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FATIGUE BEHAVIOR IN AN ALUMINUM CASTING ALLOY (A356.2): EFFECTS OF SOME DEFECTS, SDAS, HIPPING AND STRONTIUM MODIFICATION

by

Bin Zhang

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As members of the Final Examination Committee, we certify that we have read the dissertation prepared by Bin Zhang entitled Fatigue Behavior in an Aluminum Casting Alloy (A356.2): Effects of Some Defects, SDAS, Hipping and Strontium Modification and recommend that it be accepted as fulfilling the dissertation requirement for the Degree of Doctor of Philosophy.

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STATEMENT BY AUTHOR

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SIGNED: Bin Zhang
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ABSTRACT

Effects of pore, secondary dendrite arm spacings (SDAS), hot isostatic pressing (Hipping), and strontium-modification on fatigue behavior were studied in an aluminum casting alloy (A356.2). Microstructures were revealed by X-ray radiography, light microscopy and scanning electron microscopy. Small-cracks were monitored by taking replicas of the surfaces with which the cracks intersected.

As the SDAS increases from 15 to 55 μm, fatigue life decreases by a factor of 3 in low-cycle fatigue, and 100 in high-cycle fatigue. When SDAS is less than 30 μm, the pore size is below a critical size of ~ 80 μm and large eutectic constituents initiate cracks; and the initiation life is as high as 70 % of the fatigue life. As the SDAS increases beyond 30 μm, pores are the main crack-initiation sites; the initiation life is as low as 5 % of the fatigue life. Near-surface oxides initiate the fatigue crack regardless of SDAS. When crack initiated at pore and oxides, fatigue life is well correlated with the size of the initiation site and the effect of SDAS is overshadowed by the effect of pore.

Non-hipped A356.2 without Sr shows better fatigue life and the deleterious effect of pores overshadowed the beneficial effect that Sr-modification might have had. Hipping significantly increased the initiation life and small-crack propagation life of A356.2 with Sr as a result of the elimination of the porosity. However, hipping did not significantly improve the fatigue life of A356.2 without Sr. After hipping, Sr-modification is
beneficial in improving the crack initiation life, and increasing both small-crack and long-crack propagation lives.

Fracture mechanics models (Newman-Raju, and Trantina-Barishpolsky models) yielded similar results on the crack-propagation rate against the effective stress-intensity factor range. In the micro-mechanics model, the theory of continuously distributed dislocations was applied to represent crack and crack-tip plastic zone, and the propagation rate was related to the length of the crack-tip plastic zone. When the grain size is used as the characteristic length of the microstructures, the model predicts the oscillations of the propagation rates and the predicted rates agreed reasonably well with those from experiments.
CHAPTER 1. INTRODUCTION

It is well known that castings offer substantial cost savings in manufacturing, and aluminum castings can result in substantial weight reduction and enhanced fuel efficiency in automobiles. In automotive applications, aluminum castings are used as wheels, control arms, knuckles and spindles, brake calipers, cross members, and differential carriers (Nath, 1995). Many of the potential applications of aluminum castings are fatigue critical. Even more aluminum castings for structural applications would be used if the fatigue properties were improved and could be predicted.

The quality and reliability of aluminum castings are known to be highly process-dependent and microstructure-dependent. Processes such as melt-degassing, fluxing, and filtration improve the melt quality and the subsequent solidification microstructures. However, despite the urgent need in practical applications, fatigue crack-initiation and consequent propagation as small cracks are not well understood in aluminum castings. Hence, it is of paramount importance to understand the influence of microstructure on the fatigue process in aluminum castings.

In automotive applications, one widely used casting aluminum alloy is A356.2. Mechanical properties of A356.2 (Al-7Si-0.3Mg) castings have been evaluated and related to microstructural features since the late 1940s. The microstructural features include: secondary dendrite arm spacing ($SDAS$) (Spear and Gardner, 1963; Bailey, 1965;
Flemings, 1974; Radhakrishna et al., 1980; Oswalt and Misra, 1981; Meyers et al., 1983; Wickberg et al., 1984; Campbell, 1991; Boileau et al. 1997); microporosity (Closset and Gruzleski, 1982; Eady and Smith, 1986; Surappa et al., 1986; Samuel and Samuel, 1995); intermetallics (Gustafsson et al., 1986), eutectic silicon particles (Caceres and Griffiths, 1996; Wang and Caceres, 1998); and heat-treatment (Zhang and Zheng, 1996; Caceres and Wang, 1996; Schneider and Feikus, 1998; Pedersen and Arnberg, 2001).

The study of the relationship between fatigue properties and microstructures in A356 can be dated back to 1950s (Bailey, 1965), but it has been only recently that there has been a concerted impetus to use more aluminum castings in automobiles. Hence, the fatigue properties of cast aluminum alloys have not been measured and studied to the extent of the multitude of studies on tensile properties. From the point of view of fatigue life, crack-propagation dominates the low-cycle fatigue (LCF) life, whereas crack-initiation dominates high-cycle fatigue (HCF) life. Both are mainly controlled by the microstructure of materials. Although some research on the fatigue behavior of aluminum casting has established a corresponding understanding, research is still needed about the details of microstructural effects on small-crack initiation and propagation in A356.2.

Due to the widespread application of A356.2, the focus of this research has been on the fatigue behavior of A356.2 at constant amplitude loading in laboratory air environment. Other aspects such as the effect of the environment (vacuum, temperature, humidity, and corrosion), surface condition, residual stress, load interaction (variable amplitude
loading) are not considered in this dissertation. Therefore, this dissertation is a study of the effect of microstructure on the fatigue behavior of aluminum alloy castings through both experiments and computer simulations. Experimental research on microstructural effects on fatigue-crack initiation and small-crack propagation in A356.2 cast aluminum alloy has been conducted and related to the microstructures by computer calculations based on a micromechanics-model of the alloy.
CHAPTER 2. BACKGROUND

It is generally accepted that the fatigue process consists mainly of four important stages: microcrack initiation, microcrack propagation and coalescence, macrocrack propagation, and final fast fracture. In the processes of crack initiation and small-crack propagation, microstructures are very important. A review of researches on microstructural effects in fatigue related with cast aluminum alloy is made. The review described here is split into three main categories. The first subsection deals with the microstructural effects on crack initiation. The second is on microstructural effects on crack propagation including long- and small-cracks. The third subsection describes the models used in the prediction of fatigue lives. Finally a subsection regarding the proposed research in this dissertation is presented.

2.1. MICROSTRUCTURAL EFFECTS ON CRACK-INITIATION

Many studies have been conducted to understand the effects of microstructures on the fatigue crack initiation process in ductile solids. Fatigue-crack initiation is a localized process due to the heterogeneous nature of the microplastic strain of various microstructural inhomogeneities. It was found that cracks usually initiated along persistent slip bands (PSBs) (Thompson et al., 1956; Suresh, 1998), at either broken or debonded constituent particles (Lankford and Kusenberger, 1973; Morris, 1978; Kung and Fine, 1979), and grain boundary separations (Pederson, 1990). The PSBs form due to
slip heterogeneity within a given grain, where the slip system is favorably oriented with respect to loading axis. Microstructural features such as precipitate structure, grain size and morphology, and dispersoids influence the formations of PSBs. Fatigue-crack initiation may occur by particle fracture, interface failure, and slip bands emanating from the particles. Whether the particles initiate the cracks depends on the size and alignment of the constituent particles, and the maximum load applied. When coarse grains are favorably oriented to allow to accumulation of strain, the cracks initiate along grain boundaries.

A356.2 is a hypoeutectic aluminum-silicon alloy with a microstructure composed mainly of primary aluminum dendrites and eutectic constituents. The important microstructural details of the as-cast state that are related to mechanical behavior are porosity, microsegregation, inclusions, intermetallic particles, and grain size. The size, distribution and total amount of these constituents depend significantly on the melt quality and casting conditions. Reducing the secondary dendrite arm spacings (SDAS) by increasing the solidification rate improves mechanical properties more than the direct benefit of grain refining in aluminum casting alloys (Flemings, 1974; Campbell, 1991).

2.1.1. Pores and Oxide Inclusions

Although advances in melt degassing and fluxing, filtration, and casting design procedures have improved the ability to produce high quality aluminum castings, porosity and oxide inclusions are still the most nagging metallurgical aspects pertaining to the
manufacture of structural castings. Both porosity and oxides in the casting impair the ductility (elongation) of the materials (Eady and Smith, 1986; Campbell, 1991; Caceres et al., 1996) and thus the fatigue strength.

In hypoeutectic Al-Si casting alloys, fatigue cracks almost always initiate from shrinkage pores at or close to the specimen surface (Inguanti, 1985; Couper et al., 1990; Ting and Lawrence, 1993; Reinhart, 1996), and poor fatigue performance of aluminum castings is the result of both a short initiation period (Pitcher and Forsyth, 1982) and propagation period. Porosity formation in cast aluminum alloys has been widely discussed, and factors such as the decreased solubility of hydrogen in solid compared to that in the liquid, the volume shrinkage of the interdendritic liquid in the process of solidification, and the inclusion content influence the formation of porosity. Many experimental works on the physics of pore formation and its modeling have been conducted in cast aluminum alloy (Piwonka and Flemings, 1966; Campbell, 1969; Fridriksson and Svensson, 1976; Poirier et al., 1987; Fang and Granger, 1989; Tynelius et al., 1993; Mohanty et al., 1993; Sigworth, et al., 1994; Dale et al., 1998; Poirier, 1998; Easton and StJohn, 2000; Felicelli et al., 2000, Lee et al., 2001). Since pores are recognized as features found in microstructures of castings, pores are distinguished from cracks, which form after the material has been subjected to cyclic loading.

It was found that the interdendritic shrinkage porosity in the specimens almost always initiated the failure fatigue cracks, and porosity affected the fatigue life to a greater extent
than did the variations in heat-treatment (Couper et al., 1990). As the “degree of pore” (both the maximum size and the amount of the pore) increases, fatigue strength decreases (Sonsino and Ziese, 1993; Stanzl-Tschegg et al., 1993). Large surface defects (~90 μm) on the cast surface caused a reduction of 20% in fatigue life compared to the specimens with polished surfaces that had much smaller surface defects (~12 μm) (Gungor and Edwards, 1993). In another study (Jiang et al., 1999), it was shown that the period for reaching a crack length of 200 μm accounted for 60% of the total fatigue life for polished specimens but only about 30% for the as-cast specimens, as a result of shorter initiation and small-crack propagation times at a rough surface (hollows) or pores. Fatigue lives of specimens are dominated by crack propagation when pores are present as initiation sites, and the difference in fatigue behavior is principally due to the extra period of propagation in the polished specimens.

Simulations (Gall et al., 2001) showed that the driving force for fatigue-crack initiation from a pore was two orders of magnitude larger than that of a bonded inclusion of the same size. Although a shrinkage pore has been recognized to be far more deleterious than a gas pore in casting aluminum alloy (Skallerud, 1993), the simulations showed that the shape of the pore has a negligible effect on fatigue-crack initiation compared to the size of the pore.

The importance of minimizing porosity to achieve good fatigue properties has been highlighted (Major, 1998). It was found that the impact of both large pore sizes and a
large dendrite arm spacing are most severe at high stress amplitudes, with a tendency for the dendrite arm spacing effect to ebb at lower stress amplitudes. Hipping has been applied to reduce the size of the pores and has been found to significantly increase the fatigue life in aluminum castings (Lei et al., 1997; Rich et al., 1999; Mashl et al., 2000; Nyahumwa et al., 2001).

When the pore size was reduced to the point where the maximum size of pore would be no larger than the size of a nucleated crack (~20 μm) caused by persistent slip bands in pore free material, it was shown that fatigue life was not further increased (Couper et al., 1990). It was also reported, however, that pores as large as ~100 μm did not significantly reduce the fatigue life of A356 aluminum castings (Mayer et al., 1999). It seems that there is a critical size of the pore as crack-initiation site, below which other microstructural features such as eutectic constituents (silicon particles) are the crack-initiation sites.

Oxide inclusions are stress concentrators and are known to be active sites where fatigue cracks initiate. Also oxides have also been recognized to be the nucleation sites of pores during solidification and should be considered in the modeling of pore formation (Poirier 1999). The formation of oxide films has been detailed (Campbell, 1991). “Old” oxide films are formed earlier during the melting process, originating from the surface of the melt in furnace or ladles, and are relatively thick. “Young” oxide films formed during the filling of the mold and are thin and finely folded. It was shown that, oxide films,
either formed during melting or generated during the pouring process, initiated fatigue
cracks early and decreased the fatigue resistance of a cast alloy (Nyahumwa et al., 1998).
The cumulative distribution of fatigue life of cast alloys showed the fatigue lives of most
aluminum castings can probably be improved by a factor of 100 times, at least, through a
combination of metal quality and casting techniques. Based on the result that the
unfiltered but hipped castings exhibited a higher fatigue life than filtered casting, it was
reported that hipping could help to some degree to remedy the deleterious effect of oxide
films by achieving a degree of bonding across the oxide-oxide interface in the folded
films (Nyahumwa et al., 2001).

2.1.2. Eutectic Silicon Constituents
The eutectic constituent in Al-Si castings comprises a mixture of silicon particles within
the aluminum-rich solid solution. These silicon particles may initiate fatigue cracks when
more severe crack initiation sites (pores and oxide inclusions) are not present or too small
to initiate cracks. In pressurized castings, it was found that the fatigue cracks initiate at
silicon particles and the pressurized castings have a fatigue strength which is 70% greater
than the fatigue strength in sand-mold castings, where the cracks initiate at shrinkage
pores (Inguanti, 1985). It was postulated that there is a threshold casting defect size
below which fatigue properties are dependant on silicon particle size alone. In an
investigation on the initiation and propagation of cracks in strain-controlled fatigue of
squeeze-cast A356 alloys (Plumtree and Schafer, 1986), fatigue cracks initiated at the
interdendritic eutectic silicon particles either by cracking of the silicon particles or
debonding at the interface with aluminum matrix. The latter was associated with long life high-cycle fatigue at low cyclic strain ranges. In another study on the fatigue of silicon reinforced aluminum composite manufactured by powder metallurgy, cracks initiated at either the cracked silicon particles for specimens with larger particles of size of ~ 8 μm or initiated within thin ligaments of the matrix which separate the uncracked subsurface particles from the free specimen surface (Lukasak and Koss, 1993).

In the study of the fatigue of direct-chilled castings of A356 alloy, the material obtained was without pores (Odegard et al., 1990; Odegard et al., 1991). The grain size was 300 μm, the SDAS was 20 μm and the average mean silicon particle diameter was 2 μm. It was found that fatigue cracks initiated within PSBs adjacent to the silicon particles in the interdendritic regions in any of the four aging conditions (under-aged, near-peak aged, slightly over-aged and over-aged). Two mechanisms were suggested responsible for the formation of PSBs adjacent to silicon particles. One is that the silicon particles are surrounded by a stress field created during quenching due to the difference in the thermal expansion coefficient for silicon and aluminum. The other is that, due to the difference in Young’s modulus between silicon and aluminum, the applied stress gives stress gradients in the interdendritic eutectic silicon particle regions, which is high enough to induce slip and form PSBs. A significant increase in fatigue life was observed when cracks initiated at silicon particles instead of pores. Materials with approximately the same grain sizes and secondary dendrite arm spacings but containing pores show ~ 50% reduction in
fatigue life compared to the defect free material, and the fatigue strength at $10^7$ cycles was reduced by 20% (Odegard et al., 1994).

Fatigue-crack initiations at inclusion particles were also studied in ferrous alloy systems. The spheroidal or slightly ellipsoidal MnO-SiO$_2$-Al$_2$O$_3$ inclusions with sizes of 5-80 μm initiated fatigue cracks in 4340 steel, and the fatigue lives seemed to be closely related to the inclusion sizes (Lankford and Kusenberger, 1973). It was determined that cracks initiated through debonding of the inclusion, gradual growth of the debonded seam, and eventual nucleation of microcracks within the matrix, but not adjacent to the inclusion/matrix interface. Slip bands were not involved in crack initiation at surface inclusions, but they appeared to be associated with microcracks initiated by subsurface inclusions. A theoretical shear stress analysis for a perfectly bonded, rigid inclusion within a plate showed that the maximum shear stresses were located not at the interface, but rather at a distance of ~1.5 times the diameter of the inclusion, in general agreement with the experimental findings regarding that the cracks formed at the defects isolated from the interface. Although this is not a study on aluminum alloy with eutectic silicon particle as inclusions, it suggests that the debonding of eutectic silicon particles in aluminum matrix is a prerequisite for crack initiation.

Fatigue-crack initiation and microcrack propagation were examined in 2024-T4 and 2124-T4 aluminum alloy (Kung and Fine, 1979). The latter is a high purity version of 2024 and contains considerably fewer constituent particles. Results illustrated that at high
stresses fatigue cracks initiated at coarse slip lines for both alloys, while at low stresses fatigue cracks initiated from large constituents particles (above ~ 6 μm) along slip bands. The crack-initiation life of 2124 at high maximum stresses was greater than that of 2024 due to the smaller grain size of 2124, and the greater crack initiation life of 2124 at low maximum stresses was due to the smaller constituent particles. No cracked constituent particles were found to initiate fatigue cracks. This is possibly because the particles (Al₂CuMg and Al₇Cu₂Fe) are relatively spherical and small (below ~15 μm), and the local shear force applied by the matrix on the particles during cyclic loading were not enough to fracture the particles.

Simulations showed that cracked or debonded silicon particles have a maximum plastic shear strain that is two orders of magnitude higher than the bonded particles (Gall et al., 2001). A cracked particle facilitates extremely large local stresses in the broken particle halves, which will invariably leads to the debonding of a cracked particle. Particle debonding results in a local intensification of stresses in the matrix that is significantly larger that that due to particle fractures, so debonded silicon particles are asserted to be the critical inhomogeneities for crack initiation. It was shown that the particle shape has a negligible effect on fatigue crack initiation, whereas alignment and spacing in the clusters are the most dominant parameters influencing initial particle fracture and debonding (Gall et al., 2001).
In aluminum castings, the silicon particles are restricted to the eutectic constituent, which makes up the interdendritic and integranular regions that solidify last as a cluster of particles. The presence of silicon particles in a cluster may significantly influence the crack initiation mechanism. Indeed, simulations on stress distribution in a SiC-particulate reinforced A356 composite fabricated by direct casting with molten metal mixing method (Wang et al., 1993) indicated the importance of the distribution of constituents in eutectic clusters. It was shown that the stress non-uniformity is dependant on the particle shape (aspect ratio), reinforcement volume fraction, their arrangement relative to each other, and the external load. The highest von Mises stress was always found between end-to-end particles with high aspect ratios, which are closely arranged with their longitudinal dimensions aligned along the loading direction. The triaxial stress state around a particle or inside a particle cluster may change the von Mises stress and result in strain localization. As the applied stress increases, the state of the triaxial stress inside a cluster appears to promote the early particle cracking, interface debonding and void formation.

The size and morphology of the eutectic constituent particles may significantly influence the resistance to crack initiation. It was found in a particulate-reinforced aluminum alloy composite that with silicon particles of ~ 4 μm, crack initiation occurs at ~ 40% of the total fatigue life within the thin ligaments of the matrix which separate the uncracked subsurface particles from the free surface (Lukasak and Koss, 1993). In contrast, specimens containing larger silicon particles of ~ 8 μm were susceptible to crack initiation from cracked particles within only ~ 5% of the total fatigue life. So finer
silicon particles increase the crack initiation life greatly. It was also shown by a model that refining the eutectic silicon particles improves the fatigue strength of cast aluminum alloys (Fan and McDowell, 1998). In their simulations of the crack initiation life, a decreased initiation life with an increase of particle size and/or aspect ratio of the silicon particles was reported.

The use of strontium modification has been the target of numerous researches (Closset and Gruzleski, 1982; Fang and Granger, 1989; Fata-Halla, 1989; Pan et al., 1991; Shivkumar et al., 1991; Tynelius et al., 1993; Emadi et al., 1993; Schneider and Feikus, 1998; Fata-Halla et al., 1999; Anson et al. 2000; McDonald et al. 2000). On one hand, Sr-modification in A356.2 aluminum alloys changes the morphology of the silicon particles in the as-cast eutectic constituent from plate-like to fibrous and results in rounder and finer silicon particles after solution heat-treatment. Hence, modification can improve the ultimate tensile strength and percent elongation (Fata-Halla, 1989; Shivkumar et al., 1991; Fata-Halla et al., 1999). On the other hand, there are strong indications that porosity is increased by Sr-modification in A356.2 casting alloy (Fang and Granger, 1989; Tynelius et al., 1993; Zhang et al. 2000), and the porosity associated with the addition of strontium may nullify the modification effect on mechanical properties. In research on the effect of Sr-modification on the mechanical properties in A356 alloy (Schneider and Feikus, 1998), Sr-modification did not lead to significant improvements in yield strength, tensile strength, and no differences worth mentioning were detected with respect to elongation to failure after heat treatment. It seems that
when the silicon content approaches the eutectic composition, a strontium-modified alloy shows a better elongation than found in the non-modified version. It was concluded that Sr-modification is not beneficial enough to warrant its unqualified use in die-casting. It was also reported that the use of Sr-modification for aerospace castings should be discouraged (Caceres et al., 1997).

The influence of Sr-modification on the fatigue properties of a 320-T4 cast aluminum alloy showed that there was no statistically significant differences in the fatigue lives of unmodified and modified alloys (Boileau and Allison, 1996). Pores dominated the crack initiation and masked any beneficial effect that strontium may have had. When pores were not present, as a result of very low hydrogen content, the optimization of silicon morphology was shown to be beneficial to fatigue strength (Kennerknecht et al., 1997).

The influence of microstructural constituents on the fatigue of A356/A357 was studied when pores were removed through hipping (Wang et al. 2001). It was shown that large and elongated eutectic silicon particles present in the unmodified alloy result in lower fatigue lives for specimens with a secondary dendrite arm spacing of ~27 to ~77 µm.

2.1.3. Secondary Dendrite Arm Spacings (SDAS)

The SDAS is directly related to the local solidification time (Spear and Gardner, 1963). A longer local solidification time yields a larger SDAS with a coarser eutectic silicon constituents, greater amount of porosity, and larger sized intermetallics in A356 aluminum castings. It has been shown that, as SDAS decreases, the mechanical properties
of either hipped or not hipped A356-T6 alloys, especially the ultimate tensile strength and elongation, were remarkably improved (Bailey, 1965; Flemings, 1974; Wickberg et al., 1984; Campbell, 1991; Boileau et al., 1997).

There is clear evidence that the SDAS influences the fatigue performance. In early works on the relationship between fatigue life and microstructures in A356 (Bailey, 1965; Wickberg et al., 1984), it was found that as the dendrite cell size (thus SDAS) decreased, fatigue life was increased. In another study (Stephens et al., 1988), it was shown the low-cycle fatigue life of A356 with a small SDAS was superior to that with a coarser SDAS. The better fatigue properties with a smaller SDAS are due to the decreased level of porosity associated with the decreasing of the SDAS. In another study (Sonsino and Ziese, 1993), it was reported that the size of the cast cross-section (thus SDAS) has no influence on the fatigue strength when there is porosity present. It seems that the SDAS-effect was masked by the presence of the pores, and pore size appears to be the dominant factor affecting fatigue life. But when the castings are void from shrinkage and gas pores (Honma and Kitaoka, 1984), A356 alloy with a SDAS of 20 µm has a higher fatigue live than that with SDAS of 32 and 66 µm. The beneficial effect of a smaller SDAS and globular Si-particles on the fatigue properties of D357 investment castings was also reported when the controlled gas content was relatively low, about 0.085-0.118 ml H₂/100 g Al (Kennerknect et al., 1997).
In order to isolate the effect of porosity on fatigue, fatigue tests have been conducted on hipped Sr-modified A356 specimens (Boileau et al., 1997). It was shown that the SDAS does affect fatigue life; as the SDAS decreased from 90 μm to 30 μm, the fatigue life increased by a factor of three when tested at maximum stress of 138 MPa with stress ratio of -1. Fatigue tests on hipped A356 alloys were also conducted recently under a stress amplitude of 100 MPa with a stress ratio of 0.1 (Wang et al., 2001). When the SDAS was less than ~60 μm, fatigue life decreased with increasing SDAS; when the SDAS was larger than 60 μm, increasing the SDAS did not further decrease fatigue life. In fact, a slight increase of fatigue life was seen when the SDAS increased from 60 μm to ~80 μm. Oxides and persistent slip bands were identified to be the crack-initiation sites. The authors presented a detailed analysis on this phenomenon based on the interactions of the dislocations with the microstructures under static loading. Nevertheless, a smaller SDAS may result in longer crack-initiation time due to the more homogeneous microstructure of the alloy with a finer SDAS. In order to explain the increased fatigue life for the alloy with a SDAS above 60 μm, other effects such as the influence of the SDAS on the crack-closure during crack propagation must be considered.

2.1.4. Heat Treatment

The bending fatigue tests of A356 bars in the as-cast condition and after the heat treatment condition (1 hr. at 520 °C / water-quenched / 6hr. at 160 °C) were performed (Schneider and Feikus, 1998). The bending fatigue strength could be increased by heat treating but only up to $10^6$ cycles. Above that number of cycles, no measurable difference
was detected between fatigue lives of as-cast and heat treated specimens. The effect of varying the heat treatment on the fatigue behavior of A356 was also studied (Couper et al. 1990). It was found that fatigue life was insensitive to heat-treatment when the cracks initiated at interdendritic shrinkage defects.

The effect of artificial aging on the fatigue behavior in direct-chill-cast A356, nearly free of pores, has been investigated (Odegaard, 1990; Odegaard, 1991). The specimens were solutionized 5 hours at 525 °C, quenched in cold water and stored at room temperature for 1 hour before aging. The grain size of the alloy was ~ 300 μm, the SDAS was ~ 20 μm, and the average mean Si-particle diameter was ~ 2 μm. Some specimens were under-aged, aged to peak hardness, slightly over-aged, and over-aged. In the peak hardness condition coherent β"-Mg2Si needles were formed, while in slightly over-aged condition a homogeneous distribution of secondary Si particles was also found. It was shown that slightly over-aged condition had superior fatigue life compared to under-aged and over-aged specimens. Specimens in the under-aged condition showed cyclic hardening under a constant strain amplitude until a saturation stress was reached; a tendency for softening was observed, however, at the end of the fatigue process. In the near peak hardness condition, cyclic hardening to a peak stress followed by cyclic softening was observed. In the slightly over-aged condition the alloy showed moderate softening until failure. In both under-aged and near peak-aged conditions the coarse PSBs formed during fatigue, while in the slightly over-aged condition the PSBs were more rarely observed and only adjacent to the primary Si particle. In the over-aged condition very few PSBs were
observed. However, crack-initiation occurred within the PSB's adjacent to the Si particles in the interdendritic regions under all four aging conditions. It was suggested that the stress gradients created during quenching due to differences in thermal expansion coefficient and the Young's modulus for silicon and aluminum were responsible for the formation of the PSBs adjacent to the silicon particles. In under-aged to peak aged conditions, the coherent $\beta''$-Mg$_2$Si precipitates were easily cut through by the dislocations, and localization of strain into the PSBs extending through the whole grain occurred. The presence of the secondary Si-particles and the semi-coherent $\beta'$-Mg$_3$Si in the slightly over-aged and over-aged conditions prevented the formation of coarse PSB's in the matrix and made the crack-initiation more difficult.

The above discussions illustrate that a large amount of research on the fatigue behavior of aluminum castings has established a corresponding understanding on the influence of microstructure on fatigue-crack initiation. However, a good mechanistic and quantitative understanding of microstructural influences on crack-initiation and small-crack propagation has not been reached. It is believed that improved processing could result in eliminating both porosity and oxide inclusions, so that the understanding of the effect of the SDAS, silicon particle morphology and heat treatment would be important in enhancing the fatigue live. At present, some information is available on the effects of SDAS and Si-particles on fatigue behavior in aspects of crack-initiation and small-crack propagation; but additional experimental research on crack-initiation and small-crack
propagation is still needed to capture the fatigue crack-initiation and small-crack propagation mechanisms in A356.

2.2. MICROSTRUCTURAL EFFECTS ON LONG-Crack PROPAGATION

The propagation of long-cracks follows linear elastic fracture mechanics (LEFM) and have been studied in many alloys and composites. Before small-crack behavior is addressed, microstructural effects on long crack propagation related to A356 cast alloy are discussed.

2.2.1. Pores

Pores in cast alloys can initiate fatigue cracks early and lead to reduction or elimination of crack initiation life, and the total fatigue life consists mostly of the propagation life. On the other hand, pores may cause blunting of the crack tip, thus reducing the driving force for crack propagation. A more tortuous crack path in alloys with porosity results in a longer crack path and also a high level of roughness-induced crack closure, which also reduces the crack-driving force.

In a study of the effect of hipping on the fatigue behavior of cast A201-T7, it was shown that the improvement of fatigue life was attributed to the increase in the resistance to crack initiation, achieved through the removal of pores as initiation sites, but the long-crack propagation rates were unaffected by the hipping (Mocarski et al., 1991). Another investigation showed that fatigue crack propagation resistance was independent of the
porosity level in the cast alloy D357 (Ozelton, et al., 1991). The effects of different levels of porosity on the fatigue-crack propagation in cast Al-Cu alloy (A206-T7) were also studied (Rading et al., 1994). As the level of porosity decreased, the threshold stress intensity range ($\Delta K_{th}$) increased. Specimens with high level of pores showed slower increase of propagation rates with stress intensity range and tended to have lower propagation rates at higher stress intensity ranges. When hipping was used to reduce the porosity, it was found that hipping was only effective in improving the crack-propagation resistance close to the stress intensity threshold.

2.2.2. Eutectic Silicon Constituents

The effect of particle size on the long crack propagation of particulate reinforced aluminum alloy composites has been studied. Because of the similarities of the silicon particles in A356 to the particles in composites, some results in composites on the particle size effect on the long-crack propagation are reviewed.

Studies on the role of SiC particles in fatigue crack propagation in SiC/Al-Zn-Mg-Cu composites produced by powder metallurgy processing have been conducted (Shang, et al. 1988; Shang and Ritchie, 1989). It was found that, at low stress intensity ranges, composites reinforced with SiC-particles of 10 $\mu$m had a rougher fracture surface than those with finer particles of 5 $\mu$m and promoted crack closure from asperity wedging and improved crack propagation resistance when the specimens were tested under a load ratio
of 0.1. At low load ratios, increasing the particle sizes resulted in higher fatigue crack propagation thresholds for a given volume fraction of particles in the matrix. Roughness-induced crack closure at near-threshold stress intensity has been discussed (Ritchie and Suresh, 1982; Allison, 1988), and a geometric model was formulated based on the height of the fracture surface asperity, mean grain diameter and the proportion of mode II displacements (Suresh and Ritchie, 1982).

At high load ratios, when the fracture surfaces do not contact with each other, the roughness-induced crack closure effect was not present, and it was found that a smaller particle size resulted in higher stress intensity threshold levels, contrary to that at low load ratios (Shang and Ritchie, 1989). Crack trapping by smaller size particles at the crack tip was invoked to explain this phenomenon. With increasing $\Delta K^*$, it was shown that composites with coarser particles possess lower fatigue-crack propagation rates than that with finer particles. As a crack propagates, coarse particles break in the plastic zone and result in a higher density of uncracked ligaments behind the crack tip; thus coarse particles are more effective in reducing the stress intensity. At high $\Delta K$ levels approaching the fracture toughness, the propagation rates are faster in the composites with coarser particulates owing to their low toughness.

---

$\Delta K^*$ is the range of the stress intensity factor during cyclic loading.
The influence of Sr-modification on the long-crack propagation in Al-Si cast alloy was conducted (Lee et al. 1995). In a eutectic alloy (Al-12Si) the long-crack propagation rate was lower for the modified alloy than that in the unmodified alloy, and the threshold value of $\Delta K (\Delta K_{th})$ was increased after modification. The modified alloy showed better crack propagation resistance than the unmodified alloy due to increased levels of crack closure and the high probability of deflected and branched crack paths. Similar results were obtained when adding Be in A357 to modify the morphology of the iron intermetallic compound (Tan et al., 1996).

In another study on Al-10Si-3Cu die-cast alloy, it was shown that Sr-modification reduced the average silicon interparticle spacing and particle length by one-half compared to the unmodified alloy (Schaefer and Fournelle 1996). The modified alloy exhibited 10–20% higher $\Delta K_{th}$ than the unmodified alloy. No difference was observed when tested at a load ratio of 0.1. The increased crack tortuosity was responsible for the increased crack-propagation resistance.

The morphology-effect of silicon particles on crack propagation at a load ratio of 0.1 in squeeze-cast A356 alloy with a $SDAS$ of 30 μm was investigated (Kumai et al. 1996). It was found that Sr-modification did not influence the crack propagation rates in the low $\Delta K$ region despite the difference in the silicon morphology produced by the modification. The alloys exhibited a similar $\Delta K_{th}$, but the unmodified alloy had a much lower toughness and resulted in a truncation of the Paris region for the unmodified alloy. Another
investigation was also conducted on the effect of silicon particle-morphology on hipped A356 cast alloys with SDAS of 20 and 60 µm (Kumai, et al. 1999); it was found that, at higher stress intensity range, the unmodified silicon particles easily crack or debond and provide multiple cracks and accelerated rates at higher stress intensity factor range. At lower stress intensity range, the unmodified alloy showed a lower crack propagation rate, and it was suggested the increased cyclic yield strength of the unmodified A356 reduced the crack tip opening displacement and resulted in a higher crack-propagation resistance. In another study, a similar result was obtained (Katsumata et al. 1999).

2.2.3. Secondary Dendrite Arm Spacings (SDAS)

It is difficult to separate the effect of SDAS from that of silicon particles, because in cast alloys, rapid solidification results in both finer SDAS and smaller silicon particles. In the study of the influence of microstructure on the fatigue behavior of unmodified A356-T6 (Wigant and Stephens, 1987), it was shown that A356 with a SDAS of 40 µm had greater fatigue resistance and longer fatigue life than that with 85 µm. However, A356 with the two different SDAS exhibited similar values of $\Delta K_{th}$. Substantial roughness-induced crack closure was present when tested at a load ratio of 0.1, and it was reported that the fatigue crack propagation resistance of A356-T6 alloys was essentially unaffected by the SDAS.

