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Effect of C Content on the Microstructure and Tensile Properties of Lightweight Ferritic Fe-8Al-5Mn-0.1Nb Alloy

A. Zargaran[1,*], H. S. Kim[1], J. H. Kwak[2], and Nack J. Kim[1]

[1]POSTECH, Graduate Institute of Ferrous Technology, Pohang 790-784, Korea
[2]Technical Research Lab., POSCO, Gwangyang, Korea

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Microstructure and tensile properties of ferritic Fe-8Al-5Mn-0.1Nb lightweight steels with different C contents (0.005, 0.02, and 0.05 wt.%) have been investigated in the present study. It shows that the microstructure becomes more elongated along the rolling direction, i.e., increasing propensity towards unrecrystallization with an increase in C content. This is mainly due to the effect of NbC on retarding the dynamic recrystallization of ferrite during hot rolling, which is active for higher C (0.02C and 0.05C) containing alloys. In the case of the 0.05C alloy, there is an additional precipitation of κ-carbide particles, which also retard the dynamic recrystallization of ferrite during hot rolling, resulting in a much more elongated structure in the 0.05C containing alloy than in the 0.02C alloy in as-hot rolled condition. Although κ-carbide particles retard the dynamic recrystallization of ferrite during hot rolling, they play an opposite role during final annealing, i.e., promoting static recrystallization by the operation of particle-stimulated nucleation mechanism, resulting in the development of homogeneously distributed fine grains in the 0.05C alloy. As a result, the 0.05C alloy shows higher strength and larger elongation than the lower C containing alloys.

Keywords: lightweight steels, ferrite, NbC, recrystallization, grain refinement

1. INTRODUCTION

There is a large demand for lightweight alloys with a high specific strength (strength to weight ratio) and a good formability for automotive applications. Along with increasing the strength of alloys, density reduction is another approach for achieving higher specific strength. It has been known that the density of steels can be reduced by the addition of Al and therefore there is a growing interest in developing Al containing steels, so-called lightweight or low density steels [1-3]. Depending on the type of major constituent phase, these lightweight steels can be classified into several types; ferritic [1,4-13], austenitic [14,15] and ferrite-austenite duplex types [9,16,17]. Having lower mechanical properties compared to those of austenitic or duplex steels [2], ferritic lightweight steels have received less attention than other types. Although there has been no systematic study on the effect of grain size on the mechanical properties of ferritic lightweight steels, analyses of their microstructures available in literature show that their grain sizes are rather coarse (over 40 μm) in final products [5-12]. Unlike the austenitic and duplex lightweight steels, mechanical properties, particularly ductility, of ferritic lightweight steels would be strongly dependent on the grain size due to their bcc structure. One of the most effective ways to refine the grain size of steels is to use precipitates which can affect the recrystallization behavior and grain growth during thermomechanical treatments (TMTs). Microalloying additions of Nb, V, and/or Ti have been traditionally used to control the grain size of numerous high strength low alloy (HSLA) steels [18,19]. The precipitation of NbC is known to delay recrystallization of austenite during hot rolling of HSLA steels, and it is expected that NbC precipitates would also affect the recrystallization of ferrite during hot rolling of ferritic lightweight steels.

The effect of Nb on recrystallization behavior will be dependent on the contents of both Nb and C since they affect the precipitation start temperature as well as the fraction of NbC particles. However, on the other hand, it has been reported that the amount of C in ferritic lightweight steels should be kept as low as possible (less than 100 ppm) to eliminate the formation of grain boundary κ-carbide, $(Fe,Mn)_3AlC$, particles which have deleterious effect on ductility [4-6]. Therefore, the C content should be optimized when one tries to utilize the precipitation of NbC for grain size control of ferritic lightweight steels. In the present paper, the effect of different content of C on the microstructural evolution and tensile properties of Nb-added ferritic lightweight steels have

*Corresponding author: Alireza@postech.ac.kr
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been investigated.

