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The Influence of Ageing Time and Temperature on the Structure and Properties of Heat Treated A201.0 Aluminum Alloy

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Abstract*: The effect of ageing time and temperature following quenching from solution treatment at 540±2^{O}C on the micro structure and properties of A201.0 Al alloy was studied. Using thermodynamic modeling with JMatPro, the intermetallic phases that exist at the end of solification was determined. Intermetallic phases like $CuAl_2$, Al_2CuMg and Mg-Si phase contribute to the hardness of the alloy. Ageing was carried out upto a maximum duration of 10000 hours at a constant temperature of 170±2^{O}C. As ageing time increased, the hardness of the alloy increased due to the enhanced presence of Guinier-Preston (GP) zones which are coherent with the alloy matrix. With further ageing (overageing) the hardness of the alloy decreases as the precipitate loses its coherency with the alloy matrix.

***Keywords*: A201.0 Al alloy, solution treatment, thermodynamic modeling, ageing, Guinier-Preston (GP) zones, overageing**

I. INTRODUCTION

A unique combination of properties makes aluminum and its alloys the second most widely used material in the world after iron and its alloys [1]. Cast aluminium alloys are being used in the automotive and aerospace sector for the last three decades. In the cast Al alloy families the 2XX alloys offer the highest strength aluminum casting alloys available today [2]. The 2XX alloys also tend to retain their strength better than other alloy systems at elevated service temperatures [2]. High strength alloys have chemical compositions designed to provide high strength and ductility. It also implies high levels of internal soundness and microstructural refinement in case of premium engineered castings. Alloys considered premium strength consist of A201.0, A206.0, 224.0, 249.0, 354.0, A through D356.0, A through D357.0, 358.0 and 359.0 [3]. These are widely used in missile bodies, missile fins, aircraft pylons and canopies, wing flaps, speed brakes, hatch covers, hydraulic pumps, aerospace structural parts. These alloys are also used in automotive sector for suspension systems and cross-members, fuel pumps, brake valves and armored cupolas [3]. In general, heat treatments determine the final mechanical properties of aluminium alloys [1], [4], [5]. Solution heat treating, quenching and ageing are basic heat treatments for aluminum alloys. The proper selection of these heat treatments can achieve optimum combination of strength and ductility of the material. The purpose of solution heat treatment is to put the maximum practical amount of hardening solutes such as copper, magnesium, silicon into solid solution in the aluminum matrix and the purpose of quenching is to preserve the solid solution formed at the solution heat treating temperature by rapidly cooling to some lower temperature, usually near room temperature [1]. Quenching not only retains solute atoms in solid solution, but also maintains a certain minimum number of vacancies that assist in promoting the low temperature diffusion required for precipitation. The rapid quenching rates improve the strength. Furthermore, the purpose of ageing is to increase strength and resistance to corrosion by forming Guinier-Preston (GP) zones and precipitating second-phase particles from solid solution obtained from quenching [1], [6] – [9]. There are two types of ageing for aluminum alloys: natural ageing and artificial ageing [1]. Most of the heat treatable alloys exhibit age hardening at room temperature after quenching, called natural ageing. By reheating the quenched material to an elevated temperature, the solute content will be precipitated from solid solution gathering necessary energy for diffusion from heat, called artificial ageing which greatly affects the mechanical properties of aluminum alloys. Artificial ageing increases the mechanical properties of the alloy system with a decrease in ductility and toughness [10]. But overageing can lead to decrease in mechanical properties due the loss of coherency of the precipitates with the alloy matrix [10]. Hardening effects are attributable solely to the formation of a zone structure within the solid solution. Most of the strengthening occurs within a few hours after quenching at room temperature for most aluminum alloys. The mechanical properties are usually essentially stable after four days. Aluminium alloys stored under refrigeration can retard ageing [11]. The hardening observed at room temperature is attributed to localized concentrations of copper atoms forming Guinier-Preston zones, designated GP. These consist of two-dimensional copper rich regions of disk-like shape, oriented parallel to {100} planes [1]. A201.0 alloy is a precipitation hardening alloy which is subjected to a solution treatment, quenching, and a natural or artificial ageing treatment in order to obtain the optimum combination of mechanical properties.

