Journal of Alloys and Compounds 508 (2010) 315–319

Contents lists available at ScienceDirect

Journal of Alloys and Compounds

journal homepage: www.elsevier.com/locate/jallcom

# Effect of plastic deformation and semisolid forming on iron–manganese rich intermetallics in Al–8Si–3Cu–4Fe–2Mn alloy

S.G. Shabestari [a,*], M. Ghanbari [b]

[a] Center of Excellence for Advanced Materials Processing (CEAMP), School of Metallurgy and Materials Engineering, Iran University of Science and Technology (IUST), Narmak, Tehran, Iran
[b] School of Metallurgy and Materials Engineering, Iran University of Science and Technology (IUST), Narmak, Tehran, Iran

ARTICLE INFO

*Article history:*
Received 24 June 2010
Received in revised form 8 August 2010
Accepted 13 August 2010

*Keywords:*
Metals and alloys
Intermetallics
Microstructure
Metallography

ABSTRACT

The effects of plastic deformation and semisolid forming on the iron and manganese-rich intermetallics in Al–8Si–3Cu–4Fe–2Mn alloys were investigated. High volume fraction of $\alpha$-$Al_{15}(Fe,Mn)_3Si_2$ intermetallics were precipitated during melting and solidification of this alloy. The as-cast specimens were deformed plastically in rolling process to achieve 5–15% plastic deformation. Plastic deformation caused fragmentation of brittle intermetallics in Al–Si matrix because of their low formability and also produced $\alpha$-Al spheroids after reheating in semisolid temperature due to recrystallization and partial melting (RAP) process. Microstructural investigations revealed that appropriate semisolid structure with $\alpha$-Al sphericity of 93% and grain size of 64 $\mu$m was obtained after 10–15% plastic deformation of the samples and holding them in 580 °C for 15 min. Thixoforming process was performed in 580 °C up to 30% deformation on the prepared semisolid feedstocks, which caused re-distribution of the intermetallic particles in the final near net shape part. The mean equivalent diameter of iron–manganese bearing intermetallics decreased from 26–43 $\mu$m in the as-cast condition to 9–11 $\mu$m in the thixoformed samples. The aspect ratio also decreased from 1.7 in the as-cast condition to 1.2 in the thixoformed samples. Reduction of the mean diameter and aspect ratio of intermetallics and their good distribution after thixoforming process will improve tensile strength and formability of the alloy.



## 1. Introduction

In cast Al–Si alloys, intermetallic compounds with high thermal stability and hardness value are formed by entering transition elements specially iron, manganese, nickel, copper and chromium in chemical composition of these alloys [1]. Intermetallic compounds precipitate during different solidification stages because of low solid solubility of transition elements in aluminium and high affinity to react with this element.

Presence of intermetallic compounds in cast Al–Si alloys plays an important role in mechanical properties of these alloys. It is reported that presence of intermetallics in Al–Si alloys will increase the hardness and improve wear resistance to some extent [2,3]. However, mechanical properties, especially tensile strength and toughness of the Al–Si alloy containing iron impurities were deteriorated. The main reason of poor mechanical properties of cast aluminium alloys in the presence of high iron content was contributed to the large size, high aspect ratio and poor distribution of hard intermetallics in the matrix [4,5]. It is demonstrated that needle-like $\beta$-$Al_5FeSi$ compound with high aspect ratio and large size is the most harmful phase that cause poor mechanical properties of Al–Si alloys.

Based on the work done by Cacers et al. [6], it seems reasonable that decreasing the size and aspect ratio as well as uniform distribution of hard and brittle phases would eliminate their harmful effects on mechanical properties to some extent. In this situation, the hardness and wear resistance of the alloy will improve while the ductility will not decrease dramatically.

It is reported that addition of manganese and high cooling rate can strongly convert $\beta$-needles to $\alpha$-$Al_{15}(Fe,Mn)_3Si_2$ phase with less harmful effects on mechanical properties rather than $\beta$ phase [7–10].