2. EXPERIMENTAL PROCEDURE

Three alloys with nominal compositions of Fe-8Al-5Mn-0.1Nb-0.005C, Fe-8Al-5Mn-0.1Nb-0.02C and Fe-8Al-5Mn-0.1Nb-0.05C in wt.% were prepared by vacuum induction melting. After homogenization treatment at 1200 °C for 1h, they were rough-rolled to a thickness of 70 mm with a finish rolling temperature above 1100 °C, then they were reheated for 30 min at 1200 °C and hot rolled at 1100 °C with a finish rolling temperature of 1000 °C to a thickness of 20 mm. The hot rolled plates then warm rolled at 650 °C to 3 mm thick sheets. After an intermediate annealing at 850 °C for 15 min, the sheets were warm rolled again at 650 °C to 1.5 mm and they were subsequently cold rolled to 1 mm in thickness. The cold rolled sheets were finally annealed at 750 °C for 1 h. Microstructure was analyzed by optical microscopy (OM), scanning electron microscopy (SEM) and electron backscatter diffraction (EBSD). The average grain size was measured for the grains having minimum misorientation angles of 15°. Longitudinal tensile tests were conducted using the specimens with a gauge length of 12.6 mm, a gauge width of 5 mm, and a gauge thickness of 1 mm at the strain rate of $10^{-3}s^{-1}$.

3. RESULTS

3.1. Microstructure

Cross sectional optical micrographs of the alloys in as-hot rolled condition are shown in Fig. 1. Microstructures of all alloys contain very coarse grains elongated along rolling direction along with some partially recrystallized grains. The 0.005C alloy has the largest volume fraction of partially recrystallized grains among the studied alloys. The size of partially recrystallized grains is also the largest in the 0.005C alloy. While both 0.02C and 0.05C alloys show similar size and volume fraction of partially recrystallized grains, the morphology of unrecrystallized elongated grains are much more uniform in the 0.05C alloy than in the 0.02C alloy. Figure 2a shows the presence of fine (up to ~300 nm) NbC particles precipitated within ferrite matrix of the 0.02C alloy. The presence of κ-carbide particles has also been found in the 0.02C alloy, but only in limited areas. The precipitation of κ-carbide particles becomes more evident in the 0.05C alloy as shown in Fig. 2b. It also shows that the size of NbC particles is larger in the 0.05C alloy than in the 0.02C alloy and these coarse NbC particles presumably originate from the ones formed in pre-hot rolling stage due to the excess amount of C in the 0.05C alloy. It is noted that there are also very fine (up to ~ 50 nm) NbC particles besides the coarse ones in the 0.05C alloy as shown in Fig. 2c.

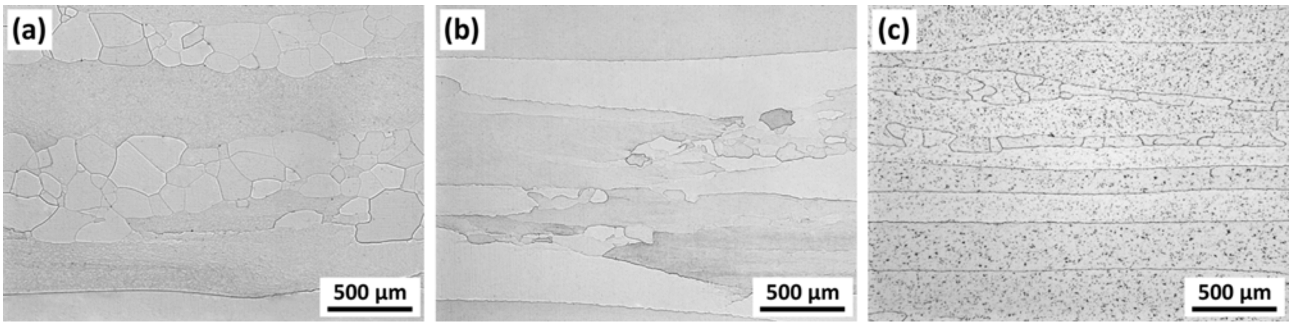

Fig. 1. Cross-sectional optical micrographs of the alloys in as-hot rolled condition: (a) Fe-8Al-5Mn-0.1Nb-0.005C, (b) Fe-8Al-5Mn-0.1Nb-0.02C, and (c) Fe-8Al-5Mn-0.1Nb-0.05C. Rolling direction: horizontal.