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A combination of alloying elements, heat treatment and ageing induce the improvement of mechanical properties. For precipitation hardening treatment the alloy system must display decreasing solid solubility with decreasing temperature, i.e., the phase diagram should exhibit a change from a single solid phase to two solid phases ($\alpha \rightarrow \alpha + \beta$) as shown in Fig. 1 for Al-Cu binary phase diagram. The alloy matrix should be relatively soft and ductile, and the precipitate should be hard and brittle. The alloy must be quenchable and a coherent precipitate must form.

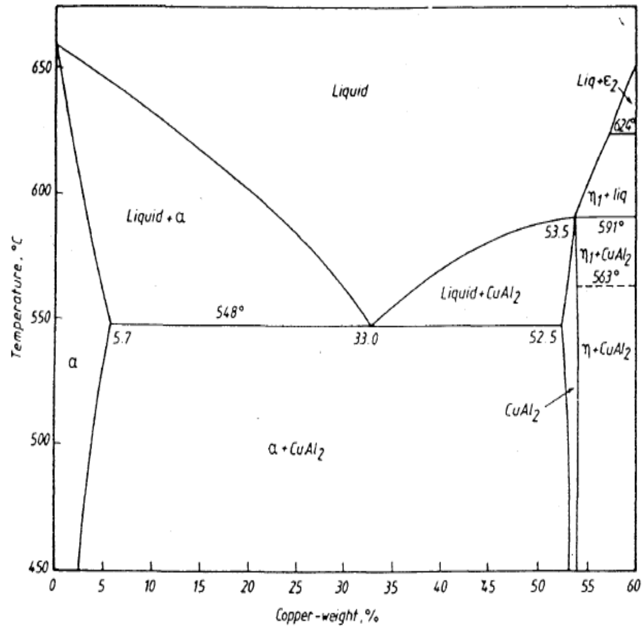

Fig. 1 The Al-Cu Binary Phase Diagram [12]

The mechanism of precipitation-hardening is based on the formation of intermetallic products during ageing treatment after the decomposition of a supersaturated solid solution (αSSS) obtained by solution treatment and quenching. During ageing treatment the two precipitation sequences are mainly responsible for precipitation hardening of this alloy type, namely [13]–[21] (αSSS) →GP zones →θ''→θ' (the metastable $CuAl_2$ phases) →θ $CuAl_2$) and (αSSS) →GPB zones → S´´→S´ (the metastable Al_2CuMg phases) → S (Al2CuMg). Depending on the alloy composition (Cu content, Cu/Mg ratio) and ageing parameter, different phase with different distribution, and consequently different material properties, can be obtained. The present work was aimed to investigate effects of ageing time and temperature treatment on the structure and mechanical properties of the A201.0 alloy. The principal mechanical property of the A201.0 alloy at the aged and overaged stages studied in this present work was hardness.

II. Materials and Methods

The experimental material is A201.0 alloy, and its chemical composition is listed in Table 1

Table 1 Chemical Composition of the A201.0 Al Alloy

Chemical element	Al	Si	Fe	Cu	Mn	Mg	Ti
wt%	Base	0.05	0.1	4.5	0.5	0.55	0.2

A. *Heat Treatment*

The heat treatment performed on the samples consisted in three successive steps. Ten cylindrical samples with volume of 1 cm^3 were treated at 540 ± 2 °C in a heat treatment furnace for 4 hours and quenched into cold water (< 20°C).

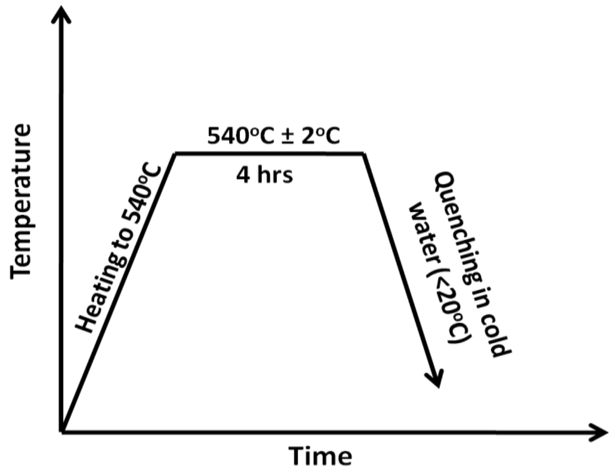

Fig. 2 Solution Treatment of the A201.0 Samples

The samples were kept in refrigeration for about 5 days to retard ageing. Ageing of the samples were carried out at 170°C according to the schedule given in Table 2.

