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Supportive strengthening role of Cr-rich phase on Al–Si multicomponent piston alloy at elevated temperature

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ABSTRACT

The microstructure and elevated-temperature strength of an Al–Si piston alloy with 0.5 wt.% Cr and 0.8 wt.% Fe addition were investigated in this paper. The addition of Cr and Fe leads to the formation of α-Al(Fe,Cr)Si phase, which maintains the crystal structure of α-Al(Fe,Mn,Cr)Si phase but has different stoichiometry and morphology from the phase in previous works. Besides, the morphology of δ-Al_3CuNi phase changes because of solid-solved Cr. These changes result in the formation of closed and semi-closed network eutectic colonies. The Cr-rich phases, acting as effective supportive strengthening phase, integrate organically with Ni-rich phases, acting as main strengthening phases. So α-Al phases are successfully encircled by eutectic intermetallic phases, and the slide of α-Al under stress at elevated temperature can be effectively hindered. Thus the ultimate tensile strength of the examined alloy is increased by 26% after 0.5 wt.% Cr and 0.8 wt.% Fe addition.

1. Introduction

Multicomponent Al–Si based casting alloys have widely applications in piston production and other automotive products, due to good castability, high strength, light weight, good wear resistance and low thermal expansion [1–3]. With the development of the world and the people's increasing awareness of environment and resources, automotive engines efficiency, greater demands have been placed on the properties of Al–Si based casting alloys, especially the elevated-temperature properties [4,5].

Alloying is one of the most important and practical methods to improve the elevated-temperature properties of Al–Si piston alloys, and many elements have been examined during the past decades [5–9]. Ni has been known as the most important element in multicomponent Al–Si piston alloys and Ni-rich phases are the main strengthening phases [2,9–12], because Ni-rich phases can maintain good mechanical properties at elevated temperature and have great contribution to the elevated-temperature strength [9]. Unfortunately, Ni is very expensive, so it is meaningful and practical to add much cheaper elements such as Fe, Mn and Cr to form thermal-stable phases for improving the elevated-temperature properties of Al–Si piston alloys. However, it needs some special methods such as spray-forming method to control the morphologies [13,14], in order to exert the strengthening effects of Fe-rich phases, which undoubtedly increases the cost.

The phase morphologies are one of the key factors of improving Al–Si piston alloys, as discussed before [9,15]. According to strengthening principle, when the intermetallic phases distribute in closed or semi-closed networks along α-Al grain boundaries, the intermetallic phases can drag grain boundaries to prevent grain boundary from sliding at elevated-temperature, and alloy strength can be greatly improved [15]. Therefore, it is significant to form a supportive strengthening intermetallic phase which can form closed or semi-closed networks with Ni-rich phases, in order to make most use of Ni-rich phases.

In this paper, Cr and Fe were added into Al–Si–Cu–Ni–Mg piston alloy, Cr-rich supportive strengthening phase organically integrates with Ni-rich phases and network-like eutectic microstructure formed, so the elevated-temperature strength of the alloy is greatly improved.

2. Experimental

The multicomponent Al–Si–Cu–Mg–Ni piston alloys were used in the study. The compositions of the alloys were listed in Table 1. Firstly, the alloy ingots were prepared in a clay-bonded graphite crucible heated by 25 kW medium frequency induction furnace, using 99.85% commercial purity Al, 98.5% commercial purity crystalline Si and 99.9% purity Cu, Ni and Mg. The melts were poured into one mold to gain ingots.

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Table 1
Chemical compositions and UTS of alloys.

Alloy grades	Elements (wt.%)								UTS at 350 °C/MPa
	Si	Cu	Ni	Mg	Mn	Fe	Cr	Al	
Group 1	13.0	3.7	3.2	1.1	<0.1	<0.2	0	Balance	78.21
Group 2	13.0	3.7	3.2	1.1	<0.1	0.8–1.0	0.5	Balance	98.61

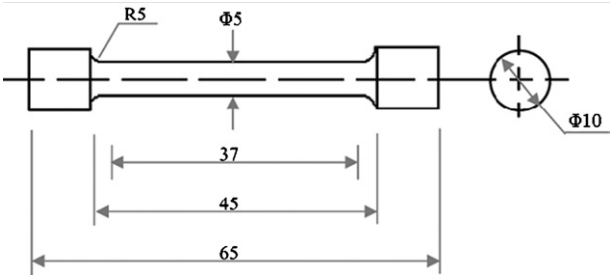

Fig. 1. Pattern dimensions for 'dog-bone' type specimen.

Then the ingots were superheated to a temperature of 750 °C in an electric resistance furnace. The temperature was measured using a digital chromel–alumel thermocouple. For the first group, 1 wt.% of Al–3.5P master alloy (Provided by Shandong Al&Mg Melt Technology Co., Ltd) was added to modify primary silicon. Then the melts were carried out using 0.5 wt.% C_2Cl_6 at 750 °C for slag-removing and degassing. 30 min after addition of Al–P master alloy, the melts were finally poured into a pre-heated (200 °C) mold at 720 °C to gain tensile test bars and specimens for microstructural analysis. For the second group, 0.5 wt.% Cr and 0.8 wt.% Fe were added before adding Al–3.5P master alloy, and Cr was added in forms of chromium agent.