Studies on unmodified eutectic Al-Si-Mg cast alloy also showed that the solidification rate effect on long-crack propagation rate was not apparent for the specimens obtained
from directionally solidified ingot (Lee et al., 1995). The microstructural effects on crack propagation of near eutectic Al-Si-Cu-Mg cast alloy were also investigated (Kobayashi et al., 1996). Two types of microstructures were obtained: a finer microstructure was produced by pouring in a mold with a pre-heat temperature of 423 K and with a modifier content of 110 ppm Ca; a coarser microstructure resulted by pouring in a mold preheated to 623 K and using a modifier content of 40 ppm Ca. Replicas were used to determine crack propagation rate and crack deflection and branching. It was concluded that the long-crack propagation rate was essentially independent of the microstructures.

Contrary to the conclusion that the long-crack propagation rate is not influenced by the fineness of the microstructure, many other studies concluded that refining the SDAS reduces the crack propagation rate at high values of ΔK. The long-crack propagation behavior of a Sr-modified squeeze-cast A356 alloy with a SDAS of 30 μm and a sand-cast A356 alloy with a SDAS of 100 μm differed (Kumai et al., 1996). It was found that the threshold stress intensity range increased from 6.0 MPa m$^{1/2}$ to 7.5 MPa m$^{1/2}$ when the SDAS increased from 30 μm to 100 μm for the modified alloy. It was found that the fracture surface of the alloy with a large SDAS was rougher than that with a small SDAS, and the size of the facets in alloy with a large SDAS was comparable with the dendrite cells and larger than that of the alloy with a small SDAS. The deflected fatigue crack propagation path and crack branching in the coarse microstructure increased the near threshold propagation property. In hipped W319 cast aluminum alloy, very similar results of the effect of SDAS on long-crack propagation were reported (Caton et al., 1999). A
similar result was also obtained recently in a study of fatigue in hipped sand-cast and strontium modified A356 with a SDAS of 52 μm and permanent cast and strontium modified A356 with a SDAS of 28 μm (Han et al., 2001). In another investigation on the effect of SDAS on hipped A356 cast alloy, a similar result was reported when the stress intensity range was above ~8.0 MPa m^{1/2} (Kumai et al., 1999). In both modified and unmodified alloys, the sand-mold cast alloy with a SDAS of 60 μm reached the final fracture at the smaller value of ΔK than the permanent mold cast alloy with a SDAS of 20 μm. In both the modified and unmodified alloys with a SDAS of 20 μm, the overall crack propagation rate compared to that of the alloys with the coarser SDAS of 60 μm. It was also found that the degree of crack closure is independent of the microstructural variations.

2.2.4. Heat Treatment

The effects of aging on the fatigue behavior in a direct-chilled A356 alloy were already mentioned in Section 2.1.4. Crack propagations in A356 without modification and subjected to resonant vibration were studied (Jiang et al., 1997). It was shown that solution heat treatment improved the ductility and crack propagation resistance. In the as-cast condition, the crack propagated preferentially through the interdendritic region, which contained acicular eutectic silicon particles. Solution heat treatment caused the crack to propagate with less deflection and branching and reduced the extent of particle breaking. In yet other studies on the aging-effect on the crack propagation, it was shown
that the crack propagation mode was different for the naturally aged \((NA)\) and under-aged \((UA)\) conditions compared to the peak-aged \((PA)\) and over-aged \((OA)\) conditions (Jiang et al., 1999; Jiang et al., 2000). In the \(NA\) and \(UA\) conditions, the cracks occur mainly through slip bands owing to the coherent interface between precipitates and the aluminum matrix. In the \(OA\) and \(PA\) conditions, the precipitates were incoherent with the matrix, so the dislocations bypassed the precipitates, and cracks through the slip bands disappeared.

### 2.3. MICROSTRUCTURAL EFFECTS ON SMALL-CRACK PROPAGATION

In a study on fatigue crack initiation and early propagation in precipitation hardened aluminum alloys (Person, 1975), it was found that cracks of a size comparable to the average grain diameter propagate much faster than the predicted rate based on large crack theories under the same applied stress intensity range, and that small cracks propagate at applied stress intensities less than the fatigue threshold. The driving force for the small-crack propagation depends on the microstructural feature where the crack initiated (Trantina and Barishpolsky, 1984). It was shown that, compared to a void, the crack driving force for a bonded, cracked inclusion is 15% higher, and the crack driving force for an unbonded inclusion is 10% lower.

The so-called anomalous behaviors of either small cracks initiated at heterogeneities at polished surfaces or at induced two-dimensional cracks have been observed in many materials (Pineau, 1984; Suresh and Ritchie, 1984; Pedersen, 1988; McClung et al.,
Generally, the small-crack effect exists when crack sizes are comparable to the characteristic length-scales in the microstructure. When the small crack sizes are comparable to the plastic zone size in front of the crack tip, the small cracks are classified as mechanically-small cracks. When the crack sizes are comparable to the extent of crack-tip shielding (e.g. crack wedging by crack closure behind the crack tip), the small cracks are classified as physically-small cracks.

In one of the studies on the difference in the propagation behavior of large and small-cracks, local crack-tip opening mechanics were characterized in a high strength aluminum alloy (Lankford et al., 1984). It was shown that both the rapid small-crack propagation rate relative to the large cracks and the absence of a small-crack threshold stress intensity were related to the larger crack-tip opening displacements and relatively low crack-opening load for a small-crack. The ratio of the plastic-zone size to crack length for a small crack was ~1, while for large-cracks the same ratio was << 1. It was concluded that linear elastic fracture mechanics for small-cracks does not apply, even when a plastic-zone correction factor and a crack-closure effect are considered.

In simulations, the variability in small-crack propagation rate was presented (Gall et al., 1997). The orientation of individual grains, with respect to the crack propagation direction, favors an increase, decrease, or even an arrest of the crack-propagation rate. It was shown that the crack-tip forward slip band size ($r_p$), crack-tip opening displacement ($\delta$), and crack closure level varied as the orientations of the slip systems were changed.
The simulations predicted the approximate upper and lower bounds on crack propagation rates for microstructurally-small cracks. In another study on the role of grain-induced local anisotropy on stress intensity factor for microstructurally-small cracks, it was also shown that the effect of local anisotropy on the stress intensity variation was significant at small-crack sizes, and the small cracks were influenced significantly by the grain-induced anisotropy (Li and Ravichandran, 1999; Ravichandran and Li, 2000).

Mechanically-small cracks propagated from notches can propagate faster than long cracks when the small-crack length is shorter than a length of some proportion of the length of the plastic zone of the notch (Leis and Forte, 1981; Lankford et al., 1984. McClung et al., 1992). The small-cracks either propagate unstably with a increasing rate or arrest completely after propagating a short distance.

Small-crack effects are very important, since numerous studies have shown that, using linear fracture mechanics and treating the pore or microstructural heterogeneity as an existing crack, often result in inaccurate estimations of fatigue life (Newman et al., 1999).

2.3.1. Pores as Notches

Cracks at notches have been treated as mechanically small-cracks. Although pores differ from notches in many ways, such as their sizes and morphology, it is still believed that
pores in cast alloys behave similar to notches, so it is necessary to understand the previous researches on small-cracks emanating from notches.

The extent of small-crack behavior has been estimated from fracture mechanics analysis, with its size being of the order of one tenth of the notch radius (Dowling 1979, Miller 1988). The stress intensity ($K_s$) for a small crack extending from an edge notch before the crack reaches the transition length is:

$$K_s = 1.12k_iS\sqrt{\pi l}$$  \hspace{1cm} (2.1)

where $l$ is the length of the small crack propagated from the notch; $S$ is the remote stress; and $k_i$ is the elastic stress concentration factor. Stress concentration factors for notches can be found in reference (Peterson, 1974).

The transition crack length, beyond which the crack is considered to be a long crack, is:

$$l_0 = c/\left[\left(1.12k_i/\sqrt{Q}\right)^2 + 1\right]$$  \hspace{1cm} (2.2)

where $Q$ is a dimensionless function of geometry; $c$ is the depth of an edge notch or half length of an internal notch; and $l_0$ is the transition size of crack propagation from the notch. Values of $l_0$ are generally a small fraction of the notch root radius $\rho$, and for moderate to sharp notches, they generally fall in the range of $\rho/20 \sim \rho/4$.

The crack-closure effects for a crack at the pore (notch) also should be considered to characterize the small-crack behavior. It was pointed out that the void height influences the calculated crack-opening stresses for small cracks (Newman 1992). As the crack
propagates the newly created crack surfaces close, and the opening stresses rapidly rise and level off to a steady state value. Crack closure at notches has been applied to cracks at pores in a cast aluminum alloy (Ting and Lawrence, 1993).

2.3.2. Silicon Particles

In the study on the fatigue behavior of smooth specimens from squeeze cast Al-Si alloy subjected to rotary-bending fatigue, it was found that the small-crack propagation was retarded around the silicon particles and enhanced in the aluminum dendritic matrix (Shiozawa et al., 1997). Thus, an improvement of fatigue strength in cast aluminum could be expected by refining the silicon particles. Also it has been shown that in aluminum cast alloys, small-cracks propagate well below the large-crack propagation threshold (Skallerud et al., 1993; Seniw et al., 1997).

The effects of the particles on small-crack behavior in SiCp/ aluminum composites were studied (Kumai et al., 1990; Kumai et al., 1992). Large perturbations in propagation rate were observed for small surface-cracks. Crack arrest took place when the crack tip reached a SiC or constituent particles. After being arrested at the particles, the cracks frequently ran around the particle-matrix interface and then propagated through the matrix. When the cracks were short, there was no evidence of fracture of particles ahead of the crack tip. The small-crack data merged with the long crack result when the crack reached a length of ~200 µm.
Bending fatigue tests with a stress ratio of \(-1\) were conducted on a Si/2024 composite manufactured by powder metallurgy (Jono and Sugeta, 1995). In specimens with Si particles of 10 \(\mu\)m, the cracks propagated more slowly than with particles of 20 \(\mu\)m at the same particle volume fraction. It was concluded that more deflective behavior of crack propagation in finer Si/2024 resulted in the increase of the intrinsic crack propagation resistance due to the longer crack path and more reduction in driving force for propagation.

Cracks were simulated to propagate through an aluminum matrix and around eutectic silicon particles in Al-Si alloy (Fan et al., 2001). The results showed that as the crack approached the particles, the maximum plastic-shear strain at the crack tip was reduced due to a blockage mechanism. However, as the crack neared a particle, the plastic shear strain range increased rapidly. As the crack encountered a particle the crack-tip displacement range \((\Delta\text{CTD})\) dropped abruptly. There was a very strong effect of particles on the small-crack fatigue behavior. Intact particles increased resistance to the small-crack propagation. Although the \(\Delta\text{CTD}\) for a crack prior to meeting the first particle was much higher for a crack with long initial crack length than for a short one, after both cracks engaged two to three silicon particles, they had very similar \(\Delta\text{CTD}\) levels and followed a common retardation pattern. The simulation also showed that cracks opened even when the applied stress was still in compression. This enhancement of crack opening likely contributes to the propagation of small cracks below the large crack threshold, and partly accounts for the rapid propagation rates of the small cracks.
2.3.3. *Secondary Dendrite Arm Spacings (SDAS)*

In one of the studies on small-crack behavior in A356 cast alloy, it was found that, in the completely reversed strain controlled uniaxial cyclic testing, the critical length of a small fatigue-crack was directly related to the SDAS; cracks about twice the SDAS displayed initially fast propagation rates (Plumtree and Schafer, 1986). At high cyclic strains, the cracks continued to propagate at a constant rate. At low cyclic strains, the cracks decelerated until they passed through the eutectic constituents.

Small-crack propagations in hipped W319 alloys with a SDAS of 23 μm and 100 μm was measured and correlated to several parameters (Caton *et al.*, 1999). The major result was that the small-crack propagation rates are much higher in the more slowly solidified alloy (larger SDAS) compared to the more quickly solidified alloy (smaller SDAS).

As mentioned before, cracks were simulated to propagate through aluminum matrix and around eutectic silicon particles in Al-Si alloy (Fan *et al.*, 2001). The crack tip displacement (CTD) was much larger for a crack tip in the middle of the dendrite cell than near the interdendritic region where the silicon particles are clustered. Hence the decrease of fatigue life that was experimentally observed for a large SDAS may be attributed to the decrease of constraint on the range of CTD as the SDAS increases.

Despite the large amount of research related to long-crack behavior in Al-Si cast alloys, investigations on the microstructural effect on small-crack behavior still needs to be
conducted to clarify problems such as the effects of pore size, silicon morphology (strontium modification), and the SDAS. Very little information on small-crack behavior of casting aluminum alloys is available, and fundamental understanding for the small crack behavior in A356.2 aluminum alloy is needed to identify the key factors that control the fatigue life.

2.4. MODELS ON FATIGUE LIFE PREDICTION

The fatigue life is defined as the total number of cycles experienced by a component under cyclic loading before it completely fractures. Fatigue mainly consists of four stages: microcrack initiation, microcrack growth and linking (small-crack propagation), long-crack propagation, and final fracture. The different design philosophies often rest on how the crack-initiation and the crack propagation stages are quantitatively treated.

Two basic techniques, linear-elastic fracture mechanics (or \textit{LEFM}) and the local-strain approach, are employed in fatigue life prediction (Suresh 1998; Holman and Liaw 1997). The latter predicts the total cycles for crack-initiation and propagation to failure for a smooth specimen. This approach focuses mainly on the resistance to fatigue-crack initiation at a stress concentration within the microstructure. The former deals primarily with the resistance to fatigue-crack propagation. It is solely a crack-propagation life method in which fatigue life is simply a calculation of crack propagation from an initial crack size (mostly the size of a pre-existent flaw) to a final crack size at failure. Even though the \textit{LEFM} approach is limited by plasticity phenomena at the crack tip, \textit{LEFM} is
still in use due to its engineering simplicity and well-established long-crack propagation information.

2.4.1. Fatigue Crack Initiation

Based on the Coffin-Manson and Basquin-Morrow equations (Suresh 1998), the local-strain approach assumes that the $S-N$ curves (stress vs. number of cycles) can be represented by:

$$
\varepsilon_a = \left( \frac{\sigma_f - \sigma_m}{E} \right) (2N_f)^b + \varepsilon_f' \left( 1 - \frac{\sigma_m}{\sigma_f} \right)^{\gamma/6} (2N_f)^c
$$

where $\varepsilon_a$ is strain amplitude; $\sigma_m$ is the average stress; $\sigma_f'$ is a fatigue strength parameter, which usually equals the true fracture strength in tensile testings; $\varepsilon_f'$ is a fatigue ductility parameter; $N_f$ is the number of cycles to failure (i.e., the fatigue life) for a defect-free member; $E$ is the elastic modulus; and $b$ and $c$ are constants.

The product of the local stress and strain at a notch or a stress concentration site is:

$$
\Delta \sigma \Delta \varepsilon = (K_f \Delta S)^2 / E = \text{const.}
$$

$$
K_f = 1 + \frac{k_r - 1}{1 + \frac{\alpha}{\rho}}
$$

where $K_f$ is the fatigue notch factor; $\Delta \sigma$ is the local stress range; $\Delta \varepsilon$ is the local strain range; $\Delta S$ is the far field stress range; $k_r$ is elastic stress concentration factor; $\rho$ is the notch radius; and $\alpha$ is Peterson constant.
The local strain approach uses the cyclic stress-strain relationship:

\[
\frac{\Delta \varepsilon}{2} = f(\Delta \sigma) = \frac{\Delta \sigma}{2E} + \left(\frac{\Delta \sigma}{2K'}\right)^{(1/n')}
\]

(2.6)

where \( K' \) is the cyclic strength coefficient; and \( n' \) is the cyclic strain hardening exponent. Now the local stress and strain can be solved through equations (2.4) and (2.6), and then the fatigue crack initiation life can be estimated by the strain-life equation (2.3).

In fatigue life prediction of steel castings using the local strain concept (Heuler et al., 1992), it was shown that the local strain concept describes the crack-initiation potential of the defects more realistically and that it is more accurate than the fracture mechanics concept. Interpretation of casting defects as cracks produces very conservative estimates in many cases.

It was proposed that total fatigue life in cast Al-Si alloys is a sum of cyclic numbers spent in four fatigue stages as follows (Fan et al., 1998): the number of cycles for initiation of a microstructurally small crack (length on the order of mean Si particle spacing, \( d \)), \( N_i \); the number of cycles required for propagation of microstructurally small crack of length of \( a \) where \( d < a < \sim 2D \) and \( D \) is the size of dendrite cell, \( N_f \); the number of cycles for propagation of a physically small crack (\( \sim 2D < a < \sim 10D \)), \( N_2 \); and the number of cycles for long crack propagation (\( a > \sim 10D \)), \( N_3 \). The Coffin–Manson law, Eq. (2.3), was applied to compute the cycles to nucleation of fatigue crack. A finite element analysis of the cyclic deformation involving realistic sets of dendrite cells obtained from digitization
of light micrographs was used, and emphasis was placed on the local cyclic strain and stress levels associated with the microstructural heterogeneity.

The fatigue-crack initiation-life of materials is influenced by the distance of sparsely distributed defects due to the interferences between the stress fields. Such was shown for the dependence of crack-initiation on the location of two holes (Song and Bae 1998). Fatigue-crack initiation-life was related to an equivalent local strain magnitude ($\varepsilon_i$) and a plastic deformation area ($A_i$).

Related scenarios are the analyses of the interactions of near-surface defects and/or microconstituents with free or loaded surface (Frantziskonis, 1995; Frantziskonis et al., 1997; Renaudin et al., 1997; Frantziskonis, 2001) and numerical simulations of small-crack propagation in a matrix with randomly placed small and large pores (Haynie, 2001).

2.4.2. Long-Crack Propagation

Life predictions using the $LEFM$ approach are based on two main relations. One considers the stress field around the advancing crack tip leading to the stress intensity factor range $\Delta K$:

$$\Delta K = F(a/w)S \sqrt{\frac{\pi a}{Q(a)}}$$  (2.7)
where $a$ is the crack length; $w$ is the specimen width; $F(a/w)$ is the boundary-correction factor; $\Delta S$ is the far-field stress range; and $Q(a)$ is the elliptical crack-shape factor. Many tabulations are available as that the stress intensity range for various configurations with cracks can be calculated (e.g., see Newman and Raju, 1983; Tada, 2000).

The steady-state crack propagation rate of an advancing crack is often written as:

$$\frac{da}{dN} = C(\Delta K^n)$$

(2.8)

where $C$ and $m$ are material constants. Thus, a life prediction for crack propagation is obtained by integration:

$$N_p = \int_{a_0}^{a_f} \frac{da}{f(\Delta K, R)}$$

(2.9)

where $N_p$ is the crack propagation life; $R$ is stress ratio; $a_0$ is the initial crack size (i.e., half length of the preexisting defect); and $a_f$ is the crack length at failure which was determined from the fracture toughness $K_{lc}$.

$$a_f = \frac{1}{\pi} \left( \frac{K_{lc} Q(a)}{F(a) S_{\text{max}}} \right)^2$$

(2.10)

where $F(a)$ is the boundary correction factor; and $S_{\text{max}}$ is the maximum far-field stress. Since the geometry factor $Q(a)$ generally changes with the advance of crack, numerical integration techniques are necessary.
Equation (2.9) has been used to simulate fatigue life for defect-containing materials by treating the total fatigue life as the propagation life for materials with known size of preexisting cracks or flaws (Couper et al., 1990; Ting and Lawrence, 1993; Gungor and Edwards, 1993; Grant Jr. et al., 1993). In fatigue life predictions (Gungor and Edwards, 1993; Grant et al. 1993; Nadot et al., 1999), it was shown that the integrated lives are accurate at high stresses. At lower stresses, however, treating the pore as an existing crack severely underestimates the fatigue life. Fatigue-life predictions using small-crack propagation data showed that the extra fatigue-crack propagation-life for polished specimens is associated with smaller initiating defects. These facts suggest that the effect of defects (pores) on small-crack behavior must be taken into account, especially when the fatigue life is $\geq 10^6$ cycles.

A number of modifications have been made to account for other factors that may affect the propagation rate predicted by equations (2.7)-(2.10). A popular method of estimating fatigue life is the efficient stress-intensity method, which considers the effects of crack closure and elastic-plastic behavior when the crack is small. The Forman Equation and the Walker Equation (Suresh, 1998) are the ones in which the stress ratio effect is taken into accounted.

Fatigue cracks may remain closed even when subjected to cyclic tensile loads. Crack-closure mechanisms have been reviewed, and the most important crack-closure mechanisms are: plasticity-induced, oxide-induced, roughness-induced, and phase
transformation-induced crack closure (Suresh and Ritchie, 1984, Allison et al. 1988). When there is a crack-closure effect, the propagation rate can be written as:

\[
\frac{da}{dN} = C(\Delta K_{\text{eff}})^n
\]  \hspace{1cm} (2.11)

where \( \Delta K_{\text{eff}} \) is the effective stress intensity factor range and is equal to \( U \Delta K \). The factor \( U \) is often related to crack closure effects:

\[
U = \frac{(S_{\text{max}} - S_{\text{op}})}{(S_{\text{max}} - S_{\text{min}})}
\]  \hspace{1cm} (2.12)

where \( \Delta K \) is the range for the stress intensity factor (see Eq. (2.7)) and \( S_{\text{max}}, S_{\text{min}} \) and \( S_{\text{op}} \) stand for maximum, minimum and crack open stress, respectively.

In a numerical calculation of fatigue life, a crack-propagation model with crack-closure was applied (Skallerud et al. 1993). The model gave somewhat nonconservative results when compared with test results. Modification of the model, taking a short-crack effect into account, gave conservative results. Multiple crack initiation sites at pores were taken into account as a series of semi-elliptical cracks of equal size located with the averaged distance between them in the model. In another study on the fatigue of A390 cast alloy, the local strain method was used to obtain the notch root stress and strain, and then a crack-propagation model was applied to estimate the fatigue live of cast aluminum under constant and variable-amplitude loading (Dabayeh et al., 1996). The fatigue-life prediction was in good agreement with test results.
Realizing that a strain or stress ratio of 0.1 cannot eliminate the crack-closure effect in fatigue-crack propagation, the effect of crack closure through the factor $U$ has been used for long cracks (Newman, 1984).

$$S_{\text{open}} / S_{\text{max}} = \frac{S_0}{S_{\text{max}}} - \frac{0.3\sigma_0 (\Delta a / a)^{1/2}}{S_{\text{max}} F}$$  \hspace{1cm} (2.13)

$$\frac{S_o}{S_{\text{max}}} = A_0 + A_1 R + A_2 R^2 + A_3 R^3 \quad \text{For } R \geq 0$$  \hspace{1cm} (2.14)

$$\frac{S_o}{S_{\text{max}}} = A_0 + A_1 R \quad \text{For } -1 \leq R < 0$$  \hspace{1cm} (2.15)

where $S_0$ is the stable crack opening stress; $R$ is stress ratio; $\Delta a$ and $a$ are the crack propagation increment and the current crack length; and $A_0$, $A_1$, $A_2$, and $A_3$ are coefficients related to the yield strength, ultimate strength, the plastic constraint factor, and the maximum stress.

For a small fatigue crack, the crack-opening stress increases from a minimum to a stable level, and the crack-closure factor $U$ decreases from 1 to the value calculated by equation (2.13). The value of $U$ for small-cracks depends on the sizes of the microstructural heterogeneities which initiated the cracks. The empirical observations on small-crack behavior may further reveal the correlation between the type of microstructural heterogeneity and the crack closure effect.

The crack-closure effect was found to be present in A356 cast alloy when the stress ratio was below a critical ratio of 0.8 (Couper and Griffiths, 1990). Equation (2.11) has been applied to estimate the fatigue life of cast aluminum alloy (Couper et al., 1990). For
solution-treated alloys, it was shown that $U$ decreases rapidly with decreasing $R$ and then levels at $\sim 0.5$ as $R$ approaches zero, while for as-cast materials, $U$ decreases to $\sim 0.8$ at $R = 0$ and extrapolates to lower values for $R < 0$. $U$ is a function of crack length and is not a constant in the short-crack regime; the selection of $U$ as a constant may induce erratic prediction of fatigue lives.

To approximate the influence of the crack-tip yielding on the crack-driving force, a portion of the Dugdale cyclic-plastic-zone length has been added to the crack length. The cyclic-plastic-zone-corrected effective stress-intensity factor is written as (Newman, et al., 1984):

\begin{equation}
(\Delta K_p)_{eff} = UF(d/w)\Delta S \sqrt{\frac{\pi d}{Q(d)}}
\end{equation}

\begin{equation}
\omega = \left(1 - \frac{S_{open}}{S_{max}}\right)^2 \frac{\rho}{4}
\end{equation}

\begin{equation}
d = a + \omega^4
\end{equation}

\begin{equation}
\rho = a[\sec\left(\frac{\pi S_{max}}{2\alpha \sigma_0}\right) - 1]
\end{equation}

where $(\Delta K_p)_{eff}$ is the cyclic-plastic-zone-corrected effective stress intensity range; $F(d/w)$ is the cyclic-plastic-zone corrected boundary-correction factor; $\omega$ is a portion of the Dugdale cyclic-plastic-zone length; $\rho$ is the plastic-zone size for a crack in a large plate; $\sigma$ is the flow stress, the average of yield strength and ultimate tensile strength; $\alpha$ is the constraint factor, equal to 1 and 3, respectively, for plane stress and plane-strain.
conditions. The model predicted a fatigue life that agreed well with the test data for aluminum alloys (2024-T3 and 7075-T6) and steel (4340) (Newman, 1999).

2.4.3. Small-Crack at Notch Model

When materials contain defects that are smaller than around 300 μm, life predictions based on LEFM may be nonconservative. This is because small cracks emanating from small defects behave very differently than do longer cracks. For example, it has been shown that the initiation and propagation of a small crack up to 1 mm in length occupies 90% of the total fatigue life (Shiozawa et al. 1997). Therefore, knowledge of the anomalous behavior of small cracks must be obtained to correctly predict the total fatigue life.

LEFM techniques are often used to model defects as cracks with simple geometric shapes. Attempts have been made to model the microstructural features and complex shapes of the defects (Trantina and Barishpolsky, 1984, Hinkle et al., 1996). One explanation of the accelerated crack-propagation rate at notches was based on the crack-closure concept. When a crack initiates from a notch root, the plastic-wake field is not fully developed. As the crack propagates, the plastic-wake develops, and the crack tip generates its own stress fields that dictate the closure behavior, causing the influence of the notch to become gradually smaller.

When the length of a crack is far smaller than the defect/notch size, the crack is strongly influenced by the stress concentration. The stress concentration at the specific
microstructural feature has been considered and used in fatigue life prediction (Trantina and Barishpolsky, 1984; Hinkle et al., 1996). A dimensionless geometric term $F(a)$ was deduced when calculating the stress intensity ranges for different crack initiation sites; it is defined as:

$$F(l) = \frac{2}{\pi} + B(1.12 k_r - \frac{2}{\pi} - 1)\left(\frac{a_0}{a_0 + l}\right)^{10} + \left(\frac{a_0}{a_0 + l}\right)^{1.8}$$  \hspace{1cm} (2.20)$$

The strain-intensity factor ($K_e$) is defined as:

$$K_e = F(l)\varepsilon \sqrt{\pi l}$$  \hspace{1cm} (2.21)$$

Here $\varepsilon$ is the applied strain; $l$ is the length of a crack extended from a void; $B$ is a constant that has the value of 1 for a void, 2 for a bonded cracked particle, and 0.3 for a debonded particle; $k_r$ is the local elastic-stress-concentration factor for the ellipsoidal void or particle without the crack; and $a_0$ is the half width of a void / particle. An effective stress intensity factor ($K_f$) was also calculated from J-integral. It was shown that the difference between the strain intensity factor and the effective stress intensity factor from a J-integral is less than 10% for materials with a power hardening exponent in the range of 0.05 ~ 0.20. The model works well for both wrought alloys and cast alloys with an $R$ ratio of 0.1 and -1.

Many others have also obtained stress intensity factors of small cracks at notches. It was found that stress fields around notches are quantitatively very similar, if the peak stress at the notch root ($k_S$) and the notch root radius ($\rho$) are the same (Schijve, 1982). For a through-thickness crack at an edge notch, the stress intensity factor was written as:
\[ K_s = (1.1215 - 3.21(l / \rho) + 5.16(l / \rho)^{1.5} - 3.73(l / \rho)^2 + 1.14(l / \rho)^{2.5})k_s \sqrt{\pi a} \]  
(2.22)

The calculation of \( K_s \) agrees with numerical calculations up to \( l / \rho \approx 1 \).

For through-thickness small-cracks at a notch, another version of the stress intensity factor was calculated based on an elastic notch-tip stress field (Kujawski, 1991). For a semi-infinite plate with an edge-notch, the stress-intensity factor for small and long cracks was written as:

\[ K_s = \left[ \frac{1.122f}{1 - 1/k_t} \left( 1 + \frac{l}{\rho} \right)^{0.5} + \left( 1 + 2 \frac{l}{\rho} \right)^{-1.5} \right] S \sqrt{\pi a_0} \]  
(2.23)

\[ K_i = 1.122 \left[ \sqrt{1 + \frac{l}{\rho} \left( \frac{2}{k_t - 1} \right)^2} \right] S \sqrt{\pi a_0} \]  
(2.24)

The value of correction factor, \( f \), is 1 for \( l / \rho < 0.2 \); and for \( l / \rho \geq 0.2 \), the value is:

\[ f = 1 + \tan \left( \frac{\pi}{2k_t} \right) \left( \frac{l}{\rho} - 0.2 \right) \]  
(2.25)

The proposed formula was used to calculate stress intensity factors for small cracks emanating from notches, and the errors from numerical solutions and approximate results are said to be less than 5%.

The accelerated propagation from the notch occurs well beyond the notch plastic zone, and the notch plasticity alone cannot account for the accelerated propagation rate; hence, a model of crack closure for cracks propagating from notches was proposed (Sehitoglu et al., 1996).
where $S_0$ is the stabilized crack opening stress for a crack propagating from a notch; $\sigma_0$ is the uniaxial yield strength; $A$, $B$, $D$ are constants related to stress ratio $R$ and stress concentration factor; and $F$ is a coefficient dependant on crack length.

The crack closure effect in small-crack propagation was also considered in another model (McEvily et al. 1991). In the wake of a newly formed crack, crack closure entails a transition from a crack-opening level of zero for a just-formed crack to a crack-opening level of $K_{op\text{max}}$ as the crack length increases to a macroscopic value. The model was written as:

$$\Delta K_{op} = (1 - e^{-k_a}) (K_{op\text{max}} - K_{min})$$  

(2.27)

where $\Delta K_{op}$ is the range of the stress intensity factor at the opening level in the transition range; $K_{op\text{max}}$ is the opening level for a macroscopic crack and therefore a function of the $R$ value; and $k$ is a material constant which reflects the rate of crack closure development with crack advancement.

A relationship, treating both the propagation of long and short fatigue cracks, was proposed in which the propagation rate in the small crack range is governed by the stress range (including plasticity), rather than by the range of the stress intensity factor, and the fatigue-crack closure in the small-crack regime. The general expression was written as:
where $K_E$ is a material constant which is used to link the endurance limit and the effective range of the stress intensity factor at the threshold level; $Y$ is the flaw shape parameter; and $\Delta K_{\text{eff}}^{th}$ is the effective range of stress intensity factor at the threshold level. This model was applied to the small-crack propagation in squeeze cast Al-Si alloy and W310 cast alloy (Ishihara et al., 1999; Ishihara et al., 2000; Caton et al., 1999). Although the model better correlates the small-crack propagation with the long-crack propagation at high levels of stress intensity range, it still cannot correlate well in the low stress intensity range.

A crack-closure at notch (CCN) model was developed to estimate fatigue life (Ting and Lawrence 1993). The effect of casting defects on the fatigue life was modeled assuming that the total life is the sum of the crack nucleation and propagation life (including the propagation of both short and long cracks). The crack nucleation life was estimated using a simplified form of equation (2.3) for long-life fatigue. A notch-stress-field boundary ($l^*$) was used to divide the approximations for the stress intensity range ($\Delta K$) of a crack from a notch and the function for effective stress intensity ratio ($U(l)$). The total propagation life was estimated by integrating the Paris power law equation from a specified nucleated crack length to the final crack length at a failure:

$$N_f = \int_0^{\Delta K_1} \frac{dl}{C_1(U_1(l)\Delta K_1)^m} + \int_{\Delta K_1}^{\Delta K_2} \frac{dl}{C_2(U_2(l)\Delta K_2)^m}; \tag{2.29}$$
\[ \Delta K_1 = \frac{1.122k_1\Delta S\sqrt{a}}{\sqrt{1 + 4.5\left(\frac{l}{\rho}\right)}} \quad \text{for } l \leq l^* \]  

(2.30)

\[ \Delta K_2 = F(D + l)\Delta S\sqrt{\pi(D + l)} \quad \text{for } l^* \leq l \]  

(2.31)

where \( l \) is the nucleated crack length; \( C_1, C_2, m_1, m_2 \) are constants; \( k \) is the elastic stress concentration factor; \( \rho \) is the radius of notch root; \( D \) is the full notch depth; and \( U_1(l) \) and \( U_2(l) \) are the effective stress intensity ratios for cracks with lengths less and greater than \( l^* \), respectively.

The success of the CCN model is attributed to its treatment of the anomalous crack growth rates of short cracks from notches through the use of crack-closure concepts. But when the defect size is smaller than a certain value, the anomalous short-crack behavior may be affected by the microstructures not described by the CCN concept, and the CCN model may not be applicable.

