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EFFECT OF ISOTHERMAL TREATMENT ON MICROSTRUCTURE OF SODIUM-MODIFIED A356.0-TYPE Al-Si-Mg ALLOY

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In the design of useful precipitation hardened alloys for engineering applications, the microstructure remains an essential tool to be altered by any proven isothermal treatment route. Based on this, the isothermal treatment consisting of multiple-step thermal treatment was developed and the microstructures as a result of the treatments were examined with optical microscope (OPM) and scanning electron microscope with energy dispersive spectroscopy (SEM-EDS) using micro-hardness as criteria. The microstructures were discussed with respect to the peak-aged alloys at different thermal conditions. The improvement in the hardness value with the double thermal ageing treatment (DTAT) developed for A356.0-type Al-Si-Mg alloy have been attributed to the finer dispersion of precipitates in the microstructure obtained in the pre-ageing stage of 105^{o}C/5 h. Sample aged at 180^{o}C/2 h (peak-aged) in the DTAT condition, the structure consists of spherodized silicon particles and grain size have been refined as a result of the thermal treatment and the SEM-EDS mapping for alloy at the peak-aged temperature and time was examined confirming the presence of the constituent alloy elements in the matrix of aluminium solid solution, even after the alloy have been reprocessed.

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1. Introduction

The use of aluminium alloys for automotive body materials and aerospace application has been driven by a number of issues, in particular weight reduction, fuel efficiency and enhanced mechanical performance. These have stimulated research and development of alloys with high strength-to-density ratio to reduce vehicle weight [1-5]. The use of Al-Si-Mg alloys in particular for automotive industry is attractive due to light weight and reasonable strength after ageing treatment [6]. The A356-type Al-Si-Mg alloy offer hardening possibilities that leads to specific properties and applications in automobile, aerospace and marine industries. Since precipitation hardening remains one of the major treatments that are usually adopted for the purpose of increasing strength/hardness in aluminium alloys [1,7-9], method(s) for improving this treatment becomes necessary. Both sodium (Na) and strontium (Sr) have historically been used for modification of Al-Si alloys. The beneficial effect of modification is connected with a change in the shape of the silicon crystals. In the normal untreated castings silicon is found as irregular plates while in the modified casting they are very small and worm shaped surrounded by ductile aluminium. The optimum Na or Sr content in a modified A356 is approximately 0.01% Na, or Sr and with less Na or Sr the modification is incomplete while with higher Na or Sr content large Si crystals are formed, besides the fine grained matrix [6]. Though many theories have been put forward to explain the process, the most probable structure is a ternary eutectic mixture Al-Si-Na, or Sr (eutectic temperature is reduced by 10% with Na as modifier).

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The microstructure and alloy constituents are necessitated to achieve optimum mechanical properties. Some of the critical microstructure features are grain size, dendritic arm spacing (DAS) and silicon morphology in the eutectic phase [9-11]. The cast Microstructure of these alloys (Al-Si) consist of a primary phase, aluminium or silicon and a eutectic mixture of these two elements. Depending on the purity of the base materials, the Al-Si alloys contain varying amounts of impurity elements such as iron, manganese, copper and zinc. In addition, copper and magnesium are often added as alloying elements to increase the strength and hardenability of the materials being cast. The impurities and alloying elements partly go into solid solution in the matrix and partly form intermetallic particles during the solidification process [12].

Through the development of technology for modifying the eutectic structure of silicon on Al-Si casting alloys, an overriding hypothesis has been that the mechanical properties of these alloys improve with changes in microstructure away from the acicular granular either to the lamellar refinement from antimony (Sb) addition or toward the fibrous structure from modification with Na and /or Sr [13]. In Kumar et al. [14], they reported that modification increases the heat transfer rate from the solidifying alloy to the mould wall which is the direct consequence of morphological transformation. In recent years chemical and thermal modifications have been used together to produce the desired casting properties [15]. The interest in this research work is to study the microstructures as a result of the novel isothermal treatment developed for the A356.0-type Al-Si-Mg alloy.

2. Experimental Procedures

The samples were produced according to the method described [16]. The composition of the produced A356.0-type Al-Si-Mg alloy is given in Table 1.