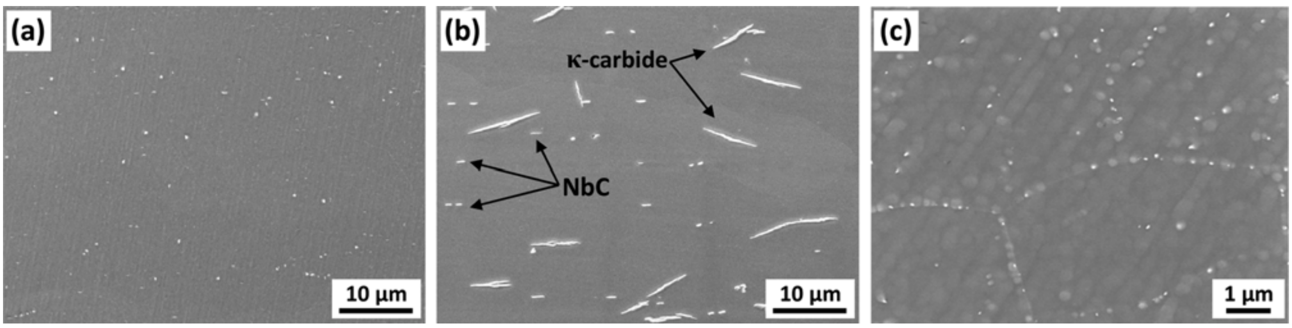

Fig. 2. SEM micrographs of as-hot rolled alloys showing the formation of (a) NbC in Fe-8Al-5Mn-0.1Nb-0.02C, (b) both NbC and κ-carbide in Fe-8Al-5Mn-0.1Nb-0.05C, and (c) fine NbC in Fe-8Al-5Mn-0.1Nb-0.05C.

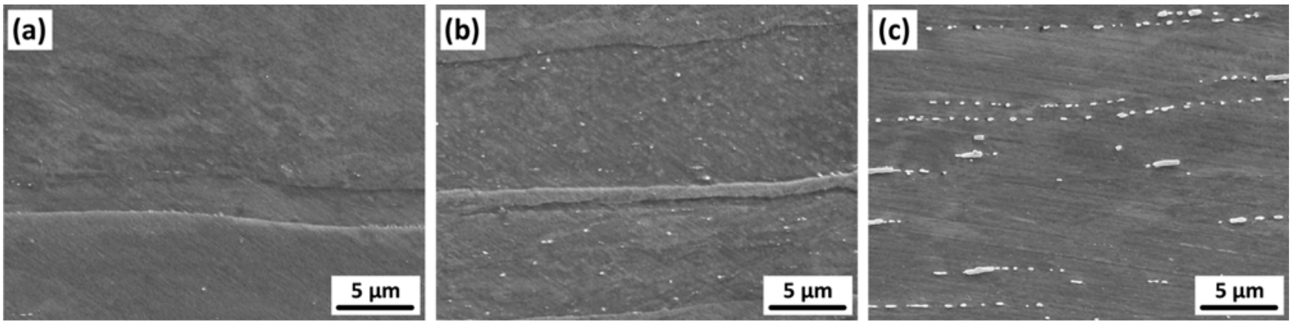

Fig. 3. SEM micrographs of as-warm rolled alloys. (a) Fe-8Al-5Mn-0.1Nb-0.005C, (b) Fe-8Al-5Mn-0.1Nb-0.02C, and (c) Fe-8Al-5Mn-0.1Nb-0.05C.

As shown in Fig. 3, microstructural analyses of the 0.005C, 0.02C and 0.05C alloys after the first warm rolling show that dynamic recrystallization does not occur during the first warm rolling. Instead grains become more elongated along rolling direction compared to the grain structure in as-hot rolled condition. The coarse needle-shaped κ-carbide particles present along grain boundaries and within matrix in as-hot rolled 0.05C alloy are broken and refined into smaller globular-shape particles by warm rolling as shown in Fig. 3c.

Subsequent annealing of the first warm rolled alloys at 850 °C results in a heterogeneous microstructure consisting of recrystallized grains near the surface of sheets and unrecrystallized at the center of sheets as shown in the inverse pole figure (IPF) maps of Fig. 4. The average sizes of recrystallized grains are almost same (~33 μm - ~38 μm) regardless of C content, suggesting that NbC and κ-carbide particles are not effective to refine the size of recrystallized grains at this TMT stage. The presence of unrecrystallized grains in a band form also indicates that the amount of deformation received during warm rolling is not sufficient for the recrystallization to occur in all alloys. Nevertheless, the microstructure of the 0.05C alloy shows the smallest volume fraction of unrecrystallized grains among the alloys.

