Precipitation Strengthening in Al-Mg-Si-Cu Alloys
I. Kalamvokas

Abstract

Quaternary alloys Al-Mg-Si-Cu application thrive in the automotive industry due to their combination of strength and lightness. Combining characteristics of both 6xxx and 2xxx series, their properties are mainly a result of a new phase that is formed in the alloy. This phase named Q, found only at equilibrium, is reported with various compositions. This phase is reported to have a precursor named Q’, a lath-shaped metastable phase that forms after peak age. At peak age, it seems that beside the β” phase carried by the ternary alloy family, another phase is formed but it is not the Q’. Research results from literature were analyzed to find the source of strength of Al-Mg-Si-Cu alloys. This is paper is an attempt to summarize up to date results about Q phase, other metastable phases found in these alloys and the possible links with the Q equilibrium phase. In this attempt, the precipitation sequence of these metastable phases is registered and conclusions about the main strengthening phases are made.
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Introduction

With the evolution of the automotive industry, the need for lighter, tougher materials is now growing stronger than ever. In search of those attributes in alloys, researches turned to aluminum alloys, known for their lightness. The lack of information though, made them wonder until the early 90s if the aluminum car was just a fantasy. Extensive research of the 6xxx series coupled with technological innovations, led to the discovery of new alloys like the quaternary Al-Mg-Si-Cu, which is now used extensively in the automotive industry. Combining the characteristic of the 6xxx and 2xxx series these alloys have distinctive properties. The main reason for these properties is the occurrence of a new phase combined with the strengthening phases of the 6xxx alloys.

At this point we should state that due to the number of investigators and the vast numbers of different phases found with the slightest changes in composition, a lot of names exist for possibly the exact same phases. The phases that concern us, though, are quite recognizable in the scientific world and their names are settled to be unanimous.

Since this new alloy is based on the 6xxx series, we first need to study what is the reason for their strength. Al-Mg-Si alloys, have been known to be influenced by the precursor phases ($\beta''$, $\beta'$) to the equilibrium phase $\text{Mg}_2\text{Si}$ ($\beta$). These phases are also found in the quaternary alloy we are interested in, providing it with the necessary strength needed for our applications. The crystal structure and morphology of this phase will be reviewed in the next segments.
The new phase that concerns us the most though is the Q phase. This phase is found only in the quaternary alloys Al-Mg-Si-Cu, and is reported with more than one stoichiometries. As with the β phase, important role is played mostly by the precursors to this phase, Q’. This is why aside from these phases, the aging process will also be reviewed extensively.

TEM analysis was crucial to the study of these phases. Presented in this paper, you will find pictures from such microscopes for the sake of explaining structural characteristics of the phases.

Commercial Al–Mg–Si–Cu alloys grouped in different phase fields. The alloys are arranged in each column in the chronological order of their development. Registration dates for the alloys with The Aluminum Association are shown within parentheses.
Morphology and crystal structure of the Q phase

The structure of the equilibrium phase Q has been thoroughly studied. Although investigators disagree about the exact composition of the Q phase, some good descriptions of its structures are given by (1). According to Arnberg L., who used single-crystal X-ray, the structure of the Q phase belongs to the hexagonal system with 21 atoms per unit cell and lattice parameters $a=10.35–10.40\text{Å}$ and $c=4.02–4.05\text{Å}$. In 2001, Wolverton (2) proposed his model of the Q phase $\text{Al}_3\text{Cu}_2\text{Mg}_9\text{Si}_7$ (14.3Al–9.5Cu–42.9Mg–33.3Si at.%).

The nanometer scale dimensions of these precipitates make quantitative analytical electron microscopy analyses challenging and susceptible to considerable uncertainties. Experiments done by (3) gave us a new estimation of the composition. Using LEAP tomography, DSC, X-ray diffraction, TEM and first-principles calculations on a W319 alloy they found that the average chemical composition of the Q-phase precipitates was $44.23 \pm 5.2\text{Al}–22.56 \pm 1.74\text{Cu}–14.89\pm 2.87\text{Mg}–17.12 \pm 1.14\text{Si}–1.08 \pm 0.44\text{Zn}$ at.%. Compared to reports by Arnberg (1), this composition was considerably richer in Al and Cu and leaner in Mg and Si.

