Journal of Babylon University/Engineering Sciences/ No.(5)/ Vol.(20): 2012

Study the Influence of Iron Content and Solution Treatment on Microstructure and Hardness of Al-Si Eutectic

Sura Abd Al-Zahrah Muhsion
Materials Eng. university of Babylon

A b s t r a c t

The present work deals with the influence of iron content and solution treatment on the microstructure and hardness of unmodified and Sr-modified eutectic Al–Si alloys. The castings were produced in cast iron permanent mold and were solution treated at 550 ˚C for 7hrs and followed by artificial aging at 160 ˚C for 6hrs, i.e., T6-temper. The microstructure changes in the ß-Al_5FeSi particle morphology were analyzed. The results indicate that dendrite arm spacing is depended on the cooling rate rather than the chemical composition. An increasing in the iron content leads to increase the hardness either in the as-cast condition or after T6-temper. The Sr-modified alloys have higher hardness than unmodified at all Fe-added values.

Keywords:
Al–Si alloys, Fe additions, Sr modification, Solution heat treatment

الخلاصة

يهتم هذا البحث بدراسة تأثيراضافة الحديد والمعاملة الحرارية المحلولية على البنية المجهرية والصلادة لسبيكة الألمنيوم – سليكون الايوتكتيكي المعدلة بإضافة السترونتيوم وغير المعدلة. تم صب النماذج في قلب من حديد الزهر وتم معاملتها محلوليا عند درجة حرارة550 درجة سليليزية لمدة سبع ساعات ثم عتقت صناعيا عند 160 درجة سليليزية لمدة ستة ساعات.لقد تم تحليل التغيرات في مورفولوجية طور بيتا– الألمنيوم – 5 حديد– سليكون الناتج عن إضافة الحديد بنسب من (0.137% لغاية 2.58%). وقد أشارت النتائج إلى إن المسافات بين الأذرع الشجيرية يعتمد على معدل التبريد أكثر من اعتماده على التركيب الكيمياوي. كما أثبتت النتائج إلى إن أي زيادة في نسبة الحديد أدى إلى زيادة الصلادة سواء كانت المصبوبة بعد السباكة مباشرة أو بعد المعاملة المحلولية. كما إن السبائك المعدلة بالسترونتيوم أعلى صلادة من الغير معدلة لجميع نسب الحديد المضافة.

1. Introduction

Iron is considered the most harmful element since its presence enhances the precipitation of many iron intermetallic phases in the form of long, intercepting platelets (or needles), and, hence, unacceptable mechanical properties [Samuel, Villeneuve, Valtiera, 2001, Shabsestari, 2004, Dinnis Cameron, Taylor John, Dahle, 2005]. These intermetallic phases seriously degrade the strength of alloys. Moreover, because they often form during solidification of the eutectic phase they may interfering with interdendritic feeding and thus promote porosity [Samuel , Villeneuve, Valtiera, 2001, Dinnis Cameron, Taylor John, Dahle, 2005]. The Al_5FeSi (ß-phase) is the most commonly observed compound which usually formed in the Al–Al_5FeSi–Si eutectic as thin plates interspersed with Si flakes or fibers; it can also be found with a rod-like morphology. If Mn is present, iron forms $Al_{15}(Fe, Mn)_3Si_2$ (ß-phase), often in the shape of Chinese script. Thus optimum control of the concentration of iron in foundry alloys is essential. It was reported that, the main factors contribute to the formation of the ß-iron intermetallic phase: (a) An Fe/Si atomic ratio close to unity, (b) a low cooling rate ~0.8 ◦C/s, and (c) low Mn and Cr concentrations [Mondolfo, 1990]. Cooling rate has a direct impact on the equilibrium kinetics and quantities of iron phase present in the microstructure. A low cooling rate favors the precipitation of the ß-phase only precise crystallographic planes. On the other hand, the α-iron intermetallic $Al_{15}(Fe, Mn)_3Si_2$ phase, promotes the diffusion of

elements towards its surfaces [Narayanan ,Samuel, Gruzleski, 1995]. This phase occurs when the cooling rate is sufficiently high, i.e., of the order of 10 ◦C/s. It has been shown that the transition from the ß to the α-iron intermetallic phase takes place at lower cooling rates if the iron concentration is less than 0.7 wt. % [Iglessis , Frantz,1977]. Some researches indicate that increasing the melt superheat or the cooling rate will increase the iron concentration needed for the formation of ß-Al_5FeSi phase [Awano , Shimizu , 1990]. It was reported that iron intermetallic phase would form in thick sections of casting where crystallization is slow, even at low Fe-concentrations ~0.15% [Yaneva, Stoichev, Kamenova, Budurov, 1984].