Figure 5 shows the IPF maps of the warm/cold rolled sheets after final annealing at 750 °C. It can be seen that the microstructure of the 0.005C alloy is heterogeneous containing both fine and coarse grains and its average grain size (~20 μm) is coarser than those of other alloys (Fig. 5a). The microstructure of the 0.02C alloy is also heterogeneous containing both fine and coarse grains (Fig. 5b). On the other hand, the 0.05C alloy has quite homogeneous microstructure with an average grain size of ~14 μm (Fig. 5c). It also shows

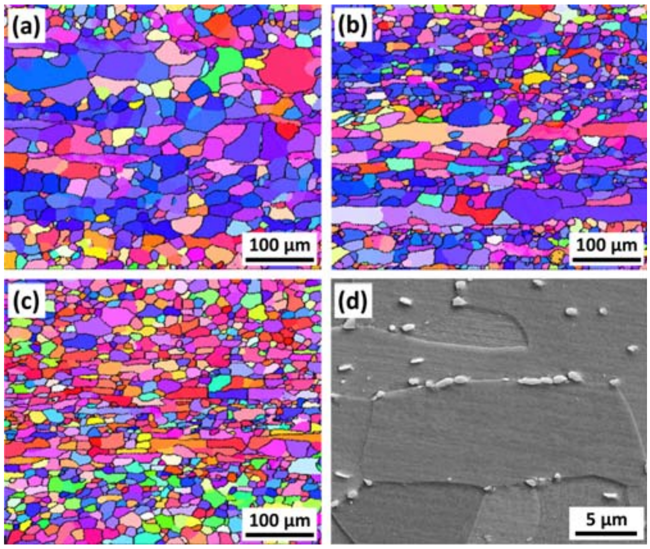

Fig. 5. Microstructure of the cold rolled alloys after final annealing: (a) IPF map of Fe-8Al-5Mn-0.1Nb-0.005C, (b) IPF map of Fe-8Al-5Mn-0.1Nb-0.02C, (c) IPF map of Fe-8Al-5Mn-0.1Nb-0.05C, and (d) SEM image of Fe-8Al-5Mn-0.1Nb-0.05C.

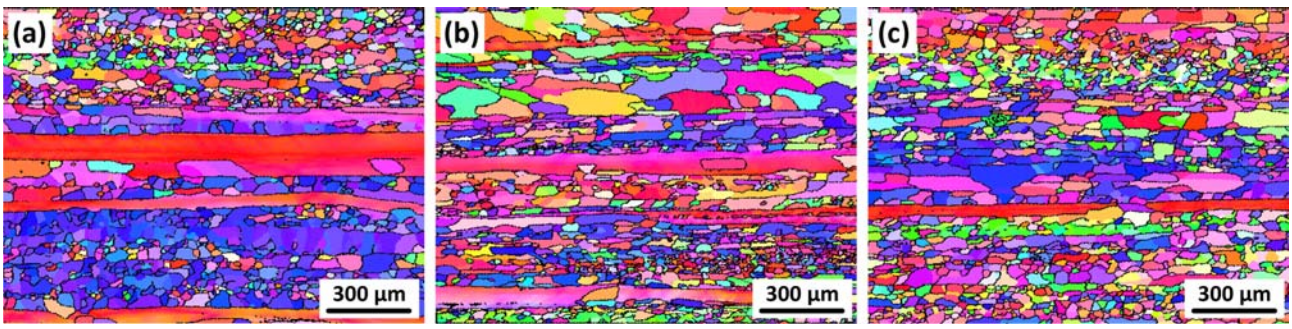

Fig. 4. IPF maps of the warm rolled alloys after intermediate annealing at 850 °C: (a) Fe-8Al-5Mn-0.1Nb-0.005C, (b) Fe-8Al-5Mn-0.1Nb-0.02C, (c) Fe-8Al-5Mn-0.1Nb-0.05C.

