Effects of Fe Content on Microstructure and Mechanical Properties of A206 Alloy

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The effects of Fe content on the microstructure (intermetallic compounds, available copper and strengthening precipitates) and mechanical properties of A206 alloy have been investigated. The network of CuAl_2 and needle-like (or platelet-like) Cu_2FeAl_7 compounds, which coexist in the α -Al inter-dendrites, characterize the typical solidification structures of A206 alloys. The volume fraction and number of Cu_2FeAl_7 needles, determined by image analysis, are proportional to the Fe content. The available copper and the supersaturated copper atoms in matrix in richer Fe alloys will decrease after the solution treatment. In DSC analysis, the smaller heat effect and the increase in peak temperature for precipitation θ' phase suggest that the amount and kinetics of precipitation are lowered in higher Fe content alloys. The relation between the mechanical properties and microstructure has been examined quantitatively. The tensile strength and hardness are mainly dominated by the volume fraction of θ' phase, and the elongation by the Cu_2FeAl_7 compound. All of these properties, especially the elongation, decrease when the Fe content increases because of reduced quantity of θ' phase and increased amount of Cu_2FeAl_7 compounds. The tensile strength decreases linearly with increasing Fe content.

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1. Introduction

The heat treatable aluminum casting A206 alloys have been widely used in aerospace and military industry owing to their excellent mechanical and physical properties.¹⁻³⁾ It is well known that the solidification structure of Al-Cu hypoeutectic alloy exists a network of Al-CuAl₂ eutectic phase at the α -Al inter-dendrites. Bäckerud *et al.*⁴⁾ have carried out a detailed study on the solidification of A206 alloy. They pointed out that the CuAl₂ eutectics would solidify at 537°C; and if the cooling rate was fast enough (4.5°C/s), the CuMgAl₂ and other complex eutectics would solidify at 500°C. Generally, alloys can be strengthened by heat treatment through solid solution treatment, quenching, and aging. The precipitated phases may include the major strengthening phase of CuAl₂ intermediate phase (θ' phase) and few CuMgAl₂ intermediate phase (S' phase). The differential scanning calorimetry (DSC) analysis has been used to investigate the precipitation kinetics of these precipitates.⁵⁻⁹⁾ The sequence of precipitation and the amount of precipitates forming at each stage during non-isothermal aging can also be obtained from this analysis. It is well known that the optimum strength and ductility can be obtained by precipitating large amount of θ' phase through T6 heat treatment. But in order to increase the stress corrosion resistance of A206 alloy, T4 or T7 is carried out instead of T6.

There are many inherent impurities in commercial aluminum alloys. Very high cost is concerned to remove those impurities. Iron is probably the most important one of these impurities since it will form intermetallic compounds. These iron compounds are hard and brittle. They may act as stress raisers or be points of weakness that reduce the strength and ductility of the alloy.^{3,10-12)} Jacob *et al.*¹¹⁾ studied Al-5%Cu alloy (0.23%Fe content) and pointed out that the iron-containing phase was a needle-like compound of Cu_2FeAl_7 . Another Chinese script of $(\text{CuFe})\text{Al}_3$ compound formed when

the Fe content reached 0.55%. Couture¹²⁾ stated that the formation of Cu_2FeAl_7 and $(\text{CuFe})\text{Al}_3$ in Al-Cu alloys decreased the supersaturated copper atoms in aluminum matrix and reduced the mechanical properties of these materials. Davidson³⁾ mentioned that the iron would form as brittle phases during the solidification of A206 alloy, and the brittle phases could not be eliminated within the range of the normal solutionizing temperature. Thus rather low Fe content (<0.1%) was required for A206 alloy. However, it is hardly to find literatures that have systematically studied on the effect of Fe content on the mechanical properties of A206 alloy. Therefore, the purpose of this study is to focus on different Fe contents in A206 alloy in order to understand thoroughly their influence on the microstructure and mechanical properties.