Some efforts have been carried out to produce Al–Si alloys containing high volume fraction of intermetallic compounds with a uniform distribution of intermetallic compounds and minimum mean particle size in the matrix using rapid solidification processing and squeeze casting [9]. The microstructure of these materials is similar to a metal matrix composite reinforced with intermetallic compounds. But few works have been done to produce

* Corresponding author. Tel.: +98 21 77240371; fax: +98 21 77240371.
*E-mail addresses:* shabestari@iust.ac.ir (S.G. Shabestari), mohsen.ghanbari@gmail.com (M. Ghanbari).

0925-8388/$ – see front matter 
doi:10.1016/j.jallcom.2010.08.043

**Table 1**
Chemical composition of 380 alloy used in this study.

| Alloying element | Si | Cu | Fe | Mn | Mg | Ni | Al |
|---|---|---|---|---|---|---|---|
| Wt% | 8.5 | 2.4 | 0.25 | 0.05 | 0.03 | 0.01 | Rem |

bulk Al–Si alloys with high volume fraction of iron-rich intermetallics.

While addition of manganese can improve the morphology of intermetallics, but it generally increases total volume fraction of intermetallics and their mean diameter size in high level of iron content and again can deteriorate mechanical properties [11]. Because of low ductility of Al–Si alloys, modification of the intermetallics by mechanical working is almost impossible and final product cannot have any application. But improvement of semisolid forming processes such as thixoforming can help us to achieve this goal [12,13].

In this investigation, the effects of plastic deformation and semisolid forming on the microstructure of as-cast Al–Si–Cu alloys with excess amounts of iron and manganese have been studied. The aim of this process is to obtain a near net shape part containing high volume fraction of Fe and Mn bearing intermetallics and having uniform dispersion of intermetallic particles. Therefore, an aluminium matrix composite reinforced with iron and manganese bearing intermetallics particles can be produced via decreasing the mean diameter and aspect ratio of these compounds in a hypo eutectic Al–Si–Cu alloy.

## 2. Materials and methods

Appropriate amount of commercial 380 aluminium ingot was charged in a preheated graphite crucible and the chemical composition is given in Table 1. It was melted in an electrical resistance furnace under the protection of coveral fluxes.

Iron and manganese were added to the melt at 800 °C using ALTAB Fe Compact (75 wt% Fe, 15 wt% Al and 10 wt% nonhygroscopic Na-free flux) and Mn compact(75 wt% Mn, 15 wt% Al, and 10 wt% nonhygroscopic Na-free flux), respectively. The melt was kept in this temperature for about 10 min for complete homogenization and dissolution of alloying elements. Then it was cooled down to 700 °C and poured into a copper mould preheated to 200 °C. Temperature variation of the melt during solidification process was measured using a calibrated K-type thermocouple and the results was recorded using a computer aided acquisition system apparatus to obtain cooling curve and first derivative of the temperature versus time to reveal the phase transformation occurred during solidification. Three specimens with the dimension of 30 mm × 30 mm × 100 mm were cut from the cast ingots and then heated up to 300 °C. They were imposed plastic deformation using two mills rolling device to achieve 5%, 10% and 15% reduction in cross section. Samples were reheated to semisolid temperature at 580 °C for 15 min and then pressed into a cylindrical shape with diameter of 35 mm and height of 20 mm using a laboratory hydraulic press. In this situation the sample were deformed up to 30%. The near net shape samples were sectioned, polished, and etched with 0.5% HF. Microstructural investigations were carried out on samples in the progressive stages of experiment by optical and scanning electron microscopy (SEM) model WEGA with back-scatter detector and EDS micro analysis. Quantitative image analysis was performed using CLEMEX image analyzer software.

## 3. Results and discussion

The chemical composition of produced alloy is presented in Table 2. As seen in this table, the iron and manganese content of the alloy increased in comparison with the chemical composition of the initial alloy. This would increase the total content of the iron and manganese containing intermetallics, specially $\alpha$-$Al_{15}(Mn, Fe)_3Si_2$ [7].

