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PROCESSES OF THE FORMATION OF THE Fe (Mn)-BEARING INTERMETALLIC PHASES IN THE Al-Fe-(Mn)-Si ALLOYS

ABSTRACT

In the present work the intermetallic phases in an Al corner of the Al-Fe- (Mn)-Si diagram were analysed in the area of the co-existence of the ternary α-AlFe (Mn) Si and quaternary α-Al(FeMn) Si compounds. The microstructure of the alloys, chemical composition of the phase constituents and thermal effects during continuous slow cooling were examined using microscopic observation, X-ray microanalysis and differential scanning calorimeter. The phase components present in the alloy AlSi2.5(Fe+Mn)12 (where Mn = 0.03, 0.25, 1.8, 3.0% mass.) were identified on the base of their chemical composition. The successive progress of the peritectic transformation $L+Al_3Fe$ (Mn, Si)$\rightarrow\alpha$-Al. + α-Al (FeMn) Si was shown as the Mn content in an alloy increased until the primary precipitation process of the quaternary α-Al (FeMn) Si phase started. The chemical composition of the phase components was estimated during the progress of the phase transformation. The content of Si in the primary Al_3Fe phase was evaluated to be in the range 0.9-2.0% mass. Simultaneously, the Mn presence (0.5-8.5% mass. increasing with an increase in the Mn content in the alloy) was stated in this primary phase. A limit of the Mn concentration in the alloy promoting the process of the primary precipitation of the α-Al (FeMn) Si phase was established as 2.5% mass. This Mn effect was enhanced by 1.3% Cr addition into quaternary alloy. The temperature and the sequence of the phase transformations during slow cooling (5K/min) of the examined alloys were identified. The temperature range of the primary precipitation of the binary, ternary and quaternary intermetallic phases, influenced by the Mn content in the alloy, was estimated.

Key words: *alumnium alloys, intermetallic phase, transition metals, microanalysis, metallographic examination, calorimetric analysis.*

INTRODUCTION

The ternary Al-Fe-Si and Al-Mn-Si systems are a base of the most of the commercial aluminium alloys. The intermetallic phase precipitates are always present in as a result of a low solubility of the alloy components in the α-Al solid solution and because of the non-equilibrium crystallisation of the materials obtained in the industrial conditions. The variety of the crystal structure and morphology of the intermetallic phases particles formed in the material destined for manufacturing is considered a serious problem in the process of the stabilisation and control of its properties. Very often in the commercial alloys the multicomponent precipitates are formed. They are secondary solid solutions of the elements in a crystal lattice of a basic phase. In the quaternary Al-Fe-Mn-Si system the series of binary, ternary and quaternary intermetallic compounds appear [1-9]:

A. **monoclinic**: C2/m $Al_3Fe\rightarrow Al_3Fe_{0.97}Mn_{0.03}$
B. **orthorhombic** : Ccmm $Al_6Fe\rightarrow Al_6Fe_{0.5}Mn_{0.5}\rightarrow Al_6Mn$
C. **hexagonal:** $P6_3/mmc$ α_H-AlFeSi $\rightarrow\alpha_H$ –AlFe(Mn)Si ------ $P6_3/mmc$ β-AlMnSi
(C1) ↑(?)
D. **cubic:** α_c-AlFeSi $\rightarrow$Im3 α_c -AlFeMnSi ------ Pm3 α_c -AlMnSi

The range of the co-existence of the ternary solid solution on a base of the binary compounds (arrays A, B) was reported in many works [1, 2, 5, 9 -11, 17, 19-22] but the hypothesis of possible isomorphism of the ternary Si–bearing compounds has not been proved yet. The mutual miscibility (C1) between α_c-AlFeMnSi (array D) and α_H-AlFeSi (array C) evoked also some contradictory opinions. The ternary phases situated in the arrays of the solid solutions (C, D), according to Munson [5] do not form an unbroken series. The cubic phase $Al_{20}Fe_5Si_2$ reported to be the metastable constituent in the ternary alloys Al-Fe-Si has not been yet confirmed by metallographic examinations. What is more, the crystallographic data given by Cooper [4] concerned the monocrystal of the chemical composition $Al_{19}Fe_4MnSi_2$ isomorphic in ternary Al-Fe-Si reported by Phragmen [9]. The hypothesis formulated previously by Munson [5] that stability of this phase was possible only under condition of a presence in the alloy other transition metals as Mn or Cr [1,3,5] was then proved by Sun and Mondolfo [10]. Both phases: hexagonal ternary α_H-AlFeSi and cubic quaternary α_c-AlFe (Mn) Si can precipitate from a liquid alloy through the following primary processes [3, 5,11-15]:

