Wear behaviour of eutectic and hypereutectic Al–Si–Cu–Mg casting alloys tested against a composite brake pad

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Abstract

The wear resistance of nine eutectic and hypereutectic Al–Si alloys with various compositions and processing routes was studied. Testing was conducted using a pin-on-disc configuration and two different testing speeds. At lower disc speed, the influence of the composition and alloy processing was very strong and a severe wear transition was observed for the alloys with low fractions of primary silicon particles. At higher disc speed, wear of all the alloys but one was very low. The high wear alloy had a high silicon content and a very large primary silicon size; the accelerated wear was caused by cracking and fracture of the large silicon particles upon impact with the abrasive disc material at the higher speed.

Keywords: Aluminium–silicon; Wear; Microstructure; Processing

1. Introduction

The high strength, low thermal expansion and high wear resistance together with excellent castability and reduced density make eutectic and hypereutectic Al–Si alloys very attractive for the transport industry. Among the most common applications are components, such as connecting rods, cylinder liners, pistons, engine blocks and air conditioner compressors [1–3].

Most of these components capitalize on the high wear resistance of these alloys, especially when hypereutectic alloys are used. The high wear resistance is mainly attributed to the presence of hard primary silicon particles distributed in the matrix. The amount of these particles becomes significant above the equilibrium eutectic composition (∼12.5% silicon [4]).

The role of the silicon content in Al–Si alloys has been investigated intensely by several researchers. Although increasing the silicon content of the alloy has usually been found to improve the wear resistance [5,6], under certain experimental conditions, high silicon content alloys have performed worse than their low silicon counterparts [7].

Furthermore, under low applied loads, associated with low coefficients of wear, it has been shown that wear resistance is not a strong function of Si content [8,9].

On the other hand, the distribution and size of the primary silicon particles, rather than the overall silicon content of the alloy, has been identified as the key parameter by some authors [10,11]. Finer particle sizes were usually found to improve wear resistance [12,13]. Then again, under different test conditions, alloys with equal silicon contents but with coarser silicon particles demonstrated wear resistance similar to alloys with finer silicon particles [14]. Besides the silicon content or the distribution of the primary silicon particles, the heat treatment state and alloying element additions are also known to significantly affect wear [6,15].

Finally, the strong influence that the test parameters have on the wear mechanisms and consequently, on the wear resistance of the alloys, make it impossible to determine the advantage of choosing one alloy over another one, unless the service conditions of the part are known [16–18].

2. Experimental procedure

The chemical compositions of the alloys together with their hardness values are shown in Table 1. Alloys H1, H2 and HS were cast in permanent metallic moulds, alloys
Table 1
Chemical composition (wt.%) and hardness of the alloys

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Processing route</th>
<th>Chemical composition (wt.%)</th>
<th>Hardness (HRB)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Si</td>
<td>Mg</td>
</tr>
<tr>
<td>H1</td>
<td>Ingot metallurgy</td>
<td>12.85</td>
<td>1.30</td>
</tr>
<tr>
<td>H2</td>
<td>Ingot metallurgy</td>
<td>12.29</td>
<td>1.30</td>
</tr>
<tr>
<td>HS</td>
<td>Ingot metallurgy</td>
<td>15.90</td>
<td>1.20</td>
</tr>
<tr>
<td>Thixo1</td>
<td>Thixo forming</td>
<td>15.30</td>
<td>0.58</td>
</tr>
<tr>
<td>Thixo2</td>
<td>Thixo forming</td>
<td>15.70</td>
<td>0.54</td>
</tr>
<tr>
<td>LF1</td>
<td>Lost foam</td>
<td>12.64</td>
<td>1.09</td>
</tr>
<tr>
<td>LF2</td>
<td>Lost foam</td>
<td>16.11</td>
<td>1.04</td>
</tr>
<tr>
<td>SQ1</td>
<td>Squeeze casting</td>
<td>16.20</td>
<td>1.16</td>
</tr>
<tr>
<td>SQ2</td>
<td>Squeeze casting</td>
<td>19.85</td>
<td>1.13</td>
</tr>
</tbody>
</table>

Thixo1 and Thixo2 were thixoformed, alloys SQ1 and SQ2 were produced using squeeze casting and finally, alloys LF1 and LF2 were cast using lost-foam technology. It is important to note that in all the alloys with the exception of HS, phosphorus was used as primary silicon refiner.

