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Effect of Melt Shearing on Morphological Change of Iron-bearing Intermetallic Phases in the Al-Si Casting Alloys

Bonghwan Kim[1], Sangmok Lee[1], Jeongmin Kim[2] and Joonsik Park[2]

[1]Liquid Processing & Casting Technology R/D Department, Korea Institute of Industrial Technology, 7-47 Songdo-dong, Yeonsu-gu, Incheon, Korea

[2]Division of Advanced Materials Engineering, Hanbat National University, 16-1 Duckmyoung-dong, Yuseong-gu, Daejeon, Korea

This study aims to investigate the combined effects of minor alloying elements and mechanical melt-shearing on the microstructure of the Al-Si casting alloys containing iron-bearing intermetallic phases. Several casting alloys were produced by conventional high pressure die casting with the variation of Fe, Mn, and Cr contents. Throughout this study, effects of melt-shearing on the microstructural evolution of die cast materials were systematically investigated. The additions of Mn and Cr were observed to transform the needle-like β-AlFeSi phase into the irregular or cross-like α-AlFeMnSi phase and rosette-type α-AlFeCrSi phase in the high pressure die casting, respectively. It was found that melt-shearing treatment before die casting practically shortened the length of the β-AlFeSi phase in the unmodified Al-Si alloys. Furthermore, the irregular or cross-like α-AlFeMnSi and rosette-type α-AlFeCrSi phases in the modified Al-Si alloys were drastically transformed into fine equi-axed morphology thanks to melt shearing. It was also confirmed that these refined microstructures led to the increase of tensile strength.

Keywords: *Al-Si casting alloy, Intermetallic phase, Modification, Die casting, Melt shearing*

1. Introduction

It has been widely known that the intrinsically low solubility of iron in aluminum alloys tends to form iron-bearing intermetallic phases. On the occasion of high iron contents in the Al casting alloys, the β-AlFeSi phase has been thought to be most harmful due to its needle-like morphology [1]. Up to date, possibilities of modification of their morphologies have been major topics owing to their deleterious effects on mechanical properties [2]. In order to overcome this serious problem, various efforts have been made including chemical treatment with specific elements such as Mn and Cr. It is now widely known that the addition of Mn promotes the script-/polygonal-shape α-AlFeMnSi phase rather than the formation of the β-AlFeSi phase [3]. In the Cr added alloys, on the other hand, the faceted dendritic α-AlFeCrSi phases are commonly recognized to form [4]. More recently, a melt-shearing process has been suggested for the effective modification method for various casting alloys based on physical mixing of the melt during solidification [5].

This work aims to understand the combined effects of chemical and physical treatments on the morphologies of various iron-bearing intermetallic phases during solidification. In other words, various levels of Fe, Mn, and Cr were added to the Al-Si casting alloy under several melt shearing conditions before die casting. Process optimization of the physical mixing was tried to have fine dispersed microstructures from various morphologies including dendritic, Chinese-script, and polygonal morphologies of various phases using the present combined techniques of chemical and physical methods.

2. Experimental procedure

Various casting alloys were designed and prepared in order to investigate the effects of chemical treatment by Fe, Mn, and Cr on the morphological change of iron-bearing intermetallic phases, as summarized in Table 1. Fig. 1 shows the schematic drawing of combined process of melt-shearing and die casting, test mold configuration, and experimental set-up of melt-shearing, respectively. The prepared alloys were melted and poured into the melt-shearing assembly (developed by Z. Fan [5]), followed by high pressure die casting (HPDC). During melt-shearing, various parameters of shearing temperature, speed (RPM), and time were controlled to optimize the microstructure of each alloy. The metallurgical and mechanical analyses were carried out with the sub-sized tensile specimens sectioned from the cast ingot as shown in Fig. 1(b). For comparison, representative microstructure and tensile properties were analyzed with the specimens obtained by conventional high pressure die casting without melt-shearing treatment.

