

Relationship between Microstructures and Tensile Properties of an Fe-30Mn-8.5Al-2.0C Alloy

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Owing to the presence of a large amount of fine (Fe,Mn)₃AlC carbides within austenite (γ) matrix, the tensile property of the Fe-30%Mn-8.5Al%-2.0%C (in mass%) alloy in the as-quenched condition was clearly superior to that of the as-quenched FeMnAlC ($C \leq 1.3\%$) alloys investigated by previous workers. After being aged at 823 K for 3 h, the present alloy could possess high yield strength up to 1262 MPa with an excellent 32.5% elongation. With almost equivalent ductility, the yield strength obtained was about 16% higher than that of the FeMnAlC ($C \leq 1.3\%$) alloys after solution heat-treatment or controlled-rolling followed by an optimal aging at 823 K. Additionally, due to the pre-existing fine (Fe,Mn)₃AlC carbides within the γ matrix in the as-quenched alloy, the aging time required for attaining the optimal combination of strength and ductility was much less than that of the FeMnAlC ($C \leq 1.3\%$) alloys aged at 823 K. When the present alloy was aged at 823 K for a time period longer than 4 h, both the strength and ductility were drastically dropped due to the occurrence of γ_0/κ (γ_0 : carbon-deficient austenite) lamellar structure on the γ/γ grain boundaries. [doi:10.2320/matertrans.M2010013]

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1. Introduction

A lot of effort has been made to develop austenitic FeMnAlC alloys as high-strength, high-ductility alloy steels. In their studies, it is seen that when an alloy with a chemical composition in the range of Fe-(26~34)%Mn-(7~11)%Al-(0.54~1.3)%C was solution heat-treated and then quenched, the microstructure of the alloy was single-phase austenite (γ).¹⁻¹⁷ When the as-quenched alloy was aged at temperatures ranging from 773 to 873 K for moderate times, fine (Fe,Mn)₃AlC carbides started to precipitate coherently within the γ matrix, and no precipitates were found on the γ/γ grain boundaries. The resulting microstructure could possess a remarkable combination of strength and ductility.⁹⁻¹⁴ However, when the aging time was increased within the temperature range, a $\gamma \rightarrow \alpha$ (ferrite) + κ reaction, a $\gamma \rightarrow \alpha + \beta$ -Mn reaction, a $\gamma \rightarrow \kappa + \beta$ -Mn reaction, a $\gamma \rightarrow \alpha + \kappa + \beta$ -Mn reaction, or a $\gamma \rightarrow \gamma_0 + \kappa$ reaction occurred on the γ/γ grain boundaries.^{13-17,22} The κ phase is also an (Fe,Mn)₃AlC carbide, which was formed on the γ/γ grain boundaries as a coarse particle. For convenience, κ' carbide and κ carbide are used to represent the (Fe,Mn)₃AlC carbide formed coherently within the γ matrix and heterogeneously on the γ/γ grain boundaries.^{11,13-17} The formation of the coarse κ carbides and β -Mn precipitates on the grain boundaries resulted in the embrittlement of the alloys.^{12-15,22}

In addition to the extensive studies of the austenitic FeMnAlC alloys with $C \leq 1.3\%$, the microstructural developments in the FeMnAlC alloys with $C > 1.3\%$ have also been given in several literatures.¹⁸⁻²⁰ In these studies, it is seen that the as-quenched microstructure of Fe-(29~30)%Mn-(7.7~9)%Al-(1.5~2.5)%C alloys was γ

phase containing fine κ' carbides.^{19,20} The fine κ' carbides were formed within the γ matrix by spinodal decomposition during quenching.²⁰ This is quite different from that observed in the austenitic FeMnAlC alloys with $C \leq 1.3\%$, in which the fine κ' carbides could only be observed in the aged alloys. When the as-quenched alloy was aged between 823 and 1073 K, the fine κ' carbides grew and a heterogeneous reaction of $\gamma + \kappa' \rightarrow \gamma_0 + \kappa$ occurred on the γ/γ grain boundaries. With increasing aging time, the heterogeneous reaction became predominant. Consequently, the stable microstructure of the alloy was found to be a mixture of ($\gamma_0 + \kappa$) lamellar structure.^{19,20} In contrast to the studies of the microstructures, information concerning the mechanical properties of the austenitic FeMnAlC alloys with $C \geq 1.3\%$ is very deficient. We are aware of only one article,¹⁹ in which the tensile properties of the Fe-26 at%Mn-15 at%Al-8 at%C (Fe-30%Mn-8.5%Al-2.0%C, in mass%) alloy were examined. However, all of their examinations were only focused on the alloy aged at 1073 K. No information was provided for the alloy aged at lower temperatures. Therefore, the purpose of this work is an attempt to investigate the relationship between the microstructures and the tensile properties of the Fe-30%Mn-8.5%Al-2.0%C alloy in the as-quenched condition and aged at 823 K for various times.