Table 1: Chemical composition of the produced A356.0-type Al-Si-Mg alloy (wt %)

Al	Si	Mg	Fe	Mn	Cr	Pb+Sn	Zn	Cu	Ti	Ni	Na
92.14	7.0	0.3	0.08	0.03	0.2	0.03	0.05	0.03	0.11	0.03	0.01

The eutectics silicon particles were modified with elemental sodium (0.01%). Sample for hardness and microstructure were sectioned from a cylindrical test bar; 22 mm by 300 mm cast alloy. All samples were then solution heat treated for 1 h at 540^{o}C,artificial ageing; double thermal ageing treatment (DTAT) and single thermal ageing treatment STAT with varying ageing temperatures of 150^{o}C,180^{o}C, 210^{o}C and ageing time of 1-5, 18, 20 h. Sample for DTAT were pre-aged at a temperature of 105^{o}C for 5 h. This was followed by ageing at 150^{o}C, 180^{o}C and 210^{o}C for 1-5, 18 and 20 h before cooling in air, then aged at 200 ^{o}C for various time, 1-4 h. Equally, samples from the solution heat treatment (SHT) temperature were step-quenched to temperature of 220^{o}C for 10, 20 and 30 s, then quenched in warm water step-quench-ageing (SQA). Finally, the quenched samples were aged to the STAT and DTAT tempers. While some sample undergo an interrupted-quenching–ageing (IQA) treatment. The schematic of the procedure for the treatments is shown in Figure 1. The hardness data was assessed by using Vickers micro-hardness tester of 500 g load with at least six measurements. The scanning electron microscopy equipped with energy dispersive spectroscopy (SEM-EDS), optical microscope (OPM) samples were electro-polished and etched with Keller's reagent. OPM and SEM machines used were digital light microscope Olympic-Japan (u25LBD) and Joel (JSM-6300) respectively.