Table 2 Ageing Schedule for A201.0 Samples

Samples	Ageing time
1 (Natural ageing)	0h
[redacted]	Effective ageing time at 170°C
2	0.5h
3	1h
4	2h
5	4h
6	10h
7	50h
8	100h
9	1000h
10	10000h

B. *Microstructural Analysis*

Microstructural analysis was carried out for specimens before heat treatment and after heat treatment. The specimens for microstructural analysis were machined to cylindrical shapes. Standard techniques were followed for the preparation for observation. Grinding was done on silicon carbide abrasive papers of various grit sizes, final polishing being done on a velvet cloth using a suspension of alumina (Al_2O_3) powder. The specimens were then washed and dried with acetone before observation. The microstructures were studied using a metallurgical microscope.

C. *Phase Analysis*

Using JMatPro the evolution of phases during solidification of A201.0 Al alloy was identified.

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The presence of intermetallic phases and their solidification was also studied. Furthermore, the stoppage/abolition of solidification of few intermetallic phases was also identified.

A. Hardness Test

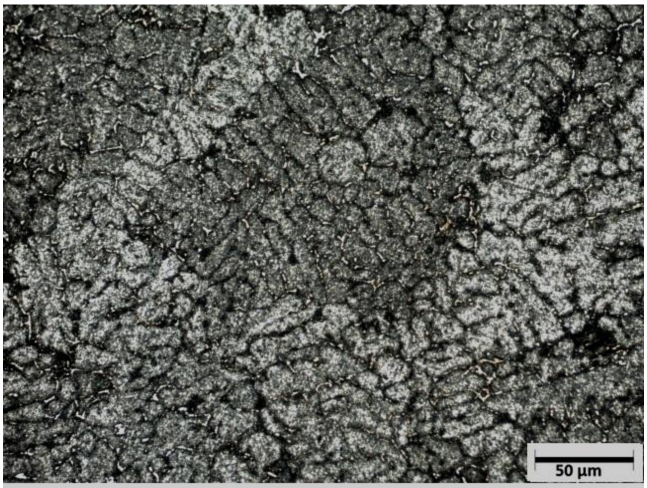
The hardness values of the samples were determined using the Brinell hardness tester. The Brinell hardness test was conducted by applying 500 kg load (L) for 30 seconds with a ball of 10 mm diameter (D) on the surface of the samples. Then the diameter of the impression (d) was measured by the scale. Then the hardness was calculated by the equation given below:

$$BHN = \frac{L}{\pi D\left(D-\sqrt{D^2-d^2}\right)} \tag{1}$$

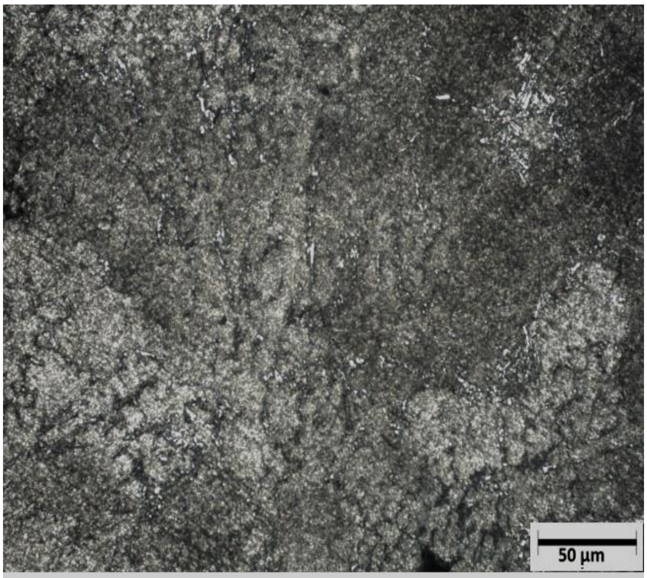
III. RESULTS AND DISCUSSIONS

A. Microstructure Evaluation and Phase Analysis

The microstructure of the A201.0 Al alloy in the as-cast and aged condition is given in Fig. 3(a) and (b) respectively.