2. Experimental Procedure

The Fe-30%Mn-8.5%Al-2.0%C (in mass%) alloy was prepared in a vacuum induction furnace by using 99.7% iron, 99.9% manganese, 99.9% aluminum and pure carbon powder. The melt was cast into a 20 × 30 × 100 mm steel mold. After being homogenized at 1523 K for 12 h under a protective argon atmosphere, the ingot was hot-rolled to a final thickness of 6 mm. The plate was subsequently solution heat-treated at 1473 K for 2 h and then quenched into room-temperature water rapidly. Aging processes were

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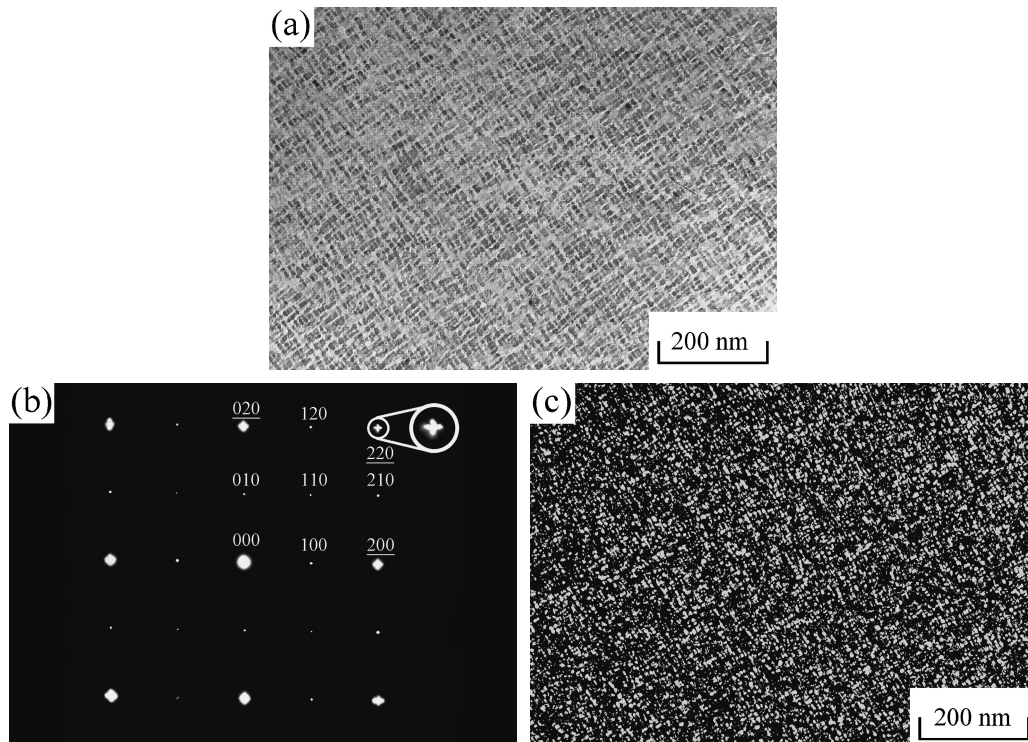


Fig. 1 Transmission electron micrographs of the as-quenched alloy: (a) BF, (b) an SADP taken from the mixed region of austenite matrix and fine κ' carbides. The foil normal is $[001]$ (hkl : γ phase; hkl : κ' carbide), and (c) $(100)_{\kappa'}$ DF image.

performed at 823 K for various times in a vacuum furnace and then quenched into water. Scanning electron microscopy (SEM) and transmission electron microscopy (TEM) were used to examine the microstructures and the tensile fracture surface as well as free surface. TEM specimens were prepared by using a double-jet electropolisher with an electrolyte of 60% acetic acid and 40% ethanol. Tensile tests were carried out at room-temperature with an Instron tensile testing machine at a strain rate of $5 \times 10^{-4} \text{ s}^{-1}$. Tensile test specimens were plates having 50.0 mm gauge length, 12.5 mm width and 5.0 mm thickness. The yield strength was measured at 0.2% offset strain, and the percent elongation was the total elongation measured after fracture.