Dry sliding wear tests were conducted using a conventional pin-on-disc testing machine. A scheme of the testing configuration is shown in Fig. 1. The pins used (Ø = 4 mm and L = 20 mm) were machined from the Al–Si alloys mentioned above and the discs from a composite brake pad developed by Honeywell for use with aluminium alloys [19]. During the tests, the change in pin height, the coefficient of friction and the temperature at 2 mm from the contact surface were measured. The temperature was measured by inserting a thermocouple in the pins.

Although no lubricants were added, it must be remembered that braking pads invariably contain several lubricants such as MoS₂, SbS₃ as well as sulphides of Cu, Sn, Sb and brass [20].

The test distance was always 2000 m and the load applied 46.51 N. The speed of the disc was set at 0.089 or 0.356 m/s, for conditions A and B, respectively. The distance was calculated according to Eq. (1).

\[
d = \pi nt
\]

(1)

where \( n \) is the number of revolutions per second, \( r \) the distance from the centre of the disc to the centre of the pin (m) (see Fig. 1) and \( t \) is the time (s).

The pins were cleaned and degreased before the tests, using the following procedure: ultrasonic bath, first with water and soap, then with ethanol and finally with acetone. The pins were dried thoroughly and kept for 30 min in an oven at 80 °C. After each test the wear debris was collected and stored for further analysis and the disc was ground with 1200 grit abrasive to avoid roughness changes between tests. The wear debris and the tested pins and discs were analysed with X-rays and also examined under a scanning electron microscope (SEM).

The pins were weighed before and after the tests to obtain the coefficient of wear, according to Eq. (2):

\[
w = \frac{\Delta m}{L}
\]

(2)

where \( w \) is the coefficient of wear (g/Nm), \( L \) the load (N), \( \Delta m \) denotes mass loss of pin (g) and \( d \) is the distance (m).

A minimum of two samples were tested for each condition. If the difference between the two samples was greater than 10%, a third and even a fourth pin were tested.

3. Results

3.1. Microstructure

The microstructures of the alloys are shown in Fig. 2. Alloys H1 and H2 both have very similar size and distribution of the eutectic and primary silicon particles. However, due to the significantly lower copper content of alloy H1 (1.4% against 4.4%), the identity and amount of the intermetallic phases are very different in the two alloys. The composition of alloy HS is similar to that of alloy H2 but with higher silicon content (15.9% against 12.3%). In alloy HS, the primary silicon particles are very coarse and their distribution in the matrix is not uniform. Both thixoformed alloys, Thixo1 and Thixo2, exhibit a very homogeneous microstructure. The aluminium phase forms globules, the primary silicon particles are fine and homogeneously
distributed in the aluminium matrix. The eutectic silicon forms small polygonal particles instead of the needle-like network observed in the other alloys.

The microstructures of both lost-foam alloys, LF1 and LF2, are very similar. The fraction of primary silicon particles is higher in alloy LF2 due to its higher silicon content. Even though the silicon content of alloy LF1 is close to the eutectic composition, it contains an unusually high fraction of primary particles. Finally, regarding alloys SQ1 and SQ2, the high solidification rate that occurs in squeeze-casting results in a very fine microstructure.

The T6 heat treatment notably increased the hardness, especially in some of the alloys (see Table 1). The microstructures of the alloys also experienced changes upon heat treatment. Most of the intermetallic phases dissolved and the eutectic silicon particles tended to agglomerate and
spheroidize (see Fig. 3). The parameters of the thermal treatments used are shown in Table 2. Since the size and the distribution of the primary silicon particles control the wear to a great extent, the volume fraction of these particles together with the mean particle size \(D_p\) and the interparticle distance \(\lambda\) were measured for all alloys (see Table 3). The large primary silicon particle size of alloy HS contrasts with the small size and high concentration of silicon particles in both thixoformed alloys. It is also important to note the high concentration of primary silicon particles in alloy LF1 considering its relatively low silicon content.

### 3.2. Coefficient of wear

#### 3.2.1. Testing condition A (lower disc speed)

The coefficients of wear of all alloys tested at condition A (0.089 m/s) are shown in Fig. 4. Comparing the wear resistance of alloys H1 and H2, it is clear that increasing the copper content lowered the coefficient of wear. The beneficial effect of silicon is also clear. Comparing alloys that share the processing route, i.e. with very similar microstructures (LF1 with LF2 and SQ1 with SQ2), the alloys with higher silicon contents performed better. In the case of alloys H2 and HS, very different microstructures prevent a clear analysis of the effect of silicon content.