$$\mathbf{Al_3Fe+L \rightarrow \alpha_H\text{-}AlFe(Mn)Si}$$
$$\mathbf{Al_6(FeMn)+L \rightarrow \alpha_C\text{-}AlFe(Mn)Si}$$
$$\mathbf{L \rightarrow \alpha_{C/H}\text{-}AlFe(Mn)Si}$$

The concentration limits for the fields of the phase α-AlFe (Mn)Si stability on the equilibrium diagrams have not been yet established correctly as well as the liquidus temperature for the quaternary alloy. Some problems due to the transition metals concentration in the intermetallic compounds, such as:

- morphology of the precipitates of determined chemical composition,
- temperature and mechanism of these intermetallic compounds formation,
- range of the transition metals concentration in alloy and in intermetallic compound determining its stability limit

are of a great importance from a point of view of the material property modification.

Some of these issues mentioned above are analysed in the present work. First of all the processes of the formation in Al-rich Al-Fe-Mn-Si alloys (2.5-3.5%Si, 12.5%Fe+Mn+Cr) of the quaternary compound α-Al (Fe Mn) Si were examined by the differential scanning calorimeter and by metallographic methods.

EXPERIMENTAL

Chemical composition of the examined alloys is given in Table 1.

Table 1. *Chemical composition of the examined alloys (% mass. Al-bal.)*

Alloy designation	Fe/Mn(+Cr)	Content of alloy components		
		Si	Mn	Fe
0-H	–	2.9	0.01	12.2
1-H	–	3.2	0.03	12.3
2-H	50	2.7	0.25	12.5
3-R	6	3.5	1.8	10.9
4-R	3	3.0	3.0	9.1
5-R	4	2.9	1.2+1.3 Cr	9,9

The examined alloys were produced from pure components: Al (99.99%) and Si, Fe, Mn of the spectral purity. They were melted in the induction furnace and poured into graphite moulds. In this manner the cooling rate was low, and the microstructure of the material could approach the near-equilibrium phase composition.

Experimental details

The microstructure of the alloys was examined on the metallographic microsections polished (end step – diamond paste -0.25μm) and then chemically etched with a reagent: 1 ml HF + 100 ml H_2O dist. The microscopic observations were carried out by means of the metallographic light microscope Neophot 32. Chemical composition of the phase components present in the alloys was estimated by means of X-ray microanalyser EDS, LINK ISIS 300, combined with the scanning electron microscope STEREOSCAN 420, by the standardless method. The thermal effects and temperature of the phase transformations occurring during slow heating and cooling of the alloy (5K/min) were registered and analysed by means of the differential scanning calorimeter Netsch DSC 404C.

RESULTS OF EXAMINATIONS

Microstructure observations

Results of the microscope observations are shown in Figs. 1a-f. One can see that the phase components of the final microstructure of the alloys were formed by peritectic transformation of the primary intermetallic phase and – by eutectic transformation in the residual liquid. In the presented figures the evolution of the alloy microstructure with an increase in the Mn content was shown. In the ternary alloy Al-Fe-Si the thin layer of the secondary phase covered the massive plates of the primary precipitates (Fig.1a). Locally the external interface (intermetallic phase /α-Al) of this layer was developing until forming profile of the Chinese script (Fig.1b). As the Mn was added to the alloy the external interface (intermetallic phase /α-Al) profile developed more and more and subsequently morphology of the secondary intermetallic phase was altered. The precipitates in form of well-shaped Chinese script, present in the alloy of 0.25% Mn (Fig.1c), became transformed to the polyhedra, initially closely adherent to the primary precipitates (Fig.1d). In the alloys of the maximum Mn content (3% mass.) and of 1.3% Cr addition, the separated compact polyhedra of the quaternary (or quinary) α-Al (FeMn+Cr) Si phase, either precipitated directly from liquid or resulted from the complete transformation of the primary precipitates could be observed (Figs. 1e,f). In the residual liquid regions (RLR) the precipitates arose from the segregated components of the alloy in form of the the needles and Chinese script of the intermetallic phases α -Al (FeMn) Si and the precipitates of the pure silicon, not built in the intermetallic phase particles. The volume fraction of the phase constituents of the eutectic in residual liquid regions was lower in the alloys of higher Mn (or Mn+Cr) content.