Fig. 1 shows the optical micrograph of the investigated alloy in the as-cast condition. Intermetallics (marked with letter A) have commonly polyhedral morphology in this condition and are in the vicinity of $\alpha$-Al (letter B) and eutectic Si (letter C) phases. EDS microanalysis of these phases demonstrates that they are $\alpha$-$Al_{15}$ (Mn, $Fe)_3Si_2$ phases, similar to what reported by others [1].

**Table 2**
Chemical composition of the as-cast alloys.

| Alloying element | Si | Cu | Fe | Mn | Mg | Ni | Al |
|---|---|---|---|---|---|---|---|
| Wt% | 8.1 | 2.07 | 3.93 | 1.84 | 0.02 | 0.01 | Rem |

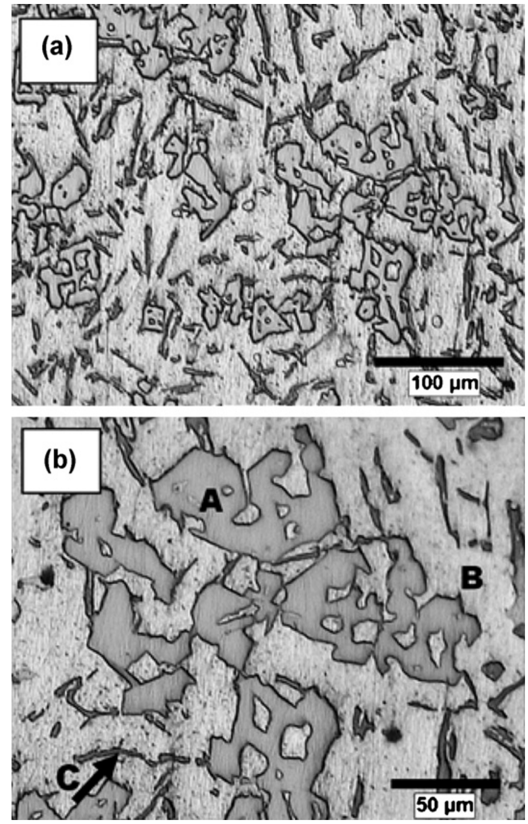


**Fig. 1.** Microstructure of the alloys in as-cast condition (a) lower magnification and (b) higher magnification.

All intermetallic compounds were in the form of primary-$\alpha$ polyhedral and no $\beta$-plates were seen in the as-cast microstructure. This is mainly due to the proper ratio of iron to manganese which is kept in 2:1. Nucleation and growth of eutectic silicon on the intermetallic compounds are clearly observed in the micrographs. In fact intermetallics are suitable nucleation sites for eutectic silicon [14]. The volume fraction of intermetallics in the final structure was measured using image analyzer and it is about 16.7 vol%.

Cooling curve and first derivative curve of temperature versus time during solidification of the alloy are shown in Fig. 2(a) and (b), respectively. Small peaks were observed before $\alpha$-Al peak that are attributed to precipitation of primary intermetallic compounds at high temperatures [15].

Determination of liquid volume fraction versus temperature is an important parameter in semisolid processing. It is calculated from the area restricted between the first derivative of untransformed curve (called as zero curve) and the actual cooling curve of the alloy at different temperatures. Zero curve is obtained from extrapolation of the initial part of the first derivative curve before the solidification starts [1]. As seen in Fig. 2(b), volume fraction of liquid phase varies to about 30–50% between 570 °C and 580 °C and this temperature range is proper for semisolid forming of the alloys [12].

The SEM micrograph of the cold rolled alloy parallel to the rolling direction (RD) is shown in Fig. 3. The phases pointed with letter (A) and (B) are cracked silicon and $\alpha$-$Al_{15}(Fe,Mn)_3Si_2$ phase in the dark $\alpha$-Al matrix, respectively.