(A)



(B)

Fig. 3 Microstructure [Magnification: 250X] of A201.0 Al alloy (a) As-Cast Condition (b) Produced by Moderately Fast Cooling from 540 ± 2 °C

The Al–Cu phase diagram tells us that, between 500°C and 580°C, the 4.4% Cu alloy is single phase: the Cu dissolves in the Al to give the random substitutional solid solution α. Below 500°C the alloy enters the two-phase field of α + $CuAl_2$. The amount of $CuAl_2$ increases with decreasing temperature. The as-cast alloy consists of dendritic microstructure [Fig. 3 (a)]. Fig. 3(b) shows the microstructure that was obtained by moderately fast cooling from 540°C to room temperature. Due to fast cooling a finer structure was produced. It is due to high nucleation rate owing to large driving force. The precipitates are small and closely spaced. They get in the way of moving dislocations and make the alloy harder [22]. Slow cooling in the furnace was avoided. In slow cooling the driving force for the precipitation of $CuAl_2$ is small and the nucleation rate is low [22]. If we cool the sample in the furnace we are allowing the sample to cool slowly which will lead to large precipitates of $CuAl_2$ in the matrix. This will lead to decrease in strength and hardness. Quenching in water will lead to small precipitates spaced closely apart leading to dislocation movement being hindered. As a result strength and hardness increases [22]. Using JMatPro thermodynamic calculations of changes in phases with temperature in equilibrium solidification conditions were made. The identity and amount of different phases in the alloy at different temperatures were predicted. From the Step Temperature calculation predicted in thermodynamic modeling using JMatPro, the evolution and abolition of different phases in the A201.0 Al alloy are clearly observed in Fig. 4 and Fig. 5. From the Step Temperature calculation predicted in thermodynamic modelling, the evolution and abolition of different phases in the A201.0 Al alloy are clearly observed in Fig. 4. It is evident that the Al phase starts to solidify first at 640^{0}C. Fig. 5, we can see that the Al-Cu-Mn phase ($Al_{20}Cu_2Mn_3$) starts to solidify at 578^{0}C and grows up to 1.95 wt%. At 385^{0}C, an Al-Cu-Mg phase (S-Al_2CuMg) starts to grow. The Al-Cu phase (Al_2Cu) starts to evolve at 496^{0}C.At 135^{0}C, this phase starts to increase from 5.2 to 5.65 wt % with the evolution of Al-Mn phase (Al_6Mn) and the abolition of Al-Cu-Mn phase. So the Al-Cu-Mn phase dissociates into those Al-Cu and Al-Mn phases. Rests of the intermetallic phases are below 1.0 wt%. Fig. 6 shows the enrichment of liquid phase with Cu and Mg, using Scheil-Gulliver modeling for non-equilibrium solidification. So the interdendritic area might be filled with intermetallic phases containing Cu and Mg. Also it can be seen that the solidification starts at 695^{0}C and completely ends at 505^{0}C.

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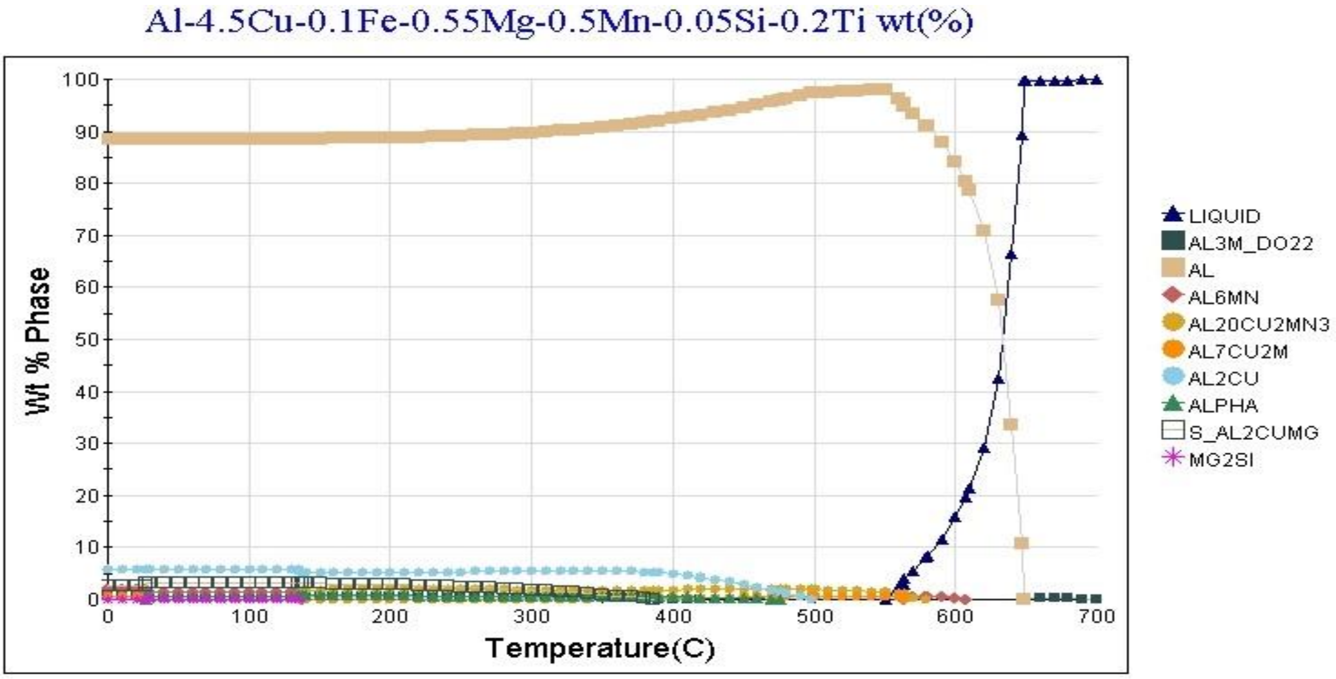