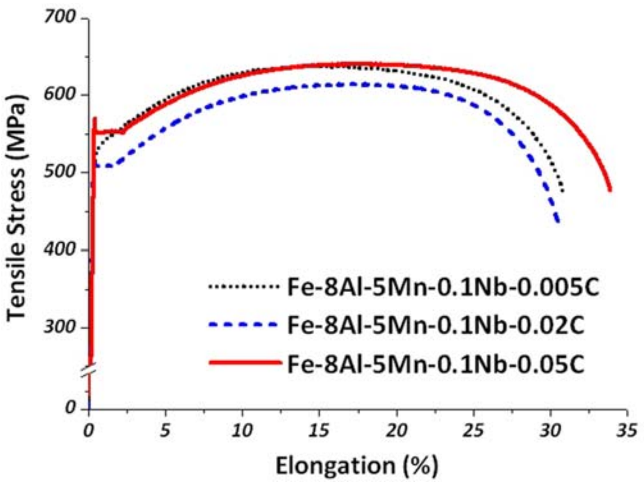

Fig. 6. Engineering tensile stress-strain curves of the alloys.

that the grain boundaries of the 0.05C alloy is decorated with globular-shape κ-carbide particles (Fig. 5d) after final annealing.

3.2. Tensile properties

Engineering stress-strain curves of the alloys with different C contents are shown in Fig. 6. While the 0.005C alloy shows continuous yielding during tensile deformation, there is an occurrence of yield point in the 0.02C and 0.05C alloys due to the well known effect of C on dislocation movement. It suggests that the C contents in the 0.02C and 0.05C alloys are in excess of the amount needed for precipitation of NbC and κ-carbide particles.

Interestingly, despite its finer grain size, the 0.02C alloy shows lower yield and tensile strengths than the 0.005C alloy. This might be due to the solid solution strengthening effect of Nb that is remained in solution without being associated with NbC. Assuming the stoichiometric composition of NbC, the amount of Nb in solution for the 0.005C and 0.02C alloys are about 0.06 wt.% and 0 wt.%, respectively. It is also found that although the 0.02C and 0.05C alloys have similar average grain sizes, the 0.05C alloy shows higher strength and larger ductility than the 0.02C alloy. The volume fractions of NbC and κ-carbide particles are quite small in both alloys (~1% and ~0.2% in the 0.05C alloy) and therefore, the difference in strength between these alloys cannot be explained by the precipitation strengthening by NbC and κ-carbide particles. As mentioned previously, the microstructure of the 0.02C alloy is heterogeneous consisting of fine and coarse grains, which is responsible for the worse tensile properties of the 0.02C alloy than the 0.05C alloy, similar to the observation made in other alloy systems [20].

There is also a large difference in the fracture mode among the alloys. Although all alloys show the occurrence of necking during tensile test, low C containing alloys, particularly the 0.005C alloy, show the occurrence of severe delamination as shown in Fig. 7. Such occurrence of severe delamination during tensile test originates from the presence of a large volume fraction of coarse elongated grains along the rolling direction in the 0.005C alloy as shown in Fig. 5a. Although the 0.05C alloy also shows the occurrence of delamination, the degree of delamination is not large since the alloy has fine and equiaxed grain structure as shown in Fig. 5c.

4. DISCUSSION

It is well known that the dynamic recrystallization of austenite in conventional steels can be retarded by the additions of strong carbide formers such as Nb, V, and Ti [18, 19]. It is expected that the additions of Nb and C to the ferritic lightweight steels can also retard the dynamic recrystallization of ferrite during hot rolling. It has been shown that the degree of dynamic recrystallization decreases with an increase in C content in the present study. That is, the propensity for unrecrystallization increases with an increase in C content. As mentioned previously, the volume fractions of NbC and κ-carbide particles increase with an increase in C content. In fact, there are only NbC particles in the 0.005C and 0.02C alloys in as-hot rolled condition, while higher C (0.05C) containing alloy has κ-carbide particles besides