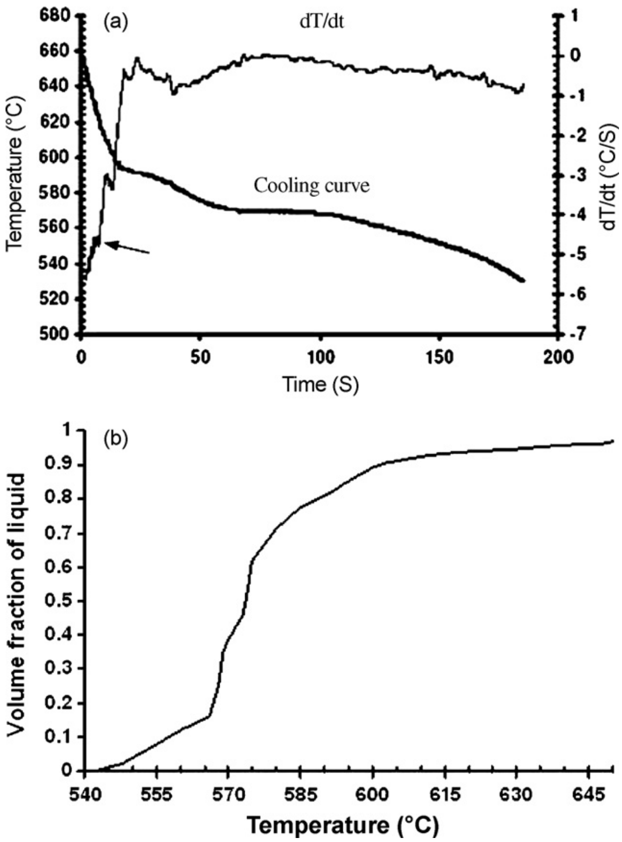


**Fig. 2.** (a) Cooling curve and first derivative of temperature versus time of alloy and (b) variation of liquid volume fraction versus temperature.

Brittle phases such as silicon particles and α-intermetallic phases undergo cracking because of applied stress and low formability of intermetallics in rolling process. Because of the larger size of intermetallics rather than silicon particles, they are more sensitive to cracking [6]. These micro-cracks can propagate in the structure of brittle phases and finally fragment them to smaller particles. This is because of limited or no ductility of ordered intermetallic compounds. Because of low ductility of the alloy, imposing high level of deformation is impossible, therefore, fragmented particles cannot be displaced, and so they remain in the place.

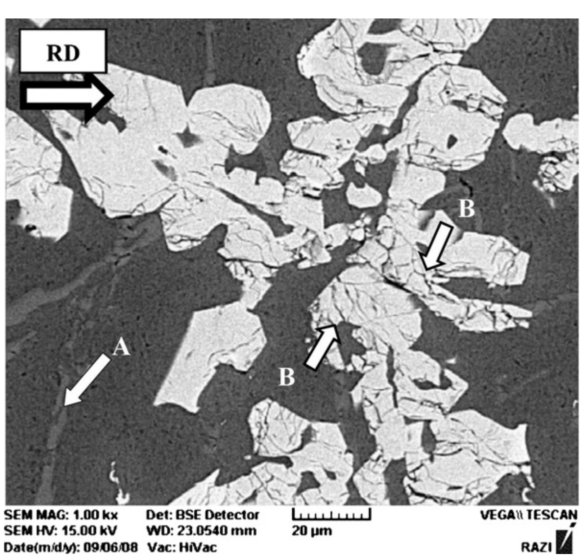


**Fig. 3.** Microstructure of the specimen showing micro-cracks in intermetallic compounds (letter B) and silicon particles (letter A) after different amounts of plastic deformation parallel to the rolling direction (RD).