One can see that iron was concentrated mainly in the primary, pseudo- binary, low-Si phase whilst the manganese and chromium – mainly in the secondary, high-Si, quaternary Al. (FeMn + Cr) Si phase either in a shape of a peritectic layer or as separate compact particles (Figs.1b, f). The high–temperature phase precipitates, both primary and secondary (peritectic) consumed all the chromium content, as the presence of this element was not stated in the residual liquid regions.

The concentration of Mn or (Mn+Cr) in the intermetallic compounds was estimated in the range:

- primary Al_3Fe (Mn) particles –0.5-8.4% mass.Mn, 3.9% mass. (Mn+Cr)
- Chinese script-0.7% mass. Mn (peritectic phase) and 0.2-2.7% mass Mn (RLR)
- adherent, peritectic polyhedra- 10.5% mass. Mn
- separate polyhedra – 15.3% mass. Mn or 6.5% mass. (Mn+Cr)

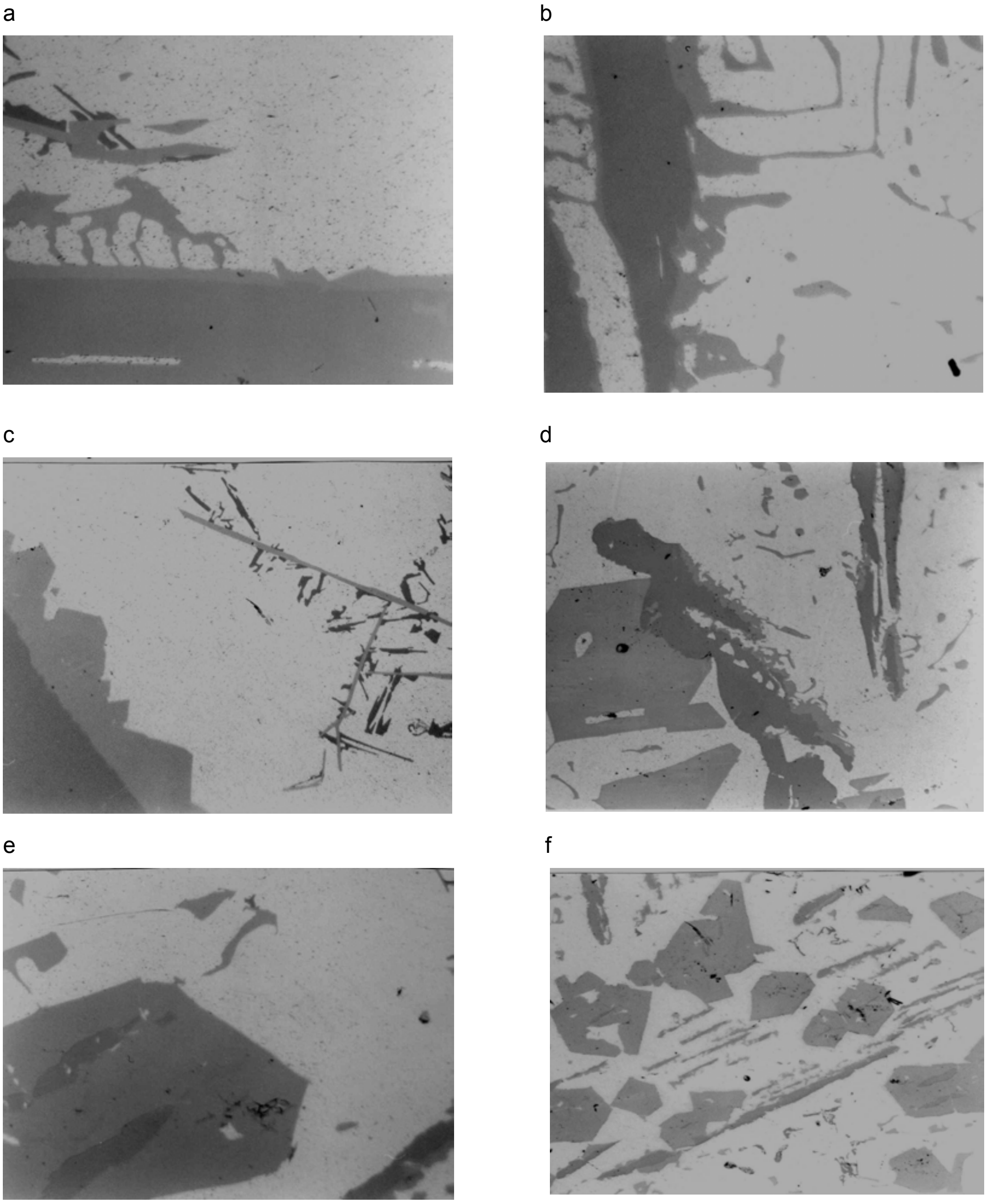

Fig.1. Microstructure of the examined alloys, LM, metallographic microsections, etched with 1%HF; a/ alloy 0-H, mag. 800x, b/ alloy 1-H, mag. 500x, c/ alloy 2-R, mag. 400x, d/ alloy 3-R, mag. 100x, e/ alloy 4-R, mag. 500x, f/ alloy 5-R, mag. 500x

***Table. 2.** Chemical composition of the intermetallic phases precipitated in the examined alloys*

Alloy designation	Phase	Chemical composition, %mass., Al-bal.					
		Si		Mn		Fe	
		%mass.	%at	%mass.	%at	%mass.	%at
0-H	Primary	0.9	1.0	-	-	41.8	25.6
	Peritectic	6.4	7.5	-	-	30.5	18.7