Table 1 Chemical compositions of the iron-rich casting alloys

Alloy	Chemical composition (wt.%)						Purpose of alloying
	Si	Cu	Fe	Mn	Cr	Al	
No. 1	7.5	2.7	**2.2**	0.13	0.05	Bal.	Base alloy (Fe-rich)
No. 2	7.7	2.8	**2.1**	**0.39**	0.04	Bal.	Effect of Fe and Mn
No. 3	7.7	2.9	**2.2**	0.12	**0.27**	Bal.	Effect of Fe and Cr

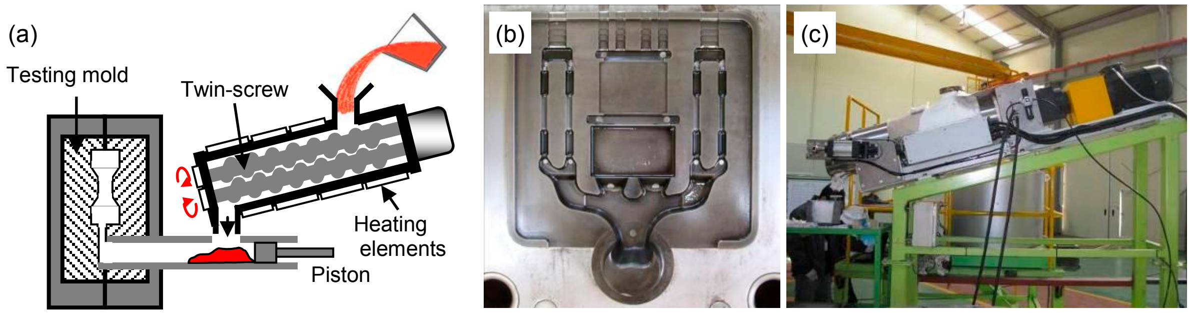

Fig. 1 (a) Schematic drawing of the melt-shearing process, (b) mold configuration of test specimen, and (c) melt-shearing assembly attached to the die casting machine.

3. Results and Discussion

In order to optimize the process parameters of the melt-shearing process, theoretical formation behavior of various iron-bearing intermetallic phases were evaluated by the J. Mat. Pro. software. The fraction of every constituent phases in No. 1~3 alloys were calculated as shown in Figs. 2(a)~(c), respectively. In No. 1 alloy, the α-AlFeMnSi and the β-AlFeSi phases are expected to form as two primary phases at about 640 and 620℃, respectively before the formation of the α-Al at about 600℃. In the Mn added alloy of No.2, the α-AlFeMnSi phase is expected to form at much higher temperature, about 660℃, with higher fraction solid, due to excessive Mn content. In the Cr added alloy of No.3, the α-AlFeCrSi phase is expected to form as the primary phase at about 660℃, followed by accompanying primary α-AlFeMnSi phase.

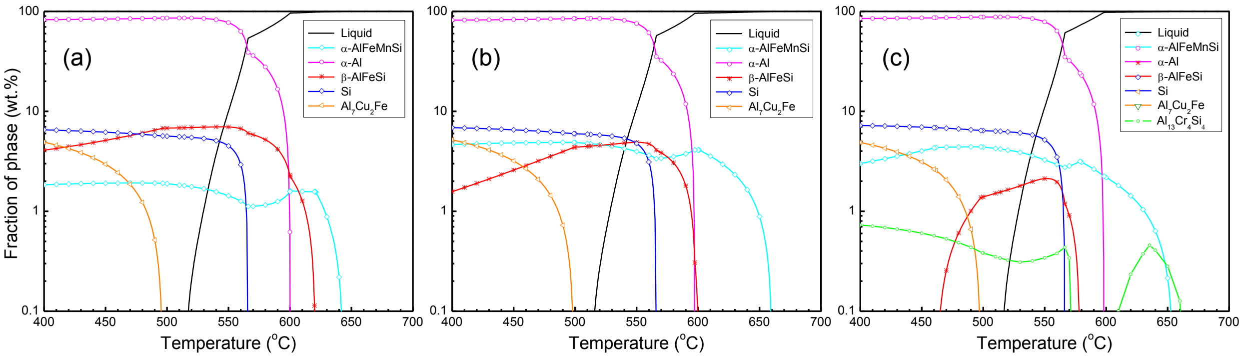

Fig. 2 Volume fractions of various phases calculated by the J. Mat. Pro. software; (a) No. 1, (b) No. 2, and (c) No. 3 alloys.