2. Experimental Procedures

Four different alloys were prepared by melting commercial alloy A206 in a graphite crucible with the addition of iron. After degassing, melt at $750 \pm 5^\circ\text{C}$ were poured into a permanent mold with the specification of ASTM B108 to prepare the test bars. The gage length and cross-sectional diameter of the test bars were 50 and 13 mm, respectively. The chemical composition of the alloys was analyzed by emission spectroscopy is listed in Table 1. These four alloys are designated as A (0.05%Fe), B (0.13%Fe), C (0.19%Fe), and D (0.25%Fe) respectively.

T71 treatment cycles were performed in accordance with AMS (Aerospace Material Specification) 4235A¹³⁾ for permanent mold castings of A206 alloy. Solid solution treatment was carried out at 515°C for 2 h then raised to 525°C for 8 h. The samples were then quenched in water at room temperature, followed by artificial aging at 185°C for 5 h. Attention was needed that the quench-delay time between solution treatment and quenching must less than 5 s in order to avoid the precipitation of coarsely equilibrium phase.

The as-cast and T71 treated samples were polished in

Table 1 Chemical composition of A206 alloys (in mass%).

Alloy	Si	Fe	Cu	Mn	Mg	Ti	Ni	Zn	Pb	Al
A	0.01	0.05	4.95	0.30	0.30	0.21	0.06	0.026	0.006	Rem.
B	0.01	0.13	4.90	0.30	0.30	0.21	0.07	0.026	0.006	Rem.
C	0.01	0.19	4.90	0.30	0.29	0.20	0.04	0.026	0.007	Rem.
D	0.02	0.25	4.93	0.30	0.29	0.22	0.03	0.027	0.006	Rem.

usual way and then etched with 0.5% HF solution to reveal the microstructure. Optical microscopy, scanning electron microscopy (SEM), electron microprobe (EPMA, JEOL-JXA8800M) and image analysis were used to examine the microstructures of these samples. Image analysis was carried out to characterize the amount of intermetallic compounds. Thirty fields (each $21379 \mu\text{m}^2$ in area) were measured from each sample, which were taken from the middle section of the test bar. The DSC instrument was a computerized SEIKO-DSC22C, which recorded the measured data and analyzed the characteristics of heat effect. Sample (about 20 mg) for DSC was solution treated and quenched, then loaded immediately into the sample pan. The heating rate was $10^\circ\text{C}\cdot\text{min}^{-1}$ and the range from 40 to 540°C . Tensile specimens were tested in accordance with the specification of ASTM B577 procedures. The tensile testing was conducted on a 100 kN-capacity computerized MTS closed-loop servo-hydraulic machine operated under stroke-control condition.

3. Results and Discussion

3.1 Microstructure

The available copper in A206 alloy depends on iron, manganese, and nickel contents; the copper combined with them is not available.¹⁴⁾ The primary constituents, which are composed of these elements, are large amounts of needle-like (or platelet-like) compound Cu_2FeAl_7 and very few amounts of manganese-containing compounds. No coarse nickel compounds were found. Other compounds that can be decomposed during solution treatment are the networks of eutectic CuAl_2 and the CuMgAl_2 . The latter is hardly to find. So a typical solidification structure of A206 alloys is mainly composed of the network of CuAl_2 and needle-like of Cu_2FeAl_7 , as shown in Fig. 1. Most of the CuAl_2 and Cu_2FeAl_7 coexist on/near the grain boundaries of α -Al matrix. The amount and number of needles increase with increasing Fe content. Figure 2 shows an SEM secondary electron image and corresponding X-ray maps of alloy D. The network is enriched in Cu and Al, and the needle is enriched in Cu, Fe and Al. Results from EDS, the composition of network ranges from 46 to 47 mass%Cu; and the needle ranges from 32 to 34 mass%Cu and from 12 to 14 mass%Fe, respectively. These agree with the stoichiometry of CuAl_2 and Cu_2FeAl_7 , respectively.

The eutectic phase CuAl_2 has almost dissolved into α -Al matrix after solution treatment. Only a few amount of it can be observed, the morphology of the residual eutectic CuAl_2 is specially shown in Fig. 3(a). Therefore, solution treatment can make the CuAl_2 dissolving to accomplish the state of homogeneity. On the other hand, this is not true for the Cu_2FeAl_7 needles since they still have the same morphol-

ogy as that in the as-cast state and disperse around the grain boundary.

