2.2 Melting and Casting

About 10 kg of alloy ingots were cut into suitable sizes for charging into the graphite crucible. The cut pieces were first mechanically cleaned using acetone to remove the adherent dust and oily materials. The ingots were preheated before charging for melting. Mild steel tools which come into contact with molten metal were suitably coated with graphite and dried properly by heating. The melting process was carried out in an 18kW electrical resistance furnace. When the crucible kept in the furnace attains 700°C, the preheated ingots were charged into it. The melting temperature was maintained at 750 ±5°C. Temperature was monitored using a Chromel–Alumel thermocouple. The alloying elements (Cu, Ni, and Mg) in the master alloy form are added in amounts calculated to obtain the desired compositions. The melt was degassed by bubbling pure, dry nitrogen gas into the melt for about 60 minutes by means of a graphite lance to remove the hydrogen and inclusions. Five near-eutectic piston alloys with varying nickel additions (2.1% to 2.8%) were employed using gravity-die casting method. The melt was poured directly into the metal moulds which were preheated to 200°C and allowed the metal to cool down to room temperature.

2.3 Heat Treatment

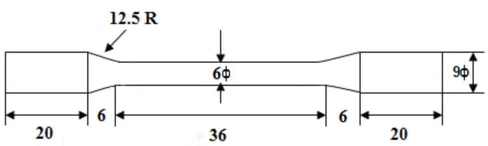
T6 heat treatment i.e., solutionizing, quenching, and ageing treatments, were carried out in a resistance heating muffle furnace at 500°C (±5°C) for 5 h and then quenching in the cold water at 20°C. The aging treatment consists of heating to 180°C (±2°C) and holding for 9 h, followed by air cooling [9].

2.4 Metallography

Metallographic specimens were polished using normal specimen polishing techniques. Leica DMRX optical microscope was employed for visualization of the internal structure of the material.

2.5 Tensile and Hardness Tests

Tensile tests were conducted using a servo-hydraulically controlled Universal Testing Machine (INSTRON model 1195-5500R). The tensile properties of the as-cast and heat-treated specimens at room temperature were measured. The standard tensile specimen used for the test is shown in fig. 1. Brinell hardness was evaluated using INDENTEC Universal Hardness Testing machine. The BHN was measured using 62.5 kg load for 30 seconds and 2.5 mm diameter indenter. The results of the tests were averaged from three determinations.



All dimensions are in mm.

Figure 1: Tensile test specimen as per ASTM E8M-04 standard

3. Results and Discussion

3.1 Microstructure

Fig. 2 shows the comparison between the microstructures of the alloy with varying nickel content in as-cast and T6 heat treatment condition. The differences in the size and shape of different features are evident. The microstructure consists of α-aluminum dendritic halos with eutectic Si and complex intermetallic compounds segregated into the inter-dendritic regions and primary Si. The features of the microstructures of the alloy in T6 condition undergo changes upon heat treatment. Microstructural observation shows that the secondary arm spacing is reduced for the heat treated alloys. Most of the intermetallic phases are partially dissolved and tend to spherodize, i.e., sharp corners have become rounded. The morphology change of the eutectic Si is obvious after heat treatment. The plate-like eutectic Si in the as cast condition is broken into small particles. That is, the Si particles break down into smaller fragments and become gradually spheroidized. The changes in size and morphology of the discontinuous silicon phase are significant since they have a direct influence on the tensile properties. It is further noticed from the microstructure that the addition of 2.28% Ni [Alloy-B] has led to an interesting microstructural change than others. The massive rod like Si particles has been changed to a fine spherical shape besides the intermetallics phases distributing evenly along the grain boundaries. Further refinement in the grain size is also noticed after T6 heat treatment. The intermetallic phases are the main elevated-temperature strengthening phases in Al–Si piston alloys, especially Ni-rich phases. At elevated temperature, the thermally stable Ni-rich phases could impose drag on boundaries and hinder the slide of α-Al grains. Therefore, Ni-rich phases can be called the main strengthening phases in Al–Si piston alloys at elevated temperature. Generally, the inter metallic phases found in these alloys are Al2Cu, Mg2Si, Al3CuNi, Al7Cu4Ni , Al3Ni and Al5CuMg8Si6 and so on [3, 10, 11].

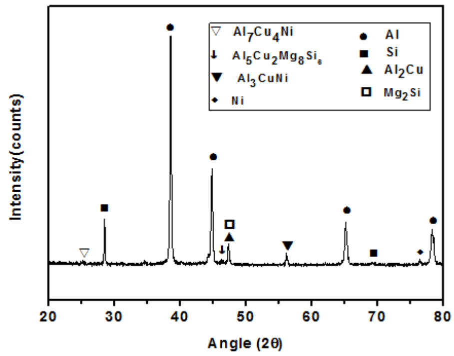


Figure 2: XRD pattern for the heat treated Alloy-B

The result of the X-ray Diffraction (XRD) analyses preformed on the heat treated alloy with 2.28%Ni shown in Figure 2 confirm that the alloy consists of different phases, viz., Al, Si, Al2Cu, Mg2Si, Al3CuNi, Al7Cu4Ni and Al5Cu2Mg8Si6.