This study was undertaken to examine in details the effect of Fe, Sr, and Fe–Sr on the ß -Al_5FeSi iron intermetallic phase characteristics, in particular, its length, thickness and surface area, as well as the influence of iron intermetallics on the porosity formation and alloy hardness.

2. Experimental procedure

Eutectic Al–12Si alloy was prepared by melting commercial aluminium in silicon carbide crucible (preheated at 200 ˚C) under a cover flux (50 wt %NaCl + 50 wt %KCl) gas furnace. The melt was brought to a temperature of 730 ˚C and then a calculated amount of commercial pure silicon, in bulk shape, was added to the melt. And degassed (to remove H2) using 1% commercial degasser, hexachloroethane (C_2Cl_6). After degassing, a part of untreated melt was poured into a cylindrical cast iron mould (18mm diameter and 150mm height) with its top open for pouring. Later, measured amounts of Fe, Ti, with/without Sr were added to the melt (Table 1). And degassed (to remove H_2) using 1% commercial degasser, hexachloroethane (C_2Cl_6). After degassing the molten metal was poured into a cylindrical cast iron mould (18mm diameter and 150mm height) heated to 200 ˚C. After solidification, three samples from each rod were cut with 20mm height from the bottom. One of them was taken for chemical analysis and the other two for microstructural characterization; one after casting and other after heat treatment. The heat treatment was carried by solution treated at 550 ˚C for 6hrs, then quenched in hot water (~60 ˚C), and followed by artificial aging at 150 ˚C for 6hrs, i.e., solid solution heat treatment and T6-temper. The purpose of this treatment is to remove solidification stresses and to change the structural morphologies of Si-particles and ß-phase. Samples for metallographic examination were grinded, polished, etched (with 0.5% HF in distilled water), and examined to determine the microstructural constituents present, and hardness measurements. For porosity measurements, approximately 12 fields for each specimen were examined at a magnification of 200×. While for ß-phase, the central surface area of the specimen was chosen and the largest 10 ß-needles were measured and the average was taken. The magnification used was depending on the size of ß-phase which increases with increasing Fe-content.

The chemical composition analysis was performed by using XRF spectroscopy (X-Met 3000 TX, horizon 600 series, model 2004). The chemical compositions of the alloys studied are shown in Table 1.

Table 1 – Chemical Composition of the Investigated Al–Si Alloys

Element	Si %	Fe%	Cu%	Mn %	Ti%	Sr	Al
A0	11.937	0.137	0.05	0.004	0.049	.....	Bal.
A0s	11.937	0.137	0.05	0.004	0.049	0.017	Bal
A1	11.862	1.190	0.046	0.002	0.042	.....	Bal.
A1s	11.850	1.188	0.049	0.003	0.043	0.017	Bal.
A2	11.930	1.502	0.047	0.004	0.043	.....	Bal.
A2s	11.930	1.502	0.047	0.004	0.043	0.017	Bal.
A3	1185	2.48	0.046	0.003	0.042	.....	Bal
A3s	1185	2.48	0.046	0.003	0.042	0.017	Bal
A4	1195	2.58	0.046	0.003	0.042	.....	Bal
A4s	1195	2.58	0.046	0.003	0.042	0.017	Bal

The microstructural investigation was carried out by Olympus microscope provided by Image J (Version 1.24o) analyzer system. Hardness was measured by Vickers process under 10 kg loads and 30 s, i.e., HV10. Six, at least, hardness measurements were taken for each cast and heat-treated specimen. The average values are shown in Table 2.