3. Results and Discussion

Transmission electron microscopy examination indicated that in the as-quenched condition, the microstructure of the alloy was γ phase containing fine κ' carbides. A typical microstructure is shown in Fig. 1(a). Figure 1(b), a selected-area diffraction pattern taken from a mixed region covering the γ matrix and fine κ' carbides, indicates that the fine κ' carbides have an $L'1_2$ structure.^{16,17,20,21} In Fig. 1(b), it is also seen that satellites lying along $\langle 100 \rangle$ reciprocal lattice directions about the (200) and (220) reflections could be observed. The existence of the satellites demonstrates that the fine κ' carbides were formed during quenching by spinodal decomposition.²⁰ This is similar to that observed by the present workers in the as-quenched Fe-30%Mn-9%Al-2.0%C alloy.²⁰ Figure 2(a) is a bright-field (BF) electron micrograph of the as-quenched alloy aged at 823 K for 3 h, revealing that the fine κ' carbides grew within the γ matrix,

and no precipitates could be observed on the grain boundaries. However, when the alloy was aged at 823 K for 4 h, some coarse precipitates started to appear on the γ/γ grain boundaries. An example is shown in Fig. 2(b). Electron diffraction examinations indicated that the coarse precipitates on the grain boundaries were κ carbides. With continued aging at 823 K, the coarse κ carbides grew into the adjacent γ grains through a $\gamma + \kappa' \rightarrow \gamma_0 + \kappa$ reaction, as illustrated in Fig. 2(c). Figure 2(d), a selected-area diffraction pattern (SADP) taken from an area covering the κ carbide marked as “K” in Fig. 2(c) and its surrounding γ_0 phase, indicates that the orientation relationship between the κ carbide and γ_0 phase is $[001]_{\kappa} \parallel [001]_{\gamma_0}$ and $(100)_{\kappa} \parallel (100)_{\gamma_0}$. With increasing aging time at 823 K, the $\gamma + \kappa' \rightarrow \gamma_0 + \kappa$ reaction would proceed toward the whole austenite grains. Consequently, the stable microstructure of the present alloy at 823 K was a mixture of $(\gamma_0 + \kappa)$ phases. A typical microstructure is shown in Fig. 3. It is clear in Fig. 3(b) that the mixture of (γ_0 phase + κ carbide) has a lamellar structure.

Figure 4 shows the tensile properties of the present alloy in the as-quenched condition and aged at 823 K for various times. In Fig. 4, it is seen that the as-quenched alloy has ultimate tensile strength (UTS) 1105 MPa, yield strength (YS) 883 MPa, and an excellent 54.5% elongation. After being aged at 823 K for 3 h, the alloy can possess the highest UTS (1395 MPa) and YS (1262 MPa) with a good elongation of 32.5%, which may be attributed to the growth of the κ' carbides within the γ matrix and no precipitates on the grain boundaries. When the alloy was aged at 823 K for 4 h, both of the strength and elongation were slightly decreased. This result was due to the over-coarsening of the κ' carbides within the γ matrix and the presence of a small amount of κ