3.2 Image Analysis

Statistical method of image analysis is a good way to characterize the amount of intermetallic compounds in terms of the Fe content. For the manganese-containing compounds, a few amounts ($<0.03\%$) of them are found in all alloys. And their effect on the available copper will not much differ since they have the same Mn content in all alloys. Conversely, different Fe contents cause different amounts of Cu_2FeAl_7 , and the copper consumption is different. So, the need is to focus on the Cu_2FeAl_7 and CuAl_2 compounds. Additionally, in quantitative stereology, the measured area fraction of a phase is equal to the volume fraction, that is, $A_A = V_V$.¹⁵⁾ Then the measured area fraction of these intermetallic compound parti-

Fig. 1 Optical micrographs of as-cast A206 alloys showing the morphologies of CuAl_2 and Cu_2FeAl_7 compounds: (a) alloy A, (b) alloy D (arrow 1: CuAl_2 , arrow 2: Cu_2FeAl_7).

Fig. 2 SEM micrographs of alloy D: (a) secondary electron micrograph, and corresponding X-ray maps for (b) Cu, (c) Fe (arrow 1: CuAl_2 , arrow 2: Cu_2FeAl_7).

Fig. 3 Optical micrographs of T71 treated A206 alloys: (a) alloy A, (b) alloy D (arrow 1: CuAl_2 , arrow 2: Cu_2FeAl_7).

cles can transfer to as the volume fraction. Table 2 shows the measured volume fraction in as-cast alloys. It is found that the fraction of needle-like Cu_2FeAl_7 increases with increasing Fe content. Consequently the fraction of eutectic CuAl_2 decreases slightly. For the Cu_2FeAl_7 , the volume fraction has no difference between as-cast and corresponding T7 treated alloys, and will not list them separately.

Assuming that all of the iron formed as Cu_2FeAl_7 , then the theoretical volume fraction of this compound was calculated in Table 2. It is found that the measured value close to the calculated value very well. It is reasonable to suggest that almost all of the iron form as Cu_2FeAl_7 and the other ionic compounds must be very few. And the volume fraction

Table 2 The measured and calculated value of the volume fraction of Cu_2FeAl_7 and CuAl_2 in as-cast A206 alloys.

Alloy	Cu_2FeAl_7 compound			Copper consumption of Cu_2FeAl_7 (mass%)	Measured volume fraction of CuAl_2
	Count/field	Measured volume fraction	Calculated volume fraction		
A	2(1)	0.26(0.08)	0.22	0.13	5.3(0.3)
B	7(2)	0.64(0.10)	0.56	0.33	5.0(0.4)
C	10(2)	0.90(0.14)	0.82	0.47	4.6(0.5)
D	13(3)	1.13(0.13)	1.08	0.59	4.3(0.3)

Thirty fields were measured from each sample (each $21379 \mu\text{m}^2$ in area). Standard deviations are listed in the parentheses.

of Cu_2FeAl_7 will be proportional to the Fe content, thus the higher Fe content with more Cu_2FeAl_7 will consume more copper. This means that the available copper in an alloy with higher Fe contents will be lower than that with lower Fe contents. In addition, a few amount of residual CuAl_2 ($<0.03\%$) are observed after solution treatment since most of them are dissolved into the α -Al matrix. It can reasonably predict that there will be more copper atoms dissolved into the α -Al matrix in lower Fe content alloys after solution treatment, and its degree of supersaturation will be higher as well.

3.3 DSC

DSC experiment was performed to evaluate the influence of Fe content on the precipitation kinetics of θ' phase in A206 alloy. The reaction peak area represents the heat effect associated with precipitation or dissolution of specific phases in aluminum alloy. The heat effect is proportional to the volume fraction of precipitation or dissolution of specific phases. The peak areas above the baseline represent the exothermic reaction associated with phase formation, such as regions I, III and IV in Fig. 4. The peak areas under the baseline represent the endothermic reaction associated with the dissolution of phases, such as regions II and V.