Table2. Hardness (HV10) Mpa

Sample No.	AS Cast		As heat treated	
	With out Sr	With Sr	With out Sr	With Sr
a	66	-----	50	-----
a1		68		52
b	74	------	57	----
b1	------	75	-----	60
c	75	-------	59	-----
c1	-------	78	------	63
d	78	-----	62	------
d1	-----	81	-------	66
e	80	-------	65	------
e1	------	83	-----	69

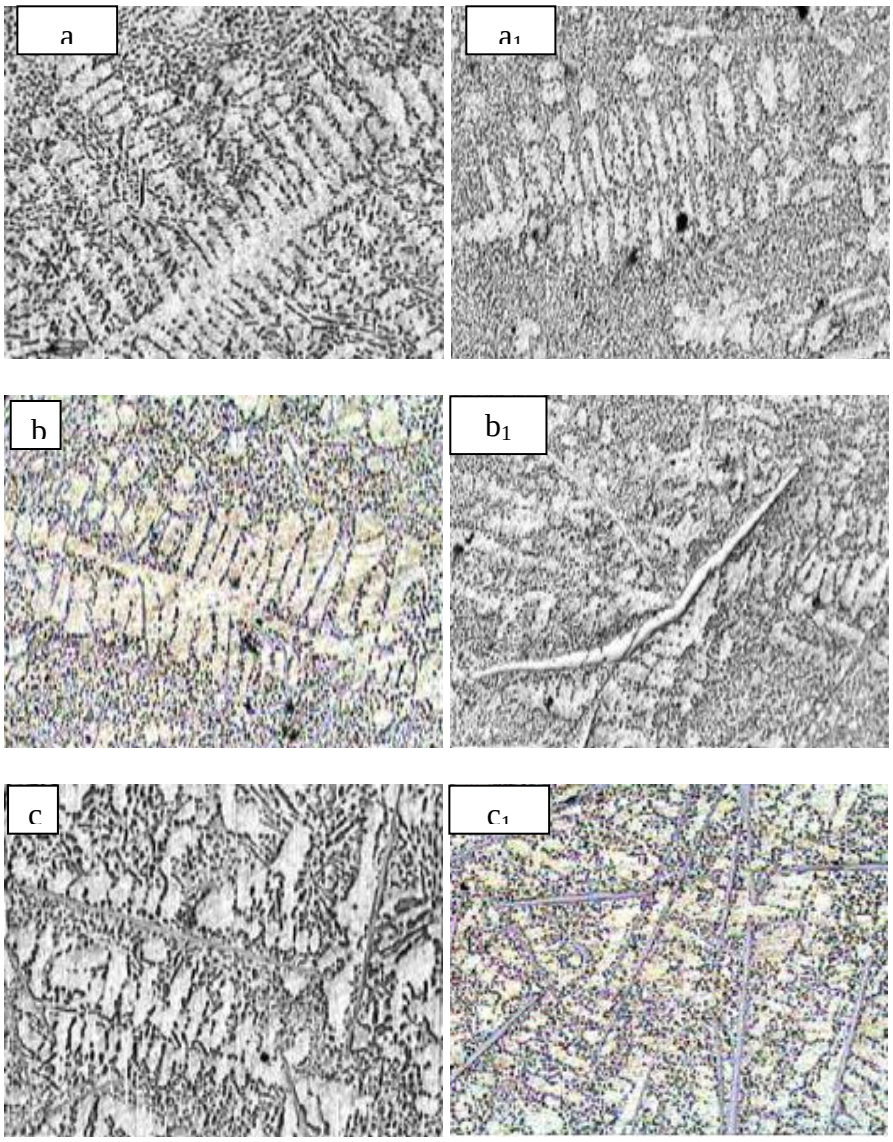
1. Results and discussion

3.1. *Microstructure*

AS recommended by the Aluminum Association (1990) [ASM Committee on Aluminum and Aluminum Alloys, 1990] the T6-temper (solution treated for 5–6hrs at

500–510 ˚C, quenching into hot water (60 ˚C), ageing for 2–5hrs at 150 ˚C, followed by air cooling) for Al-heat treatable Al-Si alloys.

This treatment is expected to result in precipitation hardening of Mg_2Si and Al_2Cu beside the change of Si-morphology. In these investigated alloys, there is no need to carry out artificial aging where Mg% was kept at <0.015 and Cu% at <0.16, as shown in Table 1. Hence, the precipitates of both Mg_2Si and Al2Cu were not detected in any specimen. But Si-particle size was a little increased. The micrographs in Fig. 1 show the microstructure of samples revealing that the secondary arm spacing (DAS) was unchanged with increasing neither Fe-content nor Sr-addition.