2.4.4. Nisitani-Goto Small-Crack Model

Small-crack propagation cannot be predicted by LEFM, since it does not account for small-scale yielding associated with small-cracks. Rather the small-crack propagation rate was thought to be determined uniquely by the term \( \sigma_a^n a \), where \( \sigma_a \) is the nominal stress amplitude, \( a \) is the small-crack length and \( n \) is a constant (Nistani et al. 1987;
Nisitani et al., 1992). Moreover, an effective and convenient method, in which the effect of mechanical properties was partly considered, was proposed:

\[
d(2a)/dN = C_j \left(\frac{\sigma_a}{\sigma_y}\right)^n 2a
\]  \hspace{1cm} (2.32)

where \(\sigma_y\) is the cyclic-yield strength, and \(C_j\) and \(n\) are material constants. The underlying assumptions for this model are that the crack-propagation rate is proportional to the plastic-zone size at the tip of the crack and a Dugdale model can be used to describe the plastic zone under unidirectional loading. The plastic-zone size was estimated as:

\[
\frac{\gamma_p}{a} \propto \left(\frac{\sigma}{\sigma_y}\right)^n
\]  \hspace{1cm} (2.33)

where \(\gamma_p\) is the plastic zone size and \(\sigma_y\) is the yield strength. The index \(n\) is 2 for \(\sigma / \sigma_y \ll 1\) and as \(\sigma / \sigma_y\) tends to unity, the value of \(n\) becomes much larger than 2.

It was found that crack propagations in two cast aluminum alloys obey Eq. (2.32) well (Shiozawa et al. 1997) when the ultimate tensile strength was used because of a correlation between the cyclic yield strength with the ultimate tensile strength (Tanaka et al. 1981). The crack-initiation life \((N_i)\) was defined as a crack length of 50 \(\mu m\), and the crack-initiation life was estimated from an evaluation of initiation sites using fracture mechanics and the statistics of extrema.
2.4.5. Microstructural Mechanics (Micro-Mechanics) Model

A crack-tip strain model for the propagation of a small fatigue-crack was applied to squeeze-cast A356 (Chan and Lankford, 1983; Plumtree and Schafer, 1986). In the model, it was assumed that a small crack nucleated at an inclusion within a grain. As the crack propagates and comes within the proximity of a grain boundary, the rate of crack propagation is reduced because the slip is restricted by the presence of the grain boundary. It was also assumed that the plastic strain range at the tip of a small crack can be described by the power law formula:

\[ \Delta \varepsilon_p = C\Delta K^n \]  

(2.34)

where \( n \) and \( C \) are constants and \( \Delta K \) is the stress intensity range.

The influence of grain boundaries on the crack-tip plastic-strain range depends on the crystallographic orientations of the neighboring grains, as well as the distance of the crack tip from the nearest grain boundary. To account for grain-boundary effects, the fractional change of the crack-tip plastic-strain due to grain boundary blocking of crack-tip slip was postulated as:

\[ \frac{\delta \Delta \varepsilon_p}{\Delta \varepsilon_p} = -k(\phi)(\frac{D - 2X}{D})^m \]  

(2.35)

\[ k(\phi) = 1 - \frac{\tau_B}{\tau_A} \]  

(2.36)

where \( m \) is constant; \( X \) is the distance of the crack-tip from the nearest grain boundary; \( D \) is the grain diameter; \( k(\phi) \) is a function related to the relative crystallographic orientation
of the two grains; \( \tau_A \) is the resolved shear stress of particular slip system in grain A where the crack nucleated; and \( \tau_B \) is the resolved shear stress of particular slip system in grain B where the crack is headed. The maximum value of \( k(\phi) \) is unity and represents the case where the slip orientation of the next grain is most unfavorable and the crack arrests. The minimum value of \( k(\phi) \) is zero and corresponds to the case where the orientation of the two grains is similar so \( \tau_A = \tau_B \). The propagation rate was then presented as:

\[
\frac{da}{dN} = \left( \frac{\Delta X'C}{\varepsilon_p^*} \right) C_1 \Delta K^{\prime} \left[ 1 - k(\phi) \left( \frac{D - 2X'}{D} \right)^m \right]
\]  

(2.37)

where \( \varepsilon_p^* \) is the critical value of local plastic strain; and \( \Delta X' \) is the size of crack tip element.

When small-crack propagation in squeeze-cast A356 was expressed using the model of Eq. (2.37), it was shown the barriers were significant at low strain ranges (Plumtree and Schafer, 1986). The model correlated well with the experimental results when the values of \( k(\phi) \) were empirically taken as 0 for the high-strain range, and 0.9 for the low-strain range used in the experiments.

Another micro-mechanics analysis for small-crack propagation in fully annealed 0.4% C steel in torsion was suggested assuming that the driving force for crack propagation was provided by the energy of the slip band (De Los Rios, et al., 1985). The assumption in the model is that as a crack propagates the change of the energy of the crack equals the change of the slip-band energy and follows the Griffith approach. The model developed
described short-crack propagation in shear, involving both the macro-mechanical and micro-crystallographic parameters, and represented the experimental observations. The effects of grain orientations and the distance of crack tip to the grain boundary on the effective slip-band length were included in a way similar to the previously discussed model (Chan and Lankford, 1983). The final equation was written as:

\[
\frac{da}{dN} = f(2\pi a D)^{1/2} \left\{1 - k(\phi) \left( \frac{D - 2X}{D} \right)^{m/2} \right\} \cdot \frac{\tau}{\mu}
\]  

(2.38)

where \( f \) is a constant for the material; \( \tau \) is half the stress range; \( \mu \) is the shear modulus; and \( a \) is the crack length.

In order to link small- and large-crack propagation laws, a generalized model was proposed (Hamm and Johnson, 1999). The small-crack propagation model was based on that developed by Wang (Wang, 1996). The model developed by Hamm and Johnson was shown to provide a good correlation to experimental results for Ti-6Al-4V under various maximum stresses at a stress ratio of 0.4. The applicability of the model is directed towards small cracks that start from geometric features and propagate through stress gradients to eventually become large cracks under normal LEFM conditions. Transition functions are used to transform small-crack plastic-zone sizes and crack-propagation law exponents to those predicted by linear elastic fracture mechanics. It was based on crack-propagation rate laws with a crack-tip plastic-zone size formulation:

\[
\frac{da}{dN} \propto \rho^k
\]  

(2.39)
where \( \rho \) is the size of the plastic zone; and \( \lambda \) is the crack-propagation exponent. The crack-propagation exponent \( (\lambda) \) and its evolution function \( (\psi') \) are written as:

\[
\lambda = \left[ m - \left( \frac{B}{2} \right) \right] \psi' + \left( \frac{B}{2} \right)
\]

\[
\psi' = \left\{ \frac{1}{2} \left[ 1 - \tanh \left( \frac{2a}{d} - \eta \alpha \right) \right] \right\}
\]

where \( m \) is the small-crack propagation law exponent; \( B \) is the Paris propagation law exponent; \( a \) is the half crack length; \( d \) is the microstructural unit size; \( \alpha \) is the plastic zone transition parameter; and \( \eta \) is the shape factor to allow flexibility in determining the shape of the crack propagation rate curve. For small cracks and large cracks in the Paris regime, the obtained propagation rates are respectively written as:

\[
\frac{da}{dN} \propto \rho_s^m
\]

\[
\frac{da}{dN} \propto \rho_l^{8/2}
\]

where \( \rho_s \) is small-crack tip plastic-zone size; and \( \rho_l \) is the large-crack tip plastic zone size.

The crack-tip plastic-zone size was expressed as:

\[
\frac{\rho}{a} = \Theta \psi + \Phi
\]

where \( \Theta \) is for the small-crack behavior; \( \Phi \) is for the large-crack behavior; and \( \psi \) is the plastic-zone evolution function that depends on the crack length, the microstructure scale-length, and an empirical parameter, \( \alpha \).
2.5. OUTLINE OF RESEARCH

Fatigue behavior is a very important engineering issue in the use of premium quality aluminum castings. Microstructural effects on fatigue behavior in A356 casting alloy needs to be further understood; an improved small-crack propagation model with microstructural effects should be applied. Most existing models and simulations assume that the initiation life of cast aluminum can be neglected in the total fatigue life; hence, the quantitative prediction of fatigue life is made by calculating the long-crack propagation life. The assumption that the initiation stage is negligible seems reasonable in the low-cycle regime, but this has not been clearly established for high-cycle fatigue with lives greater than $10^6$ cycles. Information on the effects of pores and eutectic constituents on small-crack propagation is needed. An improved comprehensive understanding of the process of fatigue-crack initiation and small-crack propagation in cast alloys is still necessary and is important for the prediction of fatigue lives. Physics-based crack initiation and small-crack propagation models should be included in the simulation of fatigue life in order to predict correctly the total fatigue life.

2.5.1. Effects of Microstructures on Small-Crack Initiation and Propagation

Pores are usually the crack initiation sites in the fatigue of cast aluminum alloys and are extremely detrimental to fatigue lives. Which aspects of the microstructural features are most important in cast aluminum and how microstructural scales influence the fatigue must be clarified. In this dissertation, fatigue lives were measured and surface replicas were taken during the fatigue tests to reveal the effects of microstructural heterogeneities
on development of small cracks, and the interactions of cracks with the interdendritic boundary and silicon particles. Specimens from directionally solidified ingots were studied for these purposes.

Little information is available on the microstructural effects in small-crack propagation behavior of cast aluminum alloy. It is recognized that, a small-crack initiating at an inclusion or a pore may decelerate when the plastic zone in front of the crack interacts with interdendritic/intergranular boundaries and/or silicon particles; a small-crack may also not close for as much of the loading cycle as do long cracks and it grows faster than expected from LEFM theory. In this research, the effects of porosity, SDAS and Sr-modification on small-crack propagation behavior have been carried out.

In previous estimations of the fatigue life in casting alloys, the transition from a small crack to a long crack was calculated by equating the short crack stress intensity factor range and the long crack intensity range. This has never been proven by experiments. In this dissertation, the transition crack size, below which the long crack propagation rate does not apply, is determined.

2.5.2. Effects of Hipping and Sr-Modification on Fatigue Behavior

Although silicon particles in Sr-modified casting alloy have been found to decelerate the small crack propagation, the use of Sr-modification is discouraged sometimes because the added porosity associated with Sr-modification nullifies the beneficial effect of the
improvement in the silicon morphology. The usual casting technology cannot produce castings without pores, but it is believed that improved processing could result in mitigating both pores and oxides. When the SDAS is relatively small, the pores may not serve as crack-initiation sites, so that the effect of silicon size and morphology (Sr-modification) would be more important in enhancing fatigue lives. Hipping can be successfully applied to close the pores; in this research, fatigue specimens from casting plates with/without Sr-modifications were hipped to avoid the porosity and then tested to compare the effect of different silicon particle morphology on small-crack propagation by observing the replicas. The effects of hipping and Sr-modification on fatigue behavior of A356.2 specimens obtained from plates cast in permanent molds are presented and discussed in connection with crack-initiation, small-crack propagation and fatigue life.

2.5.3. Modeling the Small-Crack Propagation

In this dissertation, models based on LEFM and micro-mechanics were applied to characterize the crack propagation. Since long-crack propagation is modeled reasonably well, emphasis in this dissertation is on small-crack behavior. The model for small-crack propagation rate has been correlated directly to the experimental results where the small-crack propagation rates may decelerate or accelerate when the small crack interacts with dendrite boundaries and silicon particles. Microstructure findings and small-crack information gleaned from replicas were used for evaluation of the efficiency of the model.
In view of the above, the research in this dissertation determined the microstructural effects of pore, SDAS, hipping, and Sr-modification on the fatigue behavior including crack initiation, small-crack propagation and fatigue life. Finally fracture mechanics and micromechanics models were applied to characterize small-crack propagation.
CHAPTER 3. EXPERIMENTAL DETAILS

An experimental study on effects of porosity, SDAS, Sr-modification, and hipping on the fatigue behavior of A356.2 has been conducted. To accomplish these tasks, carefully designed fatigue tests, replica experiments and microstructural analyses were conducted to reveal fatigue damage and interactions between microstructures and small-crack propagations during fatigue testing. This included the microstructural effects on fatigue behavior of specimens removed from a cast ingot with gradients in SDAS and porosity. After fatigue failure, the crack-initiation-sites were located and characterized. The small-crack propagation rates were measured by removing replicas from the specimens during cyclic loading. Specimens were also removed from plate castings made in a permanent mold; with these specimens the effects of Sr-modification, hydrogen content, and hipping on fatigue lives and small-crack propagation were investigated.

3.1. MATERIALS AND SPECIMEN PREPARATION

3.1.1. Directionally Solidified Ingots

Two types of castings were made at the Alcoa Technical Center in Alcoa Center, Pennsylvania: directionally solidified ingots and plates that were cast in a permanent mold. The casting-ingots were directionally solidified to effect a gradient in secondary dendrite arm spacing (SDAS). The mold used to cast the A356.2 ingots is shown in Fig. 3.1.
The mold was wrapped in ceramic wool, and direct water-cooling was imposed at the bottom surface. The cooling rate during solidification varied from approximately 10 Ks$^{-1}$ at the bottom to only 0.3 Ks$^{-1}$ at about 150 mm from the bottom-chill. Therefore, a wide variation in cooling rates and, hence, secondary dendrite arm spacings were obtained in a single ingot. The ingots were 95 mm wide, 28 mm deep, and 300 mm high. The chemical composition of the alloy is listed in Table 3.1.

In preparation for casting, the alloy with 0.10% Ti was melted, heated to 715° C and flux-degassed with Ar for eight minutes. A small amount of additional grain refiner (Al-3Ti-
Table 3.1. Compositions of directionally solidified Al-7Si-0.3Mg (A356.2) ingots.

<table>
<thead>
<tr>
<th>Charge</th>
<th>Si</th>
<th>Fe</th>
<th>Mg</th>
<th>Ti</th>
<th>B</th>
<th>Cu</th>
<th>Sr</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>6.85</td>
<td>0.03</td>
<td>0.36</td>
<td>0.10</td>
<td>0.0001</td>
<td>0.01</td>
<td>0.006</td>
<td>Remainder</td>
</tr>
<tr>
<td>2</td>
<td>6.87</td>
<td>0.03</td>
<td>0.37</td>
<td>0.10</td>
<td>0.0001</td>
<td>0.01</td>
<td>0.007</td>
<td></td>
</tr>
</tbody>
</table>

1B) at 0.003% Ti level and Al-15 Sr modifier at 0.008% Sr-level were added 2 min. prior to the end of fluxing. The cooling water was actuated before the melt was poured into the mold, which was tilted during pouring to reduce turbulence. The copper chill-plate at the bottom of the mold (Fig. 3.1) was removed 20 s after pouring, so that the cooling water directly impinged on bottom surface of the ingot (i.e., the chill-surface). The resulting secondary dendrite arm spacing was ~15 μm near the chill-surface and more than 50 μm in the top portion of the ingot.

3.1.2. Plate-Castings

A cast iron book-mold was used for casting grain-refined test-plates with different levels of Sr and hydrogen. The plates were 1/2 x 7 x 8 (in inches). The molds were tilted during pouring to minimize the formation of dross. Five groups of plates (three in each group), with the compositions listed in Table 3.2, were cast. The hydrogen contents of the plates were measured by the Leco method with an accuracy of 0.03 cc/100gm. The melt was flux-degassed with Ar, grain refined with an Al-3Ti-1B master alloy, and modified with Al-15 Sr alloy.
Table 3.2. Compositions of permanent-mold cast Al-7 Si-0.3 Mg (A356.2) plates.

<table>
<thead>
<tr>
<th>Group</th>
<th>Si</th>
<th>Fe</th>
<th>Mg</th>
<th>Ti</th>
<th>B</th>
<th>Sr</th>
<th>H₂ (cc/100 g)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>6.65</td>
<td>0.09</td>
<td>0.27</td>
<td>0.12</td>
<td>0.001</td>
<td>0.000</td>
<td>0.17</td>
</tr>
<tr>
<td>2</td>
<td>6.65</td>
<td>0.09</td>
<td>0.27</td>
<td>0.12</td>
<td>0.001</td>
<td>0.004</td>
<td>0.19</td>
</tr>
<tr>
<td>3</td>
<td>6.58</td>
<td>0.09</td>
<td>0.26</td>
<td>0.11</td>
<td>0.001</td>
<td>0.006</td>
<td>0.21</td>
</tr>
<tr>
<td>4</td>
<td>6.58</td>
<td>0.09</td>
<td>0.26</td>
<td>0.12</td>
<td>0.001</td>
<td>0.005</td>
<td>0.22</td>
</tr>
<tr>
<td>5</td>
<td>6.58</td>
<td>0.09</td>
<td>0.26</td>
<td>0.12</td>
<td>0.001</td>
<td>0.004</td>
<td>0.31</td>
</tr>
</tbody>
</table>

The casting-ingots, made in the mold of Fig. 3.1, were cut into slices ~ 4 mm thick, with the flat surfaces of each slice parallel to the bottom surface of the casting-ingot. The plates made in the book-mold were cut into two slices, with the cut parallel to the large flat surfaces of the plates. A portion of each slice was kept in its as-cast state for documenting the microstructures of the as-cast alloy. In order to study the effect of Sr-modification on small-crack propagation, some slices from the cast plates were first subjected to hot isostatic pressing ("hipping") to avoid the adverse influence of pores then heat-treated. The hipping conditions were 2 hours at 540°C and 100 MPa. The slices were then heat-treated to the T6 condition (540°C for 12 hours, quenched into water at less than 71°C, and aged within 2 hours at 155°C for 4 hours). The heat-treated slices were then milled to a thickness of 3 mm, while maintaining a depth of cut less than 0.3 mm per pass, and machined into tensile specimens (ASTM E8-96a), low-cycle...

* Hipping was done at Kittyhawk Products in Garden Grove, California.
† The heat-treating was done by Phoenix Heat Treating, Phoenix, Arizona.
fatigue specimens (ASTM E606-92) and high-cycle fatigue specimens (ASTM E466-82). The configurations of axial and bending fatigue specimens were as shown in Fig. 3.2. The specimens were polished using silicon carbide abrasive papers and alumina paste/colloidal silica to a final finish of 0.2 μm. The direction of the final polishing operation was along the specimen length, in order to eliminate possible stress concentration sites.

3.2. FATIGUE TESTING AND REPLICA STUDY

3.2.1. Fatigue Testing

The axial-fatigue experiments were conducted in a laboratory environment at room temperature on a closed-loop servo-hydraulic testing system (Materials Test System MTS 810). In addition to axial-fatigue tests, many specimens were tested in an instrument for reciprocating-bending fatigue. The axial-fatigue and bending-fatigue testing setups are shown in Fig. 3.3.

Most of the specimens from the directionally solidified ingots (Fig. 3.1) were tested in axial-fatigue. The specimens for low-cycle fatigue were done with a constant-amplitude, using displacement-control mode. High-cycle fatigue tests were conducted using the same system in load-control mode. The frequency was set at 10 Hz for low-cycle fatigue and 20 Hz for high-cycle fatigue. The strain ratio for low-cycle fatigue was set to -1.0 (complete reversal); the strain amplitude was 0.0047. In the high-cycle fatigue experiments, the stress ratios were -1.0, 0.1, and 0.2, with the maximum/minimum
stresses set to 85/-85 MPa, 175/17.5 MPa, and 130/26 MPa, respectively. The high-cycle stress levels were far below the 0.2% yield strength of ~220 MPa for the alloy.

Fig. 3.2. Specimen geometries used in fatigue testing (all dimensions in mm): (a) reciprocating-bending; (b) axial, high-cycle; and (c) axial, low-cycle.
Fig. 3.3. Experimental setups for fatigue testing: (a) testing instrument that was used for axial-fatigue; and (b) testing instrument that effected reciprocating-bending type of fatigue.
For specimens from the cast plates, both tensile testing and fatigue testing were conducted on a closed-loop servo-hydraulic testing machine. The tensile testing was in displacement-control mode, with a strain rate of $1.1 \times 10^3$/s.; low cycle fatigue testing was in the total-strain control mode at 0.1-0.5 Hz to avoid heating by the plastic deformation during low-cycle fatigue; and high cycle, constant-amplitude axial fatigue tests were conducted in the load-control mode at 20 Hz, which is the maximum frequency of the testing machine. An alignment fixture and a hydraulic grip were used to keep the alignment in required range. An extensometer, strain-range cartridge, and load cell provided the histories of strain and load for axial fatigue. The failure criteria were 10% drop of the tensile load for low-cycle fatigue and total fracture for high-cycle fatigue.

The bending-fatigue experiments were conducted at 20 Hz for strain ratios ($R$) of -1.0 (complete reversal) and 0.15 and at 25 Hz for $R = 0.1$. A strain gage was attached to the center of the gage section of each specimen to record the cyclic strain during the test, and a digital-storage oscilloscope was used to record the strain data from the axially mounted strain gage. Restricted by the width of the ingot, the specimens used in this research were smaller than commonly used. In order to ensure a uniform strain distribution in the gage section, a finite element analysis was performed, and many trial tests were conducted using specimens of slightly different geometries taken from a wrought aluminum alloy (5052). The final geometry is shown in Fig. 3.2 as the reciprocating-bending specimen. In the trials with the 5052-specimens, machined to the final geometry and tested in the low-cycle fatigue range, multi-fatigue cracks propagated simultaneously within the gage-
sections of the specimens. This indicated that a uniform strain-state in the gage section had, indeed, been achieved.

For the experiments with the strain ratio $R$ set to -1.0, the crank-tip deflection of the testing instrument was adjusted to obtain a maximum strain of 0.004 for low-cycle fatigue and a maximum strain of 0.0016 for high-cycle fatigue. In order to characterize the fatigue-crack initiation from the fracture surface, a strain ratio of 0.1 was also used in the high-cycle fatigue experiments and 0.15 in the low-cycle experiments to avoid the extent of damage on the fracture surfaces when the applied strain was completely reversed. The maximum strain was kept at the same level as in the case of reversed bending for high-cycle fatigue (0.0016). The maximum strain was increased to 0.0082 to achieve $R = 0.15$, when testing in the low-cycle range.

3.2.2. Replica Study of Small-Crack Behavior

Due to its reliability, a replication method (Swain, 1992) was used to detect small-crack initiation and propagation. The initiation life of a crack forming at a pore, an oxide, or eutectic constituent and the small-crack propagation were studied by viewing the replicas by light microscopy and scanning electron microscopy (SEM). Some replicas were made during both bending- and axial-fatigue experiments.

To monitor small-crack propagation, specimens polished to a 0.05 μm finish were used. The polished surfaces of fatigue specimens were lightly etched with a 0.5% HF aqueous
solution to enhance the replication of the microstructures. Fatigue cycling was interrupted periodically for stripping the replica and reapplying the replica tape. Each specimen was held under a tensile load (~60% of the maximum load) to maintain the crack in its open position while the surface was replicated. Acetone was applied to clean the surface of the specimen and to soften the surface layer of the cellulose acetate replicating-tape. Many replicas were made for each test and reviewed under the microscope from the last one to the first to trace the crack back to its initiation site. Crack length was measured from the replicas and corrected for shrinkage. Then the propagation rates were calculated using the secant method.

3.3. RADIOGRAPHY AND MICROSTRUCTURE ANALYSIS

3.3.1. Radiography

Radiographs were made before mechanical testing to reveal the porosity (maximum size and number density) in the fatigue specimens removed from the bottom to the top of the ingots, shown in Fig. 3.1. The plates cast in the permanent mold were also radiographed to avoid the apparent pores when specimens were extracted. A radiography unit with a tungsten target of an effective focal spot of 2.0 mm was used. Pores visible on the radiographs at a magnification of 20 times were counted within a grid of an area of 34.10 mm$^2$ (5.84 x 5.84 mm). The number of pores in a total of 12 grid areas within the gage section along each specimen was counted, and then the number density was averaged.

\footnote{The tape was from Ernest F. Fullam Inc. of Latham NY, and the thickness of the tape is 0.12 mm.}
The size of the largest pore in the radiograph was further measured at a magnification of 38.75 times.

3.3.2. Microstructure and Fractography

Light microscopy was conducted to examine both the specimens and the replicas. The line intercept method was used to measure the secondary dendrite arm spacing (SDAS) and the grain size on samples removed from the as-cast ingots. When grain size was measured, the polished surfaces of specimens were etched with a 5% CuCl₂ aqueous solution, rinsed with distilled water, and then swabbed with 30% dilute HNO₃ solution to remove the Cu deposits. An image analysis system were used to measure sizes and distributions of pores and of the silicon particles in the eutectic constituents. Although significant differences in the sizes, distribution, and amount of porosity were found, the morphology of the pores in the plates cast in the permanent mold was similar, in that the pores were interdendritic and never spherical. Presumably, the pores formed after solidification commenced.

The fracture surfaces of the failed fatigue samples were examined using either a light microscope or a scanning electron microscope (SEM) to investigate the crack-initiation sites, propagation paths, and fracture modes. A stereo SEM technique was also used to observe the crack path on the fractures. Both the maximum lengths and the areas of the projections of the initiation sites (porosity, oxide inclusions, and eutectics) were measured on the thermal prints of the SEM images.
CHAPTER 4. EFFECTS OF SOME DEFFECTS AND SDAS ON THE FATIGUE BEHAVIOR OF A356.2 CAST ALUMINUM ALLOY

This chapter presents a systematic investigation of the effects of the sizes of microstructural constituents, as controlled by cooling rate during solidification on the fatigue behavior of a cast and heat-treated aluminum alloy, A356.2-T6 (Al-7Si-0.3Mg). Both bending- and axial- fatigue testing were conducted. The replica method was used during fatigue to detect small-crack initiation and propagation. The characterizations of the microstructural features (pores, secondary dendrite arm spacing, grain size, and Si-particles) in A356.2 specimens removed from a directionally solidified casting ingot are presented. The microstructural effects on fatigue life, fatigue-crack initiation and small-crack propagation, and the interaction of microstructure with the propagation of fatigue cracks were studied.

4.1. MICROSTRUCTURAL VARIATION ALONG INGOT HEIGHT

Microstructures (secondary dendrite arm spacing, grain size, porosity, and Si-particle morphology) of the alloy were examined. Typical microstructures of as-cast A356.2 and T6 heat-treated A356.2 along the ingot height, at the bottom, in the middle, and at the top of the ingot are shown in Fig. 4.1. After T6 heat-treatment, the silicon particles within the eutectic constituents were spherodized as expected. At the bottom of the ingot,
Fig. 4.1. Microstructures in A356.2 cast alloy. As-cast: (a) $SDAS = 15 \, \mu m$; (b) $SDAS = 33 \, \mu m$; (c) $SDAS = 57 \, \mu m$; T6 heat-treated: (d) $SDAS = 15 \, \mu m$; (e) $SDAS = 32 \, \mu m$; and (f) $SDAS = 54 \, \mu m$. 
microstructures with smaller *SDAS* and silicon particles are seen. At the top of the ingot, microstructures with larger *SDAS* and silicon particles are found.

![Microstructures with pores and oxide films](image)

Fig. 4.2. Pores and oxide films found in the microstructures: (a) pores observed at a *SDAS* of 57 μm; and (b) pores and oxide films observed at a *SDAS* of 22 μm.

In the microstructures, both porosities and oxide films were found. More large pores were found in microstructures with large *SDAS*. Although the mold was tilted during filling to avoid turbulent filling, oxide films were still observed at the bottom of the ingot. Later it was found that these oxide films were active crack-initiation sites. Typical porosity and oxide films are shown in Figs. 4.2(a) and (b). As is shown in Fig. 4.2(b), oxide films were often found along with pores. Oxide films are active heterogeneous nucleation sites for pores during solidification, and their effect on the formation of pores had been discussed previously (Poirier, 1998).

* *SDAS* refers to secondary dendrite arm spacing.*
4.1.1. Variations of Secondary Dendrite Arm Spacings (SDAS) and Grain Size along Ingot Height.

Figure 4.3 shows the variation of SDAS along the ingot height, measured upwards from the bottom chill. Below a distance of 50 mm from the bottom chill, the SDAS increases slightly from 15 to 20 μm as the height increases. Twenty seconds after pouring, a bottom plate was pulled so that the casting butt was exposed to direct water impingement. Thus the variation in SDAS shows a change in slope at about 40 mm from the chilled surface, in Fig. 4.3. Beyond that point, the SDAS increases almost linearly from 20 μm to 60 μm at 250 mm. According to a relationship (Spear and Gardner 1963) of SDAS (μm) as a function of cooling rate \( R \) (K s\(^{-1}\)) in solidification, the cooling rate was calculated,

\[
R = \left( \frac{41.6}{SDAS} \right)^{0.34}
\]  

(4.1)

and shown in Fig. 4.3. The calculated cooling rate varies from 17 K s\(^{-1}\) at the bottom of the casting ingot to 0.4 K s\(^{-1}\) at 250 mm from the bottom.

Figure 4.4 shows the variation of grain size along the ingot height. Below a distance of 50 mm from the bottom chill, the grain size is \( \sim 417 \) μm, then the grain size increases to \( \sim 890 \) μm as the height increases to 250 mm. Unlike the SDAS, which is controlled by the solidification time or the cooling rate, grain size is influenced by a large number of factors (Campbell, 1991; Hutt and St. John, 1998).
Fig. 4.3. Variations of SDAS and cooling rate along ingot height.

Fig. 4.4. Variations of grain size along ingot height.
It has been reported in directionally solidified 6061 and 7075 wrought alloys that the increased cooling rate results in finer grain size (Chu, 2002). The effect of cooling rate on grain size had been explained by the extent of the undercooling achieved in the melt ahead of the solidification front. Increased cooling rate yields an increased undercooling of the melt and thus enables the formation of new grains (Vatne, 1999). The efficiency of heterogeneous nuclei is strongly influenced by the initial rate of development of undercooling in front of a growing grain (Easton and St. John, 2001), and thus slower solidification rate along the ingot height decreases the initial rate of the development of undercooling and decreases the efficiency of the heterogeneous nuclei and result in greater grain size along the ingot height. On the other hand, some of the crystals nucleated on the mould walls do not survive the superheat, and the number of these crystals decreases along the ingot height. The settlement of the potential heterogeneous nuclei (TiB₂) in the melt during solidification also decreases the number of nuclei in the upper part of the ingot, and results in a greater grain size.

4.1.2. Porosity Distribution as a Function of SDAS

The variations of maximum length and the number density of pores, visible on radiographs for the fatigue specimens (3-mm thick) with different SDAS, are summarized in Figs. 4.5(a) and (b). The pores are obvious when the SDAS > 28 μm; no visible pores, for specimens with a SDAS < 28 μm, could be detected even under a magnification of 40 times. As the SDAS increases from 28 to 48 μm, the images of the pores become clearer,
Fig. 4.5. Pores detected by radiography in the casting-ingot: (a) variation of maximum length; and (b) number density of visible pores.
and the maximum length increases from about 0.08 mm (80 μm) to 0.7 mm (Fig. 4.5(a)). Using the radiographic technique, no pores were detected in specimens that have SDAS less than 28 μm. Fig. 4.5(b) shows that the number density of visible pores is between approximately 0.4 to 1.1 mm$^{-3}$ for most of the specimens, with four of the number densities less than 0.2 mm$^{-3}$, grouped at a SDAS of approximately 30 μm.

Although radiography did not detect the pores at smaller SDAS, using metallographic techniques, pores were found that were 25 to 50 μm in equivalent diameter for SDAS from 16 to 30 μm, in a grain-refined and modified A356.2 ingot, which was cast in the same manner as the ingots studied here (Fang and Granger 1989). However, such small pores are not active in terms of crack initiation, as is discussed later. Results in Fig. 4.5(a) show that, as the SDAS increases (cooling rate decreases), the maximum length of the pores increases. However, when the SDAS is greater than about 34 μm, the visible number density decreases. These results are in agreement with the results reported from others (Shivkumar et al., 1991; Tynelius et al., 1993), where it was found that the maximum porosity area increases as the SDAS increases, and that, when hydrogen content is above 0.22 ml/100g in the alloy, the areal pore density decreases as the SDAS increases. It was explained that gas pores nucleated at low volume fractions of solid have longer period of growth than those pores nucleated later in the solidification. Thus, the pore population tends to be skewed towards larger pore sizes at high gas levels, and a much greater number of very small pores at lower gas levels.
4.1.3. Maximum Length of Eutectic Silicon Particle as a Function of SDAS

Figure 4.6 shows the variations of the maximum diameters of the silicon particles with the SDAS along the ingot height. A control area of 24,687 \( \mu m^2 \) was analyzed at a magnification of 1120 \( \times \). The average of the maximum diameters increases from \(~2.5 \, \mu m\) at a SDAS of \(15 \, \mu m\) to \(4.0 \, \mu m\) at a SDAS of \(~25 \, \mu m\) and then levels off at \(~4 \, \mu m\) when the SDAS is above \(25 \, \mu m\). The maximum diameter of a single silicon particle increases from \(~6.6 \, \mu m\) at a SDAS of \(15 \, \mu m\) to \(~10.0 \, \mu m\) at a SDAS of \(~25 \, \mu m\) and then levels off at a value of \(~10.0 \, \mu m\) when the SDAS is above \(25 \, \mu m\). The minimum diameter of a single silicon particle is approximately \(1 \, \mu m\) at all values of SDAS.

Fig. 4.6. Variation of diameter of Si-particles with secondary dendrite arm spacing.
4.2. ROCKWELL HARDNESS AS A FUNCTION OF SDAS

Figure 4.7(a) shows the variations of the Rockwell E hardness with the SDAS. The average hardness decreases from ~86 at a SDAS of 17 μm to ~78 at a SDAS of ~55 μm. As the SDAS increases, the standard deviation of measurements increases from 0.65 to 6.7, and the minimum hardness decreases from 85.2 to 66, indicating that specimens with large SDAS have inferior mechanical properties. The ultimate tensile strength and elongation of either hipped or non-hipped A356-T6 alloys were found remarkably improved when the SDAS was decreased (Baily, 1965; Flemings, 1974; Wickberg et al., 1984; Campbell, 1991; Boileau et al., 1997). The ultimate tensile strength and Rockwell hardness correlate closely with each other, and according to a correlation suggested by Poirier (Poirier 2002), the calculated ultimate tensile strength is shown in Fig 4.7(b). The calculated ultimate tensile strength decreases from ~ 286 MPa at a SDAS of 20 μm to ~247 MPa at a SDAS of 55 μm.