Figure 4 illustrates the precipitation sequences in the four alloys with different Fe contents. The stages of precipitation of effective strengthening phase have been identified in several published literatures.⁷⁻⁹ Jena *et al.*⁸) had shown that the precipitation peak temperature of S' phase in Al-1.53 mass%Cu-0.79 mass%Mg alloy was around at 190°C under the heating rate of $10^\circ\text{C}/\text{min}$. In present study, only a small peak was found around the same temperature, the precipitation of S' phase is not apparent. Papazian⁹) found that the precipitation peak temperature of θ' phase in 2219 alloy was around at 260°C under the heating rate of $10^\circ\text{C}/\text{min}$. This is close to the peak temperature of region III, a large

amount of θ' precipitates in this region and it is due to the major strengthening phase of A206 alloy.

Before the end of θ' phase precipitation, some of it start to transform its structure to an equilibrium phase θ and the peaks for precipitation of these two phases are obviously overlapped, as shown in regions III and IV. It is necessary to separate such overlapped peaks to characterize the precipitation heat effect of θ' phase. A numerical fitting was performed by referring the peak itself and linear baseline to reveal this work. The experimental data, within the peaks of regions III and IV, can be interpreted as a superposition of two bell-shaped functions¹⁶⁾ whose centers occur at around, for example in alloy A, 258 and 302°C as shown in Fig. 5(a). Besides, the early portion of the fitted profile will match the experimental data by taking account of the effect of GP zone dissolution, which will be an inverse bell-shaped function. Now redrawing the θ' peaks as in Fig. 5(b), the heat effect was obtained

Fig. 4 Typical DSC curves of A206 alloys after solution treatment (I: precipitation of GP zone, II: re-dissolution of GP zone, III: precipitation of intermediate phase θ' , IV: precipitation of equilibrium phase θ , V: re-dissolution of intermediate and equilibrium phases).

Fig. 5 (a) The overlapped peak of regions III and IV in Fig. 4 are composed by two bell-shaped functions (alloy A), and (b) θ' phase precipitation peaks of different Fe content alloys.

Table 3 Peak temperature, heat effect and amount of the precipitation of θ' at the heating rate of 10°C/s.

Alloy	Peak temperature ($\pm 2^\circ\text{C}$)	Heat effect (J/g)	Amount of precipitation (millimole per gram of alloy)
A	258	9.3 ± 0.6	0.27 ± 0.017
B	262	6.6 ± 0.9	0.20 ± 0.026
C	265	5.2 ± 0.7	0.15 ± 0.020
D	268	3.1 ± 0.5	0.09 ± 0.015

by integrating the bell-shaped function. The heat of precipitation of copper as the intermediate θ' phase, $\Delta H_{\text{Cu}}^{\theta'}$, was about $34 \pm 4 \text{ kJ/mol}$.⁷⁾ The amount of precipitation of θ' phase then can be obtained as shown in Table 3 and shows that it decreases with increasing Fe content. Also the precipitation peak temperature (T_p) for θ' phase raises gradually with increasing Fe content. It appears around 258°C for alloy A and raises to 268°C for alloy D. This implies that the precipitation kinetics is lowered as the Fe content increases.⁵⁻⁸⁾ The reduction in the amount and kinetics of θ' phase precipitation can be reasonably explained as due to the situation that the increasing in the volume fraction of Cu_2FeAl_7 compound will consume more Cu atoms and decrease the available content and degree of supersaturation of copper atoms in aluminum matrix after solution treatment.

3.4 Mechanical properties

The difference in microstructure of A206 alloys caused by different Fe contents has a great influence on its mechanical properties. The mechanical properties of as-cast and T71 treated alloys are shown in Table 4. The yield stress was computed at 0.2% offset strain. Elongation was measured over a 50 mm gage length.

The yield strength decreases from approximately 351 MPa for alloy A to 323 MPa for alloy D. The ultimate strength has a similar tendency. In T71 treated alloys, the strength increasing is dominated by the amount of precipitation of strengthening phase. Silcock¹⁷⁾ has observed that in the Al–3.53 mass%Cu–0.48 mass%Mg alloy, the θ' phase, GPB zone and S' phase can be observed at 190°C isothermal aging for 5 hours. But now in A206 alloy, the ratio of Cu/Mg and the observation from DSC analysis suggest that the major strengthening phase must be the θ' rather than S' . And the volume fraction of θ' phase in higher Fe content alloys will be lowered due to the lower available copper. The yield and ultimate strength of material then decrease with increasing Fe content. The relation between the Fe content and strength can be seen. The regression lines obtained from the least-square line fitting