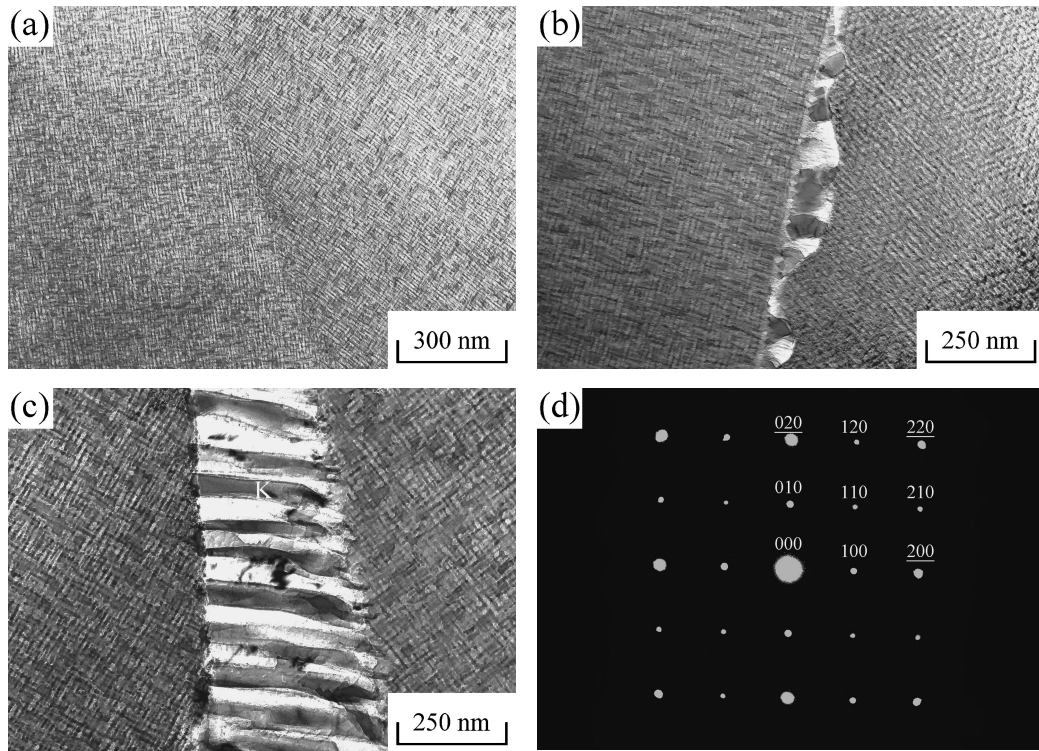


Fig. 2 (a) through (c), BF electron micrographs of the alloy aged at 823 K for (a) 3 h, (b) 4 h, and (c) 5 h, respectively. (d) an SADP taken from the κ carbide marked as "K" in (c) and its surrounding γ_0 phase. The foil normal is [001] (hkl: γ_0 phase; hkl: κ carbide).

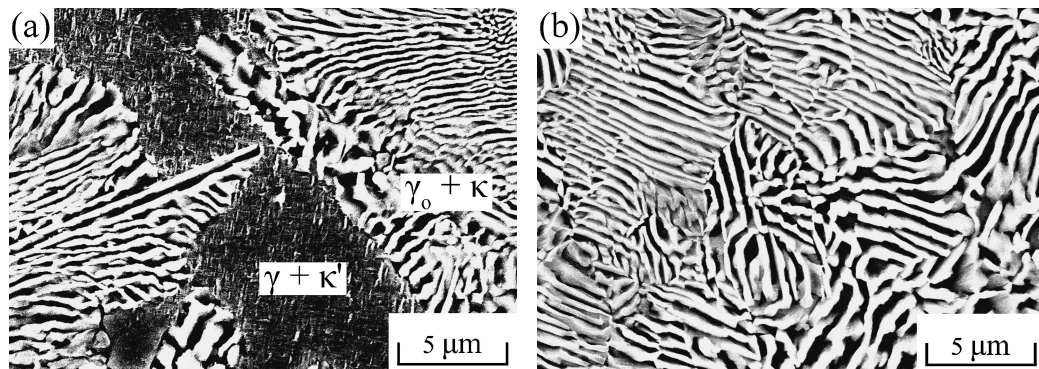


Fig. 3 Scanning electron micrographs of the alloy aged at 823 K for (a) 6 h, and (b) 7 h, respectively.

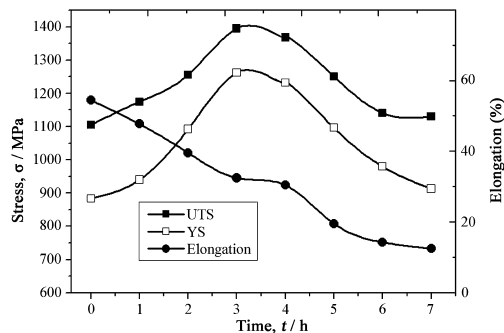


Fig. 4 Tensile properties of the alloy in the as-quenched condition and aged at 823 K for various times.

carbides on the grain boundaries. However, when the alloy was aged at 823 K for 5 h, the elongation was drastically dropped from about 30.5% to 19.6%. In order to clarify why the elongation was drastically dropped, fracture and free surface analyses were undertaken by using SEM. Figure 5(a), a fractograph of the alloy aged at 823 K for 4 h, reveals that the alloy had a ductile dimple fracture surface, and some dimples contain one or more κ carbides (as indicated by arrows). Figure 5(b) is a SEM micrograph taken from the free surface contiguous to the fracture surface, indicating that slip bands were generated over the specimen, and some isolated microvoids were formed along the grain boundaries (as indicated by arrows). It is clearly seen in Fig. 5 that the structure had a high resistance to crack propagation and exhibited self-stabilization under deformation. However, the fracture surface of the alloy aged at 823 K for 6 h revealed