The calculated ultimate tensile strength (UTS, MPa) is related with SDAS (μm) by using the Hall-Petch relationship. The relationship is

\[
UTS = 215.65 + \frac{321.26}{\sqrt{SDAS}}
\]  

(4.2)

The square of the correlation coefficient (i.e., \(R^2\)) is 0.6291.
Fig. 4.7. Variations of hardness and ultimate tensile strength with the SDAS: (a) Rockwell E; and (b) calculated ultimate tensile strength.
4.3. FATIGUE LIFE

4.3.1. Loads and Strains in Bending Fatigue

After installing the specimen on the bending-fatigue instrument, weights were used to measure the loads needed for the maximum bending-deflections for both low-cycle and high-cycle fatigue testing. Strain gages were used to measure the surface strain on the specimen. The loads normalized by the specimen thickness are shown in Fig. 4.8. The maximum strains for specimens with different SDAS are shown in Fig. 4.9.

![Fig. 4.8. Loads needed to induce deflection settings on the bending-fatigue instrument for specimens with different SDAS. For high-cycle fatigue, the maximum strain is 0.00159 ± 0.00007, the minimum strain is -0.00157 ± 0.00013, and the strain ratio is -0.987 ± 0.055. For low-cycle fatigue, the maximum strain is 0.00391 ± 0.00022, the minimum strain is -0.00388 ± 0.00022, and the strain ratio is -0.991 ± 0.023.](image-url)
Fig. 4.9. A comparison between the maximum strains measured at first cycle and calculated from loads for HCF and LCF at strain ratios of ~ -1.0.

For high-cycle fatigue, the deflection was set in the elastic deformation range. Specimens with different SDAS have the same maximum strain (Fig. 4.9), and the load needed to reach the strain (or the deflection) tends to be similar as the SDAS changes from 15 μm to 60 μm, as shown by the lower curve in Fig. 4.8.

For low-cycle fatigue, the deflection was set in the plastic deformation range. Specimens with different SDAS tend to have the same maximum strains, but specimens with small SDAS tend to show a higher load than specimen with large SDAS when SDAS is larger than 25 μm. This is in agreement with the previous result on hardness, as SDAS increases, specimens with larger SDAS show smaller hardnesses (hence, lower ultimate tensile
strength and yield strength). Consequently when loading was applied with plastic deformation, a smaller load was needed for specimens with the larger $SDAS$ than that for specimens with the smaller $SDAS$.

Assuming a linear variation of strain with distance from the neutral plane to the surface in bending and the Ramberg-Osgood stress-strain relationship (Dowling, 1993), plastic bending analysis was conducted. The maximum strains on the surface of the bending beam calculated are shown in Fig. 4.9, together with those values measured using the strain gages. For the high-cycle fatigue settings, the average maximum strain measured for all specimens with $R = -1$ regardless of the $SDAS$ is $0.00159 \pm 0.00007$; the calculated is $0.00151 \pm 0.00015$. The measured strains agree with the calculated strains at most are only 11% lower than the measured strains. For low-cycle fatigue settings, the average maximum strain measured for all specimens with $R = -1$ is $0.00391 \pm 0.00022$, but the calculated strains are $0.00300 \pm 0.00029$. These are at least 25% lower than the measured ones. The rigid specimen holder influences the deformation of the fatigue specimen and was not considered in the calculation; this could account for the large difference between the calculated and the measured strains especially in low-cycle loading condition.

The maximum, minimum, and amplitude strains measured for bending-fatigue with an average strain ratio of $\sim 0.18$ are shown in Fig. 4.10. Maximum, minimum and amplitude strains are relatively independent of the $SDAS$. The maximum strain, minimum strain,
and $R$ ratio are $0.00866 \pm 0.00035$, $0.00156 \pm 0.00355$, and $0.1789 \pm 0.0399$, respectively.

![Graph showing strain vs. secondary dendrite arm spacing](image)

Fig. 4.10. Strains measured at first cycle for LCF with maximum strain of 0.00866 and strain ratio of 0.1789.

4.3.2. Bending-Fatigue Life

In this section, references are made to the average values of the $R$ ratios and the maximum strains. The variation of the bending-fatigue life as a function of SDAS is summarized in Fig. 4.11. Under low-cycle fatigue conditions at $R = -1.0$, the fatigue life is approximately $10^4$ cycles when the secondary dendrite arm spacing is 15 $\mu$m, near the bottom-chill. The life decreases only slightly as SDAS increases to 30 $\mu$m, and the life then decreases to $\sim 3 \times 10^3$ cycles.
Fig. 4.11. Variation of bending-fatigue life with the SDAS. Arrows indicate specimens for which the tests were terminated before failure. Low-cycle fatigue specimens have fatigue lives of less than $10^5$ cycles; high-cycle fatigue specimens have fatigue lives greater than $10^5$ cycles.

As SDAS increases to about 55 µm in the upper part of the ingot. In the case of $R = 0.18$, the fatigue life fluctuates around $2 \times 10^4$ cycles when $SDAS < 30$ µm and drops to about $8 \times 10^3$ cycles when $SDAS > 30$ µm. The specimen with a SDAS of 53 µm had a fatigue life of only 200 cycles. An examination of the fracture showed that the crack initiated at a large oxide inclusion near the surface.

Under high-cycle fatigue with $R = -1.0$, the fatigue life decreases from ~1.5 million cycles at $SDAS < 30$ µm to $2 \times 10^5$ cycles at $SDAS$ of 55 µm. When the strain ratio was set to 0.1, while maintaining the maximum strain value at the same level (~0.0016) as the high-cycle condition, the fatigue life of specimens with $SDAS < 27$ µm exceeded $30 \times$
10^6 cycles without failure or observable damage. Three specimens with \(SDAS\) of 16 to 18 \(\mu m\) lasted for almost 10^8 cycles; two of the three did not fail when the tests were interrupted. The specimen with a \(SDAS\) of 25 \(\mu m\) failed at just over 1 million cycles, however. This specimen contained a large oxide inclusion on the surface, which initiated the fatigue crack prematurely. Other specimens with 20 < \(SDAS\) < 27 \(\mu m\) lasted for well over 10^7 cycles. The fatigue lives of specimens with \(SDAS\) > 29 \(\mu m\) were between 7 \times 10^5 and 2 \times 10^6 cycles. These results indicate that the amplitude of the alternating strain, not just the maximum strain, significantly affects the fatigue life of the alloy.

![Graph](image)

**Fig. 4.12.** The plastic strains when the \(LCF\) specimens failed.

When a \(LCF\)-specimen failed, the strain recorded by the strain gage is the plastic strain and is shown in Fig. 4.12. For low-cycle fatigue with \(R\) ratios of \(-1.0\) and \(-0.18\), the
residual plastic strain decreases with increase of \(SDAS\), indicating that the tendency of plastic strain localization at pores is stronger for specimens with larger \(SDAS\). Specimens with small \(SDAS\) have a relatively homogeneous distribution of plastic strain along the specimen length.

4.3.3. Axial-Fatigue Life

Figure 4.13 summarizes the fatigue lives obtained from low- and high-cycle axial tests ("push-pull") performed at a strain ratio of -1.0, or at stress ratios of -1.0, 0.1, and 0.2. The life under low-cycle fatigue decreases from an average of \(6.6 \times 10^3\) cycles at \(SDAS < 34 \ \mu m\) to an average of \(2.6 \times 10^3\) cycles for the specimens with \(SDAS > 35 \ \mu m\). The average life under high-cycle fatigue with \(R = -1.0\) decreases from \(4.2 \times 10^6\) cycles at \(SDAS < 30 \ \mu m\) to an average of \(4.3 \times 10^5\) cycles at \(SDAS > 30 \ \mu m\).

Figure 4.13 also presents the fatigue lives obtained from axial tests performed at stress ratios of 0.1 and 0.2. In order to avoid the very high number of cycles encountered in the bending tests, the maximum stresses were set to 175 MPa and 130 MPa, respectively. As shown in Fig. 4.13, similar to the previous results, the fatigue life decreases with increasing \(SDAS\). The fatigue lives of specimens tested with a stress ratio of 0.1 decreases from \(~4.0 \times 10^5\) cycles to \(~7.0 \times 10^4\) cycles for the four specimens with \(SDAS \geq 32 \ \mu m\). The fatigue lives of specimens tested with a stress ratio of 0.2 decreases from \(~10^7\) cycles at \(SDAS = 17 \ \mu m\) to less than \(10^6\) cycles for specimens with \(SDAS > 30 \ \mu m\).
During displacement-controlled low-cycle axial-fatigue, the maximum and minimum stresses respectively increase and decrease in the very early number of cycles (< 500 cycles). The saturated cyclic stresses, average stress and stress amplitude at the half fatigue life (> 500 cycles) are shown in Fig. 4.14. The cyclic hardening is about the same for all specimens regardless of the SDAS. As shown in Fig. 4.14, for specimens tested at strain ratio of -1, the average cyclic tension and compression stresses are 205 MPa and -218 MPa, respectively, indicating that the cyclic stress-strain relationship is not symmetrical for tension and compression. The unsymmetrical response of tension and compression is related to the effect of silicon particles on the work hardening rate of A356.2; the fraction of damaged silicon particles is higher in tension than in compression.
Further the fractured particles can carry the compressive load, but not the tensile load.

Fig. 4.14. Cyclic stresses for axial low-cycle fatigue at maximum strain of 0.0047 and strain ratio of -1 after saturation.

4.3.4. Crack-Initiation Time

In seven of the reciprocating-bending tests, the formation of an observable fatigue crack was indicated by the monitored reduction of the strain amplitude or the peak strain; no attempt was made to observe crack initiation with light microscopy in these tests. When the oscilloscope record showed a change of the constant amplitude of cyclic strain, which indicated a relaxed strain state, then it was assumed that relatively large fatigue cracks had formed. The corresponding number of cycles was taken as the crack-initiation life.
Fig. 4.15 shows the crack-initiation life as a percentage of total fatigue life and its dependence on SDAS under high-cycle reciprocating-bending. In specimens with small SDAS (15-30 μm), fatigue cracks initiate at ~75% of the total life, whereas for specimens with large SDAS (>50 μm), fatigue cracks can initiate at merely 15% of the corresponding total life. This indicates that the finer microstructures obtained from the higher solidification rates significantly delayed the initiation of fatigue cracks.

![Graph showing crack initiation time vs. SDAS](image)

**Fig. 4.15.** Variation of crack initiation time as percentage of total fatigue life with secondary dendrite arm spacing under reciprocating-bending.

### 4.3.5. Fatigue Life vs. Product of Maximum Stress and Strain Amplitude

Fatigue life depends on both strain amplitude and maximum strain. Smith, Watson, and Topper (Dowling, 1993) assumed that the fatigue life ($N_f$) depends on the product of the maximum stress ($\sigma_{max}$) and the strain amplitude ($\varepsilon_a$). In order to estimate fatigue life, the
relationship between $\sigma_{max} \varepsilon_a$ and fatigue life ($N_f$) is often applied (Dowling, 1993). When fatigue plots of $\sigma_{max} \varepsilon_a$ and $N_f$ are shown in Figs. 4.16(a) and (b). Specimens were not distinguished by $SDAS$ in this plot. The cyclic stress and strain relationship used was from fatigue results on specimens from cast plates (Chapter 5). The average strain of the entire specimens in the same group was used for each specimen in that group. As shown, the groups with smaller values of $\sigma_{max} \varepsilon_a$ tend to have longer fatigue lives, but clearly within each group microstructure plays a significant role and can account for more than two orders of magnitude difference in the fatigue life.

It is also shown in Figs. 4.16(a) and (b) that bending fatigue specimens have fatigue lives 4~10 times higher than that of axial fatigue specimens. This has been explained previously by other researchers (Hertzberg, 1996). In an axial fatigue specimen, there is no stress gradient inside the specimen; the entire cross section in the gage length experiences the maximum stress. In the bending case, there is a stress gradient from the surface to the inside; the thinner the specimen, the higher the stress gradient and only a small layer of material on the surface experiences the maximum stress. Thus, the probability of a crack-initiation site to result in early fatigue failure in bending specimens is less. In addition to this, bending fatigue is in deflection control and specimens experience longer crack lengths before the specimens are broken. Thus, the fatigue lives are longer for bending fatigue specimens than for axial fatigue specimens.
Fig. 4.16. Fatigue lives ($N_f$) based on the product of maximum stress ($\sigma_{max}$) and strain amplitude ($\varepsilon_a$): (a) bending fatigue; and (b) axial fatigue. (Fatigue specimens without fractures when interrupted are not included).
4.4. FATIGUE-CRACK INITIATION SITES

4.4.1. Fatigue-Crack Initiation at Near Surface Porosity

Porosity has been widely reported in casting alloys to be the major initiation site for fatigue cracks in casting alloys. In this study, this was found to be true, but detailed observations of the fracture surfaces of the bending- and axial-fatigue specimens revealed other types of initiation sites in addition to pores, depending on the SDAS and microconstituents.

Although most of the initiation sites are at or near the surface, cracks initiated from subsurface porosity in a few of the bending-fatigue specimens. Subsurface crack initiation, related to pre-existing defects and microcracks, has been reported in HCF for high-strength alloys (Umezawa and Nagai, 1997). The possibility that a subsurface pore or other defect can initiate a crack depends not only on its size, but also on its distance from the surface. The pore in Fig. 4.17(a) is approximately 360 \( \mu \text{m} \) in length, and its closest distance to the surface is about 33 \( \mu \text{m} \). Fig. 4.17(b) shows the maximum lengths of subsurface defects that initiated cracks and their distances from the surface.

For the fractures resulting from axial-fatigue testing, there were no cracks that initiated from subsurface pores, perhaps because the number density of large pores in the axial specimens was significantly less than in the bending specimens. Only one crack, initiated from a subsurface oxide inclusion (200 \( \mu \text{m} \) in size, located at 250 \( \mu \text{m} \) from the surface).
Fig. 4.17. Crack-initiation sites in HCF bending-specimens: (a) a near-surface pore, where the crack initiated in a specimen with a SDAS of 35 µm; and (b) distribution of crack-initiation sites in terms of their lengths and distances from surfaces.
A reduced cross-section was used in fatigue specimens to establish the relationship of fatigue life to pore-initiation size and the distance from the surface (Seniw et al. 1997). They found that most of the initiation sites were subsurface pores. So, whether the fatigue crack initiates from a subsurface defect depends on the extent to which the maximum stress is localized, the pore size, and the pore number density.

Fig. 4.18. Fracture surfaces of a low-cycle fatigue cracks initiated at pores: (a) $SDAS = 54 \ \mu m$, bending fatigue with maximum strain of 0.00391, $R = -1$, failed at $3.2 \times 10^3$ cycles. The left edge of the figure is 400 $\mu m$ from the surface, and the crack propagated from left to right; and (b) $SDAS = 38 \ \mu m$, bending fatigue with maximum strain of 0.00865, $R = 0.18$, failed at $8.7 \times 10^3$ cycles.

Figures 4.18(a) and (b) show fatigue cracks that initiated from pores in LCF specimens. In low-cycle fatigue, the cracks that emanate from pores show many flat paths, indicating the characteristics of slip bands; whereas high-cycle fatigue cracks rarely show flat paths.
Similar crack-initiation behavior was observed in both axial and reciprocating-bending fatigue experiments.

Fig. 4.19. Oxides revealed by SEM: (a) initiation site at subsurface oxide (SDAS = 25 μm, bending fatigue with maximum strain of 0.00159, $R = -1$, failed at $1.4 \times 10^6$ cycles); and (b) initiation site at an oxide film (SDAS = 17 μm, axial fatigue with maximum stress of 130 MPa, $R = 0.2$, failed at $5.0 \times 10^6$ cycles).

4.4.2. Crack Initiation at Oxides at or Near the Surface

Oxide inclusions or films initiated fatigue cracks in both axial- and bending-fatigue specimens. Oxides poured into molds with the melts are particularly deleterious to fatigue properties. Fig. 4.19(a) shows a crack-initiation site at an oxide which is about 75 μm from the surface of a specimen (SDAS = 25 μm). The energy-dispersive X-ray (EDX) analysis reveals the oxide to be $\text{Al}_x\text{Mg}_y\text{O}_z$ (probably the spinel of $\text{Al}_2\text{O}_3$ and $\text{MgO}$ or $\text{Al}_2\text{O}_3$ plus the spinel). On the surface of the oxide film, small particles (less than 5 μm) were detected, with different compositions: $\text{Al}_x\text{Mg}_y\text{O}_z$, $\text{Al}_x\text{Si}_y\text{O}_z$, and $\text{Al}_x\text{C}_y$ (possibly
Its shape and higher Mg content indicate that it is an "old" oxide, which existed in the melting ingot and has been described (Campbell 1995). Another old oxide with Mg was observed on the fracture of an axial-fatigue specimen (Fig. 4.19(b)). It consists of two layers of oxide films with small particles of $\text{Al}_3\text{Mg}_2\text{O}_3$ (less Si, Sr and Ti) sitting on the film.

The oxides are as deleterious to fatigue resistance as are the pores. The fatigue life of a bending specimen with a $SDAS$ of 25 $\mu$m, whose crack initiated at an old oxide, is 100 times less than that for a similar specimen with a crack initiated from a eutectic constituent and is almost as the same as that for a specimen with a crack initiated from a pore of 267 $\mu$m in maximum length. It has been stated that fatigue cracks initiate from "old" or "young" oxides, and that the fatigue lives of most aluminum alloy castings could probably be improved by a factor of 100 to 10,000 times by a combination of attention to metal quality and casting technique (Campbell et al. 1998, Wang et al. 1998).

Figure 4.20 shows a "young" oxide as classified by Campbell (Campbell, 1995) that developed on the surface of the molten alloy during pouring. This oxide initiated a fatigue crack in a specimen with a $SDAS$ of 17 $\mu$m. The oxide consists of $\text{Al}_3\text{Si}_2\text{O}_5$ with only a very slight indication of Mg, in contrast to an "old" oxide, which showed clear indications of Mg. Smaller particles sitting on the film were detected, with qualitative compositions of $\text{Al}_3\text{Si}_2\text{O}_5$, $\text{Al}_3\text{C}_2\text{Si}_5$. The fatigue life of this specimen was about the same
as in specimens in which the eutectic constituents initiated cracks. It seems that the “young” oxide is not particularly deleterious, if it is not folded and aggregated.

Fig. 4.20. Crack-initiation site at an oxide revealed by SEM (SDAS = 17.3 μm. axial fatigue with maximum stress of 175 MPa, \( R = 0.1 \), failed at \( 7.8 \times 10^5 \) cycles).

Young oxides formed during turbulent pouring were found in specimens near the bottom of the casting-ingot, even though the mold was tilted during the pouring to minimize their formation. Fig. 4.21(a) shows such an oxide film, which initiated the fatigue crack in a specimen with a SDAS of 16 μm. The film, an example of “young” oxide, is of AlₓOᵧ, with a small amount of Si and barely a trace of Mg. At a higher magnification, submicron particles of AlₓOᵧ (probably Al₂O₃) that decorated the surface of the films were found. At a much lower magnification (Fig. 4.21(b)), the oxide appears as folded films, which follow the eutectic constituents for several SDASs.
4.4.3. Crack Initiation at Eutectic Constituents

When the SDAS is less than 25 μm, the pores are relatively small (Fig. 4.4(a)). If no large oxide inclusions are present, then HCF cracks initiate exclusively from the eutectic constituents. Fig. 4.22(a) shows a rather large eutectic constituent (210 μm in maximum length), which initiated a fatigue crack at the surface. Many of the silicon particles in the eutectic constituent are debonded from the aluminum-rich phase as shown in Fig. 4.22(b); EDX analysis shows that, on the surface of the debonded silicon particle, there is a layer of aluminum. Well-organized marks left by the debonded silicon particles can also be seen in Figure 4.22(b), which indicated that the debonding was from a slip mechanism. Two other relatively large eutectic constituents which initiated the fatigue cracks are shown in Figs. 4.23(a) and (b).
Fig. 4.22. Crack-initiation site at interdendritic eutectic constituent revealed by SEM: (a) the eutectic region; and (b) a high magnification showing the debonded silicon particles in crack-initiation site ($SDAS = 29 \mu m$, axial fatigue with maximum stress of 130 MPa, $R = 0.2$, failed at $1.5 \times 10^5$ cycles).

Fig. 4.23. Crack-initiation site at interdendritic eutectic constituent revealed by SEM: (a) a portion of eutectic constituent ($SDAS = 24 \mu m$, failed at $6.58 \times 10^5$ cycles); and (b) a eutectic constituent as a secondary crack initiation site ($SDAS = 35 \mu m$, failed at $4 \times 10^5$ cycles). Both are from specimens subjected to bending fatigue with maximum stress of 0.00159, $R = -1$. 

Under low-cycle fatigue, the cracks also initiated at near-surface eutectic regions in microstructures with small SDAS (< 25μm). Figure 4.24(a) shows low-cycle fatigue-crack initiation site at large eutectic constituents. Figure 4.24(b) shows low-cycle fatigue crack initiation site at a small eutectic constituents.

![Fracture surface of a low-cycle fatigue crack initiated at a eutectic constituent located at the surface](image)

Fig. 4.24. Fracture surface of a low-cycle fatigue crack initiated at a eutectic constituent located at the surface: (a) SDAS = 21 μm, axial fatigue with maximum strain of 0.0047, R = -1, failed at 8.4 x 10^3 cycles (The surface is at the right edge of the figure, and the crack propagated from right to left); and (b) SDAS = 17 μm, bending fatigue with maximum stress of 0.00865, R = 0.18, failed at 1.7 x 10^7 cycles (The crack propagated from bottom to the top).

Since the silicon particles in the interdendritic area are stiffer than the matrix (Shivkumar et al. 1993), the stress has to be distributed in such a way that the clusters bear more load than the rest of the matrix to satisfy the deformation compatibility. Apparently, the stress concentration created in this area is enough to cause micro-slips in the matrix close to the silicon particles, which leads to the particle debonding and then crack initiation. Refining the silicon particles and the eutectic cluster size improves the homogeneous distribution
of plastic strains, and thus the crack-initiation life is increased. Sr-modified eutectic alloy
(Al-12 wt pct Si-0.35 wt pct Mg) showed better fatigue-crack growth resistance than the
unmodified version of the alloy with coarser Si-particles (Lee et al. 1996). When fatigue
behavior was investigated in squeeze-cast aluminum alloys using smooth specimens
subjected to rotary bending, crack-initiation sites were at the silicon particles within the
eutectic constituents at the surface of specimens (Shiozawa et al. 1997). They suggested
that an improvement in fatigue strength could be expected by refining the eutectic silicon
rather than increasing the static strength.

The size of a eutectic area was found to significantly affect the fatigue life. In a specimen
with a SDAS of 22 μm, a fatigue crack initiated at a relatively small eutectic constituent
of 100 μm in maximum length. The specimen exhibited the greatest fatigue life (2.49 ×
10^7 cycles) among the bending-fatigue specimens with a stress/strain ratio of -1, which is
30 times greater than that for a specimen with approximately the same SDAS but with an
initiation site of a eutectic constituent of 417 μm in maximum length.

4.5. VARIATION OF SIZE OF CRACK-INITIATION SITE WITH SDAS AND
ITS EFFECT ON FATIGUE LIFE

4.5.1. Distributions of the Crack-Initiation Sites

The maximum size and area of the fatal fatigue-crack-initiation region were measured on
each high-cycle fatigue-tested specimen. Figures. 4.25(a) and (b) and 4.26(a) and (b)
show the size and area distribution as a function of SDAS for specimens subjected to both
the axial- and bending-fatigue testing, under $R = -1, 0.1, \text{ and } 0.2$. Both the maximum length and the area distribution of the initiation site show the same trends. For discussion, we consider only the maximum length. It should be noted that the pore lengths measured at the initiation sites on the fracture surfaces were larger than the lengths detected on the radiographs (Fig. 4.5(a)), due to much better resolution obtained by the SEM images of the fracture surfaces. When the $SDAS$ is greater than $\sim 25$ to $28 \mu m$, the cracks usually initiated at pores with lengths no smaller than $100 \mu m$, for both axial and bending specimens. When the $SDAS$ is greater than $28 \mu m$, the lengths of the pores at the initiation sites increase as the $SDAS$ increases, because the pores tend to increase in size as the $SDAS$ increases (Fig. 4.5(a)). All specimens that showed visible porosity on radiographs had relatively short fatigue lives.

When the $SDAS$ is less than $\sim 25$ to $28 \mu m$, either oxide inclusions or eutectic constituents were the sites of fatigue-crack initiation. If there is no oxide near the surface, then the eutectic constituent is the site of crack initiations. Five bending-fatigue specimens with $SDAS$ below $28 \mu m$ had not broken after $2 \times 10^7$ cycles (maximum stress = 115 MPa, $R = 0.1$) when the tests were terminated. This may indicate an absence of oxide inclusions, eutectic constituents, and large pores as possible initiation sites.
Fig. 4.25. Distribution of sizes of crack initiation sites in axial-HCF specimens: (a) maximum length; and (b) area.
Fig. 4.26. Distribution of sizes of crack initiation sites in bending-HCF specimens: (a) maximum length; and (b) area.
4.5.2. Effect of Crack-Initiation Site on Fatigue Life

If pore-size is used to classify the crack initiation sites instead of SDAS, based on the relationship of pore-size and SDAS shown in Figs. 4.25(a) and 9b) and 4.26(a) and (b), then the results indicate that fatigue cracks initiate from pores when the pore-size is greater than a critical value of ~100 μm. If the pore-size is below this critical value, fatigue cracks initiate from near-surface eutectic constituents. Similar critical pore-sizes needed to initiate fatigue cracks in aluminum castings have been reported (Mayer et al. 1999).

With the porosity in the alloy, it appears that its effect on initiating fatigue cracks overshadows the effect of the eutectic constituents, as shown in Fig. 4.27 (a) for high-cycle bending fatigue with \( R = 0.1 \). But for bending fatigue with \( R = -1 \) (Fig. 4.27(b)), there is a tendency that the fatigue life decreases as the pore size increases. As the "degree of porosity" and maximum pore size increased, it was showed that the fatigue strength decreased (Sonsino and Ziese 1993).

For axial HCF testing with \( R = 0.1, 0.2 \) and \(-1\) (Figs. 4.28(a)-(c)), the fatigue life decreases as the pore or oxide size increases. Larger eutectic constituents also result in shorter fatigue life, although there are a few exceptions, as shown in Figures. 4.27(b) and 4.28(b).
Fig. 4.27. Bending fatigue life as a function of maximum length of initiation site: (a) maximum strain = 0.00166, $R = 0.1$; and (b) maximum strain = 0.00159, $R = -1.0$. 
Fig. 4.28. Axial fatigue life as a function of maximum length of initiation site: (a) maximum stress = 175 MPa, $R = 0.1$; (b) maximum stress = 85 MPa, $R = -1$; and (c) maximum stress = 130 MPa, $R = 0.2$. 
When the pore size is less than a critical size of ~100 μm, the eutectic constituent becomes the preferred initiation site, provided large oxides are absent. The larger the eutectic area, the lower the fatigue life, as a result of shorter initiation life. So, in addition to decreasing the pore size to below a critical size and eliminating oxides and oxide films in the melt, measures that would reduce the stress concentration within the eutectic constituent, such as a refinement of the size of the silicon particles and reduction of the aspect ratio of these particles, would increase fatigue life.
4.6. VARIATION OF MODIFIED FATIGUE LIFE WITH INITIAL STRESS INTENSITY FACTOR RANGE

As shown in the previous section, when present, oxide films initiated cracks and when $SDAS$ is greater than $\sim 28 \mu m$, pores are the culprits of crack-initiation. When a fatigue crack initiates at a pore, the initiation life is very short, so small-crack propagation becomes important. Although Figs. 4.27(a) and (b) and 4.28(a)-(c) show the effects of the sizes of the crack initiation sites, the quantitative relationship between fatigue life and microstructural sizes needs to be clarified. In order to do this, a crack propagation model with crack closure effect presented in Chapter 2 (Eq. 2.11) is invoked:

$$\frac{da}{dN} = C\Delta K_{eff}^m = C\left[U(a)F(a)\Delta S\sqrt{\pi a / Q(a)}\right]^m$$

(2.11)

$F(a)$ is the boundary-correction factor; and $U(a)$ is the crack closure factor. Both of these factors change slowly with crack length (Newman and Raju, 1983) and are thus assumed to be constants. $Q(a)$ is the elliptical crack-shape factor; when half penny-shape crack is assumed, $Q(a)$ is 2.464 (Newman and Raju, 1983); Also, $m$, $c$, and $\Delta S$ are constants. Integration of Eq. (2.11) yields (Appendix):

$$\left(\frac{1}{a_f^{(m-2)/2}} - \frac{1}{a_0^{(m-2)/2}}\right) = \frac{2-m}{2}\pi^{m/2}(U)^m C(\Delta S)^m (N_f - N_i)$$

(4.2)

Based on a plasticity model of a discrete surface of yielding ahead of a crack, Suresh reported that the predicted Paris exponent $m$ is 4 (Suresh, 1998); and indeed from experiments on most aluminum alloys, $m$ approaches a value of 4.0 (Basner et al., 2001).
The crack length at failure \((a_f)\) is much larger than crack-initiation site size \((a_o)\), so the term with \(a_f\) is neglected. Thus, Eq. (4.2) simplifies to

\[
\frac{(N_f - N_i)}{a_o} = \left( \frac{2}{(m-2)C} \right) \Delta K_0^{-n}
\]  

(4.3)

As presented previously when the \(SDAS\) is greater than \(-28\ \mu m\), cracks initiate from both pores and oxides. Replica experiments (discussed later) show that the crack initiation life \((N_i)\) is a very small part of the total fatigue life for cracks that initiate at pores or oxides, then Eq. (4.3) is further simplified:

\[
\frac{N_f}{a_o} \approx D \Delta K_0^{-n}
\]  

(4.4)

where the constant \(D\) is

\[
D = \left( \frac{2}{(m-2)C} \right)
\]

For the \(HCF\) specimens fatigued to fracture, the modified fatigue life \((\frac{N_f}{a_o})\) versus the initial stress-intensity factor range \((\Delta K_0)\) is shown in Figs. 4.29(a) and (b). In calculating the initial stress-intensity factor range \((\Delta K_0)\), \(\Delta K_{eff}\) in Eq. (2.11) was applied. The square root of the area of the initiation site is taken as the size of the crack-initiation site \((a_o)\). Crack closure coefficient \((U)\) is assumed 1 for \(R\) ratios of 0.1, and 0.5 for \(R\) ratio of -1 (Couper et al., 1990). \(F(a)\sqrt{\frac{1}{Q(a)}}\) is taken as 0.5 for a pore and 0.65 for an inclusion as suggested in a model of \(\Delta K\) proposed by Murakami (Murakami 1990).
Fig. 4.29. Modified fatigue life including the size of crack initiation site versus initial stress intensity factor range: (a) labeled by test condition; (b) labeled by crack initiation site.
It is shown in Figs. 4.29(a) and (b) that Eq. (4.4) correlates reasonably well the modified fatigue life \( \frac{N_f}{a_0} \) with the initial stress intensity factor range \( \Delta K_0 \). The bending and axial fatigue results are also united in terms of \( \frac{N_f}{a_0} \) and \( \Delta K_0 \) in Figs. 4.29(a) and (b).

Only when \( \frac{N_f}{a_0} \) is greater than \( 10^{11} \) (cycles/m) does the scatter seem unduly large. It should be noted that those specimens with crack initiations at eutectic constituents and small pores correspond to values of \( \frac{N_f}{a_0} \) greater than \( 10^{11} \) (cycles/m). Such specimens may not follow the assumption that the initiation life is negligible. On the other hand, Eq. (4.4) may not adequately capture small-crack behavior, as is discussed later in this chapter. Finally fitting the data of either Fig. 4.29(a) or Fig. 4.29(b) to Eq. 4.4, the parameters for crack propagation in Equation (2.11) are obtained when \( \Delta K_0 \) is in MPa m\(^{1/2}\), and \( \frac{N_f}{a_0} \) is in cycles/m; \( C \) and \( m \) are \( 2.2298 \times 10^{-10} \) and 3.8905, respectively.

4.7. SMALL-CRACK PROPAGATION IN HIGH-CYCLE FATIGUE

Three bending-fatigue specimens did not fracture for more than \( 10^7 \) cycles when tested at a maximum strain of 0.00166 and \( R = 0.1 \). They were re-polished and tested at a maximum strain of 0.00214 and \( R = 0.1 \). At the higher strain, cracks were detected. Replicas were taken on the surface of the specimens during fatigue testing to determine
the crack initiation and small-crack propagation. Table 4.1 is a summary of the fatigue lives, and type and sizes of the crack-initiation sites.

Table 4.1. Summary of specimens used for replicas

<table>
<thead>
<tr>
<th>SDAS (μm)</th>
<th>Grain Size (μm)</th>
<th>Fatigue Life (cycles)</th>
<th>Initiation Site</th>
<th>Initiation Size (μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>407</td>
<td>2.99×10^6</td>
<td>Eutectic</td>
<td>245</td>
</tr>
<tr>
<td>23</td>
<td>443</td>
<td>1.17×10^6</td>
<td>Pore</td>
<td>93</td>
</tr>
<tr>
<td>27</td>
<td>491</td>
<td>1.36×10^6</td>
<td>Pore</td>
<td>210</td>
</tr>
</tbody>
</table>

Note: All cracks were corner cracks; initiation size is the maximum length of the crack initiation site.

4.7.1. Crack Development with the Number of Cycles

The replicas of small-cracks with similar lengths from the three specimens are shown in Figs. 4.30(a), (b) and (c). All the three cracks show a “zigzag” path. Some portions of the crack paths are very straight, particularly in Fig. 4.30(c), indicating slip bands are active during crack propagation. Crack length (projection length) measurements as a function of number of cycles for three specimens are presented in Fig. 4.31. It is shown that the fatigue lives mostly consist of the cycles in forming a crack of about 2 mm. Beyond 2 mm, the cracks grow very rapidly.