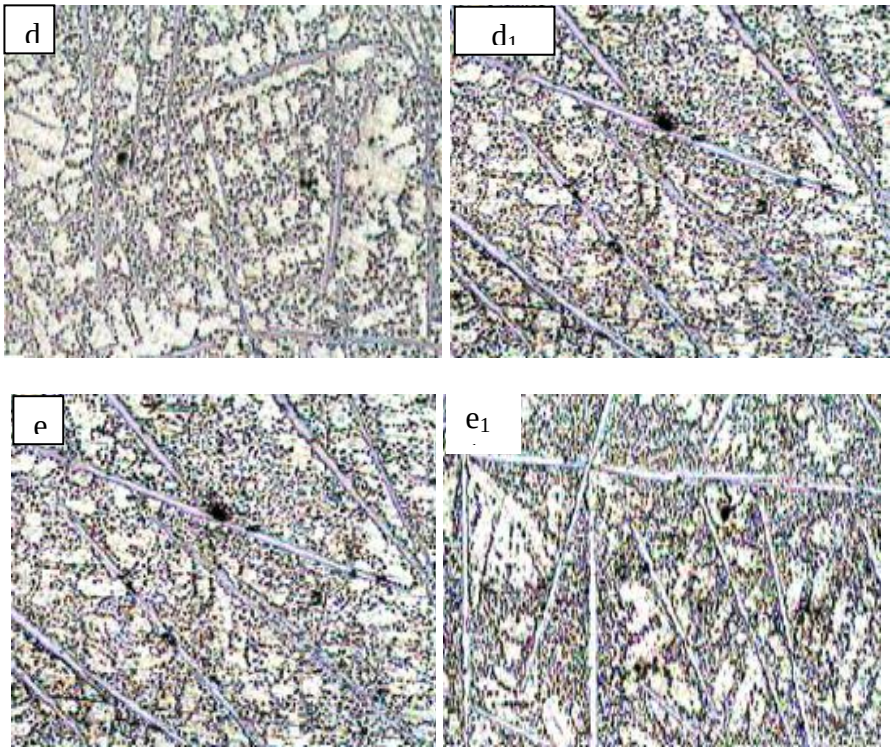


Fig. 1 – Microstructure of as-cast unmodified (a–e), and modified (a_1–e_1) alloys. The DAS value was about 21.5µm for all samples as shown in Table 3. Note that all at magnification x200.

This indicates that DAS is strongly depended on the cooling rate rather than the chemical composition.

3.1.1. ß -Phase characteristics

The variation in ß-phase average maximum length as a function of added Fe-content is shown in Fig. 2 and listed in Table 3. It is interesting to find that, the average maximum length of ß-phase in Sr-modified alloys is larger than in unmodified alloys. The variation in length of ß-phase increases with increasing Fe-content, i.e., the increment reaches up to ~ 23% at 2.58% Fe.

Table 3 –Average DAS, average dimensions of ß-Al5FeSi needles and average surface area obtained from investigated alloys

Sample No.	Iron content	Average (DAS) µm	Average length of ß-Al_5FeSi needles	Average width of ß-Al_5FeSi needles	Average surface area of ß-Al_5FeSi needles
a	0.137	18	34	2.3	78.2
a1	0.137	22	34	2.5	612.5
b	1.190	21	350	2.6	910
b1	1.190	19	449	2.9	1302.1
c	1.502	22	460	3.2	1472
c1	1.502	23	520	3.8	1976

d	2.48	24	720	8.5	6120
d1	2.48	22	850	9.8	8330
e	2.58	21	760	9.0	6840
e1	2.58	23	980	9.3	9114

This is in agreement with the work of Samuel et al.[Samuel, Villeneuve, Doty, Valtiera, 2001, Samuel, Dotty, 1996]. Fig. 3 shows the microstructure of some investigated samples that indicate the change of ß-platelets size with increasing Fe-content.