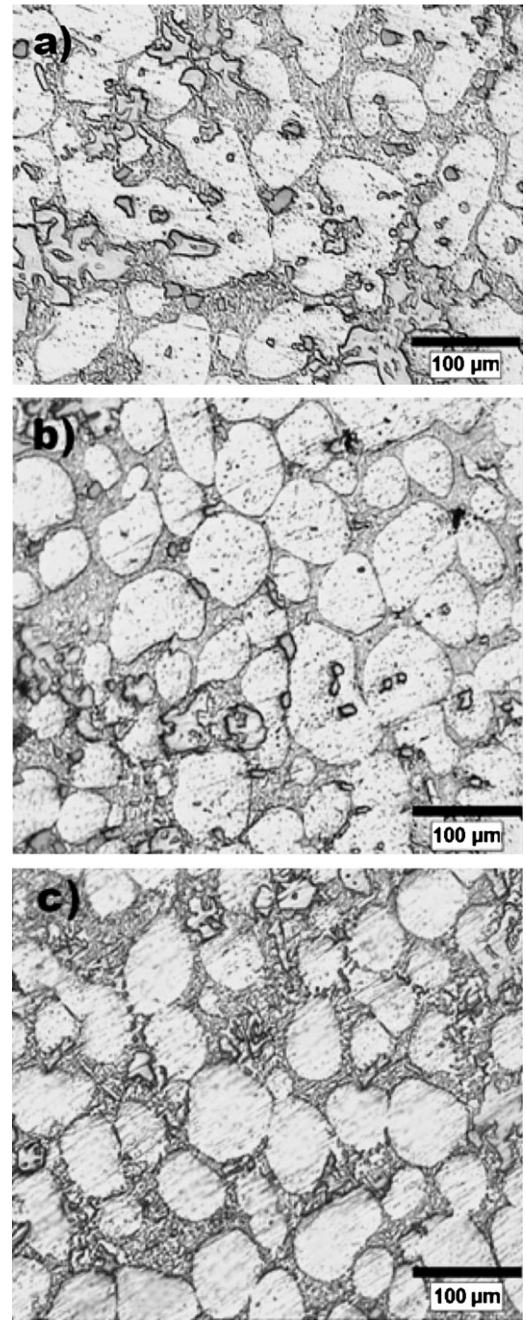


**Fig. 4.** Microstructure of the specimens after isothermal holding in 580 °C for 15 min (a) 5% deformed, (b) 10% deformed and (c) 15% deformed.

The microstructure of the specimen after isothermal holding of the deformed specimens in 580 °C for 15 min is shown in Fig. 4.

In Fig. 4(a) it is clearly seen that the sphericity of α-Al globules in specimen with 5% deformation is not sufficient and contiguity of the eutectic phase has not been reached. Entrapped eutectic phase is seen in the α-Al grains. The sphericity of α-Al and contiguity of eutectic phase were improved in samples having 10% deformation. The best sphericity of α-Al and contiguity of the eutectic phase have been obtained in sample having 15% deformation which is suitable for semisolid forming. This is mainly due to the increment of dislocation density which is basic phenomena in strain induced melt activation (SIMA) process as was discussed elsewhere [12,16]. In fact, further strain caused increase of dislocation density in α-Al

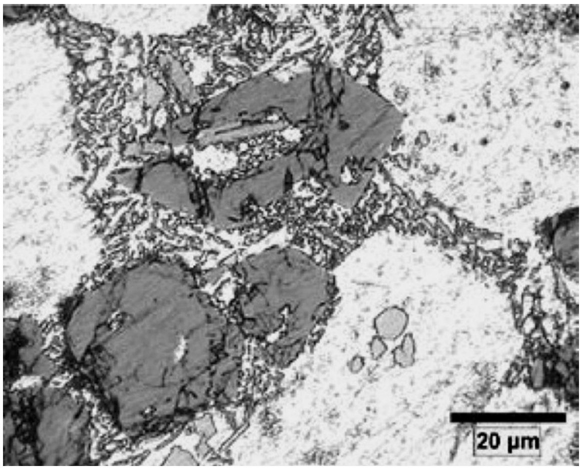


**Fig. 5.** Liquid constituent in the semisolid state in the vicinity of the fragmented intermetallics.

dendrites. By heating the sample up to semisolid temperature, dislocation structure produces subgrain boundaries and subsequently subgrains transforms to new grains (recrystallization) to decrease the Gibbs free energy of the system. Further holding the specimens at semisolid temperature, cause the penetration of eutectic liquid to new grain boundaries and produces discrete spherical grains which cause more reduction in free energy of the system.