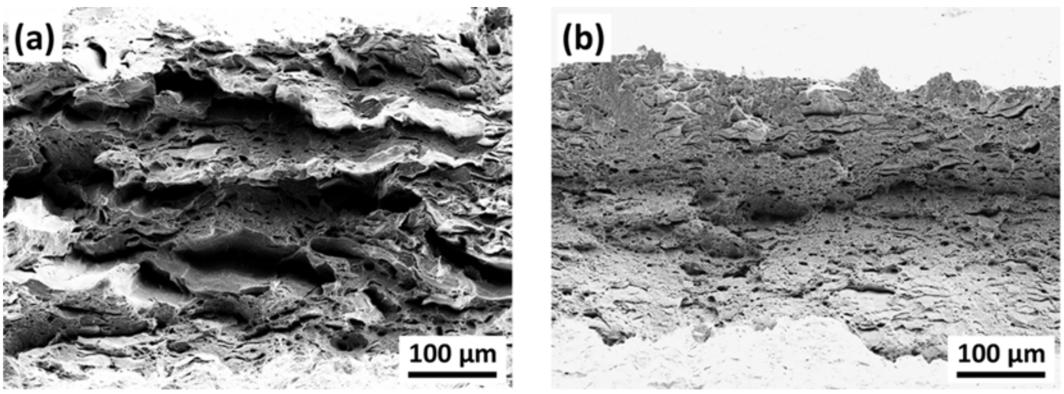

Fig. 7. Fracture surfaces of (a) Fe-8Al-5Mn-0.1Nb-0.005C and (b) Fe-8Al-5Mn-0.1Nb-0.05C.

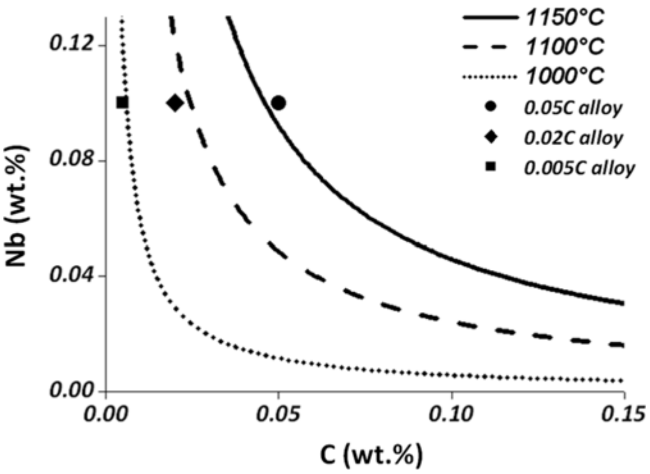

Fig. 8. Plot of NbC solubilities at various temperatures.

NbC particles. Figure 8 shows the equilibrium solubility lines of NbC at different temperatures calculated based on the data reported by Klymko and Sluiter [21]. It suggests that NbC particles can precipitate during hot rolling (at between 1000 and 1100 °C) for the 0.02C and 0.05C alloys, but cannot for the 0.005C alloy. Although the precipitation of NbC particles could occur during hot rolling at higher temperature than the one shown in Fig. 8 due to the occurrence of strain-induced precipitation, it is clear that the precipitation of NbC would be minimal for the 0.005C alloy during hot rolling, resulting in a development of a large volume fraction of partially recrystallized grains as shown in Fig. 1a. The size of partially recrystallized grains is also rather large (~160 μm) due to the insufficient number of NbC particles which can impede grain growth. It has been suggested that the dynamic recrystallization of deformed alloys can be retarded by the solute drag mechanism [22]. Although most of Nb would be in solid solution rather than being associated with NbC in the 0.005C alloy during hot rolling, the occurrence of dynamic recrystallization (albeit partial) indicates that the solute drag is not effective in retarding dynamic recrystallization in the present case. Such a phenomenon - less effectiveness of solute drag by Nb compared with grain boundary pinning by precipitated NbC - has also been observed on dynamic recrystallization of austenite in HSLA steels [19].

Although the precipitation of NbC during hot rolling can retard the dynamic recrystallization of ferrite, it appears that the number density of NbC particles is not sufficient to completely prevent the dynamic recrystallization in the case of the 0.02C containing alloy, resulting in a heterogeneous microstructure as shown in Fig. 1b. In the case of the 0.05C containing alloy, on the other hand, there is a precipitation of additional phase, κ-carbide, besides NbC. κ-carbide particles would also retard the dynamic recrystallization of ferrite during hot rolling. It has been observed that the pancake-shape ferrite grain structure can form in the κ-carbide containing Fe-8.8Al-0.1C alloy after hot rolling [12], indicating the role of κ-carbide particles in retarding the dynamic recrystallization of ferrite during hot rolling. Therefore, the microstructure of the 0.05C containing alloy mostly consists of unrecrystallized ferrite grains with only a few recrystallized grains as shown in Fig. 1c.

