Effect of die geometry on thermal fatigue of tool steel in aluminium alloy die-casting

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Abstract. In this paper, die geometry and its effect on the thermal wear of H13 steel was evaluated during aluminium die-casting. During the investigations, an innovative dipping testing device was used. The process involved a cyclic aluminium alloy-melting die and water-cooling process, which allows a dominated cycle of thermal fatigue. The H13 tool steel was first prepared with different end geometries before being subjected to the cyclic heating and cooling (in a water basin) processes. During the heating and cooling processes, the produced hardness profile, surface cracks, and the microstructure of the samples were regularly analysed after a predetermined number of cycles. The effect of the sample end geometry, sample thickness, material of the sample, and the dipping test parameters on the thermal stress was also investigated. From the metallographic investigation of the sample surface, the thermal fatigue resistance of the sample was observed to be enhanced due to the improved die steel protection from oxidation. Oxides can permeate cracks at the corrosion pits and cause the propagation of cracks through the production of a tensile stress from the increased oxide volume at the crack tip. The improved thermal fatigue resistance of the tool steel was contributed by the ductility and high yield strength of the tool steel.

1. Introduction
The production of more than a hundred thousand castings per die for alloy die-casting die in high-pressure aluminum is a general series. The geometry of such castings is complicated and made with close dimension limits, and also consists of high surface conditions [1, 2]. The service life of tools determines the production cost for the material equally influences castings and this tool service life used the design of the instrument, the heating process technology, and the process parameters. During Aluminum die-casting at a temperature range of 670-710 °C and 30-100 m/s, aluminum is melted and injected into the mold at an injection pressure of about 50 to 80 Mpa [3-6]. Several factors contribute to the low in-service tool life, such as the hot fatigue, which results to die surface heat inspection, corrosion, and soldering due to the oxidation of aluminum on the die surface, melt flow-related corrosion, heat shock-related faults, and mechanical property differences due to the heating of the material [1, 2, 7]. The stresses are attributable to the die casting process, i.e. thermos-mechanical stress. These stresses must be maintained at a low level to ensure a longer tool service life. However, such stresses can be reduced if the mold design process is optimized and a perfect heat treatment process is employed. Then, such die can be resistant to high heat shocks, asymmetrical pressure, and local material overheating which can cause serious die failure. Die-casting dies are subjected to heating and cooling cycles, exposing them to heat stress, which can alter their quantity and orientation, and consequently result in heat fatigue. The formation of surface cracks and its propagation with increasing number of cycles results in unacceptable surface casting outcomes. Surface cracking is mainly dependent on the capability of the die material to resist fatigue and this relies on the die casting
parameters and the material. Resistance to heat fatigue is inversely related to the process temperature and offers strict heat gradients to die casting parameters. Similarly, materials with medium heat resistance also have decreased heat fatigue resistance [8, 9]. In this study, the results of the finite element calculation are shown as a portion of the general analysis, heat fatigue resistance design, as well as a variety of materials with varying geometries [10-13]. An immersion experiment equipment was developed in this study for the simulation of the important aluminum alloy die-casting parameters, and for sample surface heat fatigue crack generation [3, 5, 14-19]. The specimens were exposed to cycles of heating in the molten aluminum alloy (Al 356) bath and water-cooling in a water bath. This process generates a large temperature variation within the sample and helps to control heat fatigue. The temperature of the material at dissimilar locations was detected using pre-installed thermocouples. Additionally, the stress analysis of the casting process showed effects of molten aluminum alloy temperature and the materials’ edge design on the service life of the material.

2. Experimental
The experimental work comprised of measuring the temperature at various testing areas using the ring immersion tests. Holes were made on the testing samples through which thermocouples were made attached for the temperature measurements. The temperature was determined using an advanced checking procedure and in accordance to the LabView code.

2.1 Dipping test
The fabricated dipping testing equipment used in this study to detect the materials’ heat fatigue at situations similar to the aluminium alloy die casting process. The specimens were examined as follows: dipped in a basin of molten aluminium (temperature of 700-800 °C) for 10 secs; successively cooled in a water bath (temperature of 32 °C). The fatigue loading was completed by subjecting the samples to a cyclic process of moving from a basin of melted aluminium to air (40 °C) for 6 secs and then to an emulsified water bath at 32 °C for 10 secs. The presence of oil in the water bath protects the sample surface and prevents aluminium from sticking on it. The samples used for heat fatigue evaluation were not exposed to pressure and aluminium flow. During the experiments, the sample movements were obtained with 2 PC-controlled air cylinders.