The microstructures produced by the different processing routes also affected the wear resistance. Comparing the results from alloys HS, Thixo1 and LF2, there are important differences in the wear rate of the three alloys even though their composition is nearly the same. The effect of the microstructure is also evident comparing alloys H2 and LF1.

The loss of pin height as a function of test distance is shown in Fig. 5. The wear rate of the alloys is very similar until about 500 m. Approximately at that distance, alloy

### Table 2

<table>
<thead>
<tr>
<th>Material</th>
<th>Solution heat treatment (°C)</th>
<th>Artificial ageing (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>H1</td>
<td>500 (5h)</td>
<td>180 (4h)</td>
</tr>
<tr>
<td>H2</td>
<td>500 (5h)</td>
<td>180 (4h)</td>
</tr>
<tr>
<td>HS</td>
<td>500 (5h)</td>
<td>180 (4h)</td>
</tr>
<tr>
<td>LF1</td>
<td>500 (5h)</td>
<td>180 (4h)</td>
</tr>
<tr>
<td>LF2</td>
<td>500 (5h)</td>
<td>180 (4h)</td>
</tr>
<tr>
<td>Thixo1</td>
<td>500 (5h)</td>
<td>180 (4h)</td>
</tr>
<tr>
<td>Thixo2</td>
<td>500 (5h)</td>
<td>180 (4h)</td>
</tr>
<tr>
<td>SQ1</td>
<td>500 (5h)</td>
<td>170 (6h-40 min)</td>
</tr>
<tr>
<td>SQ2</td>
<td>500 (5h)</td>
<td>170 (6h-40 min)</td>
</tr>
</tbody>
</table>

### Table 3

<table>
<thead>
<tr>
<th>Material</th>
<th>Part (mm²)</th>
<th>%Si prim</th>
<th>(D_p) (μm)</th>
<th>(\lambda) (μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>H1</td>
<td>8.4</td>
<td>0.9</td>
<td>35.3</td>
<td>3668.6</td>
</tr>
<tr>
<td>H2</td>
<td>6.0</td>
<td>1.0</td>
<td>44.2</td>
<td>5931.6</td>
</tr>
<tr>
<td>HS</td>
<td>1.5</td>
<td>2.2</td>
<td>121.0</td>
<td>5249.9</td>
</tr>
<tr>
<td>Thixo1</td>
<td>866.4</td>
<td>13.4</td>
<td>11.6</td>
<td>74.6</td>
</tr>
<tr>
<td>Thixo2</td>
<td>858.7</td>
<td>13.5</td>
<td>11.8</td>
<td>75.0</td>
</tr>
<tr>
<td>LF1</td>
<td>108.2</td>
<td>4.9</td>
<td>23.1</td>
<td>426.3</td>
</tr>
<tr>
<td>LF2</td>
<td>138.0</td>
<td>9.3</td>
<td>29.1</td>
<td>272.2</td>
</tr>
<tr>
<td>SQ1</td>
<td>336.4</td>
<td>7.2</td>
<td>15.1</td>
<td>190.8</td>
</tr>
<tr>
<td>SQ2</td>
<td>211.3</td>
<td>12.0</td>
<td>25.5</td>
<td>181.5</td>
</tr>
</tbody>
</table>

Fig. 3. Detail of the microstructure of alloy H2: (a) before and (b) after the heat treatment alloys at condition A.

Fig. 4. Average coefficient of wear of the alloys tested at condition A.
HI suffers a transition and the wear rate accelerates progressively until around 1300 m. After this distance, the wear rate stabilises and continues at a constant rate until the end of the test. The differences in overall wear coefficients can be linked to the distance at which transition takes place. In the case of both thixoformed alloys and in alloy LF2, no transition is observed.

After the thermal treatments, the wear resistance of some of the alloys, mainly alloys H1, H2, SQ1 and SQ2, is greatly improved. The wear resistance of the other alloys remains nearly unchanged. The improvement observed in the wear resistance of some of the alloys is caused by a marked delay in the transition of wear, as can be seen in Fig. 5.

In order to identify the nature of the wear transition, the wear debris as well as the pin and disc surfaces after the tests were analysed. In Fig. 6, the two kinds of particle found in the wear debris are shown. The analysis of the wear debris showed that, for alloys that did not undergo any transition, the debris was almost exclusively formed by small oxidized particles (<5 μm). The EDS and also the X-ray analysis confirmed that they consisted of aluminium and aluminium oxides together with smaller amounts of silicon fragments and also particles from the disc. In the wear debris from pins that did suffer transition in wear, along with the small particles, there were also large aluminium platelets. The amount of these platelets was higher when the transition in wear took place at early stages of the tests (especially in alloys H1 and H2).