The volume fraction of liquid phase obtained in 580 °C from thermal analysis of the alloy was about 58%. Solid constituents of the alloy in the semisolid state are composed of α-Al grains and iron and manganese-rich intermetallic compounds which are fragmented due to the plastic deformation. Eutectic melt penetrates to the defects and high energy zones to reduce the energy of the system. As shown in Fig. 5, fragmented intermetallic compounds are in the vicinity of the eutectic constituents and therefore in the semisolid state, melt penetrates to the cracks of intermetallics and surrounds them.

Forming of the semisolid mixture to a near net shape part (thixoforming) caused displacement of the fragmented intermetallics. This caused re-distribution of intermetallic compounds in the matrix and will improve mechanical properties of the alloy.

Final microstructure of the specimen after semisolid forming is shown in Fig. 6.

It is clearly seen that the size of intermetallic particles has been decreased considerably and their distribution was improved in comparison with the as-cast condition. SEM micrograph of a thixoformed sample also clearly revealed displacement of the intermetallic compounds. Displacement of these particles has been occurred due to the movement of the solid constituents into the liquid phase during thixoforming in semisolid state. Because there is no modifying element such as strontium in the chemical composition of the alloy and on the other hand slow cooling rate of the eutectic melt in thixoforming, the eutectic structure were remained unmodified. So no clear boundaries are seen between α-Al grains and eutectic structure.

The size, distribution and aspect ratio of iron and manganese-rich intermetallic compounds are important factors to determine mechanical properties of the alloy [7]. From theoretical points of view, there is relationship between mechanical properties and the microstructural features of the alloy. As reported by Cacers et al. [6] for reinforcement constituents exist in a soft matrix, probability of the crack initiation from brittle phases in the material depends on the equivalent diameter of particles, $d$, aspect ratio, $\alpha$, and average distance of the intermetallics, $\lambda$. For intermetallics in aluminium matrix, the parameter of intermetallic appearance index, *IPA*, could

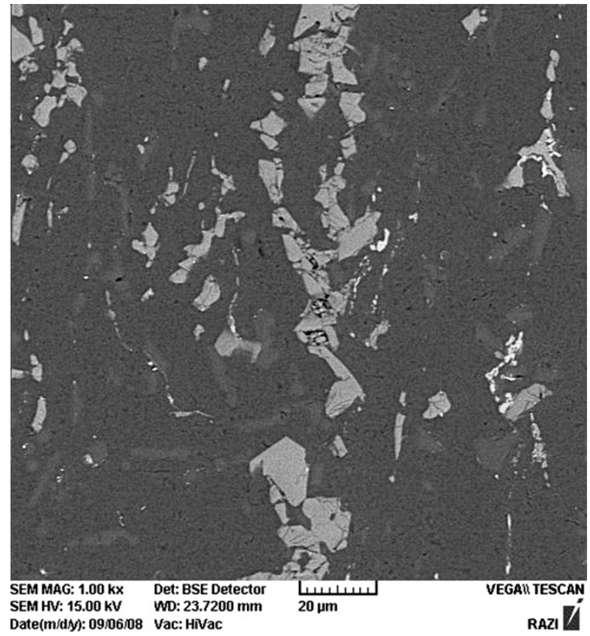


**Fig. 6.** SEM micrograph of the thixoformed sample.

be defined as follow:

$$IPA = \frac{\alpha d}{\lambda} \qquad (1)$$

Decrease in *IPA* parameter means decrease in $d$, $\alpha$ and increase in $\lambda$ value. Decrease in $d$ and $\alpha$ value, increases the strength of the material. By increasing $\lambda$ value, the strength of the material decreases but the toughness of the material increases. It is expected that by decreasing *IPA* the mechanical properties of the alloy would be improved.