<table>
<thead>
<tr>
<th></th>
<th>Si</th>
<th>Cu</th>
<th>Fe</th>
<th>Mg</th>
<th>Mn</th>
<th>Zn</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt.%</td>
<td>7.48</td>
<td>3.53</td>
<td>0.35</td>
<td>0.31</td>
<td>0.26</td>
<td>0.13</td>
</tr>
<tr>
<td>at.%</td>
<td>7.38</td>
<td>1.54</td>
<td>0.17</td>
<td>0.35</td>
<td>0.13</td>
<td>0.05</td>
</tr>
</tbody>
</table>

For their experiments they had to create a visualization of the unit cell of the Q phase based on both (1) (2), work which is presented afterwards.

It is obvious that there is still debate to be made concerning the composition and structure of the Q phase. The investigators’ work though is not made easy by the fact that the Q phase appears to be very sensitive to alloy composition.
Conventional unit cell for the Q-phase, Al$_3$Cu$_2$Mg$_9$Si$_7$ by (3). Al atoms are blue, Cu is grey, Mg is orange and Si is red.

Precipitation Sequence

As with most aluminum alloys used in the market, in order to strengthen the structure, cold quenching is required followed by aging. This process allows precipitants to form within the solid solution, increasing its strength. Dispersoids in different sizes, shapes, compositions, and structures block the gliding of dislocation arrays. Like all transformations, for the nucleation to take place an energy barrier needs to be surpassed. This energy barrier is a function of the chemical driving force for the nucleation $\Delta G_v$, the interfacial energy of the interface boundary between the 2 phases $\gamma$, and the strain energy $\Delta G_\varepsilon$ as follows.

$$\Delta G^* = \frac{16\pi \gamma_{\alpha\beta}^3}{3(\Delta G_v + \Delta G_\varepsilon)^2}$$

It is obvious then, that the energy barrier $\Delta G^*$ has a strong dependency on the interfacial energy. Thus, during the aging process, the nuclei will pass through several stages in order to maintain that barrier low.

AA6111

A study for AA6111 by (4) gives us the opportunity to understand in great detail the nucleation and growth process of two dominant phases in the quaternary alloys. This particular
alloy is commonly used in the automotive industry of North America due to its combination of strength and formability. Researchers examined AA6111 alloy using TEM and found two types of dispersoids. The first, was needle-shaped with the needles aligned with the <100> in coherency with the Al matrix. These findings are consistent with β” phase. The β” phase is the stoichiometric compound Mg₅Si₆. These needle-shaped precipitants are found along the <100> direction inside the aluminum matrix and are fully coherent with the matrix. With a monoclinic structure, the lattice parameters are a=0.77 nm, b=0.67 nm, c=0.405 nm, and γ=75 deg. The β” phase is the dominant phase in most of these quaternary alloys even after peak age. It is also found in the ternary alloys Al-Mg-Si and is what gives them their strength attributes. In the quaternary alloys we study, it is not alone in this role.

The second type of dispersoids found by the investigators, had lath morphology with a rectangular cross section when viewed from <001> direction. This lath-shaped phase is Q’. The Q’ phase is characterized by lath morphology, with a hexagonal crystal structure. The lattice parameters are a=1.04 nm and c=0.405 nm and not only is it fully coherent with the Al matrix, but it also has identical lattice parameter to the Al. As with the Q phase, the composition of Q’ is still unidentified but it is close to Al₄Cu₂Mg₆Si₇.

The findings by Wang (4) support our theory about the nucleation process and the struggle for low activation energy barrier. Those two phases are precursors of the equilibrium phases β and Q respectively, and by maintaining the coherency with the lattice, they minimize the activation energy.

In general, the precipitation sequence for AA6111 is described by the following diagram

SSS \rightarrow clusters /GP zones \rightarrow β”+Q’ \rightarrow equilibrium Q + β

To the left, BF image along <001> axis of AA6111 aged for 7 hours at 180°C
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TEM imaging of the same alloy. On the left, diffraction pattern corresponding to needle-shaped precipitates. On the right diffraction pattern corresponding to lath-shaped precipitates. Both along the <001> axis

Ternary Balanced Alloys with high/low Cu

Ternary Balanced alloy system consists of Al-Mg-Si alloys with their composition matching the β phase (Mg2Si) stoichiometry. With no Cu in the system, the precipitates formed at peak age are of β” phase (needle-shaped with monoclinic structure). With over aging they are gradually replaced by β’ (rod-shaped), on their way to β phase at equilibrium. This is done in order to minimize the energy barrier needed to be surpassed for the nuclei to form. A more round precipitant is usually less coherent with the matrix, meaning they have greater interfacial energy. This is why β” is firstly formed, followed by the rod-shaped β’ with a greater γ energy, due to its semicoherent relations with the matrix, and so on.