Based on the preliminary calculation as shown in Fig. 2, various process parameters of the melt-shearing machine were carefully determined for a specific alloy in a tailored fashion. For the modification the β-AlFeSi phase of No. 1 alloy, for example, melt-shearing temperature of 610℃, rotation speed of twin roll set of 500RPM, and melt-shearing time of 30 sec. were chosen. For the same reason, process parameters of 650℃-500RPM-30sec. for the modificaiton of the α-AlFeMnSi phase of No. 2 alloy and 650℃-500RPM-30sec. for the modificaiton of the α-AlFeCrSi phase of No. 3 alloy were determined.

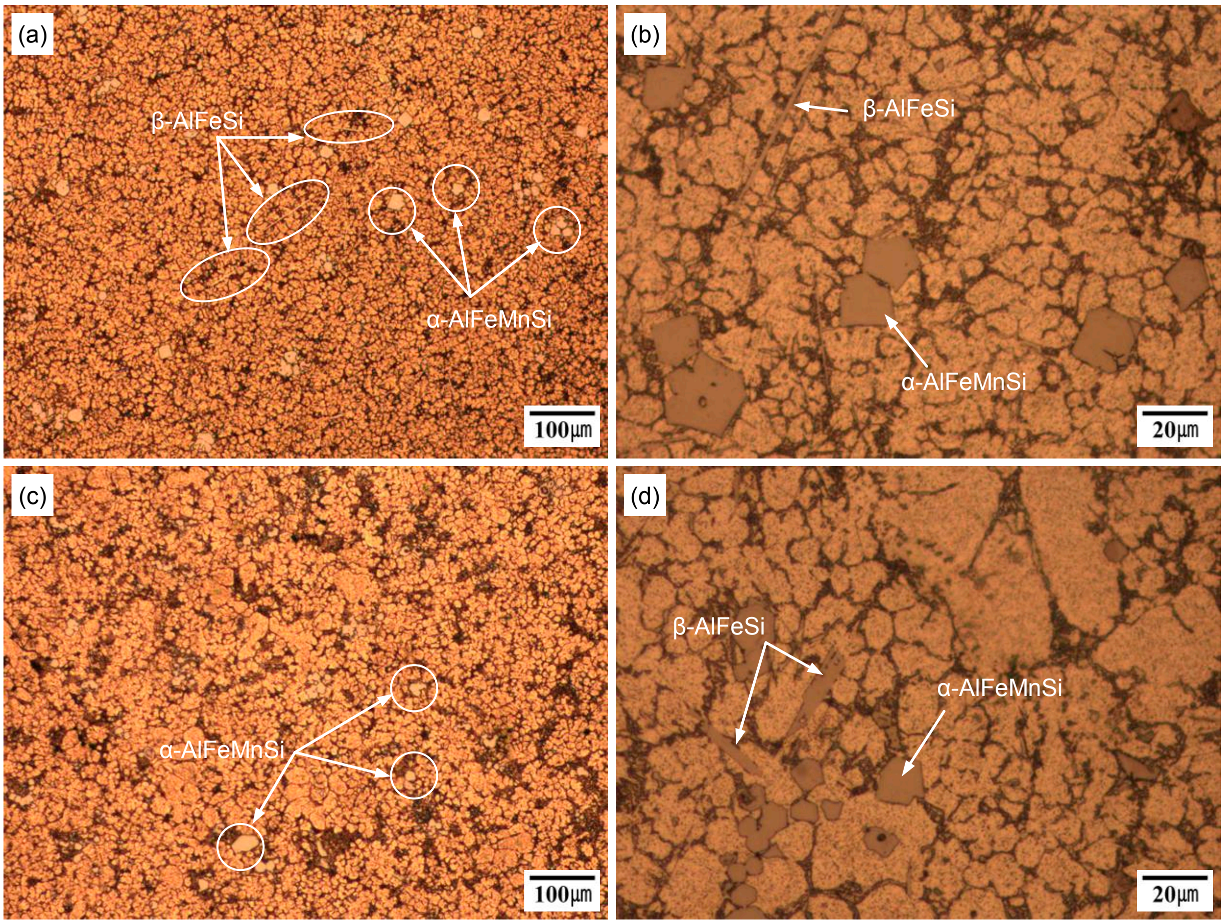

Fig. 3 Microstructure of No. 1 alloy obtained by (a) and (b) HPDC and (c) and (d) melt-shearing at 610℃ before HPDC, respectively.