### 3.2.2. Testing condition B (higher disc speed)

The average coefficients of wear of the alloys tested in condition B are shown in Fig. 7. At this testing condition (0.356 m/s), the coefficient of wear of alloys H1 and H2 was drastically reduced. In these alloys, the coefficient of wear was about 10 times smaller than in condition A. In both lost-foam and squeeze-cast alloys the coefficient of wear was also smaller at this test condition. No difference in the coefficient of wear was measured in the thixoformed alloys.

The behaviour of alloy HS is exceptional because the wear rate was higher at this testing condition and the coefficient of wear, in absolute terms, the highest of the nine alloys tested.

Thermal treatment reduced wear when testing at condition A by means of delaying or suppressing the transition in wear. Since no or nearly no such transition took place with the
higher disc speed of condition B, the thermal treatment did not have any effect on the coefficient of wear of the alloys. The loss in pin height with test distance is shown in Fig. 8.

3.2.3. Coefficient of friction and temperature

In Fig. 9, the curves corresponding to the evolution of the coefficient of friction and the temperature at condition B for two alloys are shown: alloy Thixo1 that suffered no transition in wear and alloy HS that suffered a transition in wear at around 1300 m. The evolution of the coefficient of friction was very similar in both testing conditions. During the first 10–15 m of the tests, it increased to a maximum and then smoothly decreased reaching a stationary situation after about 800–1000 m. If the pin did not undergo transition in wear, the coefficient of friction stayed constant until the end of the test. If the transition in wear took place, the coefficient of friction increased to a maximum and then it remained constant.

The temperature near the specimen surface during the tests was also monitored. The temperature followed the trend marked by the coefficient of friction. The maximum temperature measured was similar for all the alloys between 44 and 50 °C at condition A, and between 75 and 85 °C at condition B.

4. Discussion

4.1. Testing condition A (lower disc speed)

At the relatively slow sliding speeds and normal loads of this study, the frictional heat is rather low and an oxide layer can be formed on the surface of the aluminium pin during the tests [21]. As the test progresses, two processes take place simultaneously: the oxide layer breaks down, forming fine oxidized wear debris and at the same time the oxide layer grows, helped by the frictional heating; with time the competing processes reach an equilibrium situation.

SEM observations of the debris formed under oxidative wear have shown that it consists mainly of small equiaxed particles (<5 μm) and some laminar particles (50–200 μm) [22–24]. X-ray diffraction of the bulk debris demonstrated that the composition of this kind of debris is aluminium [25], silicon [24] and aluminium oxide [26].

The morphology and composition of the wear debris shown in Fig. 6 strongly suggest that before the transition mentioned earlier takes place, oxidative wear and mild abrasive wear are the main wear mechanisms for the nine alloys studied. The low coefficient of wear of the alloys before the transition also point to these mechanisms of wear.

Transitions in wear are often associated to changes in the applied pressure [27–29]. At low pressures, the wear process is limited to the oxide layer that forms on the contact surface of the alloy and the resultant wear is low. When the load is increased up to a critical value, different for each alloy and testing condition, the elastic deformation produces a progressive breakage of the oxide layer. This disruption of the layer enables a more intimate contact between the pin and the disc, producing a transition in wear from oxidative to adhesive, abrasive or others. As a consequence, the wear rate is higher after the transition. This is not surprising because in most cases tribological oxidation can reduce wear rate by up to two orders of magnitude [30].

It must be noted that in the two testing conditions used, the load was maintained constant and still a marked transition in wear was observed. Since the wear transition took place under constant load and speed conditions, the change
that produced the progressive acceleration of wear had to be related to changes on the contact surfaces during the tests. Those changes caused the rupture of the oxide layer that brought the aforementioned transition. The transition corresponds to a change in mechanism of wear from mainly oxidative to abrasive and delamination wear. Once the oxide layer disappears, the hard particles from the disc abrade the surface of the pins and produce subsurface cracks that grow under combined frictional shear stresses and normal adhesion stresses, producing platelets [31], like the one shown in Fig. 6b.

On the pin and disc surfaces, oxide particles and small fragments of silicon particles accumulated progressively. The progressive contamination of the wear track modifies the nature of the contact surface. The transfer layer formed on the wear track covers the lubricating agents from the disc, increasing the coefficient of friction. Also, the silicon particles get trapped on the disc surface, further scratching the oxide layer, and accelerating its deterioration.