It is well known that nucleation for both dynamic and static recrystallization occurs at preferred sites such as prior grain boundaries, deformation bands, and second-phase particles [22]. Finer initial grain size would lead to a finer final grain size after annealing since upon deformation the initially finer grains contain a larger number of nuclei for static recrystallization such as grain boundaries and deformation bands than the initially coarser grains. As mentioned previously, the 0.02C and 0.05C alloys have grain structure much more elongated than the one in the 0.005C alloy due to the larger volume fraction of NbC particles in the formers. Therefore, it can be expected that the 0.02C and 0.05C alloys have finer grain sizes after final annealing than the 0.005C alloy.

The important role of κ-carbide particles on promoting static recrystallization should be noted. As shown in Figs. 5a-c, the 0.005C and 0.02C alloys show heterogeneous microstructures consisting of coarse and fine grains, while the 0.05C alloy shows the homogeneous distribution of fine grains. Assuming that all of Nb is associated with NbC in the case of the 0.02C and 0.05C alloys, then any difference in the microstructure between the 0.02C alloy and the 0.05C alloy can be thought to be due to the presence of κ-carbide particles in the latter. As mentioned previously, κ-carbide particles retard the dynamic recrystallization of ferrite during hot rolling. However, these κ-carbide particles can act as nucleation sites for static recrystallization during final annealing. It is well known that coarse (over 1 μm) second-phase particles promote static recrystallization by the particle-stimulated nucleation (PSN) mechanism [22]. Moreover, these κ-carbide particles can also retard growth of recrystallized grains as shown in Fig. 5d, resulting in the development of homogeneously distributed fine grains in the 0.05C alloy.

Comparison of tensile properties of the present alloys with those of other ferritic lightweight alloys show that the present alloys have better combination of strength and ductility than others. It has been reported that addition of Mn increases the strength by ~20 MPa per 1 wt.% Mn by solid solution strengthening in ferrite [13]. Assuming an increase of 100 MPa by addition of 5 wt.% Mn, both yield and tensile strength values of Fe-8.5Al-0.1Nb alloy reported by Frommeyer *et al.* [1] and Fe-8.1Al-0.1Ti alloy reported by Rana *et al.* [6] are quite comparable to those of the present alloys. However, the present alloys show much better elongation values than other alloys. For example, Fe-8.5Al-0.1Nb

[1] and Fe-8.1Al-0.1Ti [6] alloys have the elongation values of 25% and 17%, respectively, which are much smaller than the elongation (34%) of the present Fe-8Al-5Mn-0.1Nb-0.05C alloy. It is noted that other alloys have grain sizes much coarser than the present alloy (e.g., ~90 μm for Fe-8.1Al-0.1Ti alloy [6] vs. ~14 μm for the present Fe-8Al-5Mn-0.1Nb-0.05C alloy), indicating the important role of grain size in tensile properties of lightweight ferritic steels.

5. CONCLUSIONS

In the present paper, the effect of C contents (0.005, 0.02, and 0.05 wt.%) on the microstructure and tensile properties of ferritic Fe-8Al-5Mn-0.1Nb lightweight steels has been investigated.

(1) The precipitation of NbC occurs during hot rolling of the 0.02C and 0.05C alloys, but not in the 0.005C alloy. In addition to NbC, there is also a precipitation of κ-carbide during hot rolling of the 0.05C alloy.

(2) While all alloys show the occurrence of partial dynamic recrystallization, the degree of dynamic recrystallization decreases with an increase in C content, resulting in a more homogeneous distribution of more elongated pancake-shape structure in the 0.05C alloy than the 0.005C and 0.02C alloys in as-hot rolled condition.

(3) Detailed microstructural analyses of the alloys in as-hot rolled condition show that the dynamic recrystallization of ferrite is retarded by not only NbC particles but also κ-carbide particles.

(4) While κ-carbide particles retard dynamic recrystallization during hot rolling, they play an opposite role during subsequent TMTs such that they promote static recrystallization during annealing by the operation of PSN mechanism, resulting in a more homogeneous distribution of finer grains in the 0.05C alloy than the 0.005C and 0.02C alloys after final annealing.

(5) The 0.05C alloy shows higher strength and larger elongation than the 0.005C and 0.02C alloys due to the more favorable microstructure in the former than in the latter.

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