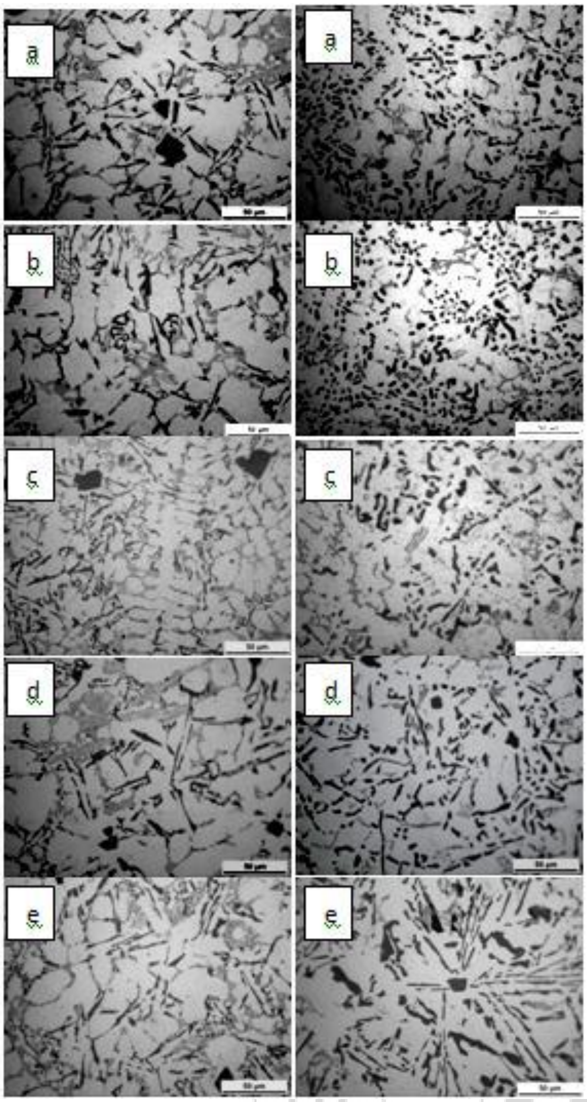


Figure 3: Optical micrographs of the permenent mould Al-Si-Cu-Ni-Mg alloy with varying nickel content before and after T6 heat treatment (a)Alloy-A (b)Alloy-B (c) Alloy-C (d) Alloy-D (e) Alloy-E

3.2 Tensile Properties

The tensile properties of the castings made in permanent moulds with different Ni content are examined in the as-cast and T6 heat treatment conditions.

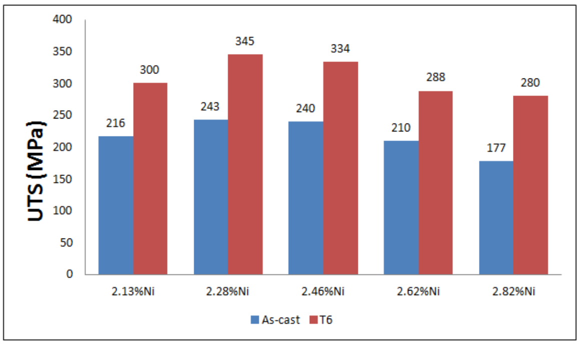


Fig. 4. Tensile properties of the alloys tested at RT.

As can be seen from fig 4, the tensile strengths are different, with the varying Ni content. The results reveal that 2.28%Ni and 2.46%Ni alloy (Alloy-B and Alloy-C) are having higher tensile strength than that of the other alloys in both as cast and T6 condition. This signifies that the addition of 2.3%Ni could be an optimum level of addition required for enhancing the strength of the alloy.

3.3 Hardness

As seen in fig.5, the hardness values for the alloys are increased after T6 heat treatment. This indicates T6 heat treatment is effective in these alloys for increasing the hardness.

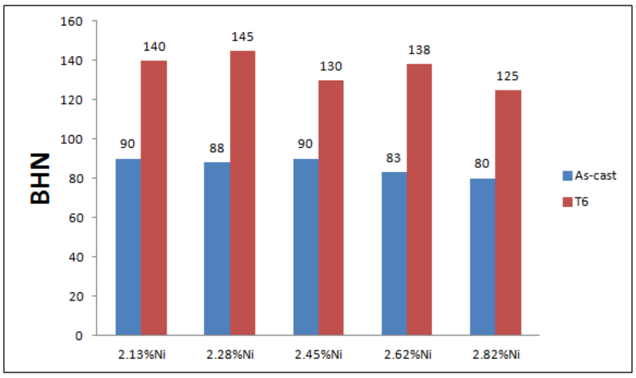


Fig. 5. Hardness values for varying Nickel content.