2.2 Test specimens
This study mainly aims to critically examine different steels that have been exposed to different surface and heat treatments in terms of their mechanical and metallurgical properties for utilization during pressure die casting processes. The study also aims to carry out a thermal fatigue test on these materials in order to find the relationship between the basic mechanical characteristics of the material, toughness, and hardness, and the materials’ thermal fatigue resistance. The heat fatigue analysis lasted for only 32 secs as the samples were cooled in water (32 °C) for 10 secs to prevent aluminium from sticking to the specimens as well as to generate heat gradients in the samples during the cooling process. The treatment cycle continues with passing the samples through the air (40 °C) for 6 secs into the molten aluminium alloy bath (about 700 °C). After about 10 secs in the molten aluminium bath, the samples are further passed through the air for 6 secs before returning them to the water bath. The cyclic movement of the samples from the molten aluminium to the air and to the water bath ensured the development of sample heat fatigue. The determined temperature cycles in the samples were used in comparing the measured heat gradient intensity at some critical points with the pressure die-casting, as well as to compute the specimens’ thermal stresses during a testing cycle.

2.3 Preparation of the samples
H13 steel was used in this study; its chemical composition is provided in Table 1. The sample was machined to geometry of outer diameter 33 mm and inner diameters of 20, 15, 10 and 55 mm using CNC milling and turning machines. The length was machined to fit the fatigue-testing machine as depicted in Figure 1. Table 2 presents the mechanical properties of the sample.
An oxide layer was generated on the sample’s surface during the experiment. The oxide layer partly covered the damage features on the sample’s surface; therefore, there is a need to employ a chemical cleaning step to ensure the removal of the samples’ outer layer without generating surface aggressions. To clean the sample’s surface, the samples were first dipped in a boiling saturated NaOH solution and allowed for 600 sec. The duration of this step strictly depends on the required level of cleaning. Next, the samples were immersed in 10 % HCl solution for 10 secs before being boiled in distilled water for another 60 secs. The last two immersion steps were aimed at terminating the reaction of NaOH as well as to remove chlorides from the surface of the sample. The surface of the samples was periodical examined using an optical and an electron microscope. After the immersion cycles, the last segment of the specimen (area clamped on the equipment) was cut off to get the actual sample for the testing processes. At last, the transverse sections of all the samples were examined and assessed for damaging features in terms of the microhardness, penetration depth, and microstructural amendments.

3. Results and discussion

3.1 Microstructure analysis
During Aluminium die-casting, some of the die parts were exposed to high temperatures and as such, undergo various levels of stress. Typically, the surface wall of samples undergoes such conditions; heat transfer in this section is in two dimensions and the amount of absorbed energy by the material is greater than the average energy absorbed by the rest of the die. They are usually the first part to fail.
There is a need to create special processes for rapid heat extraction from the surfaces, for energy dissipation within the material, and for the transfer of energy to a heat conveyer such as a cooling line. A microstructural evaluation of the oxidized specimens (Figure 2) showed that in every instance, there is an air gap or an oxide film which restricted the die steel from having a direct contact with the Aluminum. Another defect during Aluminium die-casting is impingement soldering which is just a mechanical withholding of the Aluminum by tiny scratches or pits. The appearance of pits, due to impingement soldering is different from heat checks; they occur when the semi-protective oxide surface, which developed on a die, is broken. Upon this breakage, the steel is deoxidized and upon repeated cycles, a broad shallow pit will develop [7].

![Figure 2. H13 tool steel microstructure a) before the cycle, b) 5mm, c) 10 mm, and d) 20 mm from bottom after thermal fatigue immersion of 5000 cycle sample](image)

From the SEM, two observations are noted for the oxidized samples (Figure 3); the molten phase of the surface in a graphite crucible and the cooling bath which consisted of water. A cycle duration of 32 secs was maintained, with an immersion time of 10 secs in both molten aluminum alloy and the cooling bath. The immersion time used in this experiment is dependent on the experimental setup; two holes were drilled on the sample wall at distances of 20 mm and 40 mm from the exterior surface as shown in Figure 3. Thermocouples were mounted in these holes for temperature measurement at different immersion times. The temperature data were acquired using a Data Acquisition program controlled from a personal computer. The last 10 sec of immersion was selected since the heating and cooling cycles underwent by the samples in this method was observed in the temperature range of 200 - 520 °C. A typically die casting cycle is monitored in real Teksid production plants in the temperature range of 200 - 480 °C on the external surface of the die. However, the exploited experimental conditions decide the temperature limit to be undergone by the steel layer on the surface. This means that the sample surface has to undergo serious deterioration cycles in order to accelerate the degradation mechanisms.