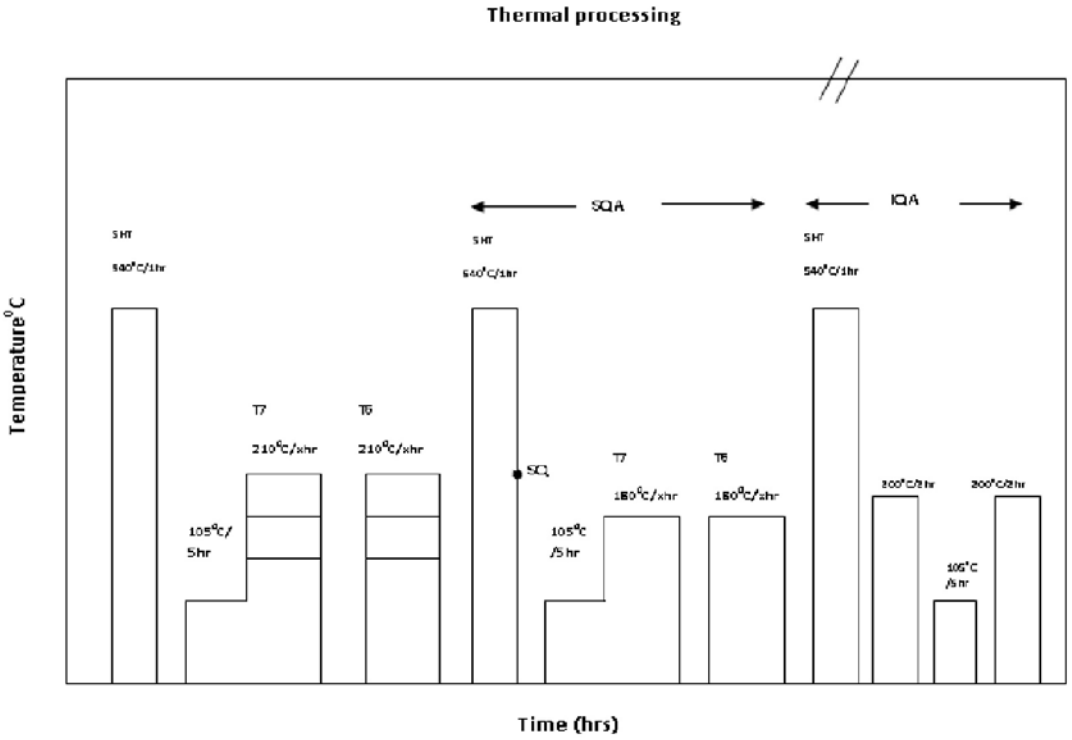


Fig. 1- Schematic diagram of the isothermal treatments used for Al-Si-Mg alloy

3. Results and discussion

The microstructures of peak aged DTAT and STAT are shown in Plates 1-2. Plates 1-2 show SQA and IQA temper micrographs. The F-temper and SHT sample micrographs are represented in Plate 4. While the SEM-EDS mapping of the peak aged sample is presented in Plate 5. Considering that the peak-aged hardness for double thermal and single thermal alloy are at 180^{o}C for 20hr (134.9) and 180^{o}C for 2hr (127.2HVN) respectively. The microstructures for these conditions can be found from SEM secondary