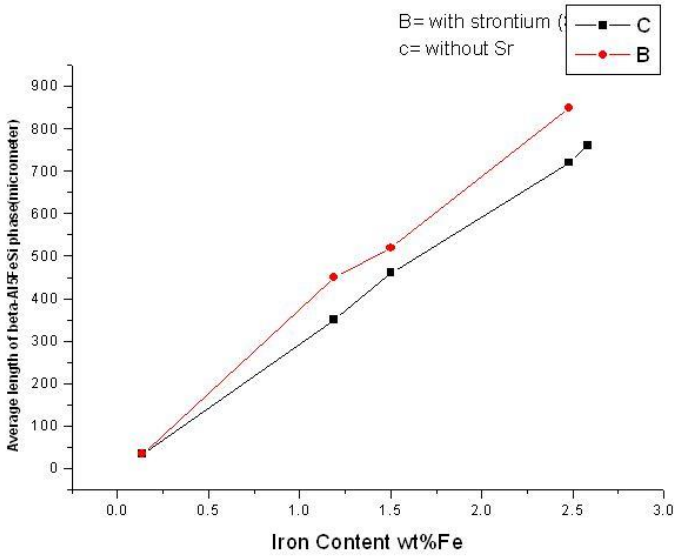
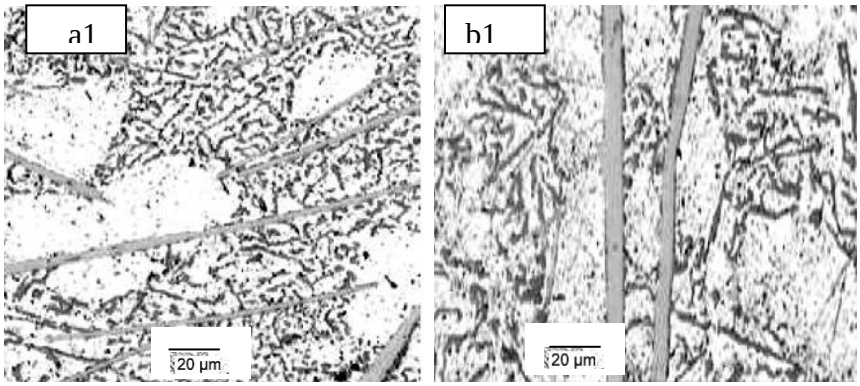


Fig.2 The relation between Fe content and length of ß-Fe



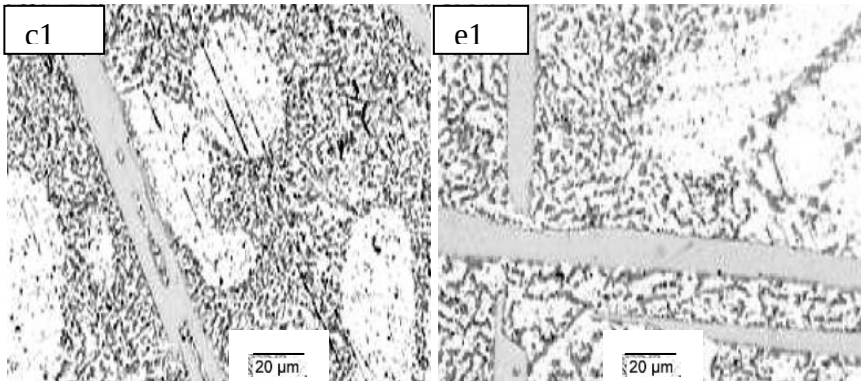


Fig. 3 shows the microstructure of some investigated samples that indicate the change of ß-platelets size with increasing Fe-content

3.1.2. Shrinkage porosity

The area percent of porosity observed under optical microscope on the measured sample surface area for different investigated alloys is shown in Fig.4.

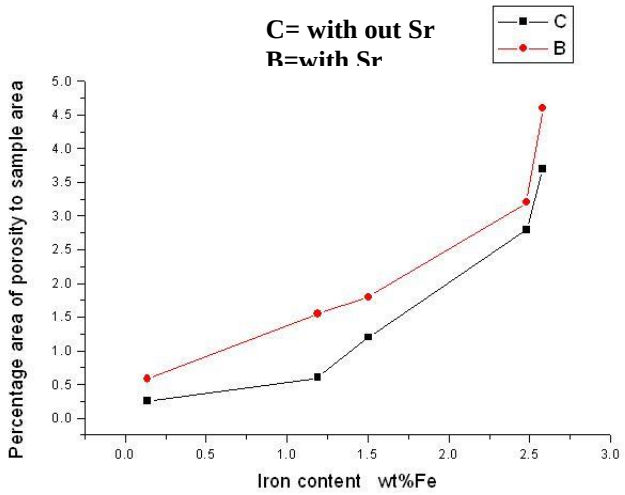


Fig.4 shows the percentage of porosity area to sample area

It should be noted here that all heats were continuously degassed prior to casting (to minimize the effect of gas and inclusion-related porosity).