	Chinese script in RLR	7.9	8.9	-	-	26.2	14.4
1-H	Primary	1.3	1.6	-	-	40.2	24.5
	Peritectic	6.5	7.5	-	-	30.1	18.6
	Chinese script in RLR	8.1	9.0	0.5	0.2	25.6	14.3
2-H/R	Primary	0.9	1,0	0,5	0,3	41,4	25,5
	Peritectic	5,5	6,0	0,7	0,4	23,6	13,1
	Chinese script in RLR	6,4	7,4	2,1	1,2	29,4	17,0
3-R	Primary	2,0	2,4	3,9	2,4	36,7	22,4
	Peritectic	7,2	8,3	10,5	6,2	21,5	12,5
	Chinese script in RLR	6,1	7,0	3,1	1,8	29,3	17,1
4-R	Primary	1,2	1,5	8,4	5,2	32,6	20,0
	Peritectic/primary	5,1	5,9	15,3	9,0	17,1	10,0
	Chinese script in RLR	6,1	6,9	4,8	2,7	23,1	13,1
5-R	Primary	1,7	2,0	2,5+1,4	1,5+0,9	35,3	21,4
	Peritectic/primary	5,4	6,2	3,6+2,9	2,1+1,8	24,9	14,4
	Chinese script in RLR						

Results of the calorimetric analysis

During the slow heating and cooling of the examined alloys (5K/min) the sequence of the thermal effects connected with the phase transformations was registered (Fig. 2a, b).

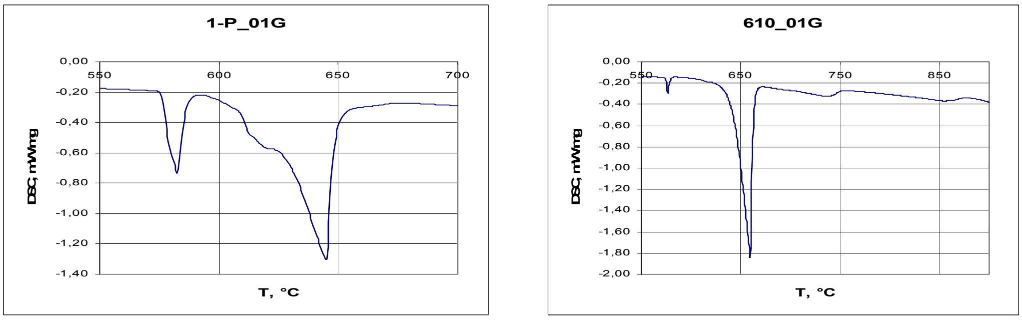

Fig. 2. Thermograms registered during heating of the examined alloys (5K/min); a/ alloy 1-H, b/ alloy 4-R

Thermogram`s shape, visible in Fig.2 was influenced by Mn (or Mn+Cr) content in the examined alloy. The high temperature thermal effects appeared when the Mn content in the alloy achieved 1.8% mass. Simultaneously as the Mn concentration in the alloy increased the thermal effects of the eutectic transformation in the residual liquid regions decreased and almost disappeared in alloy containing Mn and Cr.