SSS → clusters /GP zones → β” → β’+ β” → β

The addition of low quantity of Cu sees the birth of new phases in the precipitation process. In Al-1Mg-0.6Si-0.3Cu, a lath-shaped phase was spotted beside the rod-shaped β’ phase. (5) named this new phase B’, but from the information provided, it is clear that, this phase is no other than our well known Q’ phase. In Al-1Mg-0.65Si-0.25Cu another new phase is reported by (6), one with no structural data but, like Q’, lath-shaped. It is first observed at peak age, along with β”, and is later replaced by Q at equilibrium. This phase named L should not be confused with Q’. Although there are similarities between them, they are not the same phases. The L phase is found at peak age, while Q’ is formed with over aging. This L phase was also found in Al-0.87Mg-0.65Si-1Cu, a quaternary alloy with larger quantity of Cu. In addition, investigators found that with greater Cu quantities, the L phase numbers grew larger against the β” phase.

Excess Si with high/low Cu

Quaternary alloys with weight pct larger than what the stoichiometry of Mg2Si suggests, are named excess Si quaternary alloys. In the case of low Cu, something peculiar occurs. After peak age, a lath-shaped phase is observed coexisting with β’ precipitates. This phase shares key
characteristic of the Q’ phase, however, when equilibrium is reached, no Q phase is present. Only β and Si phases are found. For high Cu quantity alloys, β” phase found at peak age, was gradually replaced by Q’ phase. At equilibrium no β phase was found, but in its place Q and Si phases coexisted.

Ternary alloys with Excess Si

Although this alloy system contains no Cu, it would be wrong not to include it in our study. The reason why is, reports of a lath-shaped phase are made. Firstly, β” was observed at peak age together with a lath-shaped phase $\beta'_d$ formed mostly on dislocations. (6) On over aging this phase was replaced by an M phase, which has hexagonal crystal structure, lath morphology and similar lattice parameters as Q’. At equilibrium only β phase was observed. Other studies (7), suggest that more than one phases form at over aging, with at least one of them sharing similar characteristics with Q’ phase. One such phase named Type C, was characterized by lath morphology and was isostructural with the Q’ phase. Type C had been formed by a precursor called Type B, which was observed near peak age.

Crayon Findings

These findings made investigators even more interested in the origins of the Q phase. Extensive studies by Crayon (8), including TEM analysis of Al-Mg-Si-Cu alloys revealed new phases, precursors to the Q’ phase. The metastables phases they found were named QP at peak age, QC on over aging and Q’ which finally was replaced by equilibrium Q.

$$QP \rightarrow QC \rightarrow Q' \rightarrow Q$$

For a better understanding we cite pictures by (8) of the phases mentioned above.
Rod-shaped precipitates found in Al-Cu-Mg-Ag with reinforcing SiC particles

Al4Cu-1Mg-Ag alloy reinforced by Sic particles showing isolated QP precipitates
A1-4Cu-1Mg-Ag alloy reinforced by Sic particles showing a QC rod-shaped precipitate lying on $\theta'$ plates.
These results tie together all findings to date concerning Q phase, and are summarized in Table presented below, made by D.J. Chakrabarti (9)

