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Fracture behaviors of A390 aluminum cylinder liner alloys under static loading

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ABSTRACT

A390 aluminum cylinder liner alloys were fabricated by gravity-casting and squeeze-casting, and their fracture behaviors under static loading were investigated by *in situ* SEM observation in this study. The two alloy samples exhibit different morphologies and sizes of eutectic and primary Si particles, which results in diverse fracture behaviors. For the gravity-casting alloy, microcracks firstly initiate at the primary Si particles but are restricted from propagating to the matrix, and the dominant crack primarily propagates through the fracturing or debonding of eutectic Si particles. By contrast, for the squeeze-casting alloy, microcracks also firstly initiate at the primary Si particles and then propagate into matrix and link with other microcracks to form a dominant crack. In addition, the microcracks are barely observed to initiate at the eutectic Si particles in squeeze-casting alloy. All results show that the eutectic Si and primary Si act as a bridging role in gravity-casting alloy and squeeze-casting alloy, respectively. It indicates that eutectic Si particles and primary Si particles are in competition in the propagation of dominant crack. Moreover, the fractures of eutectic Si particles could release elastic strain energy, by which the strain energy at crack tips is decreased, so the propagation of the microcracks at the primary Si particles could be retarded.

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1. Introduction

Hypereutectic aluminum–silicon alloys have been strongly recommended as the preferred material for some types of automotive engine parts due to the good wear resistance and low coefficient of thermal expansion [1–4]. Typically, A390 aluminum alloy, an important representative of hypereutectic aluminum–silicon alloys, has been used for casting cylinder block and cylinder liner since the 1970s [5]. Hypereutectic A390 alloy was first introduced to the AFS Casting Congress as early as 1971 [6], and the properties of hypereutectic aluminum–silicon alloys have attracted a widely attention to researchers because of its important applications in automobile industry [7–10].

It has been reported that the aluminum alloy with high silicon content shows high wear resistance and low coefficient of thermal expansion, but poor ductility [11–13]. Because of the degradation in mechanical properties of as-cast hypereutectic Al–Si alloys, their applications in potential fields are greatly restricted [10]. Lots of work has been done by researchers to obtain a both refined and modified microstructure so as to improve the mechanical properties [14,15]. However, it is still very difficult to fabricate a production of a fine and evenly dispersed primary Si coupled with a fully modified eutectic structure [16–18]. Therefore, some previous research suggests that a more effective way for improving the mechanical properties just need the refinement of primary Si, through the addition of agents such as Cu–P alloys [19,20], Si–P alloys [21] or Al–Zr–P alloys [22]; meanwhile, another views hold that the modification of the eutectic Si is more significant, just as the widespread success of the modification in hypoeutectic alloys to enhance the ductility [23–25]. Yet the latter viewpoint has not been extensively investigated in hypereutectic alloys. Actually, the important difference of two above opinions brought out from the different understandings on the roles of primary Si and eutectic Si during the fracture process. Specifically, it is difficult to directly investigate the crack initiation and propagation in hypereutectic aluminum–silicon alloys, since the ductility of the alloys is too low to capture useful information of fracture, thereby a unified viewpoint cannot be established. Fortunately, hypoeutectic aluminum–silicon alloys show better ductility [26] and have been extensively observed by *in situ* tensile tests [27,28]. Hence, a united and systematic viewpoint, which the cracks initiate at eutectic Si particles and the aluminum matrix plays a role in blocking crack propagation, has been accepted [27]. Besides, the modification on the eutectic Si improving the mechanical properties of the hypoeutectic alloys has been successfully applied to the practical production. Therefore, some direct fracture investigations about hypereutectic aluminum–silicon alloys should be conducted in order to systematically understand the correlations among the mechanical

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properties, microstructures and fracture mechanisms. In this case, the contributions of the primary Si and eutectic Si to the fracture process could be confirmed.

In situ tensile test is a method to directly observe the specimen surface during the tensile process at a room or high temperature by scanning electron microscope (SEM) or transmission electron microscope. This method has been widely used to analyze the microfracture process in metallic materials [29–31]. In present study, *in situ* SEM tensile test was conducted to investigate the crack initiation and propagation. A390 alloys and T7 heat treatment were applied to obtain a better ductility in order to save more time to capture the fracture information. The different microstructures of A390 cylinder liner alloys were fabricated by gravity-casting and squeeze-casting for comparing the roles of the primary Si and eutectic Si during the fracture process. Furthermore, the mechanical properties of A390 alloys were analyzed and several suggestions were accordingly put forward on the basis of what roles the primary Si and eutectic Si act to optimize the mechanical properties of the hypereutectic aluminum–silicon alloys.

2. Experimental procedures

2.1. Cast A390 cylinder liner alloys

Cast A390 cylinder liner alloys were fabricated by gravity-casting and squeeze-casting in order to obtain different microstructures using the same chemical composition. The alloy was melted in a medium frequency induction furnace to ensure sufficient uniformity of alloying elements. The melt was refined with the addition of 1% Cu-10P master alloy at 800 °C and then cleaned by nitrogen gas (all composition quoted in this work are in wt.% unless otherwise stated). The chemical composition of the melt was measured by direct-reading spectrometer, and the result is listed in Table 1. The cylinder liners after the same melt treatment were fabricated by pouring the melt into the mold at 760 °C as the above mentioned processes. For both GC and SC cylinder liner alloys (finished, bore size: 81 mm, liner wall thickness: 5 mm, length: 128 mm), the permanent mold was preheated at 200 °C. The SC cylinder liner alloy was fabricated by the direct squeeze-casting process on the YQ32-100 squeeze-casting machine (die temperature is about 200 °C, casting pressure is about 20 MPa and dead time is 50 s). The same T7 heat treatments were applied to the two A390 cylinder liners (solution heat treated at 500 °C for 6 h, quenched in boiling water, and then aged for 8 h at 230 °C). For convenience, the gravity-casting and squeeze-casting cylinder liner alloys fabricated in this present study are referred to as 'GC' and 'SC', respectively.