The temperature of processes occurring during the slow heating of the alloy (of the microstructure state visible in Figs.1 a-f) is given in Table 3. One can state that temperature of the revealed phase transformation was slightly influenced by chemical composition of the alloy. The temperature of the eutectic transformation in the residual liquid regions and of the α-Al melting was rather stable, registered changes were in range of 6K and 4K respectively. The temperature of precipitation, transformation and melting of the massive primary precipitates due to the addition of Mn and Cr into alloy was much more differentiated (Tabl.3).

Table 3. *Temperature of the registered thermal effects in the examined alloys during heating 5K/min from room temperature until alloys melting*

Alloy design.	**Temperature, °C**					
	T_1	T_2	T_3	T_4	T_5	T_6
0-H, 1-H	582	615	620	644		
2-H	579	618		647		
3-R	578	618		647		918
4-R	581	619		634	747	882
5-R	576	617		659	737	886

Discussion of the results

The obtained results allowed estimating the influence of the additions of transition metals (Mn and Cr) into ternary Al-Fe-Si alloy on:

- morphology and chemical composition of the binary, ternary and quaternary intermetallic compounds,
- mechanism and temperature of the processes of the formation the phase constituents.

As it was mentioned previously ternary or quaternary intermetallic phases formed in the sequence of two-phase transformations: peritectic (at high temperature), and eutectic (at low temperature).

The primary phase Al_3Fe (MnSi) precipitated in form of the elongated plates [1-3,5]:

- directly from liquid phase: $L \rightarrow Al_3Fe$ (MnSi)
- or during an eutectic transformation: $L \rightarrow Al_3Fe$ (MnSi) +Al.

On the L/Al_3Fe (MnSi) interface, the peritectic transformation started and its products: ternary or quaternary α-Al(FeMn)Si compounds isolated primary phase from the liquid alloy.

In the ternary Al-Fe-Si alloys (0-H and 1-H) the layer of the peritectic phase developed at first in form the Chinese script. As the Mn (and Cr) content increased, the precipitates morphology of the new phase covering the plates of the primary low-Si phase became transformed. In the quaternary alloys Al-Fe-Mn-Si, an increase in the Mn content caused an appearance of the polyhedral precipitates of the high-Si quaternary α-Al(FeMn)Si phase. The observed changes in the intermetallic phase morphology due to the Mn content in teh alloy were influenced by the interaction of two factors:

- kinetic factor: Mn atoms accelerated the peritectic transformation rate so in the real cooling period it could be finished
- thermodynamic factor: Mn concentration was sufficient to change the local equilibrium conditions and primary precipitation of the quaternary α-AlFeMnSi phase started.

An increase in the width of the peritectic transformation product layer, revealed by microscopic observations, was proof of an effect of the kinetic factor.

An effect of the thermodynamic factor in the high-Mn alloys can be supported on a base of the presence of the high – temperature processes, registered during their heating and cooling course as well. They can be attributed to the primary precipitation processes of the quaternary α-Al(FeMn)Si phase [12-14].

The third effect can be considered within the framework of the crystallisation model of the process taking place on the α-Al(FeMn)Si /L interface. The attachment of the atoms to the crystal lattice of Chinese script of the α-Al (FeMn)Si phase in the low-Mn alloys occurred by the normal growth processes on the diffusive rough solidification front. In this manner the non-faceted profile of a new phase developed [16]. The faceted polyhedra of this phase, appearing in the high-Mn alloys, were the products of the lateral growth processes on the flat solidification front [1,16].

An analysis of the chemical composition of the intermetallic phases formed during slow cooling in three different zones of the alloy microstructure showed that the field of the component distribution, characteristic for the ternary Al-Fe-Si alloy, was altered in the presence of the Mn and Cr atoms (Figs. 3a-d).

In the ternary and low-Mn alloys atoms Fe were mainly accumulated in the primary precipitates of the binary Al_3Fe phase, whilst silicon and manganese (in alloy 2-H) segregated into residual liquid. This was the reason of the appearance in these regions of the Chinese script of the quaternary α_c –AlFeMn phase (Figs.3a, b).