The crack initiation and propagation lives of the three specimens of Figs. 4.30(a), (b), (c) and Fig. 4.31 are shown in Fig. 4.32. The crack-initiation life was defined as the average of the number of cycles when the crack was first captured on a replica and the number of cycles when the previous replica showed no crack. The initiation lives are 1.96×10^6,
Fig. 4.30. Small-cracks observed on replicas: (a) $SDAS = 20 \mu m$, crack length = $1890 \mu m$, replicas taken at $2.73 \times 10^6$ cycles; (b) $SDAS = 23 \mu m$, crack length = $1823 \mu m$, replicas taken at $1.05 \times 10^6$ cycles; and (c) $SDAS = 27 \mu m$, crack length = $1764 \mu m$, replicas taken at $1.05 \times 10^6$ cycles.
Fig. 4.31. Crack development with number of cycles. The SDAS shown is in μm.

Fig. 4.32. Fatigue crack initiation and propagation lives for specimens with different SDAS.
5.00 \times 10^5 and 7.00 \times 10^4 cycles for specimens with \textit{SDAS} of 20, 23, and 27 \textmu m, respectively, and are \sim 68\%, \sim 43\% and \sim 5\% of the total fatigue lives, respectively. The propagation lives are 1.03 \times 10^6, 0.7 \times 10^6 and 1.3 \times 10^6 cycles for specimens with \textit{SDAS} of 20, 23, and 27 \textmu m, respectively, and are \sim 32\%, \sim 57\% and \sim 95\% of the total fatigue lives, respectively. Among the three specimens, the one with the smallest \textit{SDAS} (20 \textmu m) has the longest initiation and total lives by far.

\textbf{4.7.2. Small-Crack Propagation Rate}

The small-crack propagation rates \((da/dN)\) for these three specimens are plotted against the crack lengths in Figs. 4.33(a), (b) and (c). As shown in Fig. 4.33(a)-(c), the overall propagation rates oscillate in the range of \(2.0 \times 10^{-10}\) to \(1.0 \times 10^{-8}\) m/cycle when the crack lengths are less than 1000 \textmu m. Oscillations persist for up to \sim 5000 to 6000 \textmu m. The propagation rate increases to \sim 4.0 \times 10^{-7}\) m/cycle when the cracks achieve a length of \sim 7.0 mm at fracture. The specimens with \textit{SDAS} of 20 and 23 \textmu m have slower small-crack propagation rates than that of the specimen with a \textit{SDAS} of 27 \textmu m when crack length is shorter than \sim 400 \textmu m.

The stress-intensity factor range indicates the driving force for crack propagation; therefore, the small-crack propagation rate as a function of the stress intensity factor range is plotted in Figs. 4.34(a), (b) and (c). Here the effective stress intensity factor range \(((\Delta K_p)_{\text{eff}})\) presented as Eq. (2.16) is used; it includes the effect of a plastic zone.
Fig. 4.33. Small-crack propagation rates against crack length: (a) $SDAS = 20 \, \mu m$; (b) $SDAS = 23 \, \mu m$; (c) $SDAS = 27 \, \mu m$. 
where \( F(d/w) \) is the cyclic-plastic-zone corrected boundary-correction factor; \( w \) is the width of the specimen; \( d \) is the corrected crack length with a portion of the Dugdale cyclic-plastic-zone length; \( U \) is the crack-closure factor.

The bending fatigue was carried out in deflection control. The strain on the surface of the specimen is constant over a long period of crack propagation, so the boundary-correction factor is therefore assumed to be constant by treating the specimen as a half-infinite plate. The maximum length of the initiation site measured is used as the crack initial length, but
when the crack initiated within a eutectic constituent, the maximum length of the silicon particle size measured in Fig. 4.6 is used.

The small-crack propagation rate is plotted against the stress intensity factor range in Figs. 4.34(a)-(c). Although the crack-closure effect and the plastic-zone size are considered in the model of Eq. 2.16, a small-crack effect is still apparent when the \((\Delta K_p)_{eff}\) is < 9.0 MPa m^{\frac{1}{2}}; in which \(da/dN\) oscillates. When the \((\Delta K_p)_{eff}\) is above ~ 5.0 MPa m^{\frac{1}{2}}, the overall propagation rate tends to increase with \((\Delta K_p)_{eff}\) (especially in Fig. 4.34(c)), and exhibits a linear relationship on a log-log plot of \(da/dN\) against \((\Delta K_p)_{eff}\). The specimens ultimately failed when \((\Delta K_p)_{eff}\) achieved ~12 MPa m^{\frac{1}{2}}. When the Paris power law is assumed for \((\Delta K_p)_{eff}\) ≥ 3.5 MPa m^{\frac{1}{2}} (corresponds to a crack length of ~ 500 μm, the approximate length scale of the grain size) in Figs. 4.34(a)-(c), the fitted parameters for crack propagation in Equation (2.11) are obtained with \(da/dN\) measured in units m/cycle and \((\Delta K_p)_{eff}\) in units MPa m^{\frac{1}{2}}; C and \(m\) are 1.264954 \times 10^{-10} and 3.597, respectively. C and \(m\) are in the same order with those estimated from modified fatigue life in Figs. 4.29(a) and (b), which are 2.2298 \times 10^{-10} and 3.8905, respectively.

4.7.3. Interactions between Crack and Microstructural Features

HCF propagation paths on the surfaces for specimens with SDAS of 17 and 30 μm are shown in Figs. 4.35(a)-(d). It is shown in Figs. 4.35(a) and (c) that the cracks propagate with no preference to propagate solely through the matrix or through the eutectic constituents. In Fig. 4.35(b), a crack passes across the dendrite cells and appears to
Fig. 4.34. Small-crack propagation rate against effective stress intensity factor range: (a) $SDAS = 20 \, \mu m$; (b) $SDAS = 23 \, \mu m$; and (c) $SDAS = 27 \, \mu m$. 
Fig. 4.34. (Continued)

Fig. 4.35. The interaction between small-cracks and microstructures: (a) and (b) $SDAS = 17 \mu m$, failed at $1.17 \times 10^6$ cycles; (c) and (d) $SDAS = 30 \mu m$, failed at $1.3 \times 10^6$ cycles. (Bending fatigue with maximum strain of 0.00159, $R = -1$).
deviate around the silicon particles in the eutectic regions. Debonded eutectic silicon particles are observed on the crack path and broken silicon particles are rarely seen when the SDAS and Si-particles are small. With a larger SDAS (30 μm), Fig. 4.35(d), a similar crack path is observed. In this instance there are some broken Si-particles within the eutectic constituents.

The SEM images of Figs. 4.36(a) and (b) show that Si-particles between the cells do not change macroscopic orientation of the crack path. In Sr-modified A356.2 alloy with Sr below 110 ppm, the eutectic grains are nucleated in the intergranular regions independent of the equiaxed primary aluminum dendrite (Dahle et al., 2001). Hence, the orientations of the aluminum in the intergranular eutectic constituents may posses different crystalline orientations from that of the primary aluminum dendrites. When a crack passes through a eutectic constituent, the orientation of the crack path changes in order to accommodate...
Fig. 4.36. The interaction between cracks and eutectic constituents when the crack length was small: (a) the variation of the orientation of the crack path when crack interacted with the intergranular eutectic constituents; and (b) the intergranular eutectic constituents at a higher magnification. \(SDAS = 17 \mu m\), bending fatigue with maximum strain of 0.00166, \(R = 0.1\), failed at \(9.11 \times 10^7\) cycles.

Fig. 4.37. The interaction between cracks and eutectic constituents when the crack length is large: (a) \(SDAS = 16 \mu m\), failed at \(1.46 \times 10^6\) cycles; and (b) \(SDAS = 37 \mu m\), failed at \(0.75 \times 10^6\) cycles. (Bending fatigue with a maximum strain of 0.00159, \(R = -1\)).
the preferred slip system. Some broken silicon particles and oxide particles in the eutectic area are observed in Figure 4.36(b).

The interactions between cracks and eutectic constituents are shown in Figs. 4.37(a) and (b) when the crack length is large. Secondary cracks at eutectic constituents are found to interact with the primary fatigue crack, especially when the primary crack propagates through the eutectic constituent, indicating that, the stress intensity factor range (driving force for crack propagation) at the crack tip is very large and the silicon particles have already been damaged (either debonded or broken) in the plastic zone at the tip of the crack before the primary crack reaches them. At the crack length shown, the silicon particles in the eutectic constituents are no longer beneficial in increasing the resistance to crack propagation. In a study on small-crack propagation, the propagation rate was retarded around the silicon particles and enhanced when it goes across the dendrite (Shiozawa et al. 1997). Above a crack length of 1 mm, however, the enhancement and retardation of crack growth rate disappears and the silicon particles do not act as obstacles to crack propagation. Based on features seen in the replicas removed during bending fatigue in this dissertation, the vacillation of propagation rates in small-cracks is related with microstructural barriers such as Si-particles and grain boundaries. The eutectic constituents are not beneficial in retarding cracks when the cracks are longer than 1 mm, in accord with the other observations (Shiozawa et al. 1997).
Microstructure-effects in tensile fracture were also studied (Wang, et al. 1998). On HCF fractures, less eutectic Si-particles are cracked than what appears on tensile fractures. Fracture paths in HCF are trans-dendritic regardless of the SDAS, while the tensile-fracture path for specimens with finer SDAS is predominantly interdendritic and for specimens with coarser SDAS, the fracture path is trans-dendritic along the cell boundaries.

4.8. SMALL-CRACK PROPAGATION IN LOW-CYCLE FATIGUE

Replicas were taken on the surface of the two specimens tested at a maximum strain of 0.00865 and $R = 0.18$ to determine the crack initiation and small-crack propagation under LCF. Table 4.2 is a summary of the fatigue lives, and type and sizes of the crack-initiation sites.

<table>
<thead>
<tr>
<th>SDAS ($\mu$m)</th>
<th>Grain Size ($\mu$m)</th>
<th>Fatigue Life (cycles)</th>
<th>Initiation Site</th>
<th>Initiation Size ($\mu$m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>17</td>
<td>370</td>
<td>$1.40 \times 10^4$</td>
<td>Si</td>
<td>10</td>
</tr>
<tr>
<td>37</td>
<td>612</td>
<td>$0.70 \times 10^4$</td>
<td>Pores</td>
<td>300/287/192</td>
</tr>
</tbody>
</table>

Notes: The crack in specimen with SDAS of 17, and 37 $\mu$m was, respectively, surface crack and corner crack.

4.8.1. Crack Development with the Number of Cycles

Replicas of small-crack development are shown in Figs. 4.38(a)-(c) and 4.39. The cracks initiated within the first 5 cycles for both specimens. The failure crack initiated at a
Fig. 4.38. Replicas showing small-crack development for specimen with a SDAS of 17 μm: (a) replica taken at 5 cycles; (b) replica taken at 800 cycles; (c) replica taken at 1600 cycles.
silicon particle in the specimen with a $SDAS$ of 17 $\mu$m (Fig. 4.38(a)). When the crack reached a length of 119 $\mu$m at 1600 cycles, micro-cracks from slip bands in front of the crack tip were observed (Fig. 4.38(c)). The failure crack of a specimen with $SDAS$ of 37 $\mu$m initiated at a pore of maximum length of 300 $\mu$m, located at the corner of the specimen (Fig. 4.39). The zigzag nature of the crack is the result of the coalescences of the micro-cracks developed along the slip bands in front of the crack tip.

Crack length measurements as a function of number of cycles for the two specimens are presented in Fig. 4.40. The fatigue life of the specimen with a $SDAS$ of 17 $\mu$m consists mostly of the cycles in propagation to a crack of ~1 mm, and this specimen also shows a higher propagation life than the specimen with a $SDAS$ of 37 $\mu$m.

Fig. 4.39. Replica showing small-crack development at slip band for a specimen with a $SDAS$ of 37 $\mu$m. Replica was taken at 3300 cycles.
4.8.2. Small-Crack Propagation Rate

The small-crack propagation rates \( (da/dN) \) for specimens with two different \( SDAS \) are plotted against the crack lengths in Fig. 4.41. As shown in Fig. 4.41, the overall \( da/dN \) of the small-cracks for specimens with two \( SDAS \) initially decrease first with increasing crack length to a rate of \( 4 \times 10^{-8} \text{ m/cycle} \) at a crack length of \( \approx 100 \mu \text{m} \), and then gradually increase from \( 4.0 \times 10^{-8} \) to \( 3.0 \times 10^{-6} \text{ m/cycle} \) with oscillations as the crack propagates from \( 100 \mu \text{m} \) to \( 6000 \mu \text{m} \) at fracture. Overall the specimen with a \( SDAS \) of \( 17 \mu \text{m} \) shows a crack that propagates slower than does the crack in the specimen with a \( SDAS \) of \( 37 \mu \text{m} \).
Fig. 4.41. Small-crack propagation rates for specimens with different $SDAS$.

When compared with the small-crack propagation rates in $HCF$, the small-crack propagating rates in $LCF$ are ~2 to 3 orders of magnitudes higher. It should be noted that the higher $da/dN$ in $LCF$ are related to the coalescence of small-cracks developed from slip bands generated in front of the crack tip, as shown by the replicas for cracks greater than ~100 $\mu$m in the specimen with a $SDAS$ of 17 $\mu$m and cracks with all lengths in the specimen with a $SDAS$ of 37 $\mu$m (Figs. 4.38(c) and 4.39). In $HCF$, plastic strains at the tip of the main crack are not large enough to induce slip bands and micro-cracks.
Fig. 4.42. Small-crack propagation rates for specimens with different SDAS.

The small-crack propagation rate is plotted against $(\Delta K_p)_{\text{eff}}$ in Fig. 4.42. The $da/dN$ of the small-crack tends to decrease with increasing $(\Delta K_p)_{\text{eff}}$ when $(\Delta K_p)_{\text{eff}}$ is $< 7.0$ MPa m$^{1/2}$, similar to the behavior in HCF. Then, with $(\Delta K_p)_{\text{eff}} \geq 7.0$ MPa m$^{1/2}$, the cracks exhibit roughly a linear relationship on a log-log plot of $da/dN$ against $(\Delta K_p)_{\text{eff}}$, following the Paris power law. In LCF, the region with a decreasing $da/dN$ for the specimens with a SDAS of 17 $\mu$m and 37 $\mu$m lasts until $(\Delta K_p)_{\text{eff}}$ of $\sim 4.0$ MPa m$^{1/2}$ and 7.0 MPa m$^{1/2}$, respectively. In HCF, the region with oscillation of the $da/dN$ persists to only $\sim 3.5$ MPa m$^{1/2}$. When the Paris power law is assumed for regions with linear relationship on a log-log plot of $da/dN$ against $(\Delta K_p)_{\text{eff}}$, Paris constants are determined with $da/dN$ measured in units m/cycle and $(\Delta K_p)_{\text{eff}}$ in units MPa m$^{1/2}$. $C$ and $m$, the fitted parameters are $7.0658 \times 10^{-10}$ and 1.5872, respectively.
4.8.3. Interactions between Crack and Microstructural Features

During LCF traces of slip bands and cracks from coarse slip bands were observed on the polished surface of the fatigue specimens. Some coarse slip bands and micro-cracks developed from the slip bands are shown in Figs. 4.43(a) and (b) and 4.44.

Fig. 4.43. Cracks developed from slip bands: (a) and (b) are two different views (SDAS = 17 μm, bending fatigue with maximum strain of 0.00391, R = -1, failed at 8.6 \times 10^5 cycles, loading axis is in vertical direction).

The coarse slip bands and micro-cracks in Fig. 4.43(a) and (b) are from the surface of a specimen with a SDAS of 17 μm, which is ~ 1 mm away from the main-failure crack. The orientation of slip bands changes from grain to grain. After polishing, most of the slip bands and micro-cracks disappeared; only the well-developed cracks from the coarse slip bands were left, indicating the slip bands were initiated on the surface. Newly formed micro-cracks initiated inside of the dendrite cells. Although Si-particles caused micro-cracks to deviate from their overall propagation paths, most cracks penetrated through several dendrite arm spacings regardless of the presence of the eutectic Si-particles.
Most of the slip-band cracks in *LCF* propagate from the surface to a depth of ~500 μm into the specimen. Figure 4.44 shows that when the crack encountered the granular boundary, the crack propagation changed to a different orientation. This indicates that intergranular eutectic constituents, instead of interdendritic eutectic constituents, are more active in influencing the propagation of the slip-band cracks.

Fig. 4.44. The interaction of slip-band crack with the granular boundary (*SDAS = 17 μm*, bending fatigue with maximum strain of 0.00865, *R* = 0.18, failed at $1.67 \times 10^4$ cycles)

As shown in Fig. 4.45 for a *LCF*-specimen with a larger *SDAS* (42 μm), crack paths are linked by micro-cracks from slip bands. Broken interdendritic and intergranular Si-particles are observed on the crack propagation path. On the crack paths of a specimen with a smaller *SDAS*, broken Si-particles were rarely observed. Although the maximum length of Si-particles and the size of the silicon cluster increase as the *SDAS* increases,
the broken Si-particles are not always the largest ones or those with large aspect ratios. In addition to the size of the silicon particles, the local stress state and the orientation of a Si-particle are factors influencing the damage of Si-particles.

Fig. 4.45. Slip-band cracks in a specimen with a SDAS of 42 μm, failed at $3.7 \times 10^3$ cycles. (Bending fatigue with maximum strain of 0.00391, $R = -1$)

4.9. SUMMARY

The fatigue behavior, including total fatigue life, fatigue-crack initiation, small-crack propagation and its interaction with the microstructures, under both low- and high-cycle fatigue-loading conditions, of a cast aluminum alloy (A356.2-T6) with a variety of SDAS and pores, were investigated. The results show the following.
Fatigue Life

When SDAS is less than ~30 μm, the fatigue life varies slightly with SDAS under low- and high-cycle axial and bending-loading conditions. As the SDAS increases beyond 30 μm, fatigue life drops by ~3 times under LCF and ~100 times under HCF at $R = 0.1$. For HCF, the effect of SDAS is overshadowed by the effect of porosity on the fatigue life.

The modified fatigue life ($\frac{N_f}{a_0}$) is well correlated with the initial stress intensity factor range ($\Delta K_o$) for both bending and axial fatigue. Therefore, in the absence of eutectic constituents, the sizes of pores and inclusions are used as parameters to predict the fatigue life of this casting alloy.

Fatigue-Crack Initiation

For HCF, when the SDAS is greater than ~25 to 28 μm, the pores with a length greater than about 100 μm are the main crack-initiation sites; the crack-initiation life is only ~5 % of the total fatigue life and fatigue propagation life dominates the fatigue life. When the SDAS is below ~25 to 28 μm, the pore size is below the critical size of ~80 to 100 μm and large eutectic constituents initiate the HCF cracks; when cracks initiate at the eutectic or small pores, the crack initiation life is as high as ~70 % of the fatigue life. In LCF, the fatigue-initiation life is negligible even when the SDAS is small. Large eutectic constituents result in a lower fatigue life than when that type of initiation site is smaller. The oxide defects initiate the fatigue crack when they are near or at the surface,
regardless of SDAS. "Old" oxide films are more deleterious to fatigue life than the "young" oxide films. Since fatigue cracks sometimes initiate at eutectic constituents in alloy A356.2, a refinement in the size and aspect ratio of the silicon particles and SDAS, would improve the resistance to the fatigue-crack initiation.

Small-Crack Propagation and the Interaction with Microstructures

For HCF, the small-crack propagation rate oscillates in a range of $2.0 \times 10^{-10}$ to $1 \times 10^{-8}$ m/cycle when the crack lengths are below 1000 µm, corresponding to an effective stress-intensity factor range of $\sim 3.5$ MPa m$^{1/2}$. Then after $(\Delta K_p)_{eff} \geq \sim 3.5$ MPa m$^{1/2}$, the crack propagation follows the Paris power. A specimen with the largest SDAS (27 µm) among the three SDAS used shows the highest propagation life. In HCF the crack propagates through the dendrites macroscopically and the tortuous HCF propagation paths are trans-dendrite regardless of the SDAS, with no preference to propagate intergranually or interdendritically. Microscopically, the orientation of crack propagation path changes more often when the crack interacts with intergranular eutectic constituents than with interdendritic eutectic constituents. Debonding is the major type of damage for eutectic Si-particles during fatigue. Silicon particle breaking is found when the SDAS is large or the stress intensity factor range is high.

For LCF, the overall propagation rate tends to decrease to $4 \times 10^{-8}$ m/cycle and then oscillates until $(\Delta K_p)_{eff}$ is $< 7.0$ MPa m$^{1/2}$. The small-crack propagation rate is $\sim 2$ to $3$ orders of magnitudes larger than that in HCF. With a larger SDAS, the propagation rate is
greater, and consequently, the fatigue life in *LCF* is shorter. The large crack-propagation rate in *LCF* is related with the coalescence of micro-cracks developed along slip bands generated at the crack tip. Multiple slip causes the zigzag-cracking in *LCF*. *LCF* cracks propagate through several dendrite cells in one orientation regardless of the presence of the dendrite cell eutectic constituents. Intergranular eutectic constituents, instead of interdendritic eutectic constituents, are more active in influencing the propagation of the slip-band cracks.
CHAPTER 5. EFFECTS OF HIPPING AND SR-MODIFICATION ON
THE FATIGUE BEHAVIOR OF A356.2
CAST ALUMINUM ALLOY

This chapter presents a systematic investigation of the effects of hydrogen content, hipping and Sr-modification on the fatigue behavior, including the fatigue life, small-crack initiation and propagation, and cyclic stress-strain response, in a permanent mold cast and heat-treated aluminum alloy (A356.2-T6). Axial fatigue testing was conducted at a stress ratio of 0.1. The microstructural features (pores, secondary dendrite arm spacing, grain size, and Si-particles) were characterized. The replication method was used during fatigue to detect small-crack initiation and propagation.

5.1. MICROSTRUCTURES

The secondary dendrite arm spacings in the permanent mold castings were ~20 μm at 10 mm from the edge of the plates and ~30 μm at the center of the plates. The measured average grain size in the permanent mold castings was ~191 μm. The measured cooling rate at the center of the plates is ~6.6 K/s. At the similar cooling rate, the grain size is ~440 μm in specimens in directionally solidified ingot. Thus the grains are much finer in permanent cast plates than grains in directionally solidified ingots. The Ti-content is very similar in both type of castings, 0.10% and 0.12 %; but the B-content in the directionally solidified ingot is 0.0001%, somewhat ten times lower than that of 0.001% in the
permanent mold cast plates. Higher B-content yields more TiB₂ heterogeneous nucleus, and then it is no wonder that the grain size is finer in the plates cast in permanent mold.

The typical T6 microstructures before and after hipping are shown in Figs. 5.1 and 5.2. Microstructures without and with Sr-modification were as expected. Although the iron contents in the alloy were only 0.09%, a few needle-like intermetallic constituents were still found.

![Microstructures for non-hipped and T6 heat-treated specimens](image)

Fig. 5.1. Microstructures for non-hipped and T6 heat-treated specimens: the microstructures shown on the upper level were not modified; the microstructures shown on the lower level were Sr-modified.
Fig. 5.2. Microstructures for hipped and T6 heat-treated specimens: the microstructures shown on the upper level were not modified; the microstructures shown on the lower level were Sr-modified.

5.1.1. Silicon Particles

The distributions of size and aspect ratios of the silicon particles before hipping are given in Figs. 5.3 (a), and 5.3 (b); no significant differences of the silicon morphology for Sr-contents from 0.004% to 0.006% were seen in plates with Sr-modification (Group 2 to Group 4). In plates without Sr-modification (Group 1), the eutectic constituents contained larger silicon particles with higher aspect ratios compared with the plates with Sr-modification.
Fig. 5.3. The distributions of size and aspect ratio of silicon particles before hipped: (a) size distribution; and (b) aspect ratio distribution.
Fig. 5.4. The maximum lengths of silicon particles in non-hipped and hipped specimens.

The maximum lengths of silicon particles in an area of 43884 $\mu$m$^2$ at a magnification of 560 times were analyzed for both non-hipped and hipped specimens (Fig. 5.4). Non-hipped and hipped specimens have similar secondary dendrite arm spacings ($SDAS$). Without Sr-modification, the average maximum length is $\sim11.4\pm9.0$ $\mu$m, and the maximum and minimum lengths measured are $\sim54.5$ $\mu$m and $\sim1.7$ $\mu$m before the specimens were hipped. After hipping, without Sr-modification, the average maximum length is $\sim9.5\pm6.3$ $\mu$m, and the maximum and minimum lengths measured are $\sim42.0$ $\mu$m and $\sim2.1$ $\mu$m. It should be noted here that the hipped specimens were already T6 heat-treated before hipping; after hipping, these specimens had to be T6-treated again. The extra time of $\sim14$ hours at $\sim540$ °C resulted in the spherodizing of silicon particles, and the aspect ratio of the silicon particles was decreased as indicated by the decrease of the
maximum length of the silicon particles. With Sr-modification, the average maximum length is \( \sim 5.5 \pm 3.3 \, \mu m \), and the maximum and minimum lengths measured are \( \sim 23.7 \, \mu m \) and \( \sim 1.7 \, \mu m \) before specimens were hipped. After hipping, with Sr-modification, the average maximum length is \( \sim 5.8 \pm 2.5 \, \mu m \), and the maximum and minimum lengths measured are \( \sim 18.0 \, \mu m \) and \( \sim 1.9 \, \mu m \). After hipping, the maximum length of silicon particles in specimens without Sr-modification decreased, and the maximum length of silicon particles in specimens with Sr-modification increased. This is the result of the extra coarsening during hipping and solutionizing.

5.1.2. Pores

The pores were healed and rarely observed in hipped specimens. In this subsection, the quantitative results are from not hipped specimens. Since the accuracy of the measurements of the hydrogen content is 0.03 cc/100 g, the hydrogen contents of groups 1 - 4 in Table 3.2 are roughly comparable, yet the porosity in the plates is very different depending on whether Sr-modification was used. The average area percentage of pores and pore size distributions in microstructures are shown in Figs. 5.5 (a) and (b). Plates without Sr-modification (i.e., Group 1) show the lowest area percentage of porosity (\( \sim 0.04\% \)) and a distribution with the most small pores. Plates with Sr-modification and the highest hydrogen content of 0.31 cc/100 g show the highest area percentage of porosity (1.67%) and a distribution with the most large pores (viz., Group 5). An average area percentage of 0.28% was measured in the plates of Groups 2, 3 and 4, which were Sr-modified and had hydrogen contents of 0.19 to 0.22 cc/100 g. The maximum size of
Fig. 5.5. The average area percentage and size distribution of the pores before hipping: (a) average percentage of pores; and (b) pore size distribution.
the pores ($\sqrt{\text{area}}$) of these five groups are 39 $\mu$m for plates without Sr and 225 $\mu$m, 164 $\mu$m, 158 $\mu$m and 236 $\mu$m for plates with Sr at hydrogen levels from 0.19 to 0.31 cc/100 g. A higher hydrogen concentration in the interdendritic or intergranular liquid during solidification increases the pressure of hydrogen and promotes the early growth of the pores, which results in the increases of amount and size of pores as shown. That Sr-modification resulted in more porosity confirms the results of the previous studies (Fang and Granger, 1989; and Tynelius et al., 1993).

5.2. TENSILE PROPERTIES

5.2.1. Tensile Properties of Non-Hipped A356.2

The ultimate tensile strength, 0.2% offset yield strength and elongation for the five groups of plates before hipping are shown in Table 5.1 and Fig 5.6. For each group of plates six tensile specimens were prepared and tested. In Table 5.1, the mean value and its standard deviation, based on six specimens, are given. Among the five groups of plates studied, plates in Group 1 (without Sr-modification and a hydrogen content of 0.17 cc/100 g) show the highest ultimate tensile strength and yield strength. Among plates in Groups 2, 3, and 4 (with Sr-modification and hydrogen contents of 0.19-0.22 cc/100 g) there are no significant differences in ultimate tensile strength and yield strength. Plates in Group 5 (with Sr-modification and a hydrogen content of 0.31 cc/100 g) show the lowest ultimate tensile and yield strengths. Although plates with Sr have higher hydrogen contents than the plates without Sr, three out of the four show higher values of
This agrees with other researchers (Closset and Gruzleski, 1982; and Schneider and Feikus, 1998) who reported that Sr-modification improves the elongation but not the strength. In this study, the strengths are slightly lower and the elongations are higher than the usually reported tensile properties of A356.2 because the magnesium content of 0.28% is somewhat low.

Table 5.1. Tensile properties of non-hipped A356.2.

<table>
<thead>
<tr>
<th>Group</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>EL (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>285±7</td>
<td>207±4</td>
<td>10.3±3.4</td>
</tr>
<tr>
<td>2</td>
<td>271±13</td>
<td>199±7</td>
<td>9.4±3.9</td>
</tr>
<tr>
<td>3</td>
<td>273±10</td>
<td>200±4</td>
<td>11.6±4.9</td>
</tr>
<tr>
<td>4</td>
<td>278±3</td>
<td>198±2</td>
<td>14.0±4.2</td>
</tr>
<tr>
<td>5</td>
<td>263±10</td>
<td>189±4</td>
<td>11.7±5.7</td>
</tr>
</tbody>
</table>

Note: UTS is ultimate tensile strength; YS is yield strength; and EL is percent elongation. The reported values are the average of six specimens.

Fig. 5.6. Tensile Properties of non-hipped specimens: (a) ultimate tensile strength; (b) 0.2% offset yield strength; and (c) elongation. (6 data in each group)
Fig. 5.6. (Continued)
Saturation of the stress-amplitude of the fatigue hysteresis and small crack behavior could be related to strain-hardening behavior. Hence, the plastic deformation of the tensile specimens was reduced to the form:

\[ \sigma_t = K\varepsilon_p^n \]  

(5.1)

where \( \sigma_t \) is the true stress, \( \varepsilon_p \) is the true plastic strain, \( n \) is the strain-hardening exponent, and \( K \) is the strength coefficient. Plots of true stress versus true strain revealed that the six specimens removed from the same group of plates had almost the same strain-hardening behavior, with the exception that specimens within a group had a rather large standard deviation in the fracture strain. The strength coefficient, \( K \), and strain-hardening exponent, \( n \), for each group, are shown in Table 5.2 along with the respective fracture strains. The strain-hardening exponent equals the maximum uniform strain (Hertzberg, 1996; Caceres, 1998). Four of the six specimens from Group 1 had a fracture strain greater than the strain hardening exponent. From Groups 2 – 4, however, eleven of the fifteen specimens had a fracture strain less than the strain hardening exponent, and in Group 5 all but one had a fracture strain less than the strain exponent. The evidence suggests that whether fracture occurs before necking depends on the amount of porosity rather than modification of the alloy. Indeed, the tensile elongation has been shown to be directly related to the pore area on the fractures (Surappa et al., 1986; Eady and Smith, 1986; Pan et al., 1991; and Caceres and Selling, 1996).
Table 5.2. The average strength coefficient ($K$), and the strain-hardening exponent ($n$) of non-hipped A356.2

<table>
<thead>
<tr>
<th>Group</th>
<th>$K$ (MPa)</th>
<th>$n$</th>
<th>True Plastic Strain at Fracture</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>377.3</td>
<td>0.0872</td>
<td>0.093±0.031</td>
</tr>
<tr>
<td>2</td>
<td>382.0</td>
<td>0.1079</td>
<td>0.083±0.026</td>
</tr>
<tr>
<td>3</td>
<td>370.3</td>
<td>0.0930</td>
<td>0.102±0.040</td>
</tr>
<tr>
<td>4</td>
<td>390.2</td>
<td>0.1108</td>
<td>0.116±0.030</td>
</tr>
<tr>
<td>5</td>
<td>371.2</td>
<td>0.1079</td>
<td>0.098±0.040</td>
</tr>
</tbody>
</table>

Note: The reported values are the average of six specimens.

There is evidence, although not conclusive, that Sr-modification improves the ductility of AlSiMg casting alloys if they can be made with minimal porosity. The tensile properties of Sr-modified and unmodified versions of AlSiMg alloys were measured (Schneider and Feikus, 1998). With 7% Si the tensile properties of Sr-modified and unmodified vacuum die-castings were the same. The castings were heat-treated at 520 °C for one hour, quenched, and aged at 160 °C for six hours. When the silicon content was increased to near-eutectic composition (Al11SiMg), however, the elongation of the Sr-modified alloy was clearly better than the unmodified alloy. Since castings of eutectic alloys are free of porosity or have considerably less porosity, this suggests that Sr-modification improves tensile ductility when porosity is absent. On the other hand, in hypoeutectic alloys (e.g., Al7SiMg) with porosity, the beneficial effect of Sr-modification on tensile ductility is largely offset by the deleterious effect of porosity. Although the elongation was increased somewhat by Sr-modification in three of the four groups in Table 5.1, it is shown in the next section that the fatigue resistance is decreased by the presence of porosity when hipping was not applied.
5.2.2. Tensile Properties of Hipped A356.2

Several hipped specimens from Groups with Sr-modification were tested for tensile properties. As shown in Table 5.3, the ultimate tensile strength, yield strength and elongation for specimens with Sr-modification were all improved after hipping. Unfortunately there were no hipped specimens without Sr-modification available for tensile testing. Nevertheless the Rockwell hardness discussed in next section indicates that the hardness of specimens without Sr-modification was not significantly influenced by hipping; thus the tensile properties of specimens without Sr-modification were probably not influenced by hipping. This suggests that Sr-modification improves tensile properties (ultimate tensile strength, yield strength and ductility) only when porosity is absent, and porosity overshadows the beneficial effect from Sr-modification on the tensile properties.

Table 5.3. Tensile properties of hipped A356.2.

<table>
<thead>
<tr>
<th>Group</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>EL (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>3</td>
<td>307±4</td>
<td>230±3</td>
<td>19.6±0.3</td>
</tr>
<tr>
<td>4</td>
<td>300±1</td>
<td>228±0</td>
<td>14.5±0.0</td>
</tr>
</tbody>
</table>

Note: UTS is ultimate tensile strength; YS is yield strength; and EL is percent elongation. The reported values are the average of two specimens.

Table 5.4. The average strength coefficient (K), and the strain-hardening exponent (n) of hipped A356.2.

<table>
<thead>
<tr>
<th>Group</th>
<th>K (MPa)</th>
<th>n</th>
<th>True Plastic Strain at Fracture</th>
</tr>
</thead>
<tbody>
<tr>
<td>3</td>
<td>425.6</td>
<td>0.1026</td>
<td>0.171±0.003</td>
</tr>
<tr>
<td>4</td>
<td>411.8</td>
<td>0.0989</td>
<td>0.126±0.007</td>
</tr>
</tbody>
</table>

Note: The reported values are the average of two specimens.
The strain-hardening parameters of hipped specimens with Sr-modification are shown in Table 5.4. All the four specimens tested had a fracture strain greater than the strain-hardening exponent. It suggests that after hipping, porosity was eliminated and the maximum uniform strain has been reached.