electron image and optical micrograph (Plates 1 and 2) respectively. From the micrographs, there are indications that the silicon flakes are little finer and that the intermetallic compound such as $Al_{12}Mg_{17}$ are responsible for hardness improvement. Microstructure generally consists of some fine dispersion of precipitates within the matrix of aluminium solid solution.

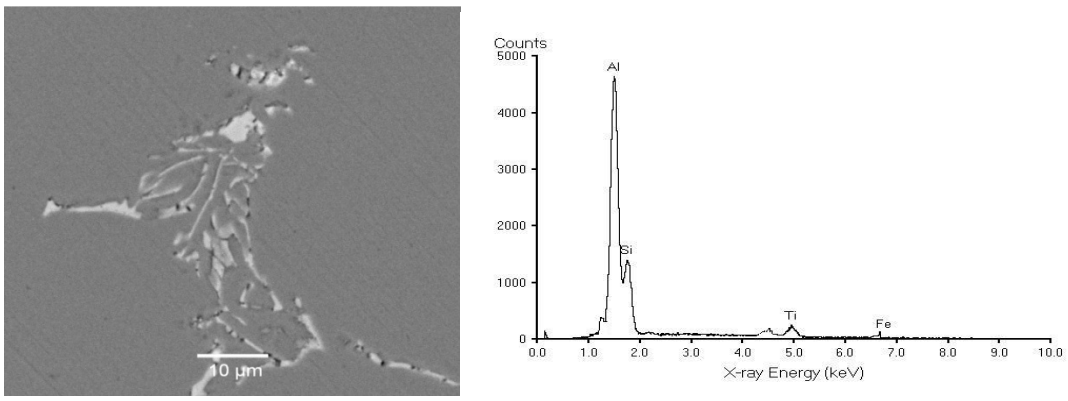


Plate 1- SEM micrograph of T7 180^{o}C /20hr with the EDS spectrum showing β''Mg_2Si and $Al_{12}Mg_{17}$.

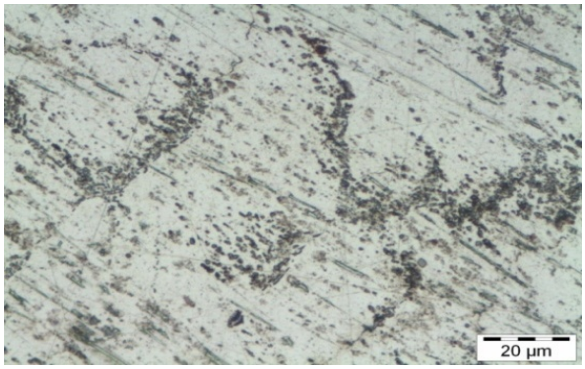
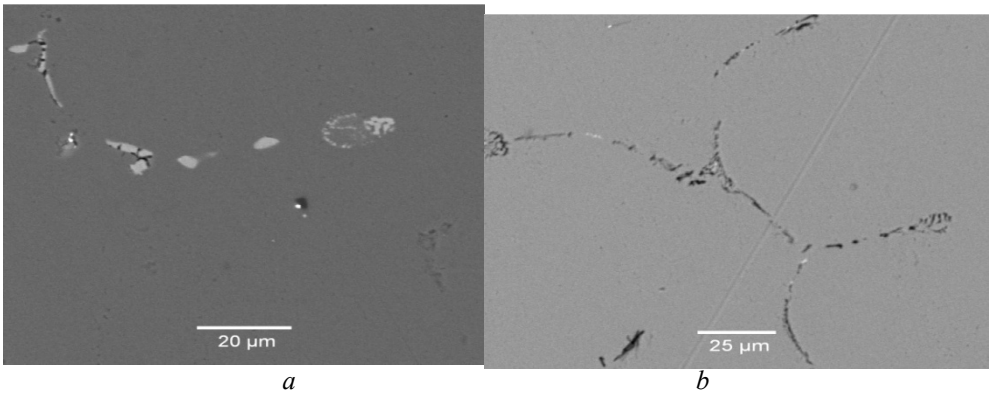
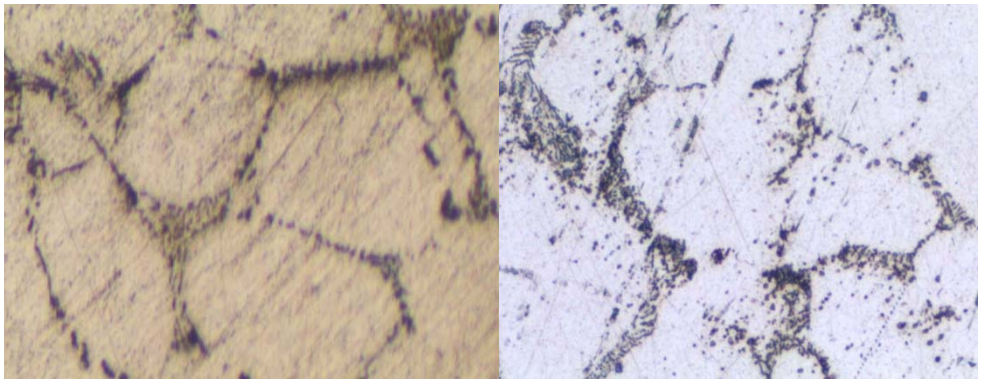