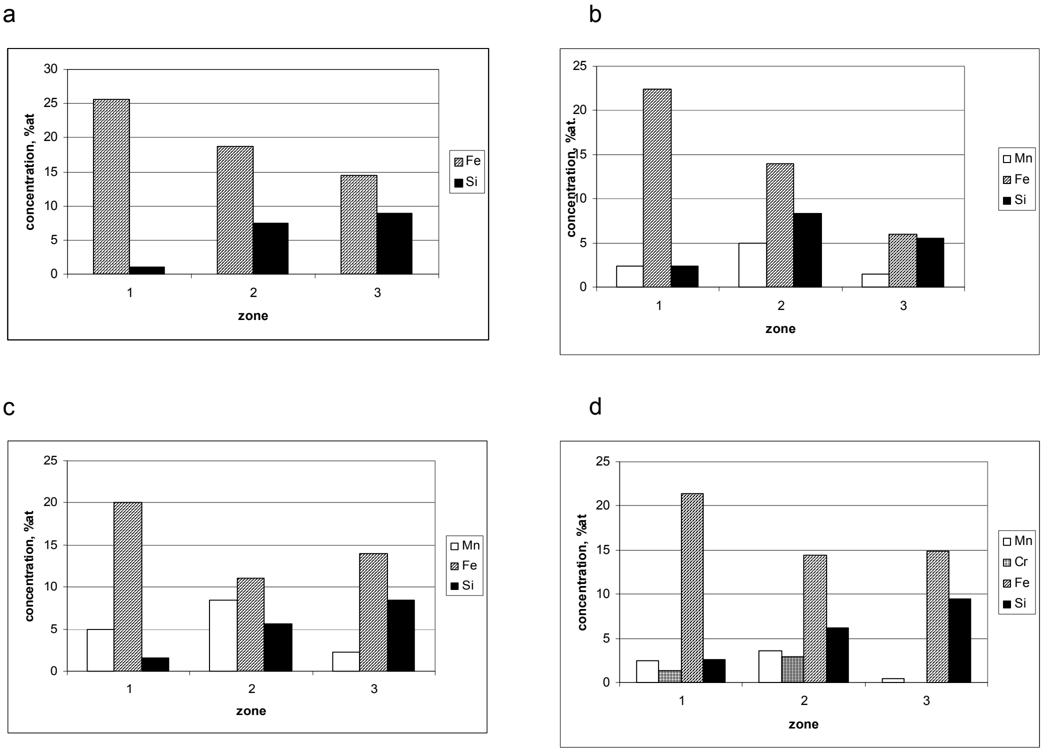

Fig. 3. Chemical composition of the phase constituents in three zones of the microstructure of the examined alloys: 1-primary low-Si phase, 2-peritectic or primary high-Si phase, 3-residual liquid region; a/ alloy 0-H, b/ alloy 2-R, c/ alloy 4-R, d/ alloy 5-R

Successive additions of the Mn content in the examined alloys resulted in some shifts in its preferential segregation area (Fig.3c). Finally the most part of atoms of the transition metals was collected in the peritectic phase precipitates, so the volume fraction of the intermetallic phases in the interdendritic regions successively decreased. The Mn content in the primary Al_3Fe estimated in the present work (5.2% at.) has gone beyond the concentration limit of 3% at, reported in the references [1-2]. Nevertheless the summary transition metals content was still of about 25% at. and general stoichiometric formula of the primary phase identified in all the examined alloys remained not modified.

An addition of Cr into alloy resulted in an further acceleration of the peritectic transformation. As it was stated previously, a chromium addition strongly promoted the cubic α_c-Al(FeMnCr)Si phase formation [5, 17]. The volume fraction of both primary and residual (not transformed) $Al_3Fe(MnSi)$ phase particles was decreasing until the primary particles pseudo-quinary phase α-AlFeMn(Cr)Si started to precipitate directly from the liquid alloy. Almost all the chromium atoms were collected in both primary $Al_3Fe(MnCrSi)$ and peritectic α-AlFeMn(Cr)Si phases. The presence of chromium in the interdendritic regions of the residual liquid was not stated (Fig. 3d).
The relationship between the transition metals contents in alloy and in the intermetallic compound is shown in Fig.4. The Fe/Mn value in the microstructure constituents did not reflect the one in the alloy in average. Although the iron, manganese and chromium belong to the transition metals group they reveal the various preferences to be accumulated in the phase constituents, leading to the changes in mechanism of formation and morphology of microstructure as well.
Silicon was present mainly in the peritectic phase and in the residual liquid. Some small amounts of Si in the primary Al_3Fe phase were stated in the alloys with the Mn. The Si concentration in this phase increased when the Mn and Cr atoms were built in the lattice of phase Al_3Fe. However, the maximum concentration of Si in the primary $Al_3Fe(MnSi)$ phase was lower than stated in other works, e.g.[18], for ternary Al-Fe-Si alloys.