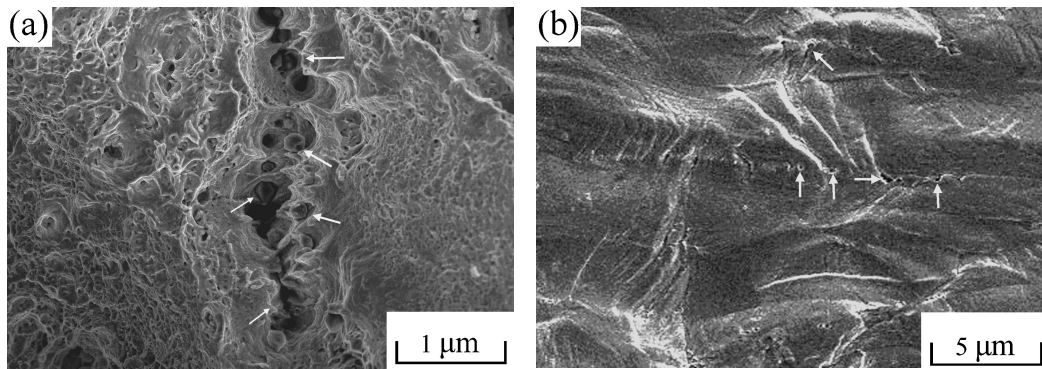


Fig. 5 Scanning electron micrographs of the alloy (aged at 823 K for 4 h) after tensile fractured. (a) fracture surface, and (b) free surface, respectively.

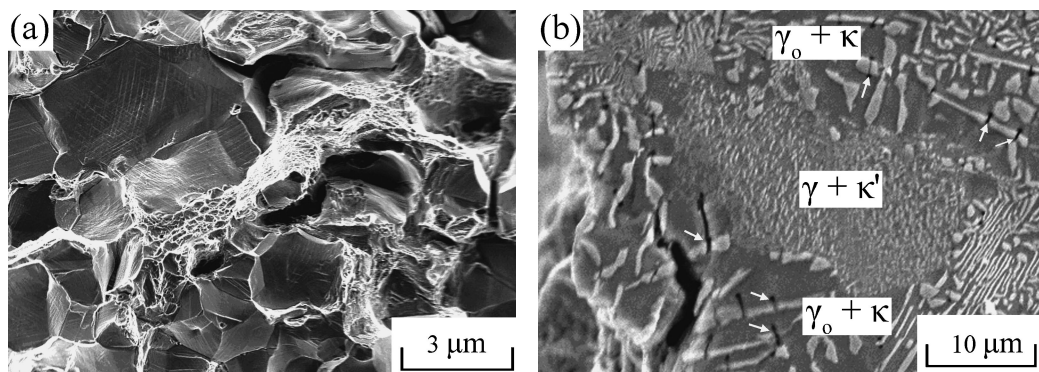


Fig. 6 Scanning electron micrographs of the alloy (aged at 823 K for 6 h) after tensile fractured. (a) fracture surface, and (b) free surface, respectively.

largely cleavage facets as well as intergranular fracture and a few dimples, as shown in Fig. 6(a). Figure 6(b) demonstrates that microcracks (as indicated by arrows) were observed only at coarse κ carbides within the γ_0/κ lamellar structure and no cracks could be observed in the $\gamma + \kappa'$ regions. Therefore, it is reasonable to believe that the existence of the γ_0/κ lamellar structure would be mainly responsible for the crack initiation, which led to the drastic drop of the ductility.