The contamination of the wear track takes place for all the alloys in a very similar way, but as shown in Fig. 5 big differences were observed in the transition distance: some alloys suffered early transitions in wear (alloys H1 and H2) some late or very late transitions (alloys SQ1, SQ2, HS and LF1) and some did not present any transition at all.

In the literature, it has been shown that additions of some alloying elements, like copper, strengthen the oxide layer, shifting the wear transition to higher loads [32] (longer distances in our case). This fact would explain the improved wear resistance of alloy H2 compared to that of alloy H1. Copper could have a double effect: first, the stronger oxide layer would delay the transition, and once the oxide layer was disrupted, the stronger matrix and higher fraction of hard Al2Cu particles would keep the wear rate low after the transition.

The coefficient of wear is often considered to be inversely proportional to the hardness of the alloy, according to the well known Archard’s law [33]. However, looking at the results for the coefficient of wear shown in Fig. 4, it is clear that hardness is not the only parameter affecting wear. For example, the hardness of alloys H1 and LF2 as-cast is nearly the same but the coefficient of wear for alloy LF2 is approximately 10 times smaller than that of alloy H1.

During the first part of the tests, all the alloys had approximately the same wear rate, and the wear curves of the alloys had a very similar slope, with the oxide layer dictating the rate of wear as shown in Fig. 5. Then the disruption of the oxide layer brought the transition in wear and the change in the slope of the wear curves, resulting in a different value for each alloy. The wear loss (µm) of the alloys after the transition as measured from the slope of the wear curves could be predicted according to a modified version of an equation proposed by Long et al. [5]:

\[ W^* = \frac{P}{HV D_p \lambda} \]

where \( W^* \) is the wear loss (µm), \( P \) the pressure (N/mm²), \( D_p \) the average primary particle size (µm), \( K \) the constant related to roughness (µm), \( HV \) the hardness (N/mm²) and \( \lambda \) is the interparticle distance (µm).

The correlation between the measured and calculated wear loss according to Eq. (3) is shown in Fig. 10. The value for the constant \( K \) was 2.5 µm. However, the equation does not take into account the role of the hard intermetallic phases, especially Al2Cu, that also help to increase the wear resistance. The hardness of these particles is high (360-780 HV [34]) and the volume fraction in some of the alloys is also high (see Table 4). In the case of alloys with low primary silicon contents, these particles may be very important. This is the case of alloy H2 for which Eq. (3) predicts a higher wear rate than the measured one. The measured wear rate for this alloy is lower because the hard Al2Cu particles help the few primary silicon particles in protecting the bulk of the material from abrasion.

In Fig. 11, the surfaces of two pins, one of alloy H1 and one of alloy LF1, after the wear test are shown. In alloy H1, with a very low silicon particle concentration, none or very few primary silicon particles are seen in the surface of the pins whilst in alloy LF1, these particles are clearly sticking out from the matrix. In both alloys, white Al2Cu particles

![Fig. 10. Relationship between wear loss measured from the slope of the wear curve after the transition and calculated wear loss.](Image 347x609 to 587x770)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>H1</th>
<th>H2</th>
<th>HS</th>
<th>Thixo1</th>
<th>Thixo2</th>
<th>LF1</th>
<th>LF2</th>
<th>SQ1</th>
<th>SQ2</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al2Cu%</td>
<td>0.1</td>
<td>3.6</td>
<td>4.0</td>
<td>4.7</td>
<td>4.2</td>
<td>4.5</td>
<td>4.4</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>Al2Cu%</td>
<td>4.2</td>
<td>1.5</td>
<td>1.5</td>
<td>1.7</td>
<td>1.3</td>
<td>1.9</td>
<td>1.8</td>
<td>–</td>
<td>–</td>
</tr>
</tbody>
</table>

Table 4 Volumetric fraction of Al2Cu particles in the alloys as-cast and after T6 heat treatment.
can be also seen sticking out from the matrix. In the alloys with high primary silicon contents, the oxide layer is also disrupted, but when this happens, the soft matrix wears preferentially, leaving the hard silicon particles to carry the load. In consequence, the wear rate of these alloys is mainly governed by the wear of the hard silicon particles. Considering this type of behaviour, it is logical that the wear rate in alloys with few primary silicon particles such as H1 and H2 is higher; the lack of a high density of primary silicon particles, forces the softer matrix to support the load and the wear rate becomes higher. In the case of alloy HS, despite the low concentration of primary silicon particles, the wear rate is low because the few silicon particles that are present (~15–20 on the pin surface) are able to carry effectively the load due to their large size ($D_p = 121\,\mu m$).