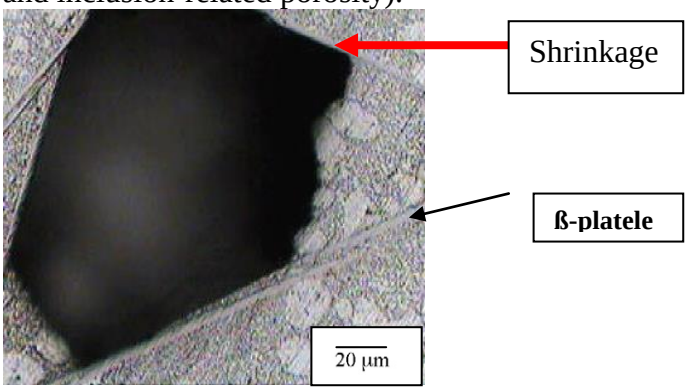


Fig.5 shows the effect of ß-platelets on the forming of the pours holes

As can be seen, no significant variation in area percent porosity is observed between unmodified and modified alloys. But, in general, the porosity curve of modified alloys lay directly over the unmodified, i.e., Sr-addition does not dramatically increase porosity. But leads to redistribution of global volume of porosity from coarse localized shrinkage into a rather uniformly distributed microporosity, as proved by Argo and Gruzleski (1988) [Argo, Gruzleski, 1988]. On other hand, porosity is directly proportional with Fe%. This is due to the precipitation of thick ß-phase platelets which, in turn, prevent the liquid metal to fill the spaces between the branched platelets as detected previously by Crepeau (1996) [Crepeau, 1996] this can be seen in Fig. 5. That shows these shrinkage holes surrounded by ß -Fe platelets.

3.2. Hardness

Fig.6 shows the effect of Fe% on the hardness of as-cast and solution heat treated (SHT) alloys. As can be seen, addition of Fe results in increase of hardness in both cases. Addition of Fe leads to increase the hardness of as-cast alloys by about 21% and for SHT by about 30%. Also it can be seen that the maximum value of hardness was in A1 alloy as a result of combined effect of the addition of Sr as refiner and modifier of Si eutectic, the addition of iron in 2.58%.

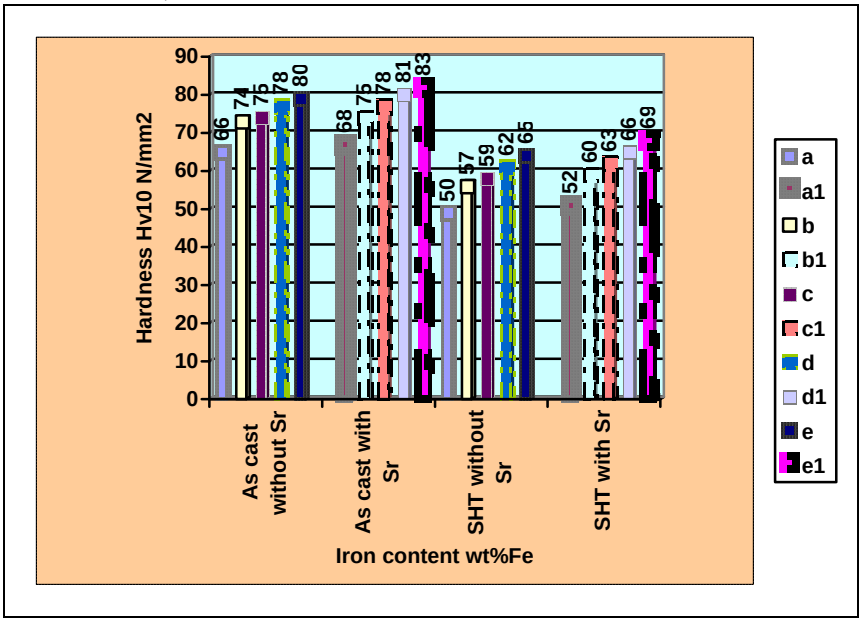


Fig.6 effect of iron content on the hardness value of as cast and SHT alloys

Also, it is worthily to notice that the hardness of as-cast alloys was shifted to lower values after SHT at all Fe%. This can be explained due to coarsening of Si particles during T6- treatment. Moustafa et al. (2003) [Moustafa, Samuel, Doty, 2003] in their work on effect of SHT and iron additive on the microstructure of eutectic Al–Si alloys found that after T6-treatment the average Si particle size was increased and its density (particles/mm^2) was decreased which in turn affect on hardness. At the same time, ß-Fe platelets need more than 24 hrs to partially dissolve and alter the hardness.