### Table 3
Precipitation sequence and precipitate structure for different 6xxx alloy compositions

<table>
<thead>
<tr>
<th>Composition</th>
<th>Peak age</th>
<th>OA 1</th>
<th>OA 2</th>
<th>OA 3</th>
<th>Equilibrium</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ternary, excess-Si</td>
<td>$\beta'' + \beta''$</td>
<td>$\beta'' + \text{Type B}$</td>
<td>Type B + Type A</td>
<td>M</td>
<td>$\beta$</td>
<td>[12]</td>
</tr>
<tr>
<td>Ternary, excess-Si (low)</td>
<td>$\beta''$</td>
<td>$\beta'' + \text{Type B}$</td>
<td>Type B + Type A</td>
<td>Type A + Type C</td>
<td>$\beta + \text{(Si)}$</td>
<td>[16-18]</td>
</tr>
<tr>
<td>Ternary, excess-Si (high)</td>
<td>$\beta''$</td>
<td>$\beta'' + \text{Type B}$</td>
<td>Type B + Type A</td>
<td>Type A + Type C</td>
<td>$\beta + \text{(Si)}$</td>
<td>[16-18]</td>
</tr>
<tr>
<td>Ternary, balanced</td>
<td>$\beta''$</td>
<td>$\beta'' + \beta''$</td>
<td>$\beta''$</td>
<td>$\beta''$</td>
<td>$\beta$</td>
<td>[7-9]</td>
</tr>
<tr>
<td>Balanced + low Cu (1)</td>
<td>$\beta''$</td>
<td>$\beta'' + \beta''$</td>
<td>$\beta'' + \beta''$</td>
<td>$\beta'' + \beta''$</td>
<td>$\beta + \text{(Si)}$</td>
<td>[1]</td>
</tr>
<tr>
<td>Balanced + low Cu (2)</td>
<td>$\beta'' + L$</td>
<td>L + $\beta''$</td>
<td>L + $\beta''$</td>
<td>$\beta'' + L$</td>
<td>$\beta + \text{(Si)} + \text{(Q)}$</td>
<td>[12]</td>
</tr>
<tr>
<td>Balanced + high Cu</td>
<td>$\beta'' + L$</td>
<td>L + $\beta''$</td>
<td>L + $\beta''$</td>
<td>$\beta'' + L$</td>
<td>$\beta + \text{(Si)} + \text{(Q)}$</td>
<td>[12]</td>
</tr>
<tr>
<td>Excess-Si + low Cu</td>
<td>$\beta''$</td>
<td>$\beta'' + \text{Lath}$</td>
<td>Q + Q'</td>
<td>$\beta + \text{Q'} + \beta + \text{(Si)}$</td>
<td>[4]</td>
<td></td>
</tr>
<tr>
<td>Excess-Si + high Cu</td>
<td>$\beta''$</td>
<td>$\beta'' + \text{Lath}$</td>
<td>Lath</td>
<td>Q + Q'</td>
<td>$\beta + \text{Q'} + \beta + \text{(Si)}$</td>
<td>[4]</td>
</tr>
<tr>
<td>Al-Mg-Si-Cu</td>
<td>$\beta'' + \text{QP}$</td>
<td>QC + QP</td>
<td>QC + QP</td>
<td>QC + QP</td>
<td>$\beta + \text{Q'} + \beta + \text{(Si)}$</td>
<td>[14,15]</td>
</tr>
</tbody>
</table>

\begin{align*}
\beta'' & \text{ Monoclinic [12]} \\
& a = 0.65 \text{ nm}, b = 0.76 \text{ nm} \\
& c = 0.405 \text{ nm}, \gamma = 70^\circ \\
L & [12] \\
& (a = 0.8, c = 0.7) \text{ nm} \\
\beta'' & \text{ Hexagonal [12]} \\
& (a = 0.705, c = 0.405) \text{ nm} \\
\beta & \text{ Cubic [12]} \\
& a = 0.642 \text{ nm} \\
Q' & \text{ Hexagonal [2]} \\
& (a = 1.03, c = 0.405) \text{ nm} \\
M & \text{ Hexagonal [38]} \\
& (a = 1.03, c = 0.405) \text{ nm} \\
\beta' & \text{ Hexagonal [1]} \\
& (a = 1.03, c = 0.405) \text{ nm} \\
\beta & \text{ Hexagonal [16]} \\
& (a = 1.04, c = 0.401) \text{ nm}
\end{align*}

Table 1: Precipitation sequence and structure for different quaternary alloys composition
Strengthening Phases in Al-Mg-Si-Cu Alloys

As we mentioned earlier, in ternary alloys we see the formation of the $\beta$ phase and its precursors, through the aging process and all the way to the equilibrium state. The phase that provides those alloys with their strength is known to be that precursor phases of $\beta$, mainly $\beta''$. In quaternary alloys however, things are quite more complicated. In the previous section of our paper, we tried to present the different metastable phases found in the Al-Mg-Si-Cu systems and the different precipitation sequences on a variety of examples. This vast number of different phases inquires further analysis to let us decide with certainty which of those belong to the strengthening phases of the system. Looking back at alloys with excess Si, researchers have found that the Si tends to make $\beta''$ precipitants form in a finer manner. Respectively, in quaternary alloys the first assumptions were that the Cu added to the alloy had the same influence as excess Si, resulting in a finer dispersion of the $\beta''$ phase.