2.2. Tensile test and microstructural analysis

The plate specimens (gage length: 25 mm; gage thickness: 3 mm; and gage width: 6.5 mm) were obtained from the cylinder liners (five plate specimens for one cylinder liner), and room-temperature tensile tests were conducted on these specimens at a crosshead speed of 0.03 mm per second. The microhardness of the matrix and overall hardness were measured by micro-Vickers (HV-100) and Vickers hardness testers (HV-10A) under 50 and 5000 g loads, respectively. The cylinder liner alloys were polished by electrolytic polishing and analyzed using a FEI Quanta 200 FEG scanning electron microscope to observe the microstructures. The measurement software (Image Pro Plus) was also applied to compare the distributions and sizes of Si particles as well as the volume fractions of Si particles between the GC alloy and SC alloy.

2.3. In situ SEM tensile test

As the A390 alloy has been used for fabricating cylinder liners of auto engine, the test specimens were obtained from cylinder liners to meet with the practical requirements. *In situ* tensile test specimens (illustrated in Fig. 1) were fetched from the identical parts of the two castings by a wire-electrode cutting machine. One v-shaped notch tip (depth: 0.3 mm) was precut at one side of the specimen for initiating the fracture. The special electrolytic polishing was used on the observational surface for capturing the surface changes by the SEM during the tensile tests. The *in situ* SEM tensile test was conducted on the SHMADZU SEM–Servopulse test machine under high vacuum and room temperature at a loading speed of 1×10^{-4} mm/s, and the changes on the specimen surface were observed simultaneously. The load was presented by stress *f* in this work, which was defined as the loading divided by the smallest transverse area of the original specimen. In this *in situ* tensile test, the microcrack initiation and propagation of dominant cracks had been focused. After the tensile tests, the fractured surfaces were observed using FEI Quanta 200 FEG SEM.

Table 1
Chemical composition of experimental A390 cylinder liner alloy (wt.%).

Alloy	Si	Cu	Mg	Fe	Ti	P	Al
A390	17.51	4.12	0.43	0.28	0.06	0.06	Bal.

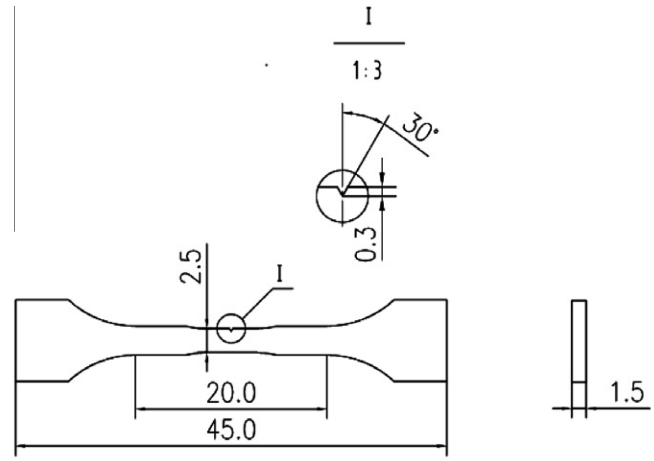

Fig. 1. Shape and size of the *in situ* SEM tensile test specimen.

3. Results

3.1. Microstructure

The results of the SEM micrographs show the morphologies of primary Si particles and eutectic Si particles in the two alloy samples, as shown in Fig. 2. The great difference between the two samples lies in the morphology and size of eutectic Si particles. The eutectic Si particles in GC alloy had a high aspect ratio and needle-shape, thin-flake morphology (Fig. 2(a)), which has been generally considered to deteriorate the tensile strength and ductility. In contrast, the eutectic Si particles with small size (Fig. 3(d)) and round morphology were homogeneously distributed in the matrix of SC alloy (Fig. 2(b)). The primary Si particles exhibited the blocky morphology in both alloys, but there were differences in the average sizes and distributions between the two samples, which are shown in Fig. 3(a) and (c). The average size of primary Si particles in GC alloy was as large as 64 μm and the size varied in a large range. However, in SC alloy, the primary Si particles were in smaller size (decreasing nearly one third than GC alloy) and finer distribution. For the eutectic Si particles, as shown in Fig. 3(c) and (d), the average size of SC alloy was only 7 μm and the distribution was concentrated, whereas the average size in GC alloy was much larger (27 μm) with varying sizes. The volume fractions of primary Si particles and the overall Si particles of the two samples are provided in Table 2, in which the SC alloy shows smaller volume fractions on both primary and overall Si particles due to the application of squeeze-casting process.

3.2. Hardness, tensile test results

The hardness, tensile test results of the GC alloy and SC alloy are shown in Table 3. The hardness of the aluminum matrix of the SC alloy is higher than that of the GC alloy, which might be caused during the squeeze-casting process. However, since more primary Si particles (11.2%) and Si particles (32.2%) were precipitated in the GC alloy, the overall hardness of GC alloy was higher. This result indicated that the GC alloy was more brittle. The yield and tensile strengths of the SC alloy were slightly higher than those of the GC

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