Fig. 4 Thermodynamic Modeling of A201.0 Al Alloy

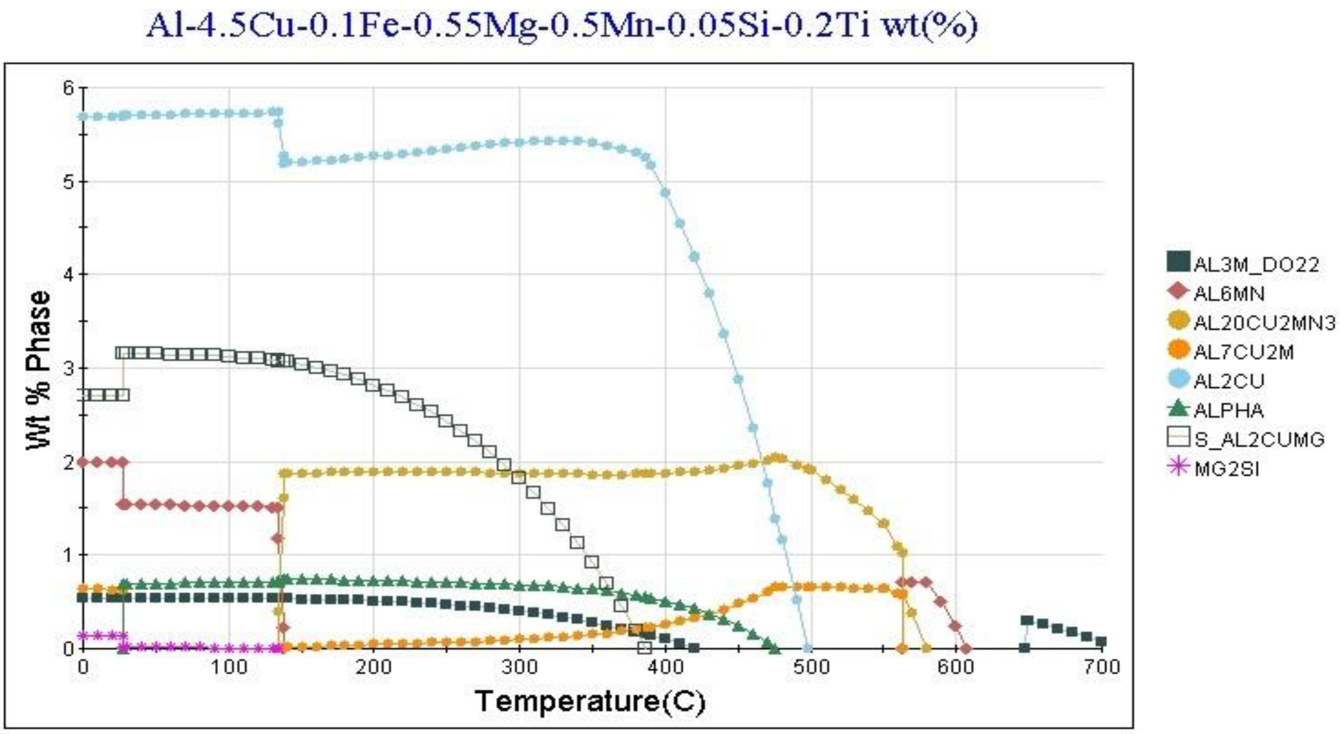

Fig. 5 Evolution and Abolition of Intermetallic Phases During Solidification

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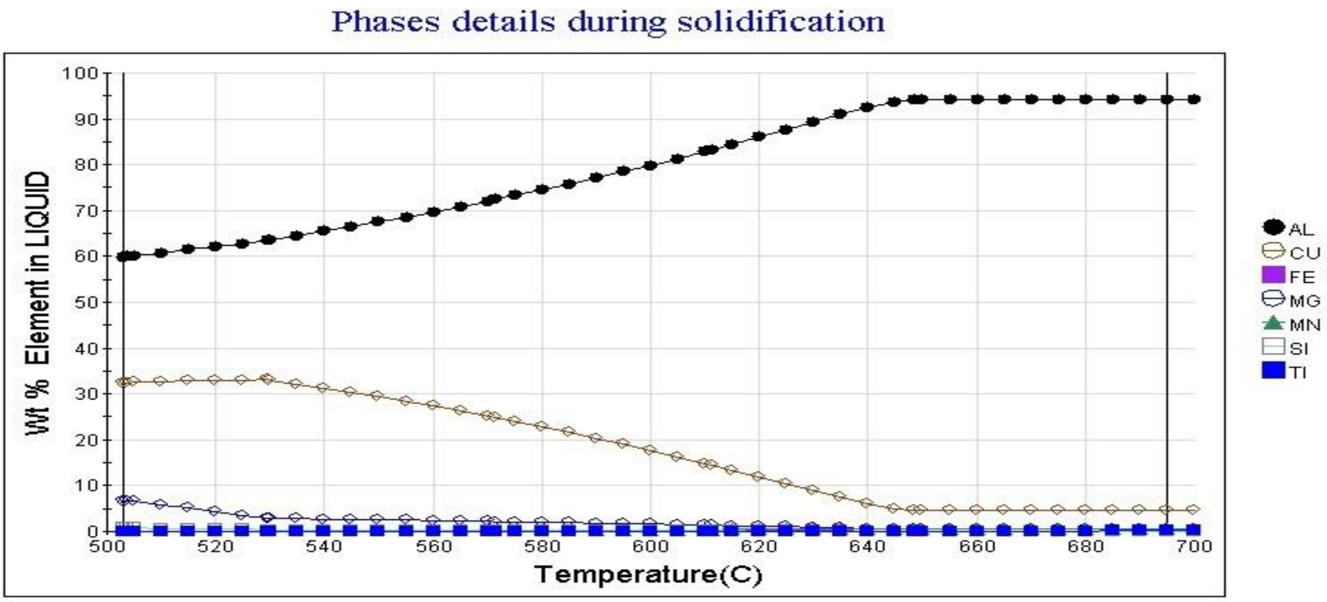

Fig. 6 Enrichment of Elements in the Liquid Phase

B. *Effect of Ageing Time and Temperature on Hardness*

Artificial ageing leads to improvement of mechanical properties. The degree of precipitation hardening is a function of ageing temperature and treatment duration. The change in hardness, during artificial ageing can be plotted as a curve (Fig. 7) with a maximum value which presents the hardness variation as a function of artificial ageing time at 170^{0}C.