In search of the strengthening phases of Al-Mg-Si-Cu (9), experiments were conducted on a collection of alloys with different hardness ratings. Out of a handful, two groups of alloys were published, but are enough for us to extract safe conclusions. Presenting at first are the volume fraction of equilibrium phases calculated by the phase diagrams of each alloy. Following are the diagrams of hardness (HRB) in relation with aging time.
For this group of alloys, it looks like the $\theta$ phase plays a significant role in hardness. A phase that was not considered before, $\theta$ phase volume increased with the addition of Cu. As the volume of $\theta$ increased, so did the hardening. And for the last alloy, S793, which had less $\theta$ phase, the hardness decreased. The diagrams might be somewhat misleading in the essence that, the hardness is not linked with the phases found in equilibrium, but the phases at peak age. Those are precursors to the ones shown on the diagrams because the composition was taken from the phase diagrams of equilibrium states. Nevertheless, the volume of equilibrium phases reflect the volume of their precursors at peak age. In this case, the true strengthening phase is a precursor metastable phase of $\theta$. It should also be mentioned that Q levels were similar in every alloy of this group.

The second group diagrams are given below.
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**Equilibrium Phases at 177°C (vol. %)**

- Q
- CuAl2
- Mg2Si
- Si

**Alloy No.**

**Rockwell Hardness (RB)**

- S798
- S799
- S800
- S801
- S802

**Aging Time at 177°C (hrs)**
In this second group of alloys, the phase with volume fraction changes is mainly the Q face. As the Q phase amount increases, the volume fraction of β and θ decreases accordingly. With the increase of Q phase, we also see an increase hardness. Since the increase of Q is accommodated by decrease of the precursors of β and θ phases, well known for their strengthening capabilities, one could argue that a precursor state to Q is a strengthening phase as well. A hasty jump to conclusions would be to acknowledge this phase to be the Q’ phase but this cannot be true. If we go back to the Precipitation Sequence chapter of this paper, we will find that Q’ phase is not formed until the peak age has been surpassed. Meaning that the Q’ phase could not be a strengthening phase, as it does not exist when maximum hardness is achieved. Taking a closer look at Table 1, the only phase that forms at peak age, beside β”, is the L phase of Segalowicz et al (6). Because though of the vast number of phases reported, i.e. Type C, Type B (7) researchers could not come to a definitive conclusion just by relying to the literature. In one of the alloys (no. S802) TEM analysis was conducted, when at peak age. Images revealed the presence of β” precipitates together with another lath-shaped phase. Looking at the images below, we see 3 dispersoids, of which 2 are lath-shaped and the other one is of β” phase.

Since equilibrium state contains Q phase, these unknown precipitates must be precursors to the Q phase. This lath-shaped phase is most likely the L phase. Therefore, the increased strength of quaternary alloys is attributed to L phase. This means that although Q phase does not provide strength to the alloy, its precursors do. Thus, the presence of Q phases at equilibrium indicates the presence of the L shape at peak age, and as the Q phase volume fraction at equilibrium increases, so does the volume fraction of L. This results in the increase of strength at
peak age. This findings also support our initial thoughts, that Q’ could not be a strengthening phase of quaternary alloys.

The Importance of Aging Time

In our previous sections, we mentioned peak age. The aging process plays a significant role in defining the overall properties of the alloy mainly the hardness. A lot of factors come into play when discussing different aging processes, i.e. time, temperature, combination of natural aging and artificial aging. In study (4) of the precipitation strengthening of quaternary alloys, researchers used AA6111 and followed two different aging procedures. A group of samples where aged at 180°C directly after the solution treatment, and the second group was force aged after 2 weeks of natural aging. Aside from TEM imaging, the investigators calculated the equivalent radius and length of the precipitates. Presented first are the images from TEM analysis of the directly aged samples. Looking at the images, we can see that the number density of β” phase is much higher that the number density of Q’, increasing during the first hour of aging. But at the 7 hour mark i.e. peak age, the number density has decreased. This decrease of the number density marks the end of the growth phase and the onset of coarsening. For years, it was thought that nucleation, growth and coarsening were separate stages of the evolution of precipitates. But if we observe the evolution of Q’, we will find that its number density increases constantly. This means that even though β” begins its coarsening, the Q’ phase is still growing. This supports recent theories of growth and coarsening happening simultaneously.