In all the alloys, many of the primary silicon particles, especially the biggest ones, are damaged during the wear test as can be seen in Fig. 12. The damage on the particles was probably caused by their collision with abrasive hard particles from the disc.

From the above explanations, the outstanding wear resistance of both thixoformed alloys and lost-foam alloy LF2 can be explained as follows: the high copper content helped the formation of a strong oxide layer. If the oxide layer was disrupted, the high concentration of primary silicon particles and their homogeneous distribution limited the extent of wear. Moreover, the small size of the silicon particles made them tougher and consequently much harder to break, limiting the abrasion of the particles by microcracking [35,36].

The thermal treatment mainly improved the wear resistance of the alloys in which the role played by the silicon particles was smaller (H1 and H2). The thermal treatment produced a harder matrix, able to withstand better the erosion from the disc. In the rest of the alloys, the effect was much smaller. The wear was limited by the low wear rate of the primary silicon particles and hardness increments of the matrix did not bring considerable changes.

4.2. Testing condition B (higher disc speed)

The minimum coefficient of friction for all alloys (before the transition) stabilized at around 0.45–0.50 at condition A, whilst at condition B the coefficient of friction was between 0.32 and 0.40.

The lower coefficient of friction was likely caused by the softening of the lubricating agents from the disc due to the higher temperature at the contact surface, thereby causing a more extensive lubrication. Similar effects of lower coefficients of friction under lubricated conditions with higher contact temperatures have been reported by other authors [37].

Looking at the results for the coefficient of wear shown in Fig. 7, it is clear that, except for alloy HS, this coefficient was very low and similar for all the alloys. In contrast with the results from condition A, no effect of microstructure or alloy composition on the coefficient of wear was observed. Moreover, the thermal treatment had no effect on the wear of the alloys.

Analyzing the wear debris, the big aluminium platelets described earlier were only found for alloy HS. In the rest of the alloys only oxidized micrometric particles were found.
The wear mechanism during this test condition was considered as purely oxidative (except for alloy HS). The more effective lubrication helped to preserve the integrity of the oxide layer and the higher temperature at the contact surface increased the oxidation kinetics favouring the formation of a fresh oxide layer during the wear tests. The wear was limited to the thickness of the oxide layer, resulting in equally low wear rates for all the alloys. If the oxide layer had been disrupted, intimate contact between the abrasive particles from the disc and the pin surface would have taken place and consequently, big differences in the coefficient of wear of the alloys would have been observed, as was previously explained for condition A.

The pins from alloy HS showed extensive damage on the primary silicon particles. When the abrasive particles from the disc impacted against the big primary silicon particles from alloy HS, the silicon particles cracked very easily. The cracks that formed propagated into the aluminium matrix, accelerating the weight loss of the pin by delamination. Also, some of the silicon fragments got trapped on the surface of the disc, increasing the chances for new impacts to happen and accelerating wear. A similar mechanism of wear has been observed in the wear of Al-Si alloys tested against a composite brake disc [38].

5. Conclusions

The results obtained show that the speed of the disc affected the wear of Al-Si alloys even when the load applied was maintained constant.

- At the lower disc speed, the wear rate for all the alloys was low and very similar. As the test went on, a wear transition was observed in most of the alloys. This transition was attributed to a progressive break down of the oxide layer. The alloys with high concentrations of silicon particles continued to wear with a slow rate even after the rupture of the oxide layer because the wear was limited by the wear rate of the hard silicon particles. The alloys with low fractions of silicon particles suffered severe wear because the softer aluminium matrix was responsible for carrying the load. The T6 thermal treatment specially improved the wear of these last alloys because the treatment increased the hardness of the matrix.

- At the higher disc speed, all the alloys but one showed very similar coefficients of wear. The coefficient of wear was very low and independent of alloy composition or processing route. The higher temperature at the contact surface caused by the higher disc speed improved the efficiency of the lubrication, thereby reducing the coefficient of friction and protecting the oxide layer. The only alloy that presented higher wear with higher disc speed had exceptionally large primary silicon particles. The abrasive particles from the disc impacted against those large silicon particles, cracking them and accelerating wear. The thermal treatment had absolutely no effect on the wear behaviour at this disc speed.

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