5.3. ROCKWELL HARDNESS

Rockwell hardness was measured for all non-hipped tensile specimens. Six data points were obtained for each of the tensile specimens in one group. The average values for each of the five groups, the measured maximum and minimum values are shown in Fig. 5.7. Among the five groups of plates studied, plates in Group 1 (without Sr-modification and a hydrogen content of 0.17 cc/100 g) show the highest average hardness. Among plates in Groups 2, 3, and 4 (with Sr-modification and hydrogen contents of 0.19-0.22 cc/100 g) there are no significant differences in the hardness. Plates in Group 5 (with Sr-modification and a hydrogen content of 0.31 cc/100 g) show the lowest hardness.

The hardness of each group of hipped castings was obtained from two specimens (Fig. 5.8). In order to compare the effect of hipping, the hardness obtained from two non-hipped specimens and that from hipped specimens with the similar SDAS are shown in Fig. 5.8. After hipping, the hardness of plates in Groups 2-5 (with Sr-modification and a hydrogen content of 0.19 – 0.31 cc/100 g) was apparently increased compared to the hardness of those not hipped; the hardness of plates with Sr-modification (Group 2 and 3) is as much as that of plates without Sr-modification and even better. This is mainly due to
Fig. 5.7. Rockwell hardness of non-hipped specimens (thirty-six data in each group).

Fig. 5.8. Comparison of Rockwell hardness between not hipped specimens and hipped specimens (twelve data in each group)
the elimination of the pores. Contrary to that the hardness was increased in Groups with Sr-modification after being hipped, the hardness of Group 1 without Sr-modification after being hipped show similar hardness to that before hipping and was not influenced by the prolonged solutionizing time (double T6-heat-treatment) during which the aspect ratio of silicon particles was somewhat decreased as indicated in Fig. 5.4. Indeed, the coarsening of the silicon particles as was not significant enough to cause variation of hardness.

5.4. FATIGUE LIFE

5.4.1. Effects of Hydrogen Content and Sr-Modification on Fatigue Life of Non-Hipped A356.2

High-cycle fatigue lives are shown in Fig. 5.9. In the high-cycle fatigue range, Group 1 without Sr-modification and the lowest hydrogen and porosity contents shows the highest fatigue lives at all stress levels and Group 5 with Sr-modification and the highest hydrogen and porosity contents tends to have the lowest fatigue lives. Specimens from other groups with Sr-modification have intermediate fatigue lives. Some specimens from Group 1 achieved more than $10^7$ cycles without fracture even at a maximum stress of 175 MPa. At maximum stresses below 175 MPa, the maximum fatigue lives for specimens from plates without Sr-modification are almost two orders of magnitude greater than those for specimens removed from plates with Sr-modification and the most hydrogen and porosity.
Fig. 5.9. High-cycle fatigue lives. The stress ratios are 0.1. The arrows pointing to the right indicate that these tests were interrupted before failure. The Sr-contents are in mass percent, and the H-contents are in cc / 100 g.

Fig. 5.10. Fracture surfaces of non-hipped specimens with Sr-modification, maximum stress = 175 MPa and $R = 0.1$. Fatigue cracks initiated at surface pores: (a) failed at $2.67 \times 10^5$ cycles; and (b) failed at $3.26 \times 10^5$ cycles.
Fractography was done on the specimens that failed under high-cycle fatigue conditions in order to identify the crack-initiation sites. Cracks initiated exclusively from pores in specimens from plates with Sr-modification (Fig. 5.10(a) and (b)). Multiple crack-initiation sites at pores were found in some of these specimens. With only two exceptions, the fatigue cracks initiated from pores in the specimens from plates without Sr-modification (Fig. 5.11(a)). One of the specimens had a fatigue crack which initiated from a cluster of silicon particles within a eutectic constituent that was about 85 μm in length (Fig. 5.11(b)). In directionally solidified ingots that were studied previously (Chapter 4), only very small pores were seen on the fracture-surfaces of specimens taken near a water-cooled surface, as a result of rapid cooling rate and good feeding. It was shown that when the secondary dendrite arm spacing was smaller than ~30 μm, the eutectic constituents acted as fatigue crack-initiation sites. In this study, no crack initiation at eutectic constituents (or at oxides) was seen before hipping, except for the two exceptions already noted. The smallest pore which initiated a fatigue crack was about ~75 μm in length.

In the low-cycle fatigue range shown in Fig. 5.12, the fatigue lives for specimens from plates without Sr-modification and the least amount of porosity tend to be higher than fatigue lives for specimens from plates with Sr-modification. Specimens from plates with Sr-modification and hydrogen contents of 0.19 to 0.22 cc/100 g appear to be equivalent, but specimens from Sr-modified plates with 0.31 cc/100 g hydrogen and the most porosity tend to have the lowest fatigue lives overall.
Fig. 5.11. Fracture surfaces of non-hipped specimens without Sr. Fatigue cracks initiated at: (a) a subsurface pore (Max. Stress = 200 MPa, and $R = 0.1$, failed at $3.13 \times 10^5$ cycles); and (b) a eutectic constituent (Max. Stress = 250 MPa, and $R = 0.1$, failed at $4.27 \times 10^4$ cycles).

Fig. 5.12. Low-cycle fatigue lives. The strain ratios are 0.1. The units for the Sr- and H-contents are as same as Fig. 5.9.
5.4.2. *Effect of Hipping on the Fatigue Life of A356.2 without Sr-Modification*

The fatigue lives of the specimens from Group 1 (Table 3.2) without Sr-modification, before and after hipping, are shown in Fig. 5.13. At a maximum stress of 250 MPa, hipped specimens show the higher fatigue lives. At the other maximum stress levels, specimens not hipped show higher fatigue lives. At all maximum stress levels, the hipped specimens have higher fatigue lives at the lower end of the fatigue band compared with specimens not hipped. Although only three randomly selected hipped specimens were tested, the hipped specimens tend to show a narrower scatter band than specimens not hipped at each of the maximum stress levels.

![Graph showing fatigue life](image)

Fig. 5.13. Fatigue lives for specimens from Group 1 without Sr-modification and hydrogen content of 0.17 cc/100g. The stress ratios are 0.1. The arrows have same meaning as in Fig. 5.9.
In order to identify the crack-initiation sites, fractography was done on the failed specimens. In the hipped specimens, the cracks initiated from either a eutectic constituent with a cluster of silicon particles or a large silicon particle, with one exception where a crack initiated at a unhealed pore with a fatigue life of $8.68 \times 10^6$ cycles at a maximum stress of 175 MPa. Figure 5.14 shows crack-initiation sites found on the fractures. It is clear that pores were not dominant in initiating cracks in hipped specimens; instead the cracks initiated at a large Si-particle (Fig. 5.14(a)) or within the eutectic constituent (Fig. 5.14(b)). Although hipping did not improve the highest fatigue lives of specimens from Group 1 without Sr-modification, hipping did improve the fatigue performance by narrowing the fatigue scatter band, as a result of the elimination of pores as crack-initiators at the lower end of the fatigue scatter band.

![Fracture surfaces of hipped specimens without Sr-modification (Group 1). Fatigue cracks initiated at: (a) a large silicon particle (failed at $3.6 \times 10^6$ cycles); and (b) a eutectic constituent (failed at $8.7 \times 10^5$ cycles). Both were tested with a maximum stress of 175 MPa and a stress ratio of 0.1.](image-url)
5.4.3. Effect of Hipping on the Fatigue Life of A356.2 with Sr-Modification

The fatigue lives of the specimens from Group 2, with Sr-modification and a hydrogen content of 0.19 cc/100 g, and Group 5, with Sr-modification and the highest hydrogen are shown in Figs. 5.15 and 5.16, respectively. Unlike specimens from Group 1 without Sr-modification, hipped specimens from Groups 2 and 5 with Sr-modification significantly improved fatigue lives at the four maximum stress levels tested. The fatigue lives at the lower end of the scatter band at each stress level were tremendously improved after hipping and were even higher than the highest fatigue lives of specimens not hipped. It should be noted that at a maximum stress level of 175 MPa, one specimen from each of Groups 2 and 5 reached a fatigue life of $\sim 10^7$ cycles without fracture when the test was interrupted. None of the specimens not hipped from Groups 2 and 5 with Sr-modification reached such a long fatigue life without fracture at a stress level of 175 MPa, indicating hipping greatly improved the fatigue lives of specimens from the plate-castings that were modified with Sr.

SEM examinations of the fractures showed that cracks initiated at either eutectic constituents or at oxides, with only one exception from Group 2, in which the crack initiated at a unhealed pore, with a fatigue life of $2.19 \times 10^5$ cycles at a maximum stress level of 200 MPa. Fig. 5.17 shows some crack-initiation sites found on the fractures. The non-hipped specimens from Groups 2 and 5, with Sr-modification, had the higher area percentages of porosity and distributions with larger pores (Group 5, especially) compared with specimens from Group 1, without Sr-modification. It is clear from Figs.
Fig. 5.15. Fatigue lives for specimens from Group 2 with Sr-modification. The stress ratios are 0.1. The arrows have same meaning as in Fig. 5.9.

Fig. 5.16 Fatigue lives for specimens from Group 5 with Sr-modification. The stress ratios are 0.1. The arrows have same meaning as in Fig. 5.9.
5.15 and 5.16 that hipping is very effective in healing the pores in specimens from Groups 2 and 5 with Sr-modification and results in a significantly improved fatigue life.

![Fracture surfaces of hipped specimens with Sr-modification (Group 2). Fatigue crack initiated at: (a) a eutectic constituent (maximum stress = 200 MPa, $R = 0.1$, failed at $5.8 \times 10^4$ cycles); and (b) oxide (maximum stress = 250 MPa, $R = 0.1$, failed at $7.1 \times 10^4$ cycles).](image)

5.4.4. Effects of Sr-Modification on Fatigue Lives

Fatigue lives of hipped specimens with and without Sr-modification are shown together in Fig. 5.18. Contrary to the previous results of non-hipped specimens, where specimens from Group 1 without Sr-modification showed the highest fatigue life and the beneficial result that Sr-modification might have had were overshadowed by the deleterious effect of pores, the specimens from Groups 2 and 5 with Sr-modification have the highest fatigue lives provided that they have been hipped. Some hipped specimens from Groups 2
and 5 achieved more than $10^7$ cycles without fracture, even at a maximum stress of 175 MPa.

Fig. 5.18. Fatigue lives for hipped specimens with and without Sr-modification. The arrows have same meaning as in Fig. 5.9. The Sr-contents are in weight percent, and the H-contents are in cc/100g.

As presented in previous sections, hipping eliminated pores as the crack initiators, and cracks almost always initiated at eutectic constituents and sometimes at oxides. The morphologies of silicon particles are very different in the eutectic constituents for specimens with and without Sr-modification. In subsection 5.1.1, it was shown that, without Sr-modification (Group 1), the eutectic constituents contained larger silicon particles with higher aspect ratios compared to microstructures of the modified alloy. Simulations (Fan and McDowell 1998) showed that refining the silicon particles should improve the fatigue strength by increasing the crack-initiation life. It is also shown that
particle shape and alignment are the most dominant parameters influencing the particle fracture and debonding in eutectic constituents (Gall et al. 2000). SEM examinations on the fractures showed that, after hipping, the crack initiators were at eutectic constituents when the alloy was Sr-modified, but the crack initiators were at some large silicon particles for specimens from Group 1 without Sr-modification. It appears that the fatigue life of Sr-modified alloys could be improved if the sizes of the eutectic constituents or if the sizes of the Si-particles were decreased. In fact, it should be noted that a solutionizing period of four hours (rather than twelve) may have been adequate, since the dendrite arm spacings in the test-castings are only ~20 to 30 μm.

5.5. EFFECT OF CRACK-INITIATION SITE ON FATIGUE LIFE

The maximum lengths of the crack-initiation sites (i.e., mostly pores) as functions of the fatigue lives at 250, 225, 200, 175, 150 and 125 MPa are shown in Figs. 5.19 (a)-(f), respectively. The fatigue life tends to decrease as the size of the initiation-site increases for tests at all maximum stresses. In Figs. 5.19 (a)-(d), the maximum sizes of the crack-initiation pores of specimens from plates without Sr-modification (i.e., Group 1) are mostly less than 200 μm and smaller than the sites in specimens from plates with modification. In Figs. 5.19 (e) and (f), the size of the initiation pore in Group 1 could not be measured because the specimens did not fail after more than $10^7$ cycles; presumably these pores are also less than 200 μm in length. In the specimens taken from the modified plates, the size of most sites varied from 150 μm to 1000 μm, and two were as large as
Fig. 5.19. The relationship between high-cycle fatigue life and the maximum size of initiation site. The maximum stresses are: (a) 250 MPa; (b) 225 MPa; (c) 200 MPa; (d) 175 MPa; (e) 150 MPa; and (f) 125 MPa. Specimens without fracture when interrupted the tests are plotted on the horizontal axial.
Fig. 5.19. (Continued)
Fig. 5.19 (Continued)

(i)

Fatigue Life, cycles

Maximum Length of Initiation Site, mm

(e)

Fatigue Life, cycles

Maximum Length of Initiation Site, mm
The maximum sizes of the crack-initiation pores on the fractures are larger than the sizes from the image analysis study (Fig. 5.5). The fatigue lives for tests at maximum stress of 125 MPa clustered together around $10^6$ cycles, but three specimens had fatigue lives of $10^7$ to $2\times10^7$ cycles without fracture when the tests were stopped. This is similar to the previous result (Section 4.1.5) of bending fatigue tests at strain ratio of 0.1 where fatigue lives for specimens initiated at different sizes of pores, which also clustered together at around $10^6$ cycles.

![Graph](image)

Fig. 5.20. The size of the initiation site and the maximum stress for high-cycle fatigue tests.

The sizes of the initiation-sites are plotted against the maximum stresses in Fig. 5.20. It is seen that the smallest pores serving as crack-initiation sites are in the range of 80 to 100
μm, all of which are associated with the unmodified specimens (Group 1). With one exception, the largest sites were found in the Sr-modified non-hipped specimens with the most porosity (Group 5).

Figure 5.21 shows the relationship between fatigue life and the initial stress intensity factor range, \( \Delta K_0 \). The calculation of \( \Delta K_0 \) was discussed in Chapter 4:

\[
\Delta K_0 = UF(\Delta S)\sqrt{a_0}
\]

where \( a_0 \) is the maximum length of the initiation-site, \( \Delta S \) is the stress range, and \( F \) is 0.5 for a pore and 0.65 for a eutectic constituent (Murakami 1990; Shiozawa, 1997). \( U \) is the crack closure coefficient, which is assumed to be 1 for \( R \) ratio of 0.1. The modified fatigue life (\( N_f/a_0 \)) increases as the initial stress intensity factor range decreases, although there is a lot of scatter in Fig. 5.21. For fatigue tests at high stress levels, the fatigue cracks initiated early, and the total fatigue life consists mostly of the propagation life (Couper et al., 1990; Gungor and Edwards, 1993; Skallerud et al., 1993; Grant et al., 1993; and Odegard and Pedersen, 1994). Based on this linear fracture mechanics model (Eq. 2.11), large pores have a greater stress intensity factor range and result in faster crack propagation rates, so less fatigue life. But at low stress levels, with fatigue lives \( \geq 10^6 \) cycles, using linear fracture mechanics and treating the pore as a propagating crack may severely underestimate the fatigue life, as has been shown in several references (Gungor and Edwards, 1993; Grant et al., 1993; and Nadot et al., 1999) where small-crack behavior is discussed. Crack-propagation life could be related to the porosity size.
Fig. 5.21. Modified high-cycle fatigue life versus the initial stress intensity factor range for specimens from cast plates.

Fig. 5.22. Modified high-cycle fatigue life versus the initial stress intensity factor range for specimens from cast plates and directionally solidified ingots.
for fatigue lives below \(\sim 10^6\) cycles, but for fatigue lives above \(\sim 10^6\) cycles, the effect of pore shape and small crack behavior must be taken into account. Nevertheless, the data in Fig. 5.21 were fitted using Eq. 4.9 and the parameters for crack propagation in Eq. 2.11 are obtained with \(N_f/a_0\) measured in units cycles/m and \(\Delta K_0\) in units MPa m\(^{1.5}\). \(C\) and \(m\) are \(1.8936 \times 10^{-10}\) and \(4.5405\) respectively.

Alloys obtained from cast plates used in this Chapter and directionally solidified ingots used in Chapter 4 have similar chemical composition and heat-treatment. When assuming they have same propagation parameters, the data in Fig. 5.21 and Fig. 4.29 were fitted together using Eq. 4.9, as shown in Fig. 5.22. The two sets of data overlap well. The parameters for crack propagation in Eq. 2.11 are obtained. \(C\) and \(m\) are \(1.38184 \times 10^{-10}\) and 4.12, respectively.

### 5.6. CYCLIC STRESS-STRAIN RESPONSE OF A356.2

The cyclic stress-strain curve provides a measure of the steady-state cyclic deformation resistance of a material (Landgraf et al., 1969). The cyclic stress-strain response of a material may be greatly different from the monotonic stress-strain curve and is dependant on the initial state (heat-treatment, and cold work), and its test conditions. The cyclic stress-strain curve is the fundamental constitutive relationship in fatigue analysis and design. Then in order to investigate whether Sr-modification affects the cyclic behavior of A356.2 and for later use in predicting fatigue lives, it is necessary to obtain the cyclic
strain-stress relationships. Thus, cyclic stress-strain responses of A356.2 from strain-controlled and load-controlled fatigue tests were studied.

5.6.1. Cyclic Stress-Strain Response of A356.2 from Strain-Controlled Fatigue

Figures 5.23(a) and (b) show the history of maximum, minimum, amplitude and mean stresses for non-hipped specimens at a maximum strain of 0.014 and a strain ratio of 0.1. For specimens without Sr-modification, at saturation, the increases of maximum stress (tension), minimum stress (compression), and the stress amplitude are 25 MPa (from 234 to 259 MPa), 51 MPa (from -199 to -251 MPa), and 40 MPa (from 216 to 256 MPa); and the decrease of mean stress is 14 MPa (from 18 to 4 MPa). For specimens with Sr-modification, the increases of maximum stress, minimum stress, and the stress amplitude are 21 MPa (from 209 to 230 MPa), 38 MPa (from -192 to -230 MPa ) and 30 MPa (from 200 to 230 MPa) and the decrease of the mean stress is 9 MPa (from 9 to 0 MPa). Similar trends were observed in specimens tested at other maximum strains.

The increases of the maximum stress, minimum stress and amplitude stress during fatigue indicate that cyclic hardening occurred during strain-controlled fatigue. However, cyclic hardening behavior depends on the microstructures. In the experimental studies, as it is not shown here, specimens with more and relatively large pores failed to show any increase of the maximum stress, although some increases of the minimum stress and the amplitude stress were observed. Thus, the presence of more and larger pores in specimens decreased the cyclic hardening effect. The decrease of the mean stress with the
Fig. 5.23. Cyclic strain hardening and mean stress relaxation for strain controlled fatigue at a maximum strain of 0.014, and a strain ratio of 0.1: (a) without Sr-modification and $H = 0.17 \text{ cc} /100 \text{ gm}$; (b) with Sr-modification and $H = 0.23 \text{ cc} /100 \text{ gm}$. 
cyclic reversal numbers indicates that cyclic relaxation occurred during strain-controlled fatigue. In this research, the strain ratio was 0.1 and tensile mean strain was produced; the large mean strains induced cyclic plasticity and the resulting mean stress gradually relaxed towards a stable value as increasing the cyclic reversal numbers. When greater strain, thus greater plastic strain, was involved in specimens with Sr-modification (Fig. 5.23(b)), at a maximum strain of 0.014, the mean stress tended to relax to zero.

The stress amplitude variations during fatigue tests were further studied at four maximum strains. Fig. 5.24 (a) shows the stress amplitude history for specimens from plates without Sr-modification (Group 1). At a maximum strain of 0.006, stress-amplitude saturation of 155 MPa occurs at ~5% of the total fatigue life. At a maximum strain of 0.008, the stress-amplitude saturation of 210 MPa occurs at ~30% of the total fatigue. At maximum strains greater than 0.008, stress amplitude increases to a maximum without saturation followed by fracture. Fig. 5.24 (b) shows the stress-amplitude history for specimens from plates with Sr-modification and a hydrogen content of 0.21 cc/100 g (Group 3). At a maximum strain of 0.006, a stress-amplitude saturation of 163 MPa occurs at ~30% of total fatigue life. At a maximum strain ≥ 0.008, the stress-amplitude increases to a maximum, without saturation, when a drop associated with fracture commences. At the higher maximum strain levels, it appears that the pores accommodate the remote strain by localizing the plastic deformation, and the specimen is fatigued to fracture at an early stage without uniform plastic deformation in the bulk specimen.
Fig. 5.24. Typical stress amplitude history during strain-control fatigue for not hipped specimens: (a) without Sr-modification and H-content of 0.17 cc/100 g; (b) with Sr-modification and H-content of 0.19 cc/100 g.
Although there is no agreement on the exact definition of the cyclic stress-strain curve, the one with some acceptance is the locus of tips of the stable hysteresis loops from several companion tests at different completely reversed constant strain amplitudes (Landgraf et al. 1969). Three methods, constant amplitude tests, multiple step tests, and incremental step tests are commonly used (Suresh, 1998). Most procedures involve plotting the locus of tips of stress-strain hysteresis loops from specimens subjected to different levels of cyclic stress or strain. Actually most of the reported cyclic strain-stress relationships were obtained by fitting the plastic strain amplitude with the stress amplitude obtained from reversed strain-controlled fatigue tests according to the following equation:

\[
\sigma_a = K' \left( \frac{\Delta \varepsilon_p}{2} \right)^{n'}
\]

where \( \sigma_a \) is the stable stress amplitude; \( \Delta \varepsilon_p \) is the cyclic plastic strain range; \( n' \) is the cyclic strain hardening exponent; and \( K' \) is the cyclic strength coefficient. It should be noted that, in doing so, the underlying assumption is kinematic hardening behavior. In fatigue life estimations, unloading and reloading during cycling is approximated as following cyclic stress-strain paths that are expanded with a scale factor of two relative to the cyclic strain-stress curve; the underlying assumption is kinematic hardening (Dowling, 1993). If the material shows kinematic hardening behavior, then it is expected that the result of fitting the stable plastic strain amplitude with the stress amplitude would be similar to that fitting the
maximum plastic strain with the maximum stress at the locus. But in this research, differences were found when fitting only the maximum stresses with the maximum plastic strains, or only the stress amplitudes with the plastic strain amplitudes. This is possibly due to the assumption of kinematic hardening. Because kinematic hardening has been used to approximate the unloading behavior of most metallic materials and it predicts Bauschinger effect (the early yielding phenomena on unloading), the early yielding may not exactly satisfy the scale factor of two in the kinematic hardening behavior. In the following discussions, the cyclic stress-strain curve was constructed by fitting the stable plastic strain amplitude with the stress amplitude, and maximum plastic strain with the maximum stress together according to Eq. (5.2).

Cyclic stress-strain responses of A356.2 specimens with and without Sr-modification are shown in Figs. 5.25(a) and (b). As previously mentioned, pores accommodate the remote strain by localizing the plastic deformation, and the specimens were fatigued to fracture without full saturation of stress and stress amplitudes. In order to exclude the effect of pores on the cyclic response, only specimens with saturated values of maximum stresses and stress amplitudes were fitted to obtain the cyclic stress-strain relationship. The cyclic strength coefficient ($K'$), cyclic hardening exponent ($n'$), and the square of correlation coefficient ($r^2$) are 331.89 MPa, 0.0526, and 0.6019 for A356.2 without Sr-modification; and 331.36 MPa, 0.0558 and 0.4098 for A356.2 with Sr-modification.
Fig. 5.25. Cyclic stress strain relationship obtained from strain-controlled fatigue tests on non-hipped specimens: (a) no Sr; and (b) with Sr.
The cyclic stress-strain relationships obtained from A356.2 without and with Sr-modification are shown together with those tensile stress-strain relationships in Fig. 5.26. A356.2 with and without Sr-modification showed very similar cyclic strength coefficients and cyclic strain hardening exponents. The cyclic strain hardening exponents are smaller than those in tensile tests, where the strength coefficient \( K \), hardening exponent \( n \) are 377.3 MPa and 0.0872 for A356.2 without Sr-modification; and 377.7 MPa and 0.104 for A356.2 with Sr-modification. Thus the strain-hardening rate during fatigue cycling was decreased when compared with that during tensile test although the initial hardening rate was increased when the plastic deformation was small. This phenomena is considered to be related to microstructural damage under different loadings. The strain hardening rate in particle-reinforced metal matrix composites was studied and found to be closely
related to the plastic flow of the matrix and the damage accumulation such as particle cracking and debonding (Corbin and Wilkinson; 1994). When particles are not effective at bearing the load, the strain-hardening rate decreases. Although the damage of silicon particles and Mg$_2$Si precipitates during fatigue are not studied in detail in this research, previous research showed that the damage accumulation in particle reinforced composites increases approximately linearly with plastic strain (Lloyd, 1991; Corbin and Wilkinson, 1994). As indicated by the decreased cyclic hardening rate during fatigue when the plastic strains are large, it is believed that more damage accumulation (particle broken and debonding) occurs during fatigue cycling than during tensile test.

5.6.2. Cyclic Stress-Strain Response from Load-Controlled Fatigue on Non-Hipped Specimens

Although strain-controlled fatigue has often been used to construct cyclic-stress-strain curves, the cyclic deformation response of A356.2 in load-controlled fatigue has not been previously studied. Thus the variations of maximum and minimum strains were recorded during load-controlled fatigue to show the effects of Sr-modification and hipping in A356.2 alloy.

Some histories of maximum, minimum, amplitude and mean strains are shown for specimens from plates without Sr-modification and with Sr-modification in Figs. 5.27(a) and (b). At a maximum stress of 250 MPa, with the increase of the cyclic numbers, the maximum stain (tension), the minimum strain (tension), and the mean strain (tension) first increase and then saturated in less than ~30 cycles; but the amplitude strain
Fig. 5.27. Strain history for load-controlled fatigue on specimens at a maximum stress of 250 MPa, and a stress ratio of 0.1: (a) without Sr (from Group 1); and (b) with Sr (from Group 2).
decreases before saturation. The tensile properties are given in Table 5.1. At 250 MPa, the maximum stress is between the yield strengths and ultimate tensile strengths of the specimens. Since the specimen from Group 1 (Fig. 5.27(a)) has a significantly greater ultimate tensile strength than that in Group 2 (Fig. 5.27(b)), it strains significantly less during fatigue. At a maximum stress of 225 MPa, similar trends were observed but with much smaller strains. As is not shown here, when the maximum stress was ≤ 200 MPa, which is about equal to the yield strengths (Table 5.1), most of the specimens did not show measurable variations of strains during fatigue. The phenomenon of the increase of strain shown in Figs. 5.27(a) and (b) is known as cyclic creep (Suresh, 1998). It takes place when the plastic deformation during the loading portion is not opposed by an equal amount of yielding in the reverse loading direction. Cyclic creep in A356.2 fatigued at room temperature has not been reported. These results show, however, that when the maximum stress exceeds the yield strength of the alloy, cyclic creep was observed in the first 30 cycles. A356.2 alloy is a cyclic strain-hardening material as previously shown in the strain-controlled fatigue, then it is expected that cyclic creep will be offset by the strain hardening.

Cyclic creep is shown in Fig. 5.28 for A356.2 without and with Sr-modification. Here cyclic-creep strain is defined as the increase of either the maximum or minimum strain from the first cycle to the cycle when the saturated stable strain is reached. It is clear that specimens with Sr-modification show more cyclic creep than specimens without Sr. The greater the maximum loading stress, the larger the difference of the cyclic creep between
Fig. 5.28. Cyclic creep for non-hipped specimens in load-controlled fatigue. The clusters of data to the right with stresses greater than 165 MPa represent the maximum stresses; the stresses less than 25 MPa represent the minimum stresses during fatigue cycles.

Fig 5.29. Cyclic stress and strain relationship from stress-controlled fatigue on non-hipped specimens.
them. At a maximum stress of 250 MPa, specimens with Sr show cyclic creep of ~ 4 times as much as those without Sr-modification. When a maximum stress is 200 MPa or less, A356.2 with/without Sr-modification shows negligible cyclic creep. The radiographs of those specimens with larger cyclic creeps than specimens with smaller cyclic creeps showed that the larger cyclic creep is associated with large pores.

After the transient period in ~ 30 cycles, the cyclic creep reached saturation. At saturation the maximum plastic strain and the strain amplitude are plotted in Fig. 5.29 and fitted according to Eq.(5.2). Also shown are the true stress-true strain curves based on tensile properties. Results from specimens with and without Sr-modification are shown in Fig. 5.29. The cyclic strength coefficient ($K'$), cyclic hardening exponent ($n'$), and the square of the correlation coefficient ($r^2$) are 375.29 MPa, 0.0932, and 0.9921 for A356.2 without Sr-modification; and 354.5 MPa, 0.0952 and 0.9866 for A356.2 with Sr-modification. In fatigue under load-control with $R$ ratio of 0.1, the cyclic creep induced by the non-zero mean tensile stress apparently decreased the cyclic strength coefficient and the cyclic hardening exponent in cyclic stress-strain curve, compared to those in the tensile stress-strain curves.

5.6.3. Cyclic Stress-Strain Relationship from Load-Controlled Fatigue on Hipped Specimens

Similar to the non-hipped case, histories of maximum, minimum, amplitude, and mean strains are shown for hipped specimens from plates without Sr-modification and with Sr-
Fig. 5.30. Strain history during load-control fatigue for hipped specimens at a maximum stress of 250 MPa and a stress ratio of 0.1: (a) without Sr (from Group 1); and (b) with Sr (from Group 2).
modification in Figs. 5.30 (a) and (b). Similar to specimens not hipped, the hipped specimens also showed cyclic creep behavior in ~ 30 cycles after fatigue started.

![Graph showing cyclic creep in hipped specimens](image)

**Fig. 5.31.** Cyclic creep in hipped specimens in load-controlled fatigue. The clusters of data to the right with stresses greater than 165 MPa represent the maximum stresses; the stresses less than 25 MPa represent the minimum stresses during fatigue cycles.

Cyclic creep is shown in Fig. 5.31 for hipped specimens without and with Sr-modification. Although pores were closed in hipped specimens with Sr-modification, these specimens with exhibit a cyclic creep of at least two times that of specimens without Sr-modification, as a result of the improvement in the tensile properties (compare Figs. 5.6(a), (b) and (c) to Table 5.3) associated with the modification of the silicon morphology. When compared with specimens not hipped (Fig. 5.28), the hipped
specimens with Sr-modification in Fig. 5.31 show about one-half of the cyclic creep; hipped specimens without Sr-modification also show less cyclic creep.

After the cyclic creep reached saturation, the cyclic stress-strain responses of hipped specimens with/without Sr, represented by stable maximum plastic strain against maximum stress, and plastic strain amplitude against stress amplitude were shown in Fig. 5.32 and fitted according to Eq. 5.2. The cyclic strength coefficient ($K'$), cyclic hardening exponent ($n$), and the square of the correlation coefficient ($r^2$) are 331.69 MPa, 0.0545, and 0.9784 for A356.2 without Sr-modification; and 317.07 MPa, 0.0515 and 0.969 for A356.2 with Sr-modification.

![Cyclic stress and strain relationship from stress-controlled fatigue on hipped specimens.](image)

Fig. 5.32. Cyclic stress and strain relationship from stress-controlled fatigue on hipped specimens.
In hipped specimens with Sr-modification, the cyclic strength coefficient and the cyclic hardening exponent (which are 317.07 MPa and 0.0515) were less than those tensile strength coefficient and hardening exponent (which are 421 MPa and 0.1028) of the hipped specimens, indicating a decreased hardening rate and greater damage accumulation during fatigue cycling when compared at the same strain level. Although the tensile strength coefficient and hardening exponent of hipped specimens without Sr-modification are not available, it is expected that the cyclic strength coefficient and hardening exponent of specimens without Sr-modification (which are 331.69 MPa and 0.05452) will be also less than those of tensile strength coefficient and hardening exponent of specimens without Sr-modification, since the tensile strength coefficient and hardening exponent of hipped specimens without Sr are greater or at least similar to those of not hipped unmodified specimens (which are 377.3 MPa and 0.0872), as result of the less pores in not hipped unmodified specimens.

When the fitted curves for hipped and non-hipped A356.2 are shown together in Fig. 5.33, it can be seen that hipping increases the cyclic yield strength of A356.2, especially the cyclic yield strength of the Sr-modified alloy as a result of the elimination of the pores. Hipping first increases the cyclic hardening rates of A356.2 as indicated by the initial hardening rates when the plastic strain is small (\(<\sim 0.005\)); and then decreases the cyclic hardening rate as indicated by the decreases of the cyclic hardening exponent (from 0.0932 to 0.0545 for A356.2 without Sr; and from 0.0952 to 0.0515 for A356.2 with Sr). The strain hardening rate was found to be closely related to the plastic flow of
the matrix and the damage accumulation such as particle cracking and debonding (Corbin and Wilkinson; 1994). Then, after hipping, large pores and micro-pores are closed, and the yield strength is increased; thus the initial hardening rates were increased as result of the effective load transfer. But as large plastic deformation is reached, particles break or debond, and then particles are not effective at bearing the load, resulting in decrease of the strain-hardening rate.

![Graph](image)

Fig. 5.33. Effect of hipping on the cyclic stress-strain relationship obtained from stress-controlled fatigue tests.

5.7. SMALL-CRACK INITIATION AND PROPAGATION

5.7.1. Small-Crack Initiation and Propagation in Non-Hipped A356.2

In this section references is made to the size of the defect (usually a pore) at which a crack has initiated. The size is a characteristic length, or simply length, taken as the
square root of the projected area of the defect. Also, the maximum length was
determined. These were measured, after failure, in images of the fractures obtained by
scanning electron microscopy.