Next, images from the naturally aged samples are presented. In these, we can observe that again β” number density is larger than the number density of Q’, but compared to the directly aged samples both numbers are much lower. This indicates that naturally aged samples are coarser than the directly aged samples. In contrast with the directly aged samples though, the naturally aged samples showed resistance to coarsening, with the coarsening of the β” phase taking place significantly later, from 7 to 24 hours of aging. These findings suggest that the nucleation rate of both precipitates is reduced when natural aging proceeds artificial aging. This is a result of the reduction of the driving force of the nucleation.
TEM imaging for samples directly aged at 180°C for 15 min (a), 30 min (b), 1 hour (c) and 7 hours (d)
TEM imaging for samples naturally aged for 2 weeks followed by aging at 180°C for 1 hour (a), 7 hours (b) and 24 hours (c)

The coarsening effect on β” phase is better understood when studying the following diagrams of frequency as a function of equivalent radius of the precipitants. Prior to peak we note that most of the precipitates in the ST (directly aged) samples have a radius of 1.25-1.5nm. At peak age, that coarsening has acted on the sample, almost no precipitants are of 1.25nm. They have significantly grown to an allocation of 1.5-2nm dispersoids. On the contrary, the precipitates found in the NA (naturally aged) samples, prior to peak are observed with radii of 1.5-2nm. Our estimations of the NA being coarser are confirmed. But at peak age, far less dispersoids have
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evolved in to 2nm or bigger precipitants. This fact also confirms that the NA samples coarsening is delayed with natural aging.

Precipitate size distributions for $\beta''$ after 1 h of aging at 180 °C

Precipitate size distributions for $\beta''$ after 7 h of aging at 180 °C
To summarize the importance of aging we study the same alloy but this time aged at 250°C for 0.5 hours to its peak. Even though the temperature is much higher the results Table 2 are quite disappointing. The yield stress of the sample aged at 250°C is much almost 100MPa lower compared to the sampled aged at 180°C for 7 hours. This punctuates the importance of aging time concerning the applications that the alloy will be used.

<table>
<thead>
<tr>
<th>Aging Condition</th>
<th>Yield Stress/MPa</th>
<th>Equivalent Radius/nm</th>
<th>Length/nm</th>
<th>Number Density/10^3 μm^-3</th>
<th>Volume Fraction</th>
</tr>
</thead>
<tbody>
<tr>
<td>7 h at 180°C</td>
<td>340</td>
<td>1.8 1.8</td>
<td>7.2 30.8</td>
<td>84.4 4.3</td>
<td>0.006 0.001</td>
</tr>
<tr>
<td>0.5 h at 250°C</td>
<td>255</td>
<td>4.4 4.7</td>
<td>21.1 89.3</td>
<td>5.6 0.8</td>
<td>0.007 0.005</td>
</tr>
</tbody>
</table>

Table 2: Comparison of Samples directly aged at 180°C and 250°C

Summary

- The quaternary alloy system inherits a lot of its properties from its 6xxx and 2xxx progenitors
- The characteristic phase of this alloy is the Q phase. The Q phase composition changes with alloy composition. Thought at first to form only in quaternary alloys, metastable phases similar to the precursor Q' are found in ternary alloys with excess Si. Prior to equilibrium, lath-shaped metastable precipitants are formed M or type C but instead of Q, they are replaced with β and/or Si.
- Although Q phase has no strengthening abilities, its precursor do. Lath-shaped precipitants are reported at peak age, named L. The greater the volume numbers of L the greater the strength of the alloy. With the β”, they are the two strengthening phases of the Al-Mg-Si-Cu alloy series.
- The size and dispersion of the precipitants play a significant role in the overall hardness. This is affected by the aging process. The correct combination of temperature and time needs to be selected to have optimum performance
References