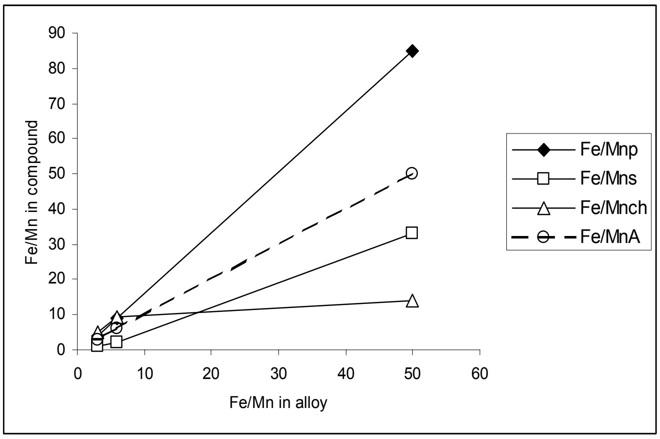

Fig. 4. Value Fe/Mn in the internmetallic compounds vs value Fe/Mn in alloy (Fe/Mnp-in plates of binary $Al_3Fe(MnSi)$phase, Fe/Mnp-in quaternary α–AlFeMnSi phase, Fe/Mnch-in chinese script in RLR, Fe/MnA-in alloy in average)

The microscopic identification of the morphology of the alloy microstructure constituents and their chemical composition were the base for determining the apparent stoichiometry (Table 4).

***Table 4.** Apparent stoichiometry of the intermetallic phases present in the examined alloys*

Alloy design.	Primary phase plates	Secondary phase layer/polyhedron	Phases in RLR	
			Ch.script	needles
0-H, 1-H	Al_3Fe	$Al_{20}Fe_5Si_2$	$Al_{15}Fe_3Si_2$	Al_5FeSi
2-H	$Al_3(FeMn)$	$Al_{11,5}(Fe\ Mn)_3Si$	$Al_{20}(FeMn)_5Si_2$	Al_5FeSi
3-R	$Al_3(FeMn)$	$Al_{11}(Fe\ Mn)_3Si_2$ $Al_{12}(Fe\ Mn)_3Si$	$Al_7(FeMn)_2Si$	Al_5FeSi
4-R	$Al_3(FeMn)$	$Al_{14}(Fe\ Mn)_{3.6}Si$ $Al_{12}(Fe\ Mn)_{3.2}Si$	$Al_7(FeMn)_2Si$	Al_5FeSi

The morphology and chemical composition (apparent stoichiometry) allowed attributing to the observed precipitates two known equilibrium phases: Al_3Fe and β-Al_5FeSi. The formulas obtained for quaternary α-AlFeMnSi phase were changing in the quite wide range (Al/M: 3.5-5, M/Si: 1.5-3.6), however the main proportions of the components were comparable with the literature data (Al/M:3-5, M/Si: 1.5-3.5) [1-5, 14, and 17, 19-22].

The results of the microscopic observations and DTA analysis allowed identifying some phase processes occurring during the final microstructure formation (Table 5). The course of the low-temperature transformation (T_1), presented in Table 5, was in a good agreement with both literature and experimental data [1-3]. The high- temperature process (T_5) was identified as a precipitation of the primary α–AlFeMn(Cr)Si phase, reported in the high- Si alloys by Zakharov [13-15]. The literature data [1-3, 18] concerning temperature of the peritectic transformation in the ternary Al-Fe-Si alloy ($L+Al_3Fe \rightarrow \alpha\text{-Al} + \alpha_H\text{–AlFeSi}$), were proved by comparable results obtained in the experiments carried out in this work. In the quaternary alloys some doubts, concerning both structure and composition of the primary phase ($Al_3Fe(Mn)$ taking part in this transformation ($L+Al_{3/6}Fe(Mn)? \rightarrow \alpha\text{-Al} + \alpha_c\text{–AlFeMnSi}$), remained to interpret. However, the sequence of the phase transformations identified in the examined alloy might lead to the microstructure state observed at room temperature after slow cooling.