In this study, the experiments were carried out for 2000 cycles and during the experiments, an oxide layer was formed on the sample surface which partly covered the damage features. Thus, a cleaning stage is required to remove the oxide layer without producing surface aggression. A chemical
cleaning process was adopted in which the sample was first immersed in a boiling saturated NaOH solution for 60 secs before immersion in 10 % HCl solution for 60 secs and boiling distilled water for another 60 secs. The last two steps were performed to terminate the reaction of NaOH and to get rid of chlorides from the surface of the sample. The surface of the samples was periodically inspected using an electron and optical microscope. After the experiments, the last segment of the specimen (area clamped to the equipment) was cut off to get the required sample for the testing processes. Finally, the samples were sectioned transversely and analyzed for damage features in terms of microstructural alterations, penetration depth, and microhardness. The layers are made up of a top 10 um thickness layer and a diffusion layer of 100 um thickness. In fact, the upper layer has a double structure formed by the iron oxide layer of 2 um thickness (Figure 3b). This layer developed into a multi-hole structure to plunge the surface pores and produce a huge mechanical keying with the potential composite layer.

Figure 3. SEM/EBS morphologies of the oxide samples, a) grain size formation, b) mechanism of oxide formation.

The microhardness profiles of the samples at different heating and cooling conditions are discussed below; the samples were also compared before and after the cyclic immersion in the molten aluminum bath. Upon observation, the first damage signs on the oxidized samples were observed after 2000 cycles. There was a formation of localized soldering points which eventually caused the stacking of the cast metal to these points and the subsequent removal of parts of the steel, resulting in surface cavity formation (Figure 4 a and b). Heat checks were clearly noted while the nitride + oxidized samples presented few but longer corner cracks.

Figure 4. The detected damages on the samples’ surface after 1,850 cycles with different magnification power.
Figure 5 shows the propagation and initiation of heat checks on the surface of the cylinder. The initialization of the damage mechanism was observed after about 2000 cycles and the density of these damages increased in the lateral dimensions as the number of cycles increased. The heat checks observed on the oxidized samples resembled those on the heat-treated samples [20]. Few long cracks were only noted in the nitride + oxidized samples likely because of the high level of residual stress resulting from the surface treatment, which may have favored the localization of stress along the edges. Furthermore, the top layer of the composite produced by the surface treatment prevented aluminum alloy from soldering on the steel.

![Image](image_url)

(a) Crack initiation and propagation in both directions in the H13.

After several cycles, there is a formation of an oxide coat on the top edge and along the specimens’ sides, which formed a crack network in the brittle scale (Figure 6). The oxidation of the metal surface beneath these cracks is more rapid compared to the positions covered by scale; hence, the same pattern is produced on the surface of the metal after further cycles. The sharp sample edge served as a stress inducer in addition to the existing temperature difference towards the center from the outer edge. Hence, the fatigue cracks were observed to initiate from the outer edge and propagate inwards. The presence of a single crack as depicted in Figure 5a has a close resemblance with an actual heat check in a die as earlier depicted in Figure 5b. This is a clear indication that the thermal cycling approach simulated the development of real fatigue cracks in die-casting. Further cycling results in the inward propagation of these cracks, making them broader at the edge and encouraging the development of more cracks.

The EDS analysis of the carbides near the surface of the H13 sample after 2000 cycles showed the abundance of Cr in the largest carbides in the samples’ microstructure while the smaller carbides are less in Cr content (Figures 6). The diffusion of Cr into ferrite is more rapid compared to most metal alloys. This results in the detection of chromium steels (Cr7C3) at a temperature of 500 °C, as shown in Tables 3 and 4 [5, 21]. A continuous softening in chromium steel normally occurs during tempering at a temperature range of 500 – 700 °C. Previous studies have also shown the formation of iron-rich chromium carbide (CrFe)7C3) during tempering of H13 steel at 700 °C [5, 20].
This prevented the formation of the surface cavities, which were observed in the heat-treated samples. Based on the earlier observations after 2000 cycles, there was a localized oxide exfoliation as shown in Figure 7, which resulted in a sustained breakage of the barrier to melt sticking. This effect was noted to be dramatically enhanced between 2000 and 3500 cycles, resulting in the formation of surface cavities, which are similar to those, observed on the samples. Furthermore, it could be observed that from 2000 to 3500 cycles, there was a change in the damage mechanism. The cleaning step which is important for the observation of the specimens’ surface, was easy for specimens exposed to 3,500 and 6000 cycles but difficult for specimens exposed to 6000 cycles. This could be attributed to the large volume of soldering points, which were formed in this prolonged test step. There is a need to conduct an analysis for a better evaluation of the actual threshold of the number of cycles prior to the development of soldering effect on the nitride-oxidized samples.