Microstructural investigation using image analyzer software revealed that intermetallic equivalent diameter and aspect ratio have been reduced from $43 \pm 9$ µm and 1.7 in as-cast condition to about $10 \pm 1$ µm and 1.2 in thixoformed condition, respectively. In this situation the *IPA* for thixoformed samples is smaller than what for as-cast condition are. Therefore, the mechanical properties of the alloy will improve by decreasing *IPA* value based on the concept described above.

## 4. Conclusion

Based on the results obtained in this research, it can be concluded that applying 10–15% plastic deformation followed by isothermal holding of cast Al–8Si–3Cu–4Fe–2Mn alloy at 580 °C for 15 min will produce fragmented $\alpha\text{-}Al_{15}(Fe,Mn)_3Si_2$ intermetallics in a globular α-Al matrix surrounded by the low melting point Al–Si eutectic. Volume fraction of intermetallics with polyhedral morphology is 16.7 vol%. In this situation, mean equivalent diameter and aspect ratio of fragmented intermetallic particles reduced from 26–43 µm and 1.7 in as-cast condition to 9–11 µm and 1.2 in the thixoformed condition, respectively. Based on the fracture mechanics theory and regarding the *IPA* concept, the mechanical properties of the produced alloy will be improved.

## References

[1] L. Backerud, G. Chai, J. Tamminen, Solidification Characteristics of Aluminium Alloys: Volume 2-Foundry Alloys, AFS/Skanalumium, Stockholm, Sweden, 1990.
[2] V. Abouei, H. Saghafian, S.G. Shabestari, M. Zarghami, Effect of Fe-Rich Intermetallics on the Wear Beheaviour of Eutectic Al-Si Piston Alloy (LM13), Mater. & Des. 31 (2010) 3518–3524.

[3] R. Taghiabadi, H.M. Ghasemi, S.G. Shabestari, Mater. Sci. Eng. A 490 (2008) 162–170.
[4] Z. Ma, A.M. Samuel, F.H. Samuel, H.W. Doty, S. Valtierra, Mater. Sci. Eng. A 490 (2008) 36–51.
[5] S.S. Sreeja Kumari, R.M. Pillai, B.C. Pai, J. Alloys Compd. 453 (2008) 167–173.
[6] C.H. Cacers, C.J. Davidson, J.R. Griffiths, Mater. Sci. Eng. A 195 (1995) 171–179.
[7] S.G. Shabestari, Mater. Sci. Eng. A 384 (2004) 289–298.
[8] S. Seifeddine, S. Johansson, I.L. Svensson, Mater. Sci. Eng. A 490 (2008) 385–390.
[9] M. Arhami, F. Sarioglu, A. Kalkanli, M. Hashemipour, Mater. Sci. Eng. A 485 (2008) 218–223.
[10] L.F. Mondolfo, Aluminium Alloys: Structure and Properties, Butterworths, 1976.
[11] J.Y. Hwang, M.J. Doty, H.W. Kaufman, Mater. Sci. Eng. A 488 (2008) 496–504.
[12] S. Ashouri, M. Nili-Ahmadabadi, M. Moradi, J. Alloys Compd. 466 (2007) 67–72.
[13] Y. Birol, J. Alloys Compd. 470 (2008) 183–187.
[14] S.G. Shabestari, S. Ghodrat, Mater. Sci. Eng. A 467 (2007) 150–158.
[15] S.G. Shabestari, S. Ghodrat, J. Can. Metall. Q. 45 (2) (2006) 207–214.
[16] Y. Birol, Mater. Sci. Forum 519–521 (2006) 1919–1924.