Figs. 3(a) and (b) show the typical microstructures of No. 1 alloy produced by the HPDC process, in low and high magnifications, respectively. As calculated in Fig. 2(a), both the α-AlFeMnSi and the β-AlFeSi phases were clearly observed in the inter-dendritic region. The β-AlFeSi phases showed the typical needle-like morphology, while the α-AlFeMnSi phases were observed in faceted shapes. Figs. 3(c) and (d) show the typical microstructures of No. 1 alloy processed by melt-shearing prior to HPDC. During the melt-shearing process at 610℃, both primary α-AlFeMnSi and the β-AlFeSi phases formed and altered its morphologies by intensive mechanical stirring. As a result, the length of the β-AlFeSi phase was considerably shortened and widened and the size of the primary α-AlFeMnSi phase was comparatively decreased. (Compare Figs. 3(b) and (d)).

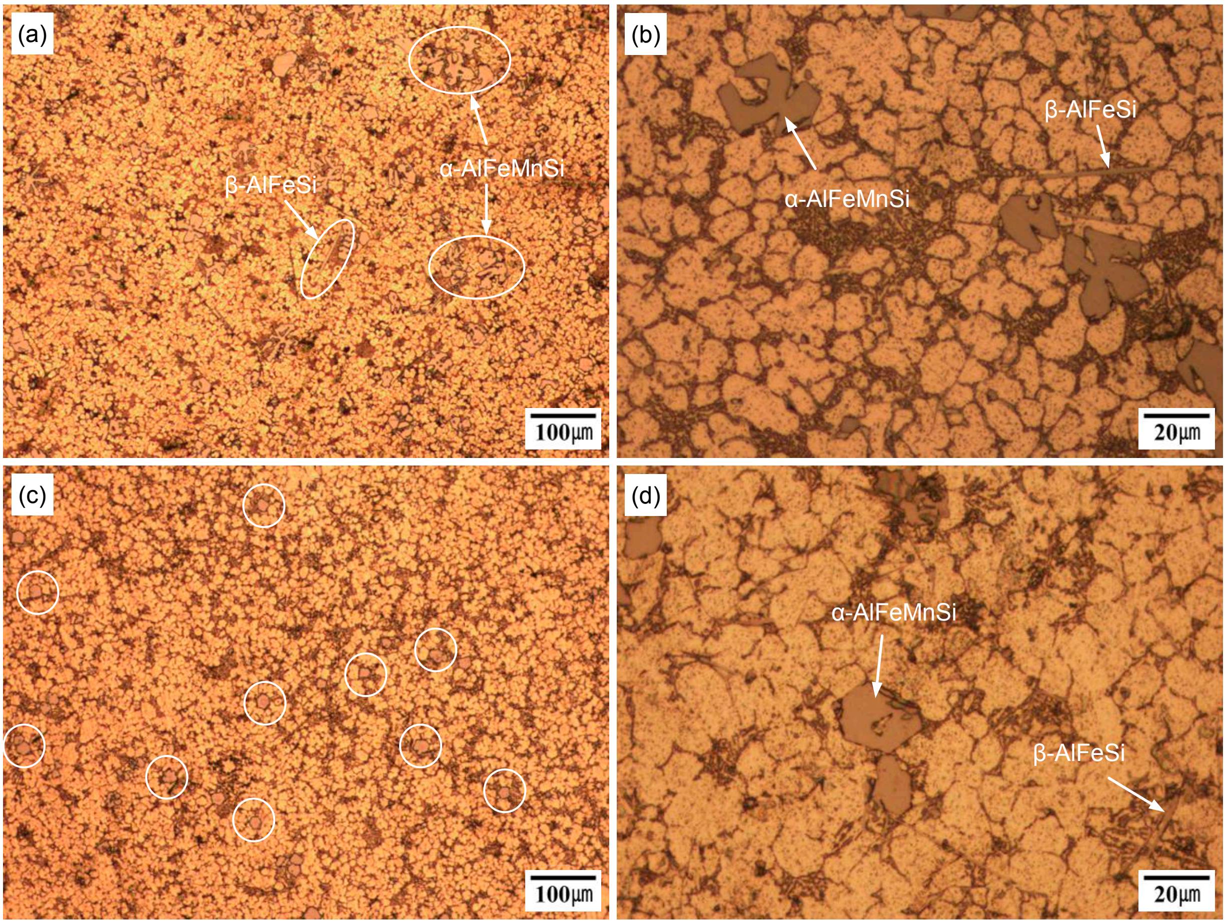

Fig. 4 Microstructure of No. 2 alloy obtained by (a) and (b) HPDC and (c) and (d) melt-shearing at 650℃ before HPDC, respectively.

Figs. 4(a) and (b) shows the microstructure of the Mn added alloy of No. 2 produced by the HPDC process, in low and high magnifications, respectively. It is suggested that the primary α-AlFeMnSi phases dominantly formed with very small fraction of the β-AlFeSi phases as shown in Fig. 4(a). The magnified view of the primary α-AlFeMnSi phase reveals branched or cross/star-like morphologies as in Fig. 4(b). This is somewhat different from No. 1 alloy which shows faceted morphology of the α-AlFeMnSi phase. The primary α-AlFeMnSi phase in No. 2 alloy was calculated to nucleate at higher temperature compared to No. 1 alloy. Therefore, this primary phase is expected to have much time to grow during solidification, exhibiting irregular shapes. Figs. 4(c) and (d) clearly show the effect of melt-shearing before the HPDC process. All the primary α-AlFeMnSi phases were effectively reduced in size and well dispersed throughout the specimen with equi-axed faceted morphology. It is suggested that the intensive melt-shearing during the nucleation and growth stages of the primary α-AlFeMnSi phase induces the turbulent flow in the vicinity of the primary α-AlFeMnSi phases, prohibiting the further growth of the primary phases. This directly results in the equi-axed faceted morphology as shown in Fig. 4(d).