Plate 2- Optical micrograph for STAT peak aged at 180°C/2hr



a *b*

Plate 3- SEM BSE image a) SQA T7 220/20sec, 180/2hr indicating $Al_8Mg_3FeSi_6$ intermetallic compound b) SEM BSE image IQA$_4$with Al_3Cr dispersiod

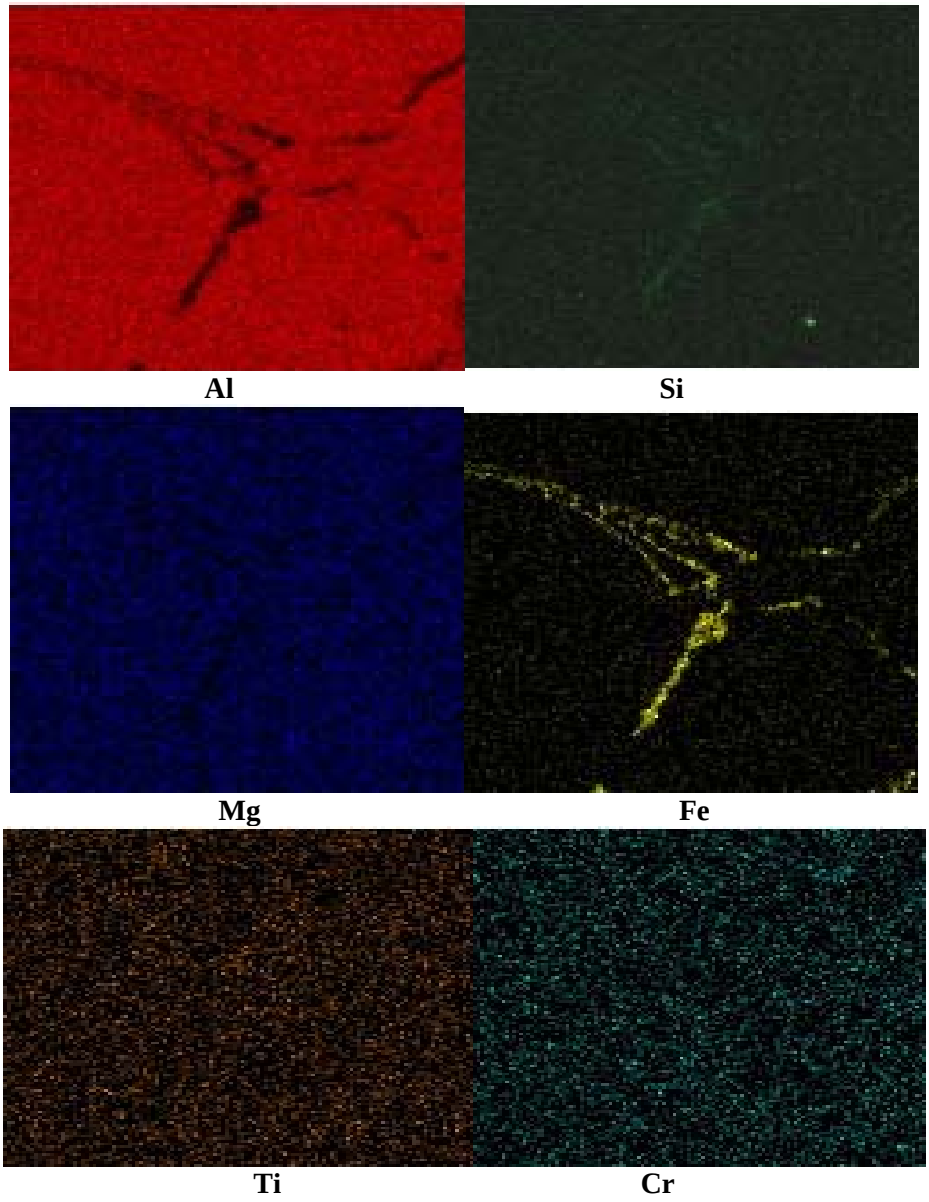


a *b*

Plate 4- Optical micrograph a) F-Temper b) SHT at 540°C for 1hr 20µm (x200)

Specifically, in the DTAT (Plate 1), the fine precipitates formed during the first-step ageing, are then retained at higher ageing temperatures and transformed to the final precipitates during the second-step ageing at temperatures and time considered. The higher values of hardness obtained for the alloys with DTAT treatment is due to the higher precipitate volume fraction (PVF) giving rises to smaller inter-precipitate spacing and the higher stress required for dislocation bowing. The maximum hardness value with interrupted-quenching-ageing (119HVN).and step-quenched-ageing (125.9HVN) alloys showed similar structures (Plate 3). The microstructure of the alloy in the F-temper (as-cast) and solution heat treatment (SHT) at 540°C for 1 h is shown in Plate 4a, and b respectively. These structures were also similar to the report [17]. Equally from the SEM

mapping of peak-aged DTAT alloy (Plate 5), the constituent alloy was confirmed to be present and have contributed to the strengthening mechanism during precipitation hardening process. However, this is not a law, but rather an indication of complete homogenization and probably a promising spherodization of the resultant microstructures developed from the thermal treatments. The as-cast microstructure consists of α-aluminium solid solution with the modified eutectics (Plate 4a). The solution heat treated sample shows spherodized silicon in the α-aluminium solid solution (Plate 4b) .Within the α-aluminium solid solution, the intermetallic phases consisting of several precipitates. These intermetallic phases are shown by the SEM study. After the various thermal treatment, for example (Plate 2); sample aged at 180^{o}C/2hr in the STAT condition, structure consist of spherodized silicon particles and grains have been refined by the thermal treatment. The Al_3Cr dispersiod are more refined, highly populated and distributed within the matrix of the aluminium solid solution and more likelihood of trace of Mg_2Si phase formation along the grain boundaries. In the microstructure of DTAT for A356.0-type Al-Si-Mg alloy at 180^{o}C (Plate 2), it is clear that more fine precipitates of β'' dispersion occurred from the pre-ageing stage of the alloy, similar to those observed [18].