This hardening effect is due to the precipitation of Al_2Cu, Al_3CuMg and Mg_2Si phases during heat treatment [4,14]. The Alloy with 2.28%Ni (Alloy-B) is having higher hardness than that of the other alloys in T6 condition. It can be seen that the strength and hardness have increased in the heat treated alloys at the expense of ductility. When the alloy is solutionized, it forms a homogeneous single phase, followed by quenching to retain the solute component in unstable state. On subsequent ageing and precipitation treatment at 180°C, the solute atoms are rejected and form a cluster as a coherent or semi-coherent precipitate. The strain field around a coherent or semi-coherent precipitate inhibits the movement of dislocations with increase in strength and hardness but decrease in ductility. From the above results, optimum microstructure and mechanical properties for both as-cast as well as heat treated specimens were obtained for the alloy containing 2.28%Ni (Alloy-B). This alloy shows 145 BHN hardness and 345 MPa UTS in the T6 heat treated condition.

4. Standardization of Aging time

4.1 Effect of ageing time on wear

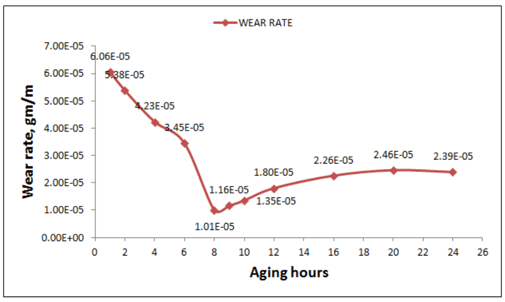


Figure 6: Effect of aging time on the wear rate of Alloy-B aged

Effect of aging time on the wear rate of the alloy is plotted in fig. 6. From the graph it is clear that when the aging time increases from 1h to 8h the wear loss decreases. A significant reduction in wear rate is obtained at the aging time of 8 hours. It is clearly visible from plot 9 the wear rate decreases from 6.06x10-5 gm/m at 1h aging to 1.01x10-6 gm/m at 8h of ageing.

4.2 Effect of aging time on hardness

Fig. 7 shows how the hardness value varies with aging time on the samples aged at 180◦C. At early stages of aging, the hardness increases with time until it reach the peak. After reaching the peak, the BHN decreases as a result of over-aging. The peak value of 145 BHN was obtained at 8 h and the hardness at 9h aging is 144 BHN. The hardness curve shown in Fig. 10 exhibit another peak at 20 h or a wavy form with aging time, resulting from the presence of several hardening phases, including Al2Cu, Mg2Si, and Al7CuNi which contribute to the precipitation hardening of the alloys.

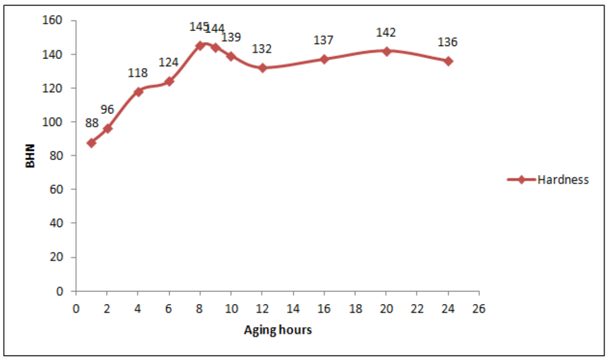


Figure 7: Hardness of Alloy-B as a function of aging time.

Artificial aging of alloy leads to the precipitation hardening, breaking the as cast dendritic structure and reducing the segregation of alloying elements and spheroidizing the silicon crystals and improved bonding between the second phase particles and matrix aluminium. Therefore, heat treated alloys show enhanced hardness. From the results, it is clear that increase in aging time from 1h to 8h enhances the hardness and wear resistance and aging for longer time leads to overaging resulting in drop in hardness values due to extensive aging, where the precipitates turns to be the equilibrium phase[9,15].

Increasing the aging time above peak aging time results the commencement of the change in the morphology of the precipitate from spherical to a mixture of spherical and elongated particles indicating the tendency of the precipitates to change from semi-incoherent to stable incoherent Al2Cu, causing a clear reduction in the alloy hardness. These observations are in good agreement with the work of M.F. Ibrahima et al. [16]. From the above results, for the developed alloy, solution treatment at 495°C for 5h and then quenching in cold water at 20°C and aging at 180 °C for 8 to 9 hours can be considered as the optimum heat treatment cycle.

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