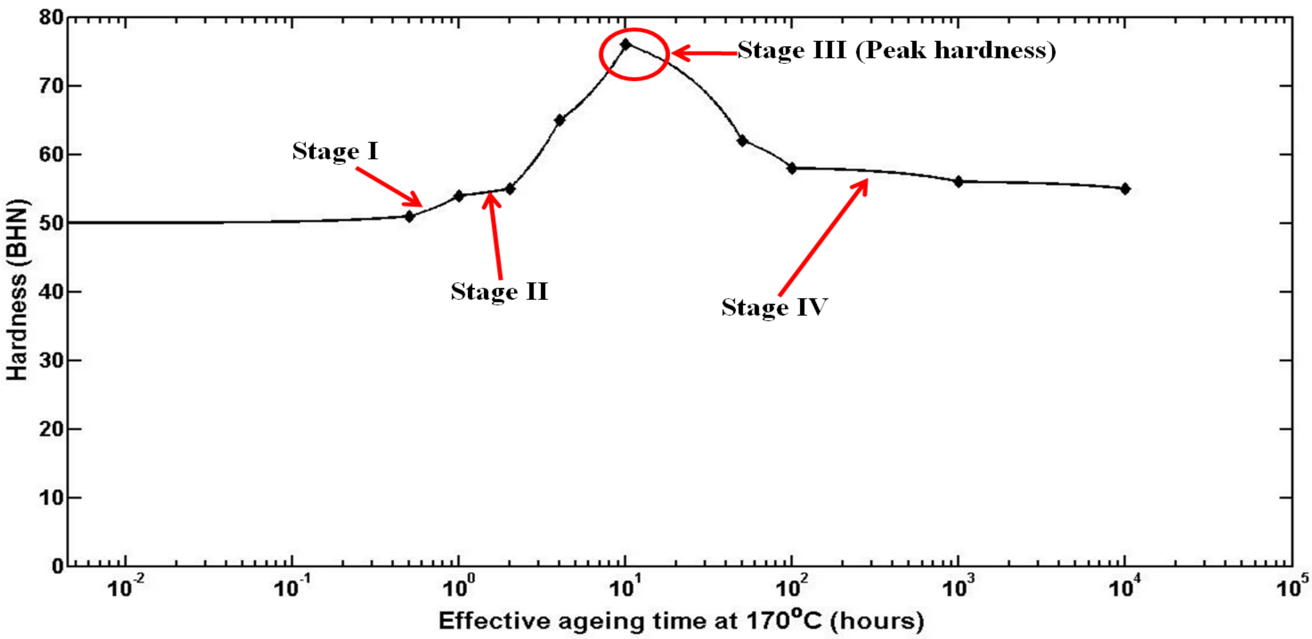

Fig. 7 Hardness as a Function of Artificial Ageing Time at 1700C

The maximum hardness peak at 170^{0}C appears after 10 hours of treatment, and a stabilization of the hardness appears respectively after 1000 hours as seen in Fig. 7. The variation of hardness as a function of the artificial ageing time showed four stages [Fig.7].

- Stage I: Initial rapid hardness, characterized by an initial rapid hardening
- Stage II: Hardness plateau, indicating the stability of the hardness
- Stage III: Second hardness, characterized by an increase in hardness (peak hardness); has been attributed to the formation of S' and S-phase as shown in Fig. 5
- Stage IV: Overageing, hardness stabilization

The first stage of hardening, characterized by an initial rapid hardening, occurs very rapidly. This stage represents approximately 50-70% of maximum hardening alloy and corresponds to the formation of Cu-Mg clusters [23] - [26]. The second stage is characterized by an increase in hardness (peak hardness) and has been attributed to the formation of the GP zone, described as short range ordering of one Cu + Mg layer and several Al layers along the {100} Al planes and S–Al_2CuMg phases [24], [27], [28].

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Between these two stages of increase in hardness, there is a "plateau", during which time the hardness does not change for many hours of treatment. The over-ageing treatment allows stabilizing the microstructure and hardness. The hardening mechanisms at work during the ageing process are solid solution hardening, Coherency stress hardening and Precipitation hardening. During solid solution hardening at the start of ageing the alloy is mostly strengthened by the 4.5 wt% of copper that is trapped in the supersaturated α. But when the GP zones form, almost all of the Cu is removed from solution and the solution strengthening virtually disappears. In coherency stress hardening the coherency strains around the GP zones and θ" precipitates generate stresses that help prevent dislocation movement. The GP zones give the larger hardening effect. In precipitation hardening, the precipitates obstruct the dislocations directly [22].

IV. CONCLUSIONS

The influence of ageing time and temperature on the structure and hardness of A201.0 alloy has been investigated. Microstructural analysis revealed change in microstructure after heat treatment of A201.0 alloy. The microstructure of A201.0 alloy changed from dendritic structure to a structure with saturated α phase with $CuAl_2$ precipitates when quenched from solution treatment temperature. By the aid of JMatPro using thermodynamic modeling, the presence of Mg-Si, $CuAl_2$ precipitates and S-Al_2CuMg phase was confirmed and the presence of Cu enriched phase in the liquid further confirmed the presence of Cu intermetallic phases. These phases are primarily responsible for the hardness of the alloy. The hardness of the A201.0 alloy as a function of ageing time at a particular temperature followed a bell shaped curve revealing four stages of variation. The peak hardness was achieved in the fourth stage after 10 hours. Overageing the alloy stabilized the microstructure and hardness after a period of 1000 hours.

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