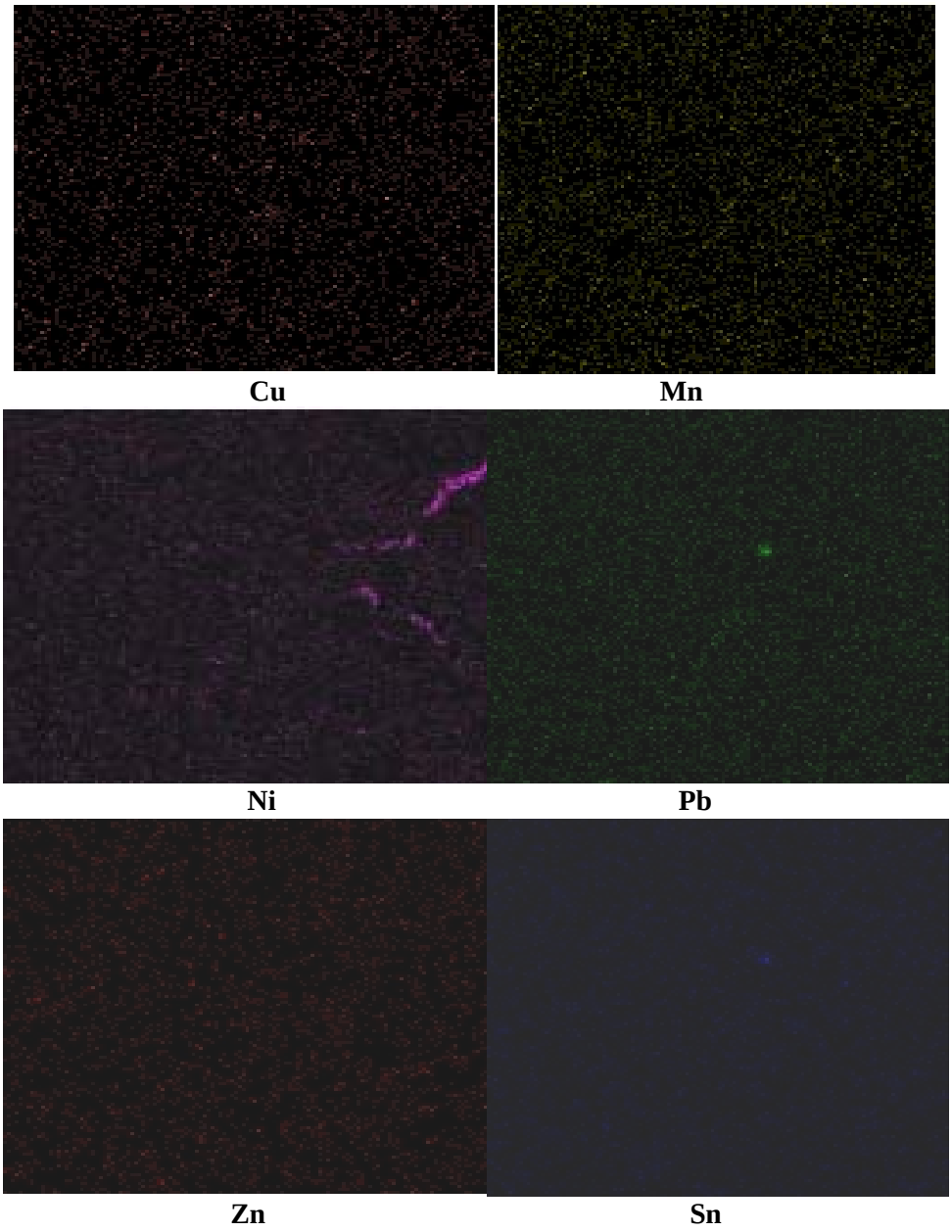


Plate 5- SEM-EDS mapping for peak-aged DTAT A356.0-type Al-Si-Mg sample at 180/20hr with Fe, Ti, Cr, Cu, Mn, Ni, Pb, Zn and Sn element.

This was also responsible for the improved hardness value of the alloy. Considering the hardness of SQA alloy obtained at 220^{o}C/ for 20 seconds, the structure (Plate 3a) is similar to the one obtained at the peak-aged temperature of 180^{o}C in the DTAT condition. The microstructure obtained from IQA_4 (Plate 3b) also consists of breakdown of inter-dendritic network of dispersed silicon eutectic in the α-aluminium solid solution. The increase in the hardness of this temper is however attributed to this network [19-20]. Ageing of the alloy has resulted in a pronounced agglomeration of undesolved silicon crystallites. The commonest structure for most of the alloy aged at prolongs ageing time consists of dissolution of the silicon and partially refined grain size. This however differs in size and morphology as regards ageing temperature and time studied. Although in order to have full microstructural characterization of the alloy, an extensive study using transmission electron microscopy (TEM) analysis is required. The TEM study of the morphology and crystal orientation of this group of alloy upon multiple-step thermal ageing treatment is ongoing in one of our work.

4. Conclusions

The sizes and shapes of silicon particles and primary alpha phase in the structures as a result of the isothermal processes are responsible for the hardness variations in the studied A356.0-type Al-Si-Mg alloy.

The improvement in hardness associated with the DTAT developed for A356.0-type Al-Si-Mg alloy especially at the peak-aged stage have been attributed to the finer dispersion of precipitates obtained in the pre-ageing stage of 105^{o}C/5hr which was further retained and fully transformed to finer precipitates at higher ageing time.

Equally, the SEM mapping microstructures has shown that the constituent alloying elements contributed to the increase in hardness value within the studied conditions.

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