On the basis of the preceding results, some discussion is appropriate. According to the previous studies in the Fe-(26~34)%Mn-(7~11)%Al-(0.54~1.3)%C-(0~1.75%)(Nb+V+Mo+W) alloys,^{1-17,21-25)} it is seen that the as-quenched microstructure was single γ phase or γ phase with (Nb,V)C carbides. Depending on the chemical composition, the alloys in the as-quenched condition show various UTS ranging from 814 to 993 MPa, YS ranging from 423 to 552 MPa and elongation from 72 to 50%.^{4-9,21)} By the optimal aging treatments between 773 and 873 K for moderate times, a high density of fine κ' carbides started to precipitate coherently within the γ matrix and no precipitates were formed on the grain boundaries. The resulting microstructure could lead to the optimal combination of the mechanical strength and ductility. With an elongation better than about 30%, the values of 953~1259 MPa for UTS and 665~1094 MPa for YS could be attained.^{9-14,21-25)} Obviously, owing to contain a large amount of fine κ' carbides within the γ matrix, the tensile property of the present alloy in the as-quenched condition is not only superior to that of the as-quenched Fe-

(26~34)%Mn-(7~11)%Al-(0.54~1.3)%C-(0~1.75%)(Nb+V+Mo+W) alloys but comparable to that of the aged alloys. Furthermore, when the present alloy was aged at 823 K for 3 h, the UTS and YS could reach up to 1395 and 1262 MPa, respectively, with a good elongation of 32.5%. Compared to the previous studies, it is found that with almost equivalent ductility, the present alloy possesses yield strength about 16% higher than the Fe-(26~34)%Mn-(7~11)%Al-(0.54~1.3)%C-(0~1.75%)(Nb+V+Mo+W) alloys after the solution heat-treatment or controlled-rolling followed by the optimal aging.^{9-14,21-25)} The reason is probably that due to higher carbon content in the present alloy, a greater amount of κ' carbides could be formed within the γ matrix during aging. Additionally, in the previous studies of the austenitic FeMnAlC ($C \leq 1.3\%$) alloys aged at 823 K, it was generally concluded that the aging time required for attaining the optimal combination of strength and ductility was about 15~16 h.^{10-14,21-25)} Whereas, the aging time of the present alloy aged at 823 K was only about 3 h, which was attributed to the pre-existing fine κ' carbides within the γ matrix in the as-quenched condition.

Finally, it is worthwhile to point out that, in the previous study of the Fe-26 at%Mn-15 at%Al-8 at%C alloy (the chemical composition of the alloy is similar to that of the present alloy),¹⁹⁾ Kimura *et al.* reported that when the alloy was solution heat-treated at 1373 K for 1 h and then furnace cooled to room temperature, the alloy exhibited almost zero ductility due to a lot of coarse κ carbides on the γ/γ grain

boundaries. In order to improve the ductility, it was proposed in their study that the alloy was solution heat-treated followed by water quenching, and subsequent aging at 1073 K for up to 120 h. The resulting microstructure was a mixture of ($\gamma_0 + \kappa$) lamellar structure. By forming the γ_0/κ lamellar structure, the elongation could be improved to be about 12% with YS around 675 MPa and UTS around 1125 MPa.¹⁹ Obviously, the values of both strength and ductility are much lower than those obtained in the present study, and even lower than those obtained in the austenitic FeMnAlC ($C \leq 1.3\%$) alloys.⁴⁻¹⁴ Therefore, it is reasonable to deduce that the $\gamma + \kappa'$ structure is thought to be more favorable for both strength and ductility than the $\gamma_0 + \kappa$ lamellar structure.

4. Conclusions

The relationship between the microstructures and the tensile properties of the Fe-30%Mn-8.5%Al-2%C alloy was investigated. The obtained results are as follows:

- (1) The as-quenched microstructure of the present alloy was γ phase containing fine κ' carbides. The fine κ' carbides were formed within the γ matrix by spinodal decomposition during quenching. The tensile property of the as-quenched alloy was far superior to that of the as-quenched FeMnAlC ($C \leq 1.3\%$) alloys.
- (2) After being aged at 823 K for 3 h, the present alloy could possess a high yield strength up to 1262 MPa with an excellent 32.5% elongation. With almost equivalent ductility, the yield strength obtained was about 16% higher than that of the FeMnAlC ($C \leq 1.3\%$) alloys after solution heat-treatment or controlled-rolling followed by an optimal aging at 823 K.
- (3) Due to the pre-existing fine (Fe,Mn)₃AlC carbides within the γ matrix in the as-quenched alloy, the aging time required for attaining the optimal combination of strength and ductility was much less than that of the FeMnAlC ($C \leq 1.3\%$) alloys.
- (4) When the alloy was aged at 823 K for a time period longer than 4 h, the γ_0/κ lamellar structure was formed on the γ/γ grain boundaries. The γ_0/κ lamellar structure was mainly responsible for the crack initiation, which led to the drastic drop of the ductility.

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