Test bars were then heat-treated in the process: solution treated at 490 °C for 3 h; water quenched; aging treated at 200 °C for 8 h and cooled in air. The test bars were machined to 'dog-bone' type specimens (shown in Fig. 1), and then tested by electronic all-purpose test machine at 350 °C. The tensile strength data of each alloy reported below is an average of four tensile specimens.

Specimens for metallographic microstructure observation were cut from tensile bars after heat-treatment. Metallographic specimens were polished in the usual manner and final polishing was carried out with fine magnesia powder by hand. The analysis was performed using X-ray diffraction (XRD, Rigaku D/max-rB), JSM-6380LA scanning electron microscope and field emission scanning electron microscope (FESEM).

3. Results

3.1. Elevated temperature tensile strength

The ultimate tensile strengths (UTS) of the two groups at 350 °C are shown in Table 1. It can be seen that group 1 has the average UTS of 78.21 MPa, which was increased up to 98.61 MPa by 26% in group 2 with 0.5 wt.% Cr and 0.8 wt.% Fe addition.

3.2. Microstructural analysis

Fig. 2 shows the microstructure of the used Al–Si–Cu–Ni–Mg piston alloy, which has relatively high elevated-temperature strength. The main intermetallic phases include δ-Al_3CuNi, γ-Al_7Cu_4Ni, π-$Al_8FeMg_3Si_5$, Q-$Al_5Cu_2Mg_8Si_6$ and M-Mg_2Si. The phases can be identified from the EDS and XRD results, since they have specific stoichiometries [2,10]. The δ-Al_3CuNi and γ-Al_7Cu_4Ni phases have much bigger contributions to the elevated-temperature properties of Al–Si piston alloys, owing to their

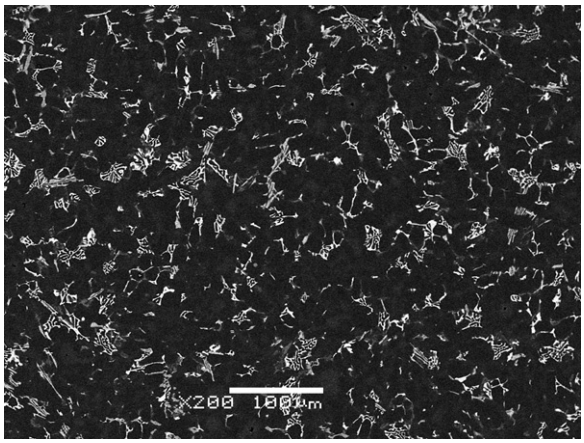

Fig. 2. Microstructure of the group 1 Al–Si–Cu–Ni–Mg piston alloy.

better thermal stability, mechanical properties, morphologies and distributions [9,16]. While Q-$Al_5Cu_2Mg_8Si_6$, M-Mg_2Si and π-$Al_8FeMg_3Si_5$ are not thermal stable enough, they become dissolved or coarsened when exposed in elevated temperature [7,15]. In Fig. 2, most Ni-rich phases are bright skeleton-like, and there is no network-like eutectic microstructure in this alloy.

Fig. 3 shows the XRD pattern of group 2 Al–Si piston alloy with 0.5 wt.% Cr and 0.8 wt.% Fe addition. From the XRD results, it can be seen that only one new phase formed after 0.5 wt.% Cr and 0.8 wt.% Fe additions, which is α-Al(Fe,Cr)Si phase. Because of seven kinds of elements in the second group, according to Gibbs phase rule (Eq. (1)), eight kinds of stable phases can form in this alloy.

$$P = C + 2 - F \quad (1)$$

where P represents the number of phases in system, C represents the number of elements in system, and F represents the degree of freedom.

So it can be confirmed the α-Al(Fe,Cr)Si phase is the only newly formed phase. Its diffracted peaks are in well accordance with the α-Al(Fe,Mn,Cr)Si phase in previous works [10,11,13]. However, there is no Mn addition in alloy of group 2. So it can be concluded that the new phase maintains the crystal structure of α-Al(Fe,Mn,Cr)Si phase, but the stoichiometry is different from the phase in previous works.

Fig. 4 shows the microstructure of group 2 Al–Si–Cu–Ni–Mg piston alloy with 0.5 wt.% Cr and 0.8 wt.% Fe. Compared to the microstructure of group 1 without Cr and Fe addition, the morphologies and distributions of intermetallic phases have experienced big changes. The eutectic colonies in group 1 alloy are desultory and α-Al phases are not encircled by intermetallic phases, since the main strengthening Ni-rich phases have skeleton-like morphology. While in the group 2 alloy, most of the intermetallic phases exhibit strip-like morphology, and the eutectic colonies are formed by network-like intermetallic phases and encircled α-Al phases, as marked in SEM micrographs of Fig. 4.

Then EDS analyses were conducted to further study the microstructure of the group 2 alloy, which are shown in Figs. 5 and 6. As can be seen from Fig. 5, the closed network eutectic colonies which encircle α-Al phases are mainly δ-Al_3CuNi phases.