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Formation of grain boundary ferrite in eutectoid and hypereutectoid pearlitic steels

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ABSTRACT

Formation of grain boundary ferrite (GB-α) in high carbon pearlitic steels deteriorates ductility of steel wires. Since it is generally believed that fully pearlitic structure forms in eutectoid steels, GB-α formation in eutectoid or even hypereutectoid steels is curious. Therefore, effects of temperature and carbon content on the formation of GB-α in Fe-1mass%Mn-(0.75 and 1.05) mass%C alloys transformed isothermally at temperatures ranging from 873 K (600 °C) to 973 K (700 °C) were investigated to clarify the formation mechanism of GB-α. It was found that volume fraction of GB-α increases with decreasing transformation temperature, carbon content and prior austenite grain size. Pearlite nucleated at prior austenite grain boundary usually grows into only one of austenite grains separated by the grain boundary, and GB-α forms on the other austenite grain. Orientation analyses revealed that GB-α and pearlitic-α hold the same orientation, indicating that one α grain grows as pearlite into one of austenite grains, and as GB-α into the other austenite grain. It was shown that such morphology difference is caused by the difference in orientation relationship (OR) between ferrite and austenite such that near K–S OR and non K–S OR correspond to GB-α and pearlite, respectively. Consequently, it was proposed that suppression of cementite nucleation at ferrite/austenite boundary holding near K–S OR is a reason for the formation of GB-α in the transformation at low temperature. Furthermore, degenerated pearlite (DP) and Widmanstatten ferrite (WF) is formed as well as GB-α at lower temperature, and OR dependency of GB-α, WF, DP and pearlite were clarified.

1. Introduction

Demands for high strength pearlitic steel wires have being increased in order to develop long-span bridges or to reduce weight of tires for automobiles. The steel wires are produced by heavily cold drawing. In order to increase strength of steel wires, increases in carbon content, drawing strain and strain hardening ratio are required. On the other hand, strengthening of steel wires deteriorates ductility and delamination, a fracture along longitudinal direction of wires during torsion straining, makes further strengthening of them difficult. It has been proposed that anisotropic texture development by drawing [1,2], excess carbon in ferrite (α) by dissolution of cementite (θ) during drawing [3] and formation of thin grain boundary ferrite (GB-α) along prior austenite (γ) grain boundaries [4,5] are reasons for delamination.

There are a lot of literature reporting the formation of fully pearlitic microstructure when a steel is transformed in the condition (1) in Fig. 1(a) where an alloy composition is below both A_{e3} and A_{cm} lines, and thus γ is supersaturated with respect to both α and θ [6]. In the condition (2) in Fig. 1(a), α and pearlite mixed structure evolves [7,8], which is a typical microstructure of medium carbon steels transformed by slow cooling. On the other hand, a few literature reported that GB-α can be formed in the condition (1) or even in the condition (3), where carbon content is higher than A_{e3} composition.

Hillert [9] reported that abnormal α forms when a carburized steel with carbon content higher than the eutectoid composition are transformed initially just above A_1 temperature to form grain boundary cementite (GB-θ), subsequently transformed just below A_1 temperature. He proposed that carbon content in γ nearby θ becomes lower than A_{e3} composition, leading to the nucleation of α. Later, Chaiguangsri and Edmonds [10] also reported the formation of coarse α grains in hypereutectoid steel transformed just below A_1 temperature. They supposed that driving force for transformation is not sufficient for co-operative growth of (α + θ) lamellae and then abnormal α is developed as a result of non-cooperative growth of α and θ.

On the other hand, even when hypereutectoid steels are transformed at much lower temperature, e.g. at temperatures ranging