Replicas were taken from non-hipped specimens with Sr-modification at maximum
stresses of 125, 150, and 175 MPa, at a \( R \) ratio of 0.1. Two replicas are shown in Figs.
5.34(a) and (b). The cracks in Fig. 5.34(a) initiated at a surface pore of a maximum length
of \(~346\ \mu m\). The crack in Fig. 5.34(b) initiated at a corner pore with a maximum length
of \(~173\ \mu m\).

Fig. 5.34. Small-cracks from non-hipped specimens with Sr: (a) replica taken at \(1.30 \times 10^5\) cycles; maximum stress = 175 MPa, failed at \(1.37 \times 10^5\) cycles; (b) replica
taken at \(3.7 \times 10^5\) cycles; maximum stress = 150 MPa, failed at \(4.6 \times 10^5\)
cycles.
Table 5.5 is a summary of the fatigue lives, and type and sizes of the crack-initiation sites. Two specimens have two cracks initiated independently at two different pores. The lengths of the cracks that become the fatal cracks in their later stages were monitored.

Table 5.5. Cycles when replicas were taken from specimens of Sr-modified A356.2 and characteristics of the crack-initiation sites

<table>
<thead>
<tr>
<th>Maximum Stress (MPa)</th>
<th>Total Fatigue Life (cycles)</th>
<th>Fatigue Crack-Initiation Site</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Defect</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>125</td>
<td>1.64 \times 10^5</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>1.89 \times 10^3</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>1.30 \times 10^6</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>(9.69 \times 10^5)</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>6.70 \times 10^5</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>(5.05 \times 10^5)</td>
<td>P</td>
</tr>
<tr>
<td>150</td>
<td>4.57 \times 10^5</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>9.5 \times 10^4</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>(8.5 \times 10^4)</td>
<td>P</td>
</tr>
<tr>
<td>175</td>
<td>1.37 \times 10^5</td>
<td>P</td>
</tr>
<tr>
<td></td>
<td>5.52 \times 10^4</td>
<td>O</td>
</tr>
<tr>
<td></td>
<td>1.36 \times 10^5</td>
<td>P</td>
</tr>
</tbody>
</table>

Note: P and O represent pore and oxide, respectively; SC and CC stand for surface crack and corner crack, respectively. The numbers in parentheses are the number of cycles captured before the specimen fractured or interrupted.

Crack length measurements as a function of number of cycles are presented in Figs. 5.35 (a)–(c). It is shown that the fatigue lives mostly consist of the cycles in forming a small
Fig. 5.35. Crack development during fatigue. The maximum stresses are: (a) 125 MPa; (b) 150 MPa; and (c) 175 MPa. (The lengths of the crack-initiation sites are in μm)
Fig. 5.35. (Continued)

Fig. 5.36. Fatigue crack initiation and propagation lives. The maximum stresses are: (a) 125 MPa; (b) 150 MPa; and (c) 175 MPa. (The maximum length and the length (parenthesis) of the crack-initiation sites are in μm)
Fig. 5.36. (Continued)
crack of about 1 mm. For these specimens, the fatigue crack achieved a length of \( \sim 3.0 \) to \( 4.0 \) mm before the final fracture. The crack-initiation and propagation lives are shown in Fig. 5.36. The crack-initiation life is defined as the average of the number of cycles when the crack was first captured on a replica and the number of cycles when the previous replica showed no cracks.

When tested at a maximum stress of 125 MPa, the specimen with its crack initiated at a pore of length of 215 \( \mu \text{m} \) had an initiation life of \( 9.7 \times 10^5 \) cycles, which was \( \sim 74\% \) of the total fatigue life. With the larger pores, the initiation life tends to decrease; e.g., the pores of lengths of \( \sim 549 \) and 593 \( \mu \text{m} \), the initiation life is only 3 to 6 \% of the total fatigue life. At a maximum stress of 150 MPa, the specimen with its crack initiated at a pore of length 116 \( \mu \text{m} \) has an initiation life of only \( 5 \times 10^3 \) cycles which is \( \sim 1\% \) of the fatigue life. When the length of the crack-initiation pore is 275 \( \mu \text{m} \), the initiation life is also \( 5 \times 10^3 \) cycles which is about 5\% of the fatigue life. At maximum stress of 175 MPa, the specimen with its crack started at a large oxide film of with a length 668 \( \mu \text{m} \) (the maximum length of 1333 \( \mu \text{m} \)) has an initiation life of only \( 3 \times 10^3 \) cycles which is about 5\% of the fatigue life. Tested at the same maximum stress, the specimen with crack at a pore of a length 247 \( \mu \text{m} \) has an initiation life of \( 1 \times 10^4 \) cycles which is about 7\% of the total fatigue life. Hence, the cracks that initiate at the smaller defect have a greater initiation lives.
When the maximum stress is greater than 125 MPa, pores with lengths greater than ~116 μm, have an initiation life is less than ~ 10% of the total fatigue life. Small-crack propagation life dominates, and the crack-initiation life can be neglected. At a maximum stress of 125 MPa, a pore has a length 215 μm, then, the initiation life is as high as 70% of the total fatigue life. Thus, when pores are eliminated, the fatigue-crack initiation life is increased, and the total fatigue life is greatly improved.

5.7.2. Small-Crack Initiation and Propagation in Hipped A356.2

In order to study the effect of Sr-modification on small-crack behavior, surface replicas were made on hipped specimens made with and without Sr-modification at maximum stresses of 200 and 225 MPa. Since the non-hipped specimens without Sr have small and significantly less pore volume, they showed similar microstructure and Rockwell hardness with those of hipped specimens without Sr.

Two replicas of small-cracks from specimens without and with Sr are shown in Figs. 5.37(a) and (b). The crack in Fig. 5.37(a) initiated from a large silicon particle of a maximum length of ~35 μm in a specimen without Sr. The crack in Fig. 5.37(b) initiated from a eutectic constituent of maximum length of ~116 μm on the fracture in a specimen with Sr. The crack lengths as a function of cycles are shown in Figs. 5.38 (a) and (b). Again, the fatigue lives mostly consist of the cycles in forming a small crack of about 1 mm; the fatigue cracks achieved lengths of less than ~3 mm before the final fractures.
Fig. 5.37. Small-cracks from specimens: (a) without Sr (replica was taken at $7.54 \times 10^4$ cycles; maximum stress = 200 MPa, failed at $2.4 \times 10^5$ cycles); and (b) with Sr (replica was taken at $1.0 \times 10^5$ cycles; maximum stress = 200 MPa, failed at $5.8 \times 10^5$ cycles).
Fig. 5.38. Crack development during fatigue. The maximum stresses are: (a) 200 MPa; and (b) 225 MPa. (P, E, O refer to pore, the eutectic constituent and oxide, respectively; the length of crack initiation sites is expressed in μm.)
Table 5.6 is a summary of the fatigue lives and type and sizes of the crack-initiation sites for specimens tested at maximum stress levels of 200 and 225 MPa. The crack initiation and propagation lives of the eight specimens in Table 5.6 are shown in Fig. 5.39 (a) and (b). The crack-initiation life was defined as described before. At a maximum stress of 200 MPa (Fig. 5.39(a)), the two specimens, with their cracks initiated at pores of lengths 163 and 80 µm were not hipped and not modified. They have the shortest crack-initiation lives; one has an initiation life of only $1.5 \times 10^4$ and the other of $2.0 \times 10^4$ cycles, which are respectively $\sim 13.5\%$ and $\sim 8\%$ of the total fatigue lives. The hipped specimen with no Sr, with its crack initiated at a eutectic constituent with a length 42 µm, has a fatigue initiation life of $7.5 \times 10^4$ cycles, which is $\sim 33\%$ of the total fatigue life. The two hipped
Fig. 5.39. Fatigue crack initiation and propagation lives for specimens with replicas. The maximum stresses are: (a) 200 MPa; and (b) 225 MPa. (The maximum length and length (in parenthesis) of crack initiation sites are in μm)
specimens with Sr-modification have the highest total lives; one of them with its fatal crack initiated at an oxide of length 36 μm has a fatigue initiation life of $1.5 \times 10^5$ cycles, which is ~ 48% of the total fatigue life; the other one with a crack initiated at a eutectic constituent of length 89 μm has a fatigue initiation life of $4.0 \times 10^4$ cycles, which is ~ 7% of the total fatigue life. Among the three unmodified specimens, one of the non-hipped specimens has the highest propagation life, followed closely by the hipped specimen. One of the two hipped specimens with Sr-modification has the highest propagation life, by far.

With a maximum stress of 225 MPa, (Fig 5.39(b)), the specimen from Group 1, with the crack initiated at a eutectic silicon particle of length of 28 μm, had an initiation life of about zero. This specimen was neither modified nor hipped. A hipped specimen from Group 1, with its crack initiated at a eutectic silicon of length 20 μm, has a fatigue initiation life of $1.55 \times 10^4$ cycles, which is ~ 14% of the total fatigue life. The hipped specimen with Sr-modification, with its crack initiated at an oxide of length 51 μm had the highest initiation life of $7.99 \times 10^4$ cycles, which is ~ 70 % of the total fatigue life. By hipping, the crack initiation life was increased, specimen with Sr-modification showed the highest initiation life. The non-hipped specimen without Sr-modification showed the highest propagation life, and the hipped specimen with Sr-modification had the least propagation life. These three specimens showed similar total fatigue life. At a maximum stress of 225 MPa, the greater plastic strain introduced in the Sr-modified A356.2 as shown in Fig. 5.32 might cause greater accumulated damage to the Si-particles in the long period of crack initiation and thus, lower propagation life result. Indeed, as shown in
the next subsection, the small propagation rate in Sr-modified A356.2 at a maximum stress of 225 MPa was not improved by the Sr-modification.

5.8. SMALL-CRACK PROPAGATION RATE IN A356.2

It is known that small-cracks in the fatigue of alloys grow much faster than predicted from large-crack data, which are correlated on the basis of linear-elastic fracture mechanics (LEFM) (Person, 1975). At present, only a few studies (Plumtree and Schafer, 1986; Seniw et al., 1997; Shiozawa et al., 1997) reported on the small-crack behavior in A356 aluminum casting alloy. In this section, small-crack propagation rate in non-hipped A356.2 specimens and the effects of Sr-modification on small-crack propagation rate in hipped specimens from A356.2 castings made in a permanent mold are presented.

5.8.1. Small-Crack Propagation Rate in Non-Hipped A356.2 with Sr-Modification

The small-crack propagation rates \( \frac{da}{dN} \) tested at maximum stresses of 125, 150 and 175 MPa for not hipped specimens with Sr-modification are plotted against the crack lengths in Figs. 5.40(a)-(c). Similar trends are observed at maximum stresses of 125, 150, and 175 MPa. When the cracks initiate at pores, and the crack lengths increase from ~5 \( \mu \text{m} \) to 300 \( \mu \text{m} \), the overall propagation rates of oscillate in a range of \( 3.0 \times 10^{-10} \) to \( 1.0 \times 10^{-8} \) m/cycle. Then with crack lengths greater than 500 \( \mu \text{m} \), the propagation rate increases rapidly to ~\( 1 \times 10^{-6} \) m/cycle, and the cracks achieve a length of ~3.0 mm at fracture. At a maximum stress of 175 MPa, when the small-crack initiated at a large oxide of length
Fig. 5.40. Small-crack propagation rate versus crack length in specimens with Sr-modification. The maximum stresses are: (a) 125 MPa; (b) 150 MPa; (c) 175 MPa. The numbers in the legends are the lengths of the defects from which the cracks originated. P represents pore and O represents oxide as the crack-initiation sites.
of 668 μm (a maximum length of 1333 μm), the small-crack has a higher of $\sim 2 \times 10^{-9}$ to $5 \times 10^{-8}$ m/cycle.

5.8.2. Effect of Sr-modification on Small-Crack Propagation Rate in A356.2

Research on the influence of Sr-modification on the long-crack propagation has been conducted by using compact tension (CT) specimens on related alloys. It was reported that in a eutectic alloy (Al-12 wt. pct. Si) the long-crack propagation rate was slower for modified alloy than that in unmodified alloy (Lee et al. 1995). In another study on an Al-
Si-Cu die-cast alloy, it was shown that strontium modified alloy exhibited a higher fatigue crack growth threshold compared to an unmodified alloy when tested at a load ratio of 0.5; no difference was observed when tested at a load ratio of 0.1 (Schaefer and Fournelle 1996). Regarding crack propagation in an alloy similar to A356, it was found that with a high-stress intensity factor range the unmodified silicon particles accelerated the crack propagation, while with a low stress-intensity factor range the unmodified alloy showed a lower crack propagation rate (Kumai et al. 1999).

The small-crack propagation rates for specimens without strontium are plotted against the crack lengths in Figs. 5.41(a) and (b). At a maximum stress of 200 MPa (Fig. 5.41(a)), when the cracks are less than about 200 μm, the propagation rates oscillate and tend to increase in a range of $10^{-9}$ to $6 \times 10^{-9}$ m/cycle, with the exception of the propagation rate of $1.3 \times 10^{-10}$ m/cycle at a crack length of 50 μm. Then when the cracks are more than 500 μm, the propagation rate increases rapidly to $\sim 1 \times 10^{-5}$ m/cycle when the cracks achieve final fracture. The two cracks with the slowest propagation rates when the cracks are smaller than 500 μm initiated at a pore of length 80 μm and at a Si-particle of 13 μm. At a maximum stress of 225 MPa (Fig. 5.41(b)), when the cracks are less than 200 μm, the propagation rates oscillate in a range of $4.0 \times 10^{-10}$ to $1.0 \times 10^{-8}$ m/cycle. Then when the cracks achieves 500 μm, the propagation rate increases rapidly to $\sim 3 \times 10^{-6}$ m/cycle, when the cracks achieve a length of $\sim 1.0$ mm and fracture. The specimen with a fatal crack that initiated at a Si-particle of length 28 μm shows the slowest small-crack propagation rate among the six cracks analyzed in Fig. 5.41(b). This particular crack
Fig. 5.41. Small-crack propagation rates for specimens without Sr. The maximum stresses are: (a) 200 MPa; and (b) 225 MPa. The length of the crack-initiation sites are indicated on the figure in μm. L and R stand for two cracks emanated from the initiator in opposite directions. P represents pore and E represents eutectic constituents as the crack-initiation sites.
Fig. 5.42. Small-crack propagation rates for specimens with Sr-modification. The maximum stresses are: (a) 200 MPa; and (b) 225 MPa. The length of the crack-initiation sites are indicated on the figure in μm. L and R stand for the two cracks emanated from the initiator in opposite directions. O represents oxides and E represents eutectic constituents as the crack-initiation sites.
At a maximum stress of 200 MPa, it is apparent by viewing Figs. 5.41(a) and 5.42(a) that cracks in specimens with Sr-modification propagates slower than the cracks in the unmodified specimens, when the small-cracks are smaller than ~200 µm. At a maximum stress of 225 MPa, by viewing Figs. 5.41(b) and 5.42(b), specimens with and without Sr-modification, however, have similar small-crack propagation rates when the small-cracks are shorter than 200 µm. This will be further discussed in the next chapter (Chapter 6). When the cracks are longer than 500 µm, the slope of $\frac{da}{dN}$ versus crack-length curve is greater for the unmodified specimens than that for Sr-modified specimens. Hence, in addition to improving the resistance to crack initiation, Sr-modification is also beneficial in improving the resistance to crack propagation.

The oscillation of the small-crack propagation rate for cracks less than ~500 µm may be related to the variation of the driving force for small cracks. In a casting alloy, there are heterogeneous microstructures; hence, it appears that in order to model small-crack propagation, the micromechanics at the microstructural level, as suggested by others (Gokhale and Yang 1999; Fan et al., 2001), should be effected. As discussed in the next section (Section 6.1.2), the stress-intensity factor range corresponding to small-cracks less than ~200 µm is below the threshold stress-intensity factor range obtained from long-crack propagation on compact tension (CT) specimens, and microstructural effects are more important than in long-crack propagation.
5.9. SUMMARY

The effect of hydrogen content, hipping and Sr-modification on the fatigue behavior, including the fatigue life, small-crack initiation and propagation, and cyclic stress-strain response has been studied in a permanent cast aluminum alloy (A356.2-T6) with a SDAS of ~20 to 30 μm. The results show the following.

Effects of Hydrogen Content and Hipping

In non-hipped A356.2 alloy, specimens without Sr-modification and a hydrogen content of 0.17 cc/100 g showed the best fatigue life as a result of containing the least amount of porosity. Specimens with Sr-modification (0.004-0.006 %) and hydrogen contents of 0.19-0.22 cc/100 g had higher fatigue lives than specimens with a hydrogen content of 0.31 cc/100 g. Pores as small as 75 μm initiated fatigue cracks in high-cycle fatigue, and porosity was found to be the main culprit in crack initiation in the non-hipped alloy. Infrequently, crack-initiation sites were at eutectic constituents with rather large silicon particles in non-hipped A356.2 without Sr-modification.

For A356.2 alloy with Sr-modification, hipping significantly increased the initiation life and small-crack propagation life as a result of the elimination of the porosity. However, hipping did not significantly improve the fatigue life of A356.2 without Sr-modification, as a result of the lower hydrogen and porosity content of castings without Sr-modification. Silicon particles within the eutectic constituents and oxides were the
predominant crack-initiation sites in hipped A356.2. Finer silicon particles and smaller sizes of eutectic constituents are expected to improve the resistance to crack initiation.

Studies on the cyclic stress-strain relationship of A356.2 showed that, in strain-controlled fatigue with a strain ratio of 0.1, strain-hardening and mean stress relaxation occurred, and in stress-controlled fatigue with stress ratio of 0.1, cyclic creep occurred. The strain-hardening rate in fatigue loading was decreased compared with that in tensile loading, indicating more damage accumulation occurs in fatigue loading than in tensile loading. When pores are present, pores accommodate the remote strain by localizing the plastic deformation and specimens fatigued to fracture without full-saturation of strain hardening during strain-controlled fatigue testing. Hipping increased the cyclic yield strength of A356.2 alloy.

Effect of Sr-modification

In non-hipped A356.2 alloy, the expected gains in ultimate tensile strength and yield strength by Sr-modification were overshadowed by the detrimental effect of an increase in porosity associated with Sr-modification. In most cases, however, the tensile elongation was increased somewhat by Sr-modification. The fatigue life of non-hipped A356.2 with Sr-modification was lower than that in non-hipped A356.2 without Sr-modification. Possible improvements on fatigue behavior from Sr-modification in non-hipped A356.2 are overshadowed by the presence of more pores that accompany the Sr-modification.
The hipped A356.2 with Sr-modification showed a higher fatigue life than hipped A356.2 alloy without Sr, contrary to the result from fatigue testing of non-hipped A356.2. The fatigue performance of A356.2 with Sr-modification is better than unmodified A356.2 when pores and oxides are not the crack initiators. After hipping, when the small-cracks are less than 200 μm, the propagation rate is about one order of magnitude slower in modified A356.2 than in unmodified A356.2. When cracks are longer than 500 μm, the propagation rates are also slower in modified alloy than in unmodified alloy. Thus, in addition to improving the toughness and increasing crack-initiation life of A356.2, Sr-modification increased both small-crack and long-crack propagation lives as a result of the decreased small-crack and long-crack propagation rates. Sr-modification is beneficial in improving the resistance to crack-initiation and propagation in hipped A356.2.

Studies on the cyclic stress-strain relationship showed that A356.2 with Sr-modification had higher hardening value of stress than A356.2 without Sr-modification, indicating that the refined Si-particles experienced less accumulation damage than did the unmodified Si-particles during fatigue. In load-controlled fatigue, A356.2 with Sr-modification showed overall larger scale of cyclic creep than A356.2 without Sr-modification.
CHAPTER 6. MODELS FOR SMALL-CRACK PROPAGATION IN A356.2 CASTING ALUMINUM ALLOY

This chapter presents the models for characterizations of the small-crack propagation rates. Two established fracture mechanics models (Newman-Raju model, and Trantina-Barishpolsky model) are applied to correlate small-crack propagation rates. One micromechanics model is applied to relate the small-crack propagation rate to the maximum plastic zone length through the application of the Paris-power law. In the micromechanics model, the theory of continuously distributed dislocations is applied to represent the plastic zone at the tip of the crack in A356.2 cast alloy.

6.1. FRACTURE MECHANICS MODELS

The stress-intensity factor range is often used as the driving force for crack propagation; therefore, the small-crack propagation rate as a function of the stress intensity factor range is summarized to reveal small-crack behavior. When a crack-closure effect is considered, plots showing the relationship between the crack propagation rate and the effective stress-intensity factor range may be obtained using several established models (Newman and Raju, 1983; Trantina and Barishpolsky 1984). In the first subsection, a stress intensity solution (Newman and Raju, 1983; Newman, 1984; Newman et al., 1999) considering effects of crack closure and a cyclic-plastic-zone is used to correlate small-crack propagation rates. In the second subsection, an elastic–plastic stress intensity
solution (Trantina and Barishpolsky, 1984) considering microstructural features is applied.

6.1.1. Newman and Raju Model

When one considers the stress field around an advancing crack tip, a widely accepted elastic stress-intensity factor range (Newman and Raju, 1983) is used. The Newman-Raju model (NR model) has previously been discussed in Chapter 2 and is rewritten here:

\[
(\Delta K_p)_{eff} = UF(d/w)\Delta S \sqrt{\frac{\pi}{Q(d)}}
\]

\[
\omega = \left(1 - \frac{S_{opp}}{S_{max}}\right)^2 \frac{\rho}{4}
\]

\[
d = a + \omega\frac{\rho}{4}
\]

\[
\rho = a[\sec(\frac{\pi S_{max}}{2a\sigma_0}) - 1]
\]

\[
U = \frac{(S_{max} - S_{opp})}{(S_{max} - S_{min})}
\]

where \((\Delta K_p)_{eff}\) is the cyclic-plastic-zone-corrected effective stress-intensity-factor range. To approximate the influence of the crack-tip yielding on the crack-driving force, a portion of the Dugdale cyclic-plastic-zone length has been added to the crack length. \(S_{max}, S_{min}\) and \(S_{opp}\) stand for maximum, minimum and crack-open stresses, respectively. \(F(d/w)\) is the cyclic-plastic-zone corrected boundary-correction factor; \(\omega\) is a portion of the Dugdale cyclic-plastic-zone length; \(\rho\) is the plastic-zone size for a crack in a large
plate; \( \sigma_0 \) is the flow stress, the average of yield strength and ultimate tensile strength; \( \alpha \) is the constraint factor, equal to 1 and 3, respectively, for plane-stress and plane-strain conditions; \( a \) is the crack length including the half-length of the initiation-site; \( w \) is the specimen width; \( \Delta S \) is the far-field stress range; \( Q(d) \) is the elliptical crack-shape factor. Since the geometry factor \( Q(d) \) generally changes with the advance of crack, numerical integration techniques are necessary. The factor \( U \) is related to plasticity-induced crack-closure effects, and the effect of crack closure through the factor \( U \) has been used for long cracks (Newman, 1984).

The small-crack propagation rates \( (da/dN) \) tested at maximum stresses of 125, 150 and 175 MPa for non-hipped specimens with Sr-modification are plotted against the stress intensity factor range \( ((\Delta K_p)_{eff}) \) in Figs. 6.1(a)-(d). At all maximum stresses, when cracks initiated at pores, the small-crack propagation rate oscillates in a range of \( 3.0 \times 10^{-10} \) to \( 1.0 \times 10^{-8} \) m/cycle when \( (\Delta K_p)_{eff} \) is smaller than \( \sim 3.0 \) MPa m\(^{1/2}\). The initial region in which \( da/dN \) oscillates is defined as small-crack behavior. When \( (\Delta K_p)_{eff} \) is above \( \sim 3.0 \) MPa m\(^{1/2}\), although there are still oscillations in the propagation rate, the propagation rate tends to increase with the \( (\Delta K_p)_{eff} \) until the specimen is fractured. This region follows the Paris power law, and is defined as long-crack behavior. At fracture, an average \( (\Delta K_p)_{eff} \) of \( \sim 11.0 \) MPa m\(^{1/2}\) is reached. At a maximum stress of 175 MPa, when the crack initiated at an oxide film, the value of \( da/dN \) oscillated and decreased from \( 5.0 \times 10^{-8} \) to \( 2.0 \times 10^{-9} \) m/cycle as \( (\Delta K_p)_{eff} \) increased from 4.5 to 5.5 MPa m\(^{1/2}\). Then \( da/dN \) increased as \( (\Delta K_p)_{eff} \)
Fig. 6.1. Small-crack propagation rates with the effective stress-intensity factor range according to the Newman-Raju model for non-hipped specimens with Sr-modification, the maximum stresses are: (a) 125 MPa; (b) 150 MPa; (c) 175 MPa; and (d) all together. The length of the initiation site is shown in μm. P, O indicate that the initiation-site was at a pore, oxide.
Fig. 6.1. (Continued)

\[ \frac{da}{dN}, \text{m/cycle} \]

\[ \frac{(\Delta K_{p})_{eff}}{F_{m}} \text{MPa.m}^{1/2} \]

\[ \Delta 175 \text{MPa} \]

\[ \Delta 150 \text{MPa} \]

\[ \Delta 125 \text{MPa} \]

\[ 144 \text{P} \]

\[ 247 \text{P} \]

\[ 668 \text{P} \]

\[ 108 \text{P} \]

\[ 423 \text{P} \]
increased until the specimen completely fractured at \((\Delta K_p)_{\text{eff}}\) of \(~10.0\) MPa \(m^{1/2}\). As shown in Fig. 6.1(d), the crack propagation rate at a maximum stress of 125 MPa increases faster than that at maximum stress of 150 MPa, and 175 MPa at the same \((\Delta K_p)_{\text{eff}}\). This is because there are more large pores in specimens tested at a maximum stress of 125 MPa.

The small-crack propagation rates \((da/dN)\) tested at maximum stresses of 200 and 225 MPa for hipped specimens with/without Sr-modification are plotted against the effective stress-intensity factor range \((\Delta K_p)_{\text{eff}}\) in Figs. 6.2(a) and (b). At a maximum stress of 200 MPa (Fig. 6.2(a)), when \((\Delta K_p)_{\text{eff}}\) is smaller than \(~3\) MPa \(m^{1/2}\), the small-crack propagation rates oscillate in a range of \(1.3 \times 10^{-10}\) to \(1.25 \times 10^{-9}\) m/cycle, and \(7.0 \times 10^{-10}\) to \(4.8 \times 10^{-9}\) m/cycle, respectively, in A356.2 with and without Sr-modification. It is important to note that the small-crack propagation rate in A356.2 is at least \(~3\) times slower in the Sr-modified alloys than in unmodified alloy. When \((\Delta K_p)_{\text{eff}}\) is above \(~3.0\) MPa \(m^{1/2}\), the crack-propagation rate increases with the increase of \((\Delta K_p)_{\text{eff}}\) until the specimens fracture at \((\Delta K_p)_{\text{eff}}\) of \(~17.5\) MPa \(m^{1/2}\), and \(~12.3\) MPa \(m^{1/2}\), respectively, for specimens with Sr-modification and without Sr-modification. Again the Sr-modified specimens, after hipping have somewhat better \((i.e.\ lower)\) propagation-rates when the cracks can be considered as large-cracks.
Fig. 6.2. Small-crack propagation rate versus effective stress-intensity factor range according to Newman-Raju model for hipped specimens with/without Sr-modification. The maximum stresses are: (a) 200 MPa; (b) 225 MPa.
At maximum stress of 225 MPa (Fig. 6.2(b)), when $(\Delta K_p)_{\text{eff}}$ is smaller than $\sim 3 \text{ MPa m}^{1/2}$, the small-crack propagation rates oscillate in a range of $1.0 \times 10^{-9}$ to $1.0 \times 10^{-8} \text{ m/cycle}$. and $5.0 \times 10^{-10}$ to $1.0 \times 10^{-8} \text{ m/cycle}$, respectively, with and without Sr-modification. At this very high stress-level, however, there is no significant difference found in small-crack rates for alloys with and without Sr-modification. When $(\Delta K_p)_{\text{eff}}$ is above $\sim 3.0 \text{ MPa m}^{1/2}$, the crack propagation rates tends to increase with an increase of $(\Delta K_p)_{\text{eff}}$ until the specimens fracture at $(\Delta K_p)_{\text{eff}}$ of $\sim 13.3 \text{ MPa m}^{1/2}$, and $\sim 7.3 \text{ MPa m}^{1/2}$, respectively, for specimens with Sr-modification and without Sr-modification. The benefits of Sr-modification is apparent once again for the large-crack behavior.

The small-crack propagation rates at the maximum stresses of 200 and 225 MPa are plotted together in Figs. 6.3(a) and (b). Although the plastic-zone and the crack closure effects were already included in Newman-Raju model, the crack-propagation rate at 225 MPa is still greater in specimens with Sr-modification than that at 200 MPa at the same $(\Delta K_p)_{\text{eff}}$ level, as seen in Fig. 6.3(a); but the crack propagation rates at 220 MPa and 225 MPa are very similar in specimens without Sr-modification, as shown in Fig. 6.3(b).

It should be noted here from Figs. 5.31 and 5.32 that, for hipped A356.2 with Sr-modification, the cyclic creep and the saturated plastic strain at 225 MPa are higher than those at 200 MPa, respectively, by $\sim 0.00020$ and $\sim 0.00150$; for hipped A356.2 without Sr-modification, the cyclic creep and the saturated plastic strain at 225 MPa are higher than those at 200 MPa by $\sim 0.00005$ and $\sim 0.00111$. Researches (Dighe et al.,
Fig. 6.3. Variation of the small-crack propagation rates with the effective stress-intensity factor range according to the Newman-Raju model for hipped specimens: (a) with Sr; (b) no Sr.
2002) on the damage of Si-particles in A356 alloy showed a strain of 0.005 in uniaxial tension causes ~1% of Si-particles to become either broken or debonded, and the damage of the Si-particles increases with the tensile strain. At a maximum stress of 225 MPa in hipped specimen with Sr-modification, a remote strain of ~ 0.0045 is introduced, and it is believed that much higher cyclic creep and plastic strain at 225 MPa than at 200 MPa for A356.2 with Sr-modification caused greater amount of cyclic damages to Si-particles and resulted higher propagation rate at 225 MPa than at 200 MPa. The more cyclic damages at 225 MPa for A356.2 with Sr-modification might be also responsible for the fact that Sr-modification did not improve the small-crack propagation resistance in A356.2 alloy at a maximum stress of 225 MPa as shown in Figure 6.2 (b).

Since there are oscillations in small-crack propagation rates when $(\Delta K_p)_{\text{eff}}$ is smaller than ~ 3.0 MPa m$^{1/2}$, and the Newman-Raju model can not describe small-crack behavior. Therefore, the small-crack propagation rates are averaged to a constant value of $(da/dN)_0$ independent of the intensity factor range until an arbitrarily chosen value of $(\Delta K_p)_{\text{eff}} < 3$ MPa m$^{1/2}$. As shown previously, when $(\Delta K_p)_{\text{eff}}$ is greater than ~ 3 MPa m$^{1/2}$, crack-propagation rate tends to increase with $(\Delta K_p)_{\text{eff}}$, and it is believed that the crack-propagation rate merges with that of long-crack propagation rates in compact tension (CT) specimens, when the long-crack propagation rate follows the Paris-power law relationship as shown by Eq. (2.11):

$$\frac{da}{dN} = C((\Delta K_p)_{\text{eff}})^m$$  \hspace{1cm} (2.11)
Crack-propagation parameters in non-hipped A356.2 with Sr modification, hipped A356.2 with and without Sr-modification are shown in Table 6.1. When the Newman-Raju model was used, either the maximum length or the length of the initiation-site can be treated as the initial crack length. By fitting results from using either the maximum length or length of the initiation-site are presented in Table 6.1. Results obtained from using the length and the maximum length are very similar in hipped A356.2 alloys since the initiation sizes are small. A large difference is present only for non-hipped A356 with Sr on the C values. The long-crack propagation constants in this research are in accordance with those in a recent report (Basner et al., 2001), where the long-crack propagation parameters in A356 alloy were summarized.

Table 6.1. Crack-propagation parameters from Newman-Raju model

<table>
<thead>
<tr>
<th>Materials</th>
<th>Input</th>
<th>Small Crack</th>
<th>Long Crack</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>$(\frac{da}{dN})_0$, m/cycle</td>
<td>$C$</td>
</tr>
<tr>
<td>With Sr</td>
<td>Length</td>
<td>1.882$x10^{-9}$</td>
<td>8.227$x10^{-11}$</td>
</tr>
<tr>
<td>No HIP</td>
<td>Maximum Length</td>
<td>1.555$x10^{-9}$</td>
<td>2.369$x10^{-11}$</td>
</tr>
<tr>
<td>With Sr</td>
<td>Length</td>
<td>8.641$x10^{-10}$</td>
<td>2.423$x10^{-11}$</td>
</tr>
<tr>
<td>HIP</td>
<td>Maximum Length</td>
<td>8.453$x10^{-10}$</td>
<td>2.405$x10^{-11}$</td>
</tr>
<tr>
<td>No Sr</td>
<td>Length</td>
<td>2.142$x10^{-9}$</td>
<td>7.029$x10^{-11}$</td>
</tr>
<tr>
<td>HIP</td>
<td>Maximum Length</td>
<td>2.142$x10^{-9}$</td>
<td>5.254$x10^{-11}$</td>
</tr>
</tbody>
</table>

Note: Paris constants (C and m) were obtained with $da/dN$ measured in units m/cycle and $(\Delta K_p)_{eff}$ in units MPa m$^{1/2}$.