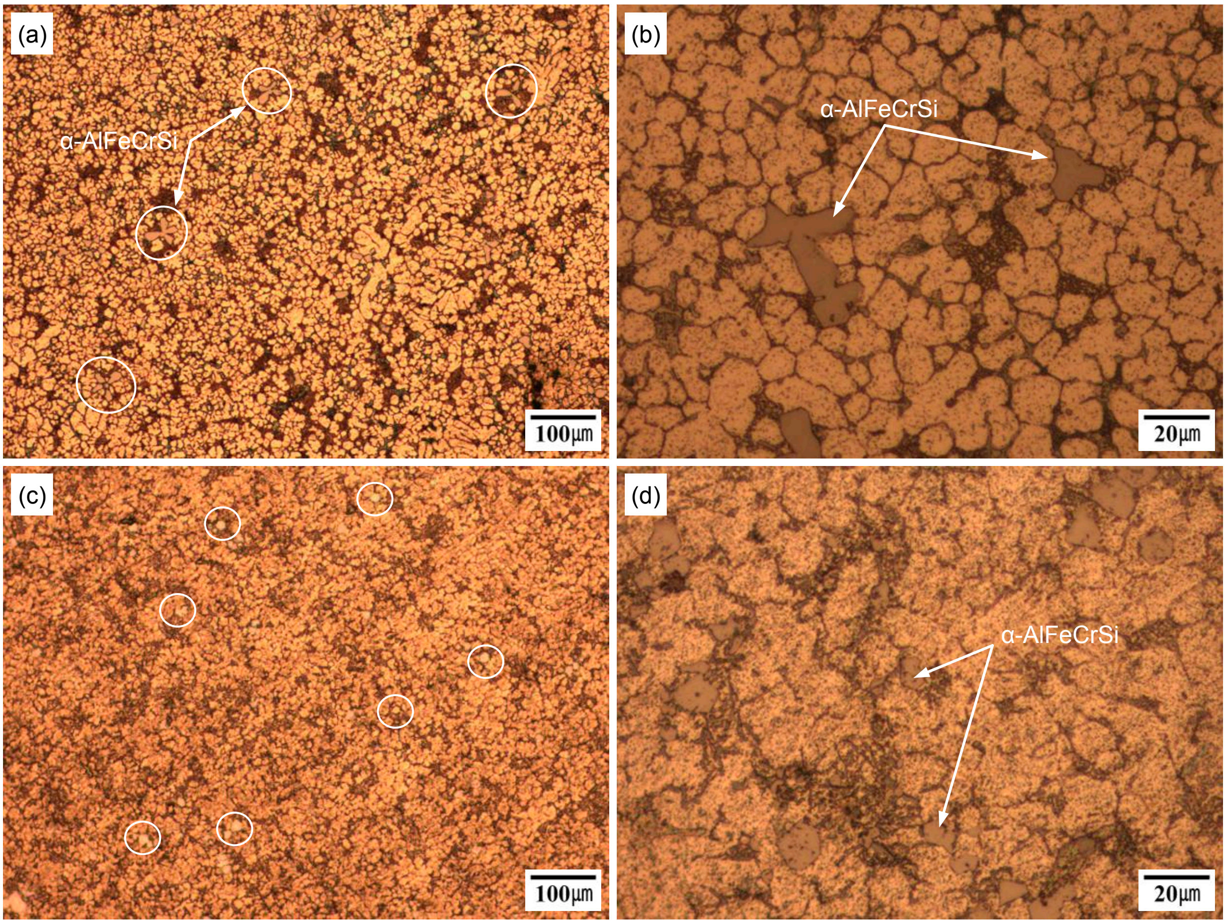

Fig. 5 Microstructure of No. 3 alloy obtained by (a) and (b) HPDC and (c) and (d) melt-shearing at 650℃ before HPDC, respectively.

Typical microstructures of the Cr added alloy of No. 3 produced by the HPDC process were shown in Figs. 5(a) and (b) in low and high magnifications, respectively. In the microstructure of Fig. 5(a), the typical dendritic α-Al phase was observed with the primary α-AlFeCrSi phase having rosette-type morphology as indicated by white circles. The formation of the primary α-AlFeCrSi phase was well expected by the J. Mat. Pro. software as shown in Fig. 2(c). The primary α-AlFeCrSi phase was reported to have a fine dispersed morphology replacing the formation of the β-AlFeSi phase in the Cr added Al-Si-Fe alloy [4]. However, in the present study, the primary α-AlFeCrSi phase showed a rosette-type morphology having 3-/4-fold rotational symmetry in 2-dimensional surface of microstructure. In the microstructure of No. 3 alloy, it should be noted that the β-AlFeSi phase was not observed to form into the deleterious large needle-shape as both primary and eutectic, implying that the primary α-AlFeCrSi phase successfully replaces the β-AlFeSi during solidification. The microstructures of the Cr added alloy of No. 3 processed by melt-shearing at 650℃ before HPDC are shown in Figs. 5(c) and (d). The primary α-AlFeCrSi phases are observed to be well dispersed equi-axed faceted grains. It is suggested that the rosette-type growth of the primary α-AlFeCrSi phase was effectively prevented by the turbulent flow in the vicinity of the primary α-AlFeCrSi phase during the melt-shearing process.

The tensile properties of No. 1~3 alloys produced both by HPDC and combined process of melt-shearing and HPDC were evaluated and summarized in Fig. 6. The tensile strengths of No. 1~3 alloys produced by HPDC was evaluated to be 230~240MPa, as shown in Fig. 6(a). Despite some modification features of the harmful primary β-AlFeSi phase in No. 2 and No. 3 alloys (compare Figs. (a) and (b) of 3, 4, and 5) with Mn and Cr additions, respectively, it is found that there are small differences among the tensile strengths of No. 1~3 alloys produced by HPDC.