4. Conclusions

A study of the effect of iron content on the formation of ß - Al_5FeSi and porosity in Al–Si eutectic alloys was carried out. From the results obtained the following was concluded:

(1) Increasing Fe-content from 0.137 to 2.58 wt. % has no effect on DAS which remains about 22µm in all conditions.

(2) Modification of eutectic Al–Si alloys by Sr results in the precipitation of long, thick ß -platelets more than unmodified alloys.
(3) Increasing Fe-content up to 2.58% leads to increase the variation in length to 25% and in width to 4.8%.
(4) Sr-modified alloys have 0.5 to 1% more surface porosity than unmodified alloys.
(5) ß -Fe platelets prevent the liquid metal to fill the spaces between the branched platelets.
(6) Addition of Fe leads to increase the hardness of as-cast and SHT alloys by a linear relationship.

5. References

Argo, D., Gruzleski, J.E., 1988. Porosity in modified aluminum alloy castings. AFS Trans. 96, 316–366.

ASM Committee on Aluminum and Aluminum Alloys, 1990. Properties and Selection: Non Ferrous and Special Purpose Materials, ninth ed. Metal Handbook. vol. 2, 160–168

Awano, Y., Shimizu, Y., 1990. Non-equilibrium crystallization of AlFeSi compound in melt-superheat Al–Si alloy castings. AFS Trans. 98, 889–895.

Crepeau, P.N., 1996. Effect of iron in Al–Si casting alloys: a critical review. AFS Trans. 95, 361–366.

Dinnis Cameron, M., Taylor John, A., Dahle, K., 2005a. As-cast morphology of iron-intermetallics in Al–Si foundry alloys. Scripta Mater. 53, 955–958.

Dinnis Cameron, M., Taylor John, A., Dahle, K., 2005b. Effect of Fe-content on fatigue crack initiation and propagation in a cast aluminum-silicon alloy (A356-T6). Mater. Sci. Eng. A386, 396–407.

Iglessis, J., Frantz, C., Gantois, M., 1977. Conditions de formation des phases de fer dans les alliages Al–Si de puret ´e commerciales.

M´emoires Scientifiques de la Revue de M´ etallurgie 73 (4), 237–242.

Mondolfo, L.F., 1990. Manganese in Aluminium Alloys. The Manganese Centre, Neuilly sur seine, France, pp. 1–35.

Moustafa, M.A., Samuel, F.H., Doty, H.W., 2003. Effect of solution heat treatment and additives on the hardness, tensile properties and fracture behaviour of Al–Si (A413.1) automotive alloys. J. Mater. Sci. 38, 4523–4534.

Narayanan, L.A., Samuel, F.H., Gruzleski, J.E., 1995. Dissolution of iron intermetallics in Al–Si alloys through non-equilibrium heat treatment. Metal. Mater. Trans. A 26A, 2161–2174.

Samuel, A.M., Samuel, F.H., Villeneuve, C., Doty, H.W., Valtiera, S., 2001. Effect of trace elements on _-Al_5FeSi characteristics, porosity and tensile properties of Al–Si–Cu (319) cast alloys. Int. Cast Metals Res. 14, 97–120.

Samuel, A.M., Samuel, F.H., Dotty, H.W., 1996. Observation on the formation of ß-Al_5FeSi phase in 319 type Al–Si alloys. J. Mater. Sci. 31, 5529–5539.

Shabsestari, S.G., 2004. The effect of iron and manganese on the formation of intermetallic compounds in aluminum-silicon alloys. Mater. Sci. Eng. A 383, 289–298.

Yaneva, S., Stoichev, N., Kamenova, Z., Budurov, S., 1984. Quaternary iron-containing phases in Al–Si cast alloys. Z.