Fig. 6 Relations between Fe content and tensile strength for A206 alloys.

show a good relationship between strength and Fe content as the form $\sigma = a\text{Fe}\% + b$ ($0.05 \leq \text{Fe}\% \leq 0.25$), as shown in Fig. 6, where σ is ultimate or yield strength in MPa. The constants a and b for the case of ultimate strength are -160.4 and 433.8 respectively, and for the yield strength case are -136.7 and 358.6 respectively.

In general, the ductility of material increases as the strength decreased. But this does not apply to the present situation. The elongation decreases rapidly when the Fe content is increased. This may be due to the brittle rupture or decohesion occurring rather easily and consequently acting as crack initiation site when the intermetallic compound is under stress. Some microcracks are seen on those sites of needle-like Cu_2FeAl_7 intermetallic compound that are away from the main crack in alloy D, as shown in Fig. 7. This will deteriorate the ductility of material. However, higher Fe contents with more of needle-like intermetallic compound will deteriorate the ductility of material more deeply. In this connection, the SEM secondary electron images of the rupture surface of tensile specimens in Fig. 8 will provide another useful information. Figure 8(a) shows the fracture morphology of alloy A, which exhibits a typical ductile dimple rupture on the rupture surface. As the Fe content increases in alloys B and C, the fracture feature will appear less ductile rupture surfaces and more brittle cleavage surfaces. The dimples are hardly to find in alloy D, as shown in Fig. 8(b), which exhibits almost brittle cleavage surfaces. These results are pieces of evidences to explain the less elongation of A206 alloy with higher Fe content.

The data listed in Table 4 shows that the hardness is higher for those alloys with higher Fe contents under as-cast con-

Table 4 Tensile properties of T71 treated specimens and the hardness of A206 alloys.

Alloy	Yield strength (MPa)	Ultimate strength (MPa)	Elongation (%)	Hardness (R_B)	
				As-cast	T71
A	351(1.7)	427(3.4)	9.0(0.8)	38.4(1.4)	66.0(1.5)
B	340(2.8)	411(2.3)	8.0(0.6)	40.8(1.2)	63.2(1.4)
C	335(3.1)	401(4.5)	6.2(1.2)	42.8(0.9)	60.6(1.0)
D	323(3.8)	396(4.1)	5.6(1.5)	45.3(0.6)	58.3(0.8)

Standard deviations are listed in the parentheses.

Fig. 7 Optical micrograph of crack path morphology of alloy D (arrows: microcracks of Cu_2FeAl_7).

dition. Since the strengthening phase θ' does not precipitate yet in as-cast alloys, so it is reasonable to conclude that the higher hardness is due to the more amount of intermetallic compounds such as CuAl_2 and Cu_2FeAl_7 for richer Fe alloys. After T71 heat treatment, a large amount of strengthening phase θ' precipitate, and the hardness of alloys is then improved remarkably. More of θ' phase precipitates in lower Fe content alloys, therefore the hardness raises higher than that of richer Fe alloys.

4. Conclusions

It can be clarified that the influence of Fe content on the tensile properties of A206 alloy through the examination of microstructure and DSC analysis. Base on the results of our study, the conclusions are as follows:

- (1) The volume fraction and number of Cu_2FeAl_7 compounds are proportional to the Fe contents. The available copper decreases with increasing volume fraction of Cu_2FeAl_7 compound.
- (2) In DSC analysis, the amount and kinetics of θ' phase precipitation decrease with the decreasing available copper.
- (3) The mechanical properties are mainly dominated by the volume fraction of θ' phase and Cu_2FeAl_7 intermetallic compound where the tensile strength and hardness are due to the former and the elongation is due to the latter. Hence the richer Fe alloys that contain less θ' phase and more Cu_2FeAl_7 compound will exhibit inferior mechanical properties, especially the elongation. The tensile strength decreases linearly with increasing Fe content.

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Fig. 8 SEM micrographs of fracture surface of tensile specimens: (a) alloy A, (b) alloy D.

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