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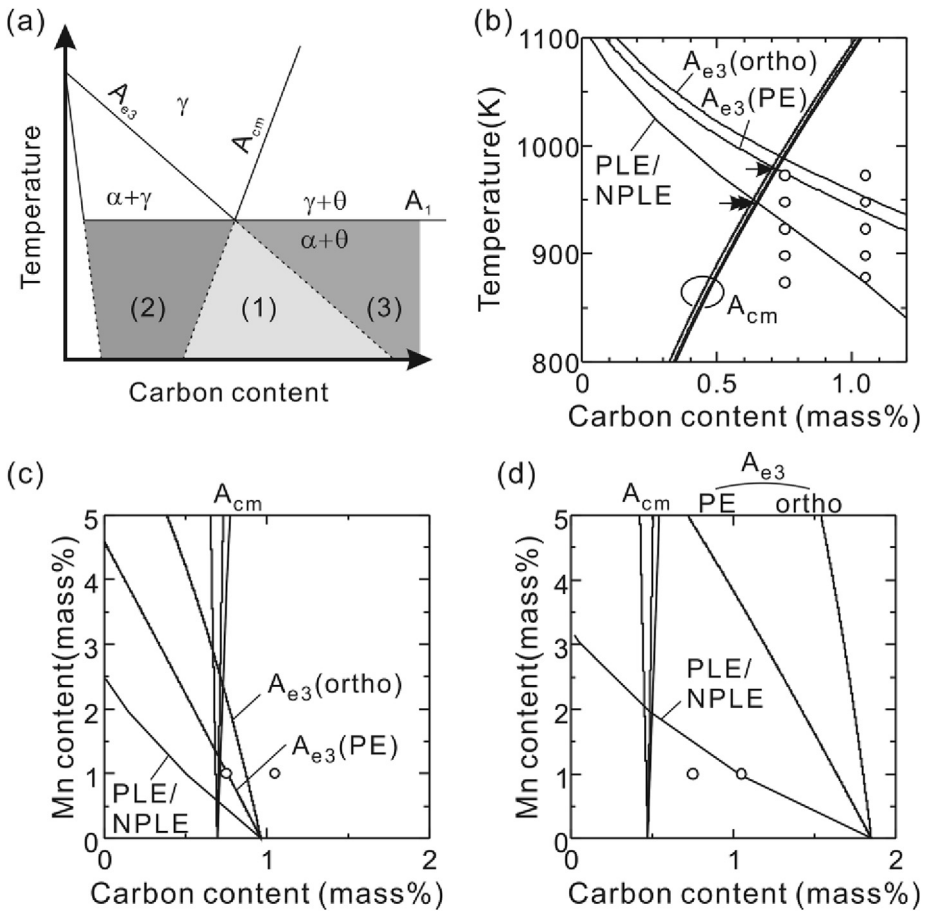

Fig. 1. (a) Schematic Fe–C phase diagram showing transformation condition for ferrite and pearlite, (b) isopleth phase diagram of Fe–1Mn–C system, (c) and (d) isothermal phase diagram of Fe–Mn–C system at 873 K (600 °C) and 973 K (700 °C), respectively. Circles in (b), (c) and (d) show the conditions employed in this study. Acm lines consist of ortho, PE and NPLE/PLE from low to high carbon content. Single and double arrows in (b) represent eutectoid compositions and temperatures for para equilibrium and negligible partitioning local equilibrium (NPLE).

from 823 K (550 °C) to 923 K (650 °C) which are typical temperatures for the patenting process in the production of steel wire, the formation of GB-α was reported [11–13]. However details of the formation of GB-α at the lower transformation temperatures are still unclear. Since it is generally believed that fully pearlitic structure forms in eutectoid steels, the formation of GB-α in the eutectoid and hypereutectoid steels is in particularly curious.

Therefore, the present study aims to investigate the effects of carbon content and transformation temperature on the isothermally transformed microstructure in high carbon steels, and to clarify the formation mechanism of GB-α.

2. Experimental procedure

Fe-1 mass% Mn-(0.75 and 1.05) mass% C steels were used and hereafter they are noted as 0.75C and 1.05C alloys, respectively. 1 mass% Mn was added to increase hardenability of the alloys to prevent transformation during cooling. The chemical compositions of the alloys were shown in Table 1. Fig. 1(b), (c) and (d) represent isopleth of a Fe–1Mn–C system and isothermal sections of a Fe–C–Mn system at 873 K (600 °C) and 973 K (700 °C), respectively. Equilibrium calculations were conducted by using ThermoCalc with TCFE5 database where full equilibrium (ortho) and quasi-equilibriums assuming no macroscopic partitioning of Mn, that are para equilibrium (PE) and negligible partitioning local equilibrium (NPLE), were considered. Carbon content of the 0.75C alloy is close to the eutectoid one under PE as indicated by a single arrow in Fig. 1(b) and hypereutectoid under local equilibrium as indicated by a double arrow. In order to eliminate segregation of Mn, hot-rolled plates were homogenized at 1453 K (1180 °C) for 345.6 ks in Ar encapsulated silica tube. Small rectangular specimens, $2 \times 4 \times 12$ mm^3, were cut from the homogenized specimens. They were austenitized at 1273 K (1000 °C) for 1.8 ks subsequently quenched into a salt bath and isothermally transformed at temperatures ranging from 1003 K (730 °C) to 873 K (600 °C) for various time of periods followed by quenching into water. Transformation temperatures employed in this study are above and below eutectoid temperature under local equilibrium indicated by the double arrow in Fig. 1(b). Nominal prior γ grain sizes in the 0.75C and 1.05C alloys evaluated by a planimetric procedure are 168 μm and 209 μm, respectively. In order to investigate effects of γ grain size, austenitization at 1173 K (900 °C) for 0.6 ks or 1373 K (1100 °C) for 1.8 ks was also employed for the 0.75C alloy, resulting in the nominal γ grain sizes of 78 μm and 617 μm, respectively.

After the heat treatments, the specimens were cut, mechanically polished by emery paper, diamond sprayed buff and colloidal

Table 1
Chemical compositions in mass% of the alloys used in this study.

Alloy	C	Si	Mn	P	S	N	Fe
0.75C	0.74	<0.003	1.00	<0.002	0.0008	<0.002	Bal.
1.05C	1.04	<0.003	0.99	<0.002	0.0009	<0.002	Bal.