**Figure 6.** The EDS elemental analysis of structures near the surface of H13 after 2000 cycles

**Table 3.** EDS-identified element within and around the crack

| Element | Are(counts) | APP.CONC | % weight | % weight σ | Atomic % |
|---------|-------------|----------|----------|------------|----------|
| C       | 317.0       | 1.2790   | 15.3330  | 1.3870     | 42.5680  |
| O       | 473.0       | 1.6310   | 10.4150  | 1.2200     | 21.7060  |
| Si      | 205.0       | 0.1520   | 1.5110   | 0.2740     | 1.7930   |
| Cr      | 246.0       | 0.4670   | 3.0680   | 0.4490     | 1.9670   |
| Fe      | 2270.0      | 6.2370   | 44.7940  | 1.7020     | 26.7450  |

**Table 4.** EDS-identified element within and around the formed crack and carbide

| Element | Are (counts) | APP.CONC | % weight | % weight σ | Atomic % |
|---------|-------------|----------|----------|------------|----------|
| C       | 361.0       | 1.4550   | 16.5490  | 1.3370     | 40.3780  |
| O       | 795.0       | 2.7420   | 17.1380  | 1.3150     | 31.3930  |
| Si      | 146.0       | 0.1080   | 1.0430   | 0.2510     | 1.0880   |
| Cr      | 193.0       | 0.3660   | 2.4440   | 0.4340     | 1.3780   |
| Fe      | 2069.0      | 5.6860   | 41.6530  | 1.6210     | 21.8580  |
3.2 Microhardness Measurements
A Leitz microhardness-testing unit was used to measure the microhardness values diagonally from the transversely sectioned samples and from the longitudinal section toward the base metal using a 500 g load. The first indentation was very close to the sample edge while the rest of the readings were taken at 50 um increments. There was a significant softening of the samples’ edge, as the hardness value was RC 46.5 before and after cycling. These specimens exhibited a poor thermal fatigue resistance due to the presence of a brittle surface layer in the region from the surface, which reduced the hardness values from the reverted austenite in the steel. Figure 8 presents the microhardness values on each thermal fatigue specimen.

Figure 7. (a, b, c) The appearance of cavity on the samples’ surface after 1850 cycles
The formation of this austenite can be due to the high-temperature region of the thermal fatigue cycling. The thermal cycling process could not soften the hardened areas because of the lower temperatures away from the edge, which cannot match the high temperature at the edge. The microhardness at some location is high but such locations are brittle as well. However, there were no cracks and the observed high hardness did not cause any apparent reduction in the thermal fatigue resistance. This increased hardness may have a role in the enhanced thermal fatigue resistance of the stress-free samples compared to the coated samples. Meanwhile, the observed hardness value of RC 36 is still less than the initial hardness of stress-free H-13 dies steel.

4. Conclusions

1) Steel softening is the most significant cracking initiation index especially for configurations without severe stress concentrators. This has been the case for less thermal fatigue damages when the conditions encouraged lower surface temperature, which maintained the strength and hardness of the sample. A high yield strength value indicates a higher materials’ resistance to plastic deformation, but elevated surface temperatures can result in a deeper material softening. It seems that a condition for thermal fatigue cracking damage propagation is a reduction in strength before the crack front.
2) The maximum temperature during die-casting usually occurs in the wall sections where the capacity of the material to absorb and transfer heat from the surface differs. Similarly, a high temperature and a prolonged resident time are necessary for the similarity when die casting large components, especially when the die is exposed to high temperatures over prolonged periods. The results of the experiments showed the need to reduce cracking when the cooling line is located near the surface. The results further indicate a temperature threshold beyond which the thermal fatigue damage is less. The location of a cooling line near the surface will cause a shift in the maximum temperature to lower values, and still maintains the level of stress value. Meanwhile, a reduction of the maximum surface temperature by positioning the cooling lines too proximal to the surface may be hindered by the high hoop stress level generated at the cooling line.

3) Softening is minimal in the presence of strong carbide-formers such as molybdenum, vanadium, and chromium as they will preserve a fine distribution of carbides. The most susceptible to growth among the 3 elements is chromium-rich carbide.

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