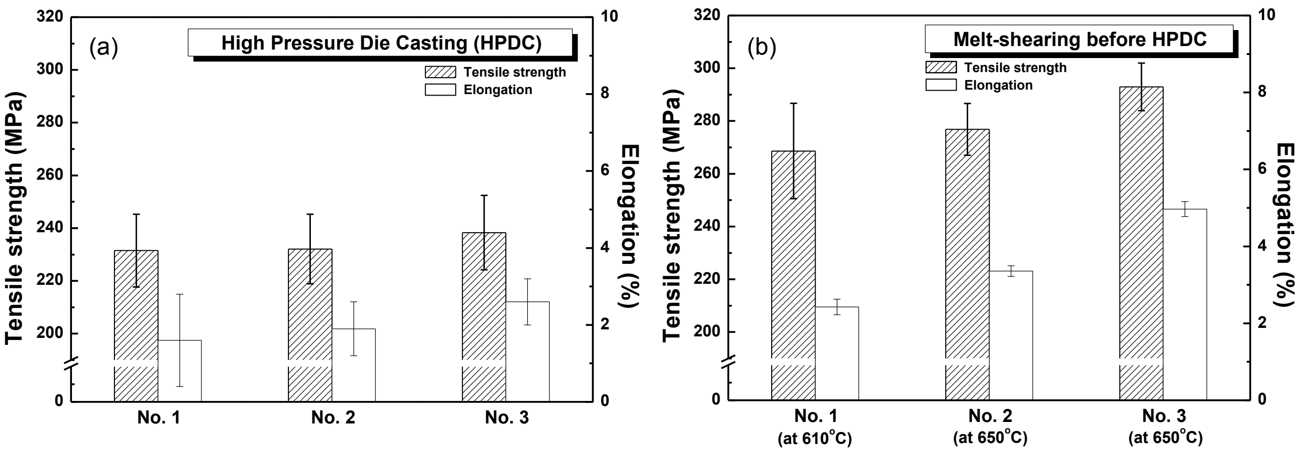

Fig. 6 Tensile properties of No. 1~3 alloys produced by (a) HPDC and (b) melt-shearing before HPDC.

It is found that the tensile strength increased up to average 268MPa by melt-shearing of No. 1 alloy at 610℃. This means that shortening the β-AlFeSi phase and size reduction of the primary α-AlFeMnSi phase effectively improves the tensile properties of No. 1 alloy. The tensile strength of No. 2 alloy melt-sheared at 650℃ was increased up to average 277MPa, owing to the modification of the irregular or cross-/star-like primary α-AlFeMnSi phase into the fine dispersed equi-axed faceted grains despite the existence of small amount of the eutectic β-AlFeSi phase. The tensile strength of No. 3 alloy melt-sheared at 650℃ was evaluated as about 293MPa, which was the highest strength among the test casting alloys containing various iron-bearing intermetallic phases.

4. Conclusions

Melt-shearing before HPDC was proved to be effective on the modification of various iron-bearing intermetallic phases, resulting in the increase of tensile strength. The specific effects of melt shearing on various alloys are summarized as follows.

(1) In the Fe-rich alloy of No. 1 (base alloy), the length of the primary β-AlFeSi phase was shortened and the size distribution of the faceted primary α-AlFeMnSi phase decreased by melt-shearing, slightly improving tensile strength.

(2) In the Mn added alloy of No. 2, the irregular or cross-/star-like primary α-AlFeMnSi phase was effectively modified into fine dispersed equi-axed grains, leading to the tensile strength increment up to 277MPa.

(3) In the Cr added alloy of No. 3, the rosette-type primary α-AlFeCrSi phase was changed into fine dispersed equi-axed grains. Moreover, the formation the β-AlFeSi phase was entirely replaced by the formation of the α-AlFeCrSi phase during both HPDC only and the melt shearing before HPDC. The resultant tensile strength was evaluated as the highest value up to 292MPa.

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