Table 5. *Microstructure formation processes possibly occurring during slow cooling (5K/min) of the examined alloys*

Alloy design.	thermal effect T_5	thermal effect T_4	thermal effect T_3	thermal effect T_2	thermal effect T_1	final microstructure state
0H, 1-H	-	$L \rightarrow \alpha\text{–Al} + Al_3Fe$	$L+Al_3Fe \rightarrow \alpha\text{–Al} + \alpha_H\text{–AlFeSi}$	$L+\alpha_H\text{–AlFeSi} \rightarrow \beta\text{–AlFeSi} + \alpha\text{–Al}$	$L \rightarrow Si+\beta\text{–AlFeSi} + \alpha\text{–Al}$	α–Al, Al_3Fe, α_H–AlFeSi, Si, β–AlFeSi
2-H	-	$L \rightarrow \alpha\text{–Al} + Al_3Fe$	$L+Al_3Fe(Mn) \rightarrow \alpha\text{–Al} + \alpha_H\text{–AlFe(Mn)Si}$		$L \rightarrow Si+\beta\text{–AlFeSi} + \alpha\text{–Al}$	α–Al, Al_3Fe, α_H–AlFe(Mn)Si, Si, β–AlFeSi
3-R	-	$L \rightarrow \alpha\text{–Al} + Al_3Fe(Mn)$	$L+Al_3Fe(Mn) \rightarrow \alpha\text{–Al} + \alpha_c\text{–AlFeMnSi}$		$L \rightarrow Si+\beta\text{–AlFeSi} + \alpha\text{–Al}$	α–Al, Al_3Fe, α_c–AlFeMnSi, Si, Al_5FeSi
4-R	$L \rightarrow \alpha_c Al(FeMnCr)Si$	$L \rightarrow \alpha\text{–Al} + Al_3Fe(Mn)$	$L+Al_3Fe(Mn) \rightarrow \alpha\text{–Al} + \alpha_c\text{–AlFeMnSi}$		$L \rightarrow Si+\beta\text{–AlFeSi} + \alpha\text{–Al}$	α–Al, Al_3Fe, α_c–AlFeMnSi, Si, Al_5FeSi
5-R	$L \rightarrow \alpha_c Al(FeMnCr)Si$	$L \rightarrow \alpha\text{–Al} + Al_3Fe(MnCr)$	$L+Al_3Fe(MnCr) \rightarrow \alpha\text{–Al} + \alpha_c\text{–AlFeMnCrSi}$		$L \rightarrow Si+\beta\text{–AlFeSi} + \alpha\text{–Al}$	α–Al, Al_3Fe, α_c–AlFeMnCrSi, Si, Al_5FeSi

CONCLUSIONS

1. The processes of the formation of the quaternary α-Al (FeMn) Si phase were strongly influenced by either Mn or Cr atoms present in the liquid Al-rich alloy. The visible effect of the increasing Mn (or Mn+Cr) content on the peritectic transformation rate ($Al_3Fe + L \rightarrow \alpha\text{–Al (FeMn) Si}$) was stated. The microscopic observations as well as the results of thermal analysis have proved the assumption of the primary precipitation of this phase in high –Mn alloys.
2. The mechanism of formation of the quaternary α-Al (FeMn) Si phase varied in presence of other transition metals atoms in the ternary Al-Fe-Si alloy. The faceted polyhedra of α-Al(FeMnCr)Si phase, arose at the beginning in form of the products of the peritectic

transformation and finally - precipitated directly from the liquid alloy. These processes were promoted by the transition metals addition both Mn and Cr.

3. An addition of 3% mass. Mn to the Al-Fe-Si alloy of 2.5% mass. Si caused the start of the primary precipitation of the α-Al(FeMn)Si phase.
4. Obtained results indicated the possibility to solve of about 5% at Mn in the Al_3Fe phase formed in Al-rich alloy in the near–equilibrium state.

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