Widmanstätten ferrite-like structure and was thought to be a structure wherein the degree of curving of the αq boundary increases markedly. A similar microstructure was reported by Shibata and Asakura in Fe–0.002%C steel. They proposed that such a structure is formed during cooling when γ to α transformation continues at relatively low temperatures and the transformation mechanism approaches bainitic transformation. When the steel is cooled at a higher rate, the transformation temperature decreases and therefore it is thought that the number of α grains having severely curved interfaces increases. Figure 17 shows the change of such a structure with cooling rate. The degree of curving and the frequency of this structure seem to increase with the cooling rate, as speculated above. From the appearance of this microstructure, it is not ideal to use αq for this microstructure, although the present authors have not found an appropriate name for this microstructure. The shape of an α grain of this microstructure is acicular, but it is not appropriate to call this microstructure acicular ferrite, because the term acicular ferrite has already been used for the microstructure of fine cementite and fine martensite dispersed in a ferrite matrix of low-carbon steel heavily deformed in the non-recrystallization temperature range.

5. Conclusion

The γ→α transformation behavior during cooling in ul-
tralow-carbon steels was observed under a high temperature optical microscope. The results can be summarized as follows.

1. The ferrite is mainly $\alpha_q$, $\alpha_p$ was partially observed. As the cooling rate was increased, the Widmanstätten fer-
rite-like structure and microstructure having severely
ragged boundaries were also observed.

2. The velocity of the $\alpha/\gamma$ interface was $1 \sim 9 \times 10^{-4}$
m/s which increased with the cooling rate and decreased
upon increasing the carbon content, the velocity of Fe–
0.004C steel was about two times larger than that of Fe–
0.01% C steel.

3. The ratio of the diameters of $\gamma$ and $\alpha$ was around
1.2 which is fairly small compared to the ratio reported for
commercial low-carbon steels.

4. It was often observed that the $\alpha/\gamma$ interface propa-
gates across the prior $\gamma$ grain(s) during $\gamma \rightarrow \alpha$ transfor-
mation. As the result of such growth behavior of $\alpha_q$ in ultralow-carbon steel, (a) low nu-
cleration rate of $\alpha_q$ on the $\gamma$ grain boundaries and (b) the
ease with which the $\alpha_q/\gamma$ interface crosses the $\gamma$ grain
boundaries are conceivable, together with (c) the high ve-
locity of the $\alpha_q/\gamma$ interface.

5. At almost all regions where the $\alpha_q/\gamma$ interfaces go
across the $\gamma$ boundaries, the curvature of the interface does
not change. This result indicates almost all $\alpha_q/\gamma$ interfaces
to be incoherent.

6. The $\gamma \rightarrow \alpha$ transformation start temperature ($F_s$)
ranges from 845 to 813°C and decreases with an increase in
the cooling rate. The phase diagram calculated using
ThermoCalc shows that this temperature range corresponds
to the single-phase region of $\alpha$.

7. On the basis of the results (5) and (6), the $\gamma \rightarrow \alpha_q$ tran-
sformation is thought to occur through massive transfor-
mation. It is incorrect to call only the transformation
showing a plateau on the cooling rate versus $F_s$ curve mas-
sive transformation.

8. The usage of terms of morphology in the name of the transformation mechanism produces confusion. Therefore, terminologies such as equi-axed ferrite transformation and massive ferrite transformation are not recom-
mended.

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