The fitted propagation rates obtained by using the lengths of the crack-initiation sites are presented together in Fig. 6.4 in order to compare the effects of hipping and Sr-
modification. As can be seen, the small-crack propagation rate is the slowest in hipped A356.2 with Sr-modification. Non-hipped A356.2 with Sr-modification and hipped

A356.2 without Sr-modification shows similar small-crack propagation rates. Long-crack propagation rates are similar in hipped and not hipped A356.2 with Sr-modification and are slower than that in hipped A356.2 without Sr-modification. Hipped A356.2 without Sr-modification shows greater crack-propagation rates than hipped A356.2 with Sr-modification. Therefore, hipping decreased the small-crack propagation rate and improves the small-crack propagation life in A356.2 alloy with Sr-modification. Sr-modification decreases both small-crack and long-crack propagation rates and improves the fatigue crack propagation life after hipping. Previously, it was shown that crack

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**Fig. 6.4. Effects of hipping and Sr-modification on crack-propagation rate from Newman-Raju model.**

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propagation rates are similar in non-hipped and hipped A356.2 without Sr; thus before hipping, similar small-crack propagation rates are observed in A356.2 with and without Sr-modification. Sr-modification decreases the long-crack propagation rate and improves the long-crack propagation life when hipping was not conducted.

6.1.2. Trantina and Barishpolsky Model

In Newman-Raju model used in previous section, the size of the crack-initiation site is treated as the initial crack size. In doing so, the stress-concentration effect on the initial small-crack propagation is neglected. When the length of the crack is smaller than the size of the crack-initiation site, the crack is strongly influenced by the stress concentration. The Trantina-Barishpolsky model (Trantina and Barishpolsky, 1984) considers the stress concentration at specific microstructural features and has been used for estimating $\Delta K$ for small cracks initiated at pores or inclusions. Their initial model did not include the effects of crack closure and finite width correction. However, these effects can affect the small-crack propagation significantly. So the plasticity-induced crack closure effect ($U$) as in Newman-Raju model is included. The finite width correction ($f(a/w)$) for surface and corner cracks is also included. Then the effective stress intensity factor in the modified Trantina-Barishpolsky model (TB model) is expressed as:

$$\Delta K_{\text{eff}} = U(a)F(a)\Delta S\sqrt{\pi a f(a/w)}$$  \hspace{1cm} (6.1)

where $F(a)$ is:

$$F(a) = \frac{2}{\pi} + B(1.12k_f - \frac{2}{\pi} - 1)(\frac{a_0}{a_0 + a})^{10} + (\frac{a_0}{a_0 + a})^{18}$$  \hspace{1cm} (2.20)
Here \( a \) is the length of a crack extended from the void or particle; \( B \) is a constant that has the value of 1 for a void, 2 for a bonded cracked Si-particle, and 0.3 for a debonded Si-particle; \( k_i \) is the local elastic-stress-concentration factor for the ellipsoidal void or particle without the crack; and \( a_0 \) is the half width of the void or particle.

The small-crack propagation rates for non-hipped specimens with Sr-modification are plotted against \( \Delta K_{eff} \) in Figs. 6.5(a)-(d). The TB model shows similar trends of variations of crack propagation as the Newman-Raju model (i.e., NR model) in previous subsection. Overall at all the maximum stresses, the small-crack propagation rate oscillates in a range of \( 4.0 \times 10^{-10} \) to \( 1.3 \times 10^{-8} \) m/cycle when \( \Delta K_{eff} \) is smaller than \( \sim 3 \) MPa m\(^{1/2} \), very similar to that observed when using NR model. Then with \( \Delta K_{eff} \) above \( \sim 3 \) MPa m\(^{1/2} \), the propagation rate increases with \( \Delta K_{eff} \) until the cracks achieve an average \( \Delta K_{eff} \) of \( \sim 11 \) MPa m\(^{1/2} \) at fracture in the region previously defined as long-crack behavior. When compared to the NR model, the TB model shows a smaller initial \( \Delta K_{eff} \) in the region of small-crack behavior with \( \Delta K_{eff} < \sim 3 \) MPa m\(^{1/2} \). In the region of long-crack propagation, the TB model gives similar relationship of \( da/dN \) with \( \Delta K_{eff} \) as does the NR model.

The small-crack propagation rates \( (da/dN) \) tested at maximum stresses of 200 and 225 MPa for hipped specimens with/without Sr-modification are plotted against the stress intensity factor range \( (\Delta K_{eff}) \) in Figs. 6.6(a) and (b). Similar results to that from the NR model are observed. At a maximum stress of 200 MPa, when \( \Delta K_{eff} \) is smaller than \( \sim 3.0 \) MPa m\(^{1/2} \), the small-crack propagation rates oscillate in a range of \( 1.3 \times 10^{-10} \) to \( 2.0 \times 10^{-9} \)
Fig. 6.5. Small-crack propagation rates versus effective stress-intensity factor range according to TB model for non-hipped specimens with Sr-modification. Maximum stress is: (a) 125 MPa; (b) 150 MPa; (c) 175 MPa; (d) all together. The length of the initiation site is shown in µm. P and O indicates that the crack-initiation was at a pore or oxide, respectively.
Fig. 6.5. (Continued)
Fig. 6.6. Small-crack propagation rate versus effective stress-intensity factor range according to TB model for hipped specimens with/without Sr. Maximum stress is: (a) 200 MPa; (b) 225 MPa.
m/cycle, and $7.0 \times 10^{-10}$ to $5.0 \times 10^{-9}$ m/cycle, respectively, in A356.2 with and without Sr-modification, with only one exception of $1.5 \times 10^{-10}$ m/cycle at $\Delta K_{\text{eff}}$ of ~ 1.7 MPa m$^{\frac{1}{2}}$.

When $\Delta K_{\text{eff}}$ is above ~ 3.0 MPa m$^{\frac{1}{2}}$, crack propagation rate tends to increase with the increase of $\Delta K_{\text{eff}}$ until the specimen is fractured at $\Delta K_{\text{eff}}$ of ~ 23.0 MPa m$^{\frac{1}{2}}$, and ~ 11.0 MPa m$^{\frac{1}{2}}$, respectively, for specimens with Sr-modification and without Sr-modification. and the $\Delta K_{\text{eff}}$ at fracture are greater in A356.2 with Sr-modification. At maximum stress of 225 MPa, when $\Delta K_{\text{eff}}$ is smaller than ~ 3.0 MPa m$^{\frac{1}{2}}$, the small-crack propagation rates oscillate in a range of $1.0 \times 10^{-9}$ to $1.0 \times 10^{-8}$ m/cycle, and $5.0 \times 10^{-10}$ to $1.0 \times 10^{-8}$ m/cycle, respectively, in A356.2 with and without Sr-modification, and no apparent difference in small-crack propagation rate for alloys with and without Sr-modification is observed. When $\Delta K_{\text{eff}}$ is above ~ 3.0 MPa m$^{\frac{1}{2}}$, crack propagation rate tends to increase with the increase of $\Delta K_{\text{eff}}$ until the specimen is fractured at $\Delta K_{\text{eff}}$ of ~ 12.0 MPa m$^{\frac{1}{2}}$, and ~ 7.0 MPa m$^{\frac{1}{2}}$, respectively, for specimens with Sr-modification and without Sr-modification, and the $\Delta K_{\text{eff}}$ at fracture is greater in A356.2 with Sr-modification.

The small-crack propagation rates for A356.2 with and without Sr-modification are plotted, respectively, in Figs. 6.7(a) and (b). Similar to the NR model, crack closure effect is included in the TB model; nevertheless, the crack-propagation rate at 225 MPa is also greater in specimens with Sr-modification than that at 200 MPa at the same $\Delta K_{\text{eff}}$ level. But as shown in Fig. 6.7(b), the crack propagation rates at 220 MPa and 225 MPa are very similar in specimens without Sr-modification.
Fig. 6.7. Variation of small-crack propagation rates with the effective stress intensity factor range according to the TB model for hipped specimens: (a) with Sr; (b) without Sr.
It should be noted that the small cracks emanating from pores propagate well below the effective threshold stress intensity factor range \((\Delta K_{eff})_m\) obtained from the long-crack propagation measurements using compact tension (CT) specimens, which is \(\sim 1.4 \text{ MPa m}^{1/2}\) for A356 alloy (Couper and Griffiths, 1990; and Skallerud et al., 1993).

Table 6.2. Parameters for crack-propagation from Trantina-Barisholpsky model.

<table>
<thead>
<tr>
<th>Materials</th>
<th>Input</th>
<th>Small Crack</th>
<th>Long Crack</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>(\frac{da}{dN})_0, m/cycle</td>
<td>(C)</td>
<td>(m)</td>
</tr>
<tr>
<td>With Sr HIP</td>
<td>Length</td>
<td>2.044x10^-9</td>
<td>2.060x10^-10</td>
</tr>
<tr>
<td></td>
<td>Maximum Length</td>
<td>2.092x10^-9</td>
<td>1.823x10^-11</td>
</tr>
<tr>
<td>With Sr HIP</td>
<td>Length</td>
<td>9.537x10^-10</td>
<td>7.124x10^-11</td>
</tr>
<tr>
<td></td>
<td>Maximum Length</td>
<td>9.043x10^-10</td>
<td>6.274x10^-11</td>
</tr>
<tr>
<td>No Sr HIP</td>
<td>Length</td>
<td>2.238x10^-9</td>
<td>7.933x10^-12</td>
</tr>
<tr>
<td></td>
<td>Maximum Length</td>
<td>2.166x10^-9</td>
<td>6.646x10^-12</td>
</tr>
</tbody>
</table>

Note: Paris constants \((C\) and \(m\) were determined with \(\frac{da}{dN}\) measured in units \(m/cycle\) and \((\Delta K)_{eff}\) in units MPa m\(^{1/2}\).

As in the NR model, the small-crack propagation rates in TB model were also averaged to a constant value \((\frac{da}{dN})_0\) until an arbitrarily chosen value of \(\Delta K_{eff} < 3.0 \text{ MPa m}^{1/2}\).

When \(\Delta K_{eff}\) is greater than \(\sim 3.0 \text{ MPa m}^{1/2}\), the long-crack propagation rates was fitted according to the Paris-power law (Eq. 2.11). The fitted crack-propagation parameters in non-hipped A356.2 with-Sr modification, hipped A356.2 with and without Sr-modification are shown in Table 6.2. Results are similar to that from NR model. Fitted propagation rates obtained by the length of the initiation site are presented together in
Fig. 6.8 in order to compare the effects of hipping and Sr-modification. Regarding the effects of hipping and Sr-modification on crack-propagation and fatigue life, the TB model and the NR model show very similar conclusions.

![Graph showing crack-propagation rate vs. stress intensity factor range](image)

Fig. 6.8. Effects of hipping and Sr-modification on crack-propagation rate from the TB model.

6.2. MICRO-MECHANICS MODEL

In the previous section, the Newman-Raju model and Trantina-Barishposky model that utilize the effective stress intensity factor range, considering the plastic zone at the tip of the crack, was applied to relate small-crack propagation rate. Such models, however, does not account for the oscillation of small-crack propagation rates at stress intensity factor range below the threshold since they are continuous mechanics based models. Small-
crack behavior in A356.2 cast alloy is related to the intergranular and interdendritic eutectic constituents, where Si-particles and aluminum with different orientations are present. To model the oscillations of small-crack propagation rate, micrstructural effects must be accounted for. Here, a micro-mechanics model is applied.

6.2.1. The Micro-Mechanics Model

The theory of continuously distributed dislocations was applied to represent the crack and the plastic zone at the tip of the crack (Bilby et al., 1963; Weertman 1966; Taira et al., 1978; Navarro and de los Rios, 1987; Wang, 1996), as shown in Fig. 6.9. When a crack initiates intergranularly at an inclusion /second phase particle, it is assumed that the slip bands extend across the grains and the whole grain undergoes slip. In A356.2 cast alloy,
intergranular boundaries and some interdendritic cell boundaries are barriers to the movement of dislocations in the plastic zone at the crack tip. Thus, as a crack propagates but with the plastic zone still being blocked by the intergranular boundaries, the ratio $a/c$ increases towards a critical value. The critical value corresponds to a stress concentration ahead of the plastic zone that is high enough to activate an appropriate dislocation source in the next grain of a different orientation.

The stress concentration ahead of the plastic zone was presented as (Navarro and de los Rios 1988):

$$S(x_0) = \frac{\sigma}{\sqrt{2(x_0 - 1)}} \left[ 1 - \frac{2\sigma}{\pi\sigma} \cos^{-1} n \right] + \sigma \left( 1 + \frac{2}{\pi} \arctan \frac{-n\sqrt{2(x_0 - 1)}}{\sqrt{1 - n^2}} \right)$$  (6.2)

where $n = \frac{a}{c}$, and $c = iD/2$, ($i = 1, 3, 5 \ldots$). The dimensions $a$, $c$, and $D$ are shown in Fig. 6.9; $i$ is the number of half grains that the crack has spanned; $\sigma$ is the applied stress amplitude; $\sigma_f$ is the ultimate tensile strength; and $x_0 = (\rho_0 + \epsilon)/c$. Here $\rho_0$ is a fixed distance ahead of the plastic zone. The critical value of $n$ when slip transmission occurs was derived for the case of asymmetrical loading (Wang 1996):

$$n_c' = \cos \left( \frac{\pi}{2} \frac{\sigma_{\text{max}} - \sigma_{FL}/\sqrt{i}}{\sigma_f} \right)$$  (6.3)

where $\sigma_{\text{max}}$ is the maximum stress; and $\sigma_{FL}$ is the maximum stress at fatigue limit. At a crack length of $a$, the value of $n$ is less than the critical value ($n_c'$):
where $\Delta \varepsilon_p$ is the plastic tensile strain range, $K'$ and $n'$ are the cyclic strength coefficient and strain hardening exponent.

In the small-crack model (Eq. 6.7) proposed by Wang (Wang 1996), and the enhanced Wang-model proposed by Hamm and Johnson (Hamm and Johnson 1999), Equation 6.9 is incorrect. The correct equation for the plastic strain amplitude is:

$$\Delta \varepsilon_p = 2\left(\frac{\Delta \sigma}{2K'}\right)^{(1/n')}$$  \hspace{1cm} (6.11)

When equation 6.11 for the plastic strain amplitude was applied in the model, however, the predicted propagation rates were three orders of magnitude smaller than the measured rates.

The long-crack propagation rate described by the power law shows that crack propagation rate is proportional to the power of the size of plastic zone. It follows that the small-crack propagation rate is also related to the plastic zone size. Hence, the crack propagation rate is then related to the crack-tip plastic zone size by the Paris-power law equation:

$$\frac{da}{dN} = C(\Delta K_{eff} (a))^n$$  \hspace{1cm} (2.11)

The cyclic plastic zone size ($r_{cp}$) can be expressed as (Dowling, 1993):

$$r_{cp} (a) = \frac{1}{\pi} \left(\frac{\Delta K(a)}{2\sigma_y}\right)^2$$  \hspace{1cm} (6.12)
where \( \sigma_y \) is the yield strength of the material. Then, the stress intensity factor range in terms of the yield strength and cyclic plastic zone size is:

\[
\Delta K(a) = 2\sigma_y \sqrt{\pi r_{cp}(a)} \quad (6.13)
\]

It should be noted here that the cyclic plastic zone size \( r_{cp} \) is half the value of the maximum plastic zone size at the crack tip \( \rho_{max} \):

\[
r_{cp}(a) = \frac{1}{2} \rho_{max}(a) \quad (6.14)
\]

Since the Trantina-Barishpolsky model in Section 6.1.2 showed very similar result to the Newman-Raju model on the effective stress-intensity factor range and the Trantina-Barishpolsky model used the length of a crack emanated from the initiation site as the crack length, then the Trantina-Barishpolsky model is chosen to calculate the effective stress-intensity factor range. The effective stress-intensity factor range is

\[
\Delta K_{eff} = U(a) F(a) \Delta S \sqrt{\pi a f(a/w)} \quad (6.1)
\]

After combining Eqs. (2.11), (6.1), (6.13) and (6.14), the final result is

\[
\frac{da}{dN} = \left(2^{n/2} C \sigma_y^n \pi^{m/2} \right) \left(U^n(a) F^n(a) f^n(a/w) \right) \rho_{max}^{n/2} \quad (6.15)
\]

where \( U(a) \) is related to the crack closure effect, and the plasticity-induced crack closure model (Eq. 2.12) is used; the \( F(a) \) is dimensionless geometry term (Eq. 2.20); \( f(a/w) \) is the correction coefficient for the finite width of either corner and surface crack. The crack propagation rate is now related to the maximum plastic zone length \( \rho_{max} \) by Eq. (6.15),
and the crack-propagation rate is proportional to the power of the maximum plastic zone length.

6.2.2. Application of the Micro-Mechanics Model in Bending-Fatigue

In this section, experimental results on the small-crack propagation in bending fatigue are used to validate the small-crack model. Since the cyclic strength coefficient ($K'$) and strain-hardening exponent ($n'$) are not available from directionally solidified ingots, $K'$ and $n'$ in Section (5.6.1) from cast plates with similar compositions are used. $K'$ and $n'$ are 331.36 MPa and 0.0558, respectively. Then, the maximum stress, stress amplitude are calculated using Eq. (2.6). The maximum stress at fatigue limit under a stress ratio $R$ of 0.1 is 175 MPa for hipped specimens (Section 5.4). The ultimate tensile strength ($UTS.$ MPa) changes with the $SDAS$ (μm) following the Eq. (4.1) proposed in Section 4.

$$UTS = 215.65 + [0.32106 \times SDAS]^{0.5}$$

(4.1)

$C$ and $m$ are obtained by fitting the data in Section 4.7.2 using the Trantina-Barishpolsky model. The microstructure barriers in A356.2 are the intergranular and interdendritic boundaries. Both boundaries are delineated with Si-particles. In the application of the model, both types of microstructural barriers were considered. Both the average secondary dendrite arm spacing ($SDAS$) and the average grain size ($D$) in A356.2 were measured. Parameters used in calculating the small-crack propagation rate are shown in Table 6.3.
Table 6.3. Parameters used in calculating the small-crack propagation

<table>
<thead>
<tr>
<th>Maximum Strain</th>
<th>Strain Ratio</th>
<th>SDAS (µm)</th>
<th>D (µm)</th>
<th>Maximum Stress (MPa)</th>
<th>Stress Range (MPa)</th>
<th>Stress Ratio</th>
<th>C</th>
<th>m</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.00214</td>
<td>0.10</td>
<td>20</td>
<td>407</td>
<td>149.8</td>
<td>134.8</td>
<td>0.100</td>
<td>2.666 × 10⁻¹¹</td>
<td>3.236</td>
</tr>
<tr>
<td></td>
<td></td>
<td>23</td>
<td>443</td>
<td>149.8</td>
<td>134.8</td>
<td>0.100</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>27</td>
<td>491</td>
<td>149.8</td>
<td>134.8</td>
<td>0.100</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.0080</td>
<td>0.15</td>
<td>17</td>
<td>370</td>
<td>245.7</td>
<td>429.3</td>
<td>-0.747</td>
<td>3.519 × 10⁻⁹</td>
<td>1.906</td>
</tr>
<tr>
<td></td>
<td></td>
<td>37</td>
<td>612</td>
<td>245.7</td>
<td>429.3</td>
<td>-0.747</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Notes: Stress values and stress ratios are calculated using cyclic stress-strain relationship from the strain settings in bending fatigue. C and m are obtained by fitting the data in Section 4.7.2 using Trantina-Barishpolsky model, with da/dN measured in units m/cycle and (ΔK)eff in units MPa m¹/².

Crack propagation rates predicted from the micro-mechanics model are presented in Figs. 6.10(a), (b) and (c) for high-cycle fatigue, and Figs. 6.11(a) and (b) for low-cycle fatigue.

When the microstructure barriers are taken as SDAS, the predicted small-crack rates when cracks are smaller than ~500 µm are consistently 2 to 3 orders of magnitude lower than those obtained from experiments, indicating that the SDAS is not a good length-scale in predicting small-crack propagation. When the microstructure barriers are taken as the grain boundaries, the predicted rates agree reasonably well with those from experiments. Replicas and microstructures showed that interdendritic eutectic constituents are less influential than intergranular eutectic constituents on the crack propagation paths.

For high-cycle fatigue, the predicted propagation rates are consistently in agreement with those from experiments; for low-cycle fatigue, the predicted propagation rates are consistently lower than those from experiments. Low-cycle fatigue crack propagation is
Fig. 6.10. High-cycle fatigue crack propagation rates predicted from the micro-mechanics model with grain size assumed to be the approximate length scale: (a) $SDAS = 20 \, \mu m$, $D = 407 \, \mu m$; (b) $SDAS = 23 \, \mu m$, $D = 443 \, \mu m$; and (c) $SDAS = 27 \, \mu m$, $D = 491 \, \mu m$. 
Fig. 6.11. Low-cycle fatigue crack propagation rates predicted from the micro-mechanics model with the grain size assumed to be the approximate length scale: (a) $SDAS = 17 \mu m$, $D = 370 \mu m$; and (b) $SDAS = 37 \mu m$, $D = 612 \mu m$. 
through the coalescence of micro-cracks developed from slip bands in front of the crack tip (as shown in Fig. 4.39). The model assumes only the primary crack is present, secondary cracks at the tip of the crack and the coalescence of micro-cracks are not considered. Even so the micro-mechanics model on small-crack model is reasonable and considered effective.

6.2.3. Application of the Micro-Mechanics Model in Axial-Fatigue

In this section, experimental results on the small-crack propagation in axial-fatigue are used to validate the small-crack model. In the calculation of the crack-propagation rate, the maximum stress at fatigue limit under a stress ratio $R$ of 0.1 is 175 MPa for hipped specimens (Section 5.4.3). The ultimate tensile strength and the yield strength are the
measured average values (Section 5.2). C and m are obtained fitting the data (Section 6.1.2). The calculations in Section 6.2.2 showed that grain size instead of the secondary dendrite arm spacing is a better microstructural parameter and the results of the calculations are reasonably close to the experimental data. In this calculation, the average grain size was measured, which is ~191 μm in grain refined A356.2 castings. Some important parameters used in the calculations are shown in Table 6.4.

Table 6.4. Parameters used in calculating the small-crack propagation

<table>
<thead>
<tr>
<th></th>
<th>Max. Stress (MPa)</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>C</th>
<th>m</th>
</tr>
</thead>
<tbody>
<tr>
<td>With Sr</td>
<td>125</td>
<td>263</td>
<td>189</td>
<td>2.060×10^{-10}</td>
<td>3.1756</td>
</tr>
<tr>
<td></td>
<td>150</td>
<td>274</td>
<td>199</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>175</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>With Sr, HIP</td>
<td>200</td>
<td>304</td>
<td>229</td>
<td>7.124×10^{-11}</td>
<td>3.6430</td>
</tr>
<tr>
<td></td>
<td>225</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No Sr, HIP</td>
<td>200</td>
<td>285</td>
<td>207</td>
<td>7.933×10^{-12}</td>
<td>5.4173</td>
</tr>
<tr>
<td></td>
<td>225</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Notes: R = 0.1; D =191 μm. C and m are obtained by fitting the data in Section 6.1.2 using Trantina-Barishpolsky model, with da/dN measured in units m/cycle and (ΔK)_{eff} in units MPa m^{\frac{1}{2}}.

The predicted fatigue crack propagation rates for non-hipped A356.2 with Sr-modification, at maximum stresses of 125, 150, and 175 MPa are shown in Fig. 6.12(a)-(c). It is shown that, at the same crack length, the predicted fatigue crack propagation rate tends to increase with the maximum stress. For small-cracks shorter than ~1000 μm (three times the grain size), oscillations of several orders of magnitude are observed on the predicted fatigue crack propagation rate when the plastic zone tip meet the
Fig. 6.12. Axial fatigue crack propagation rates predicted from the micro-mechanics model: (a) maximum stress = 125 MPa; (b) maximum stress = 150 MPa; and (c) maximum stress = 175 MPa. $R = 0.1$. 

The predicted fatigue crack propagation rates for hipped A356.2 with, and without Sr-modification, at maximum stresses of 200 and 225 MPa are shown in Figs. 6.13(a)-(d). For A356.2, either with or without Sr-modification, the predicted propagation rates agree reasonably well with those experimental data, except perhaps the predicted crack propagation rate at the first intergranular boundary. When the small-crack length is shorter than half the length of the grain size, the calculation shows that the small-crack propagation rate is smaller in A356.2 without Sr-modification than that in A356.2 with Sr-modification. This is because the effect of the morphology of the Si-particles on the local stress field when the crack is small is not included in the model.

Fig. 6.13. Axial fatigue crack propagation rates predicted from the micro-mechanics model. At maximum stress of 200 MPa: (a) with Sr; (b) no Sr; at maximum stress of 225 MPa: (c) with Sr; and (d) no Sr.
Fig. 6.13. (Continued)
The original Paris-power law was formulated for ideal flat cracks in isotropic materials, and it is understandable to neglect the oscillation of the propagation rate. In most engineering materials, however, poly-crystals with different orientations and secondary phases are present; then the crack-tip plastic-zone size is strongly influenced by the microstructural features that effect the oscillation of the driving force. Even in single crystal of Ni-based alloy, the oscillation of the crack propagation rate was reported (Zhang et al., 2002). The oscillation of the crack propagation rate is present in the long-cracks. However, when the propagation rate is calculated using the incremental polynomial method (ASTM E 647, 1995), the oscillation of the crack propagation rate is averaged out.

Fig. 6.13. (Continued)
The oscillation of the crack propagation rate is well simulated when the crack length is shorter than a length of several times of the grain size. With the crack length greater than a length of several times of the grain size, the oscillation of the crack propagation rate ebbs because, as the crack grows, the size of crack-tip plastic zone exceeds the characteristic length of the microstructural features.

The predicted effect of Sr-modification is contrary to the experimental observation. In the small-crack range, it appears that the morphology of the Si-particles will influences the local stress-strain field and interacts with the crack-tip plastic zone size. Thus, future endeavors need to include the effect of the morphology of the Si-particles and possibly the secondary dendrite arm spacing in order to capture the small-crack behavior.

6.3. SUMMARY

Well-accepted fracture mechanics models, considering effects of plasticity-induced crack-closure, crack-tip plastic-zone, the effect of the microstructure on stress concentration, are applied to correlate crack-propagation rate with effective stress intensity factor range. In order to model the microstructural effects on small-crack propagation, a micro-mechanics model was applied. The results show the following.

*Fractural Mechanics Models for Small-Crack Propagation in A356.2*

Fracture-mechanics models were applied to correlate crack-propagation rates. The Newman and Raju model, which considers the crack closure and cyclic plastic zone, and
the Trantina and Barishpolsky model, which considers the stress concentrations for the microstructural features, yielded similar results on the crack propagation rate against the effective stress intensity factor range. It was shown that small-cracks propagate well below the effective threshold stress intensity factor range obtained from long-crack propagation measurements.

**Micro-Mechanics Model for Small-Crack Propagation in A356.2**

In the micro-mechanics model, the theory of continuously distributed dislocations was applied to represent the crack and the plastic zone at the tip of the crack in A356.2 cast alloy. The small-crack propagation rate was related to the maximum plastic zone length. When the grain size instead of SDAS was used as the characteristic length of the microstructures, the model predicts the oscillations of the propagation rates, especially when the crack is short. The oscillations ebb as the crack grows. The predicted rates agreed reasonably well with those from experiments, confirming the experimental observation that intergranular eutectic constituents and/or grain boundaries are more influential than interdendritic eutectic constituents on the crack propagation paths. Further endeavors need to include the effects of the Si-particle morphology and possibly the secondary dendrite arm spacings.
CHAPTER 7. CONCLUSIONS AND RECOMMENDATIONS
FOR FUTURE STUDY

7.1. CONCLUSIONS

Fatigue behavior (fatigue life, small-crack initiation and propagation) was studied and fracture mechanics and micro-mechanics models were applied to calculate small-crack propagation rate in A356.2. Radiography, replicas, light microscopy and SEM were used to characterize microstructures and fatigue damages. The results show the following.

Effects of Porosity and SDAS

Fatigue behavior of specimens removed from directionally solidified ingots with a gradient of porosity and SDAS was studied. Fatigue life decreases by a factor of 3 in low-cycle fatigue and by a factor of 100 in high-cycle fatigue as the SDAS increases from 15 to 55 μm. When SDAS is less than 30 μm, the pore size is below the critical size of 80 μm and large eutectic constituents initiate cracks; and the crack initiation life is as high as 70% of the fatigue life. As the SDAS increases beyond 30 μm, pores greater than 100 μm are the main crack-initiation sites; the crack-initiation life is as low as only 5% of the total fatigue life. The oxide defects initiate the fatigue crack when they are near or at the surface, regardless of SDAS. When fatigue crack initiated at pore and oxides, fatigue life is well correlated with the size of the crack initiation site and the effect of SDAS is overshadowed by the effect of pore on the fatigue life.
Effects of Hipping and Strontium-Modification

Fatigue specimens with SDAS of 20 to 30 µm were from permanent-mold cast-plates. Non-hipped A356.2 without Sr shows better fatigue life as a result of less and small pores. The deleterious effect of pores overshadowed the beneficial effect that Sr-modification might have had on improving the fatigue behavior. Hipping significantly increased the fatigue life of A356.2 with Sr as a result of the elimination of the pore as crack initiation site and a slower small-crack propagation rate. However, hipping did not significantly improve the fatigue life of A356.2 without Sr, as a result of the lower hydrogen and porosity content of castings without Sr. After hipping, Sr modification is beneficial in improving the crack initiation life, and increasing both small-crack and long-crack propagation lives. Hence, hipped A356.2 with Sr had a higher fatigue life than hipped A356.2 without Sr. Si-particles within the eutectic constituents and oxides were the predominant crack-initiation sites in hipped A356.2.

Small-Crack Propagation in A356.2

Replicas experiments showed that small-crack propagation rate oscillates in a range and then follows the Paris power law. Crack propagates through the dendrites macroscopically and the tortuous paths are trans-dendrite regardless of the SDAS, with no preference to propagate intergranually or interdendritically. Fracture mechanics models (Newman-Raju, and Trantina-Barishpolsky models) yielded similar results on the crack-propagation rate against the effective stress-intensity factor range. In the micro-mechanics model, the small-crack propagation rate was related to the length of the plastic
zone. When the grain size instead of SDAS was used as the characteristic length of the microstructures, the model predicts the oscillations of the propagation rates, and the predicted rates agreed reasonably well with those from experiments, confirming the experimental observation that intergranular eutectic constituents are more influential than interdendritic eutectic constituents on the crack propagation paths.

7.2. RECOMMENDATIONS FOR FUTURE STUDY

Based on the present work, in order to understand the fundamentals of fatigue behavior in aluminum casting alloys, the following recommendations are made for further studies.

In aluminum casting alloy of normal commercial quality, pores and oxides reduce fatigue life and overshadow the effects of SDAS, Sr-modification. So in the future study, fatigue behavior of aluminum casting alloy without pores and oxides have yet to be understood. The effects of solidification microstructure and heat-treatment on cyclic stress-strain relationship, small-crack initiation and propagation in aluminum castings free from pores and oxides, specifically the effects of secondary dendrite arm spacing (SDAS), grain refinement, and heat-treatment needs to be clarified.

The microstructural constituents interact with plastic deformation, and influence strain hardening behavior, Bauschinger effect, and the rate of damage accumulation by particle cracking and debonding during dynamic cyclic-loading. It is unclear about the effects of SDAS and grain refinement on cyclic hardening behavior and cyclic damage
accumulation, which will greatly influence the small-crack initiation and propagation. Only when these effects are clarified, the optimization of mechanical properties and fatigue performance will be obtained. Heat-treatment determines the structure of precipitates, and the cyclic plasticity and strength, thus the small-crack behavior. In defect-free specimens, fatigue life will mainly consist of small-crack initiation and propagation, it is important to determine the heat-treatment effect on small-crack behavior.

Although it was shown that small-cracks propagate faster than the predicted rate based on long-crack theories, and small-crack propagation rate oscillates when the crack tip interacts with secondary phase particles and grain boundaries, but the influences of length scales of $SDAS$ and grain size on small-crack propagation need to be studied. Micro-mechanics model to predict small-crack propagation, considering grain size, $SDAS$ and Si-particle is needed in high-cycle fatigue.
APPENDIX: MODIFIED FATIGUE LIFE WITH INITIAL STRESS-INTENSITY FACTOR RANGE

In this appendix, the relationship between the modified fatigue life with the size of the crack initiation site is deduced based on the Paris-power law for the long-crack propagation. It has been stated in the Paris-power law:

\[
\frac{da}{dN} = C\Delta K_{\text{eff}}^m
\]  

(A.1)

where the \( \frac{da}{dN} \) is the crack propagation rate; \( C \) is a constant and \( m \) is the slope on the log-log plot of \( \frac{da}{dN} \) with \( \Delta K_{\text{eff}} \). When the effects of crack-closure, boundary-correction and crack-shape are considered, the effective stress intensity factor range, \( \Delta K_{\text{eff}} \) is expressed as:

\[
\Delta K_{\text{eff}} = U(a)F(a)\Delta S\sqrt{\pi a} / Q(a)
\]  

(A.2)

where \( F(a) \) is the boundary-correction factor; and \( U(a) \) is the crack closure factor. Both of these factors change slowly with crack length (Newman and Raju, 1983) and are thus assumed to be constants. \( Q(a) \) is the elliptical crack-shape factor; \( \Delta S \) is the remote-field stress range; Hence substituting equation (A.2) into equation (A.1):

\[
\frac{da}{dN} = C(U(a)F(a)\Delta S\sqrt{\pi a} / Q(a))^m
\]  

(A.3)

\[
da = C(U(a)F(a))^m(\Delta S)^m\pi^{m/2}a^{m/2}(Q(a))^{-m/2}dN
\]  

(A.4)

\[
(U(a)F(a))^{-m}(Q(a))^{m/2}a^{-m/2}da = \pi^{m/2}C(\Delta S)^m dN
\]  

(A.5)

After integration of equation (A.5) from the initial crack size (the size of the crack initiation site) to the final crack length at fracture, the following equation is obtained:
\[ \int_{a_0}^{a_f} \left( \frac{Q(a)}{U(a)F(a)} \right)^{m/2} a^{-\gamma/2} da = \int_{a_0}^{a_f} \pi^{\gamma/2} C(\Delta S)^m dN \]  \hspace{1cm} (A.6)

\( F(a) \) is the boundary-correction factor; and \( U(a) \) is the crack closure factor. Both of these factors change slowly with crack length (Newman and Raju, 1983) and are thus assumed to be constants. \( Q(a) \) is the elliptical crack-shape factor, when half penny-shape crack is assumed, \( Q(a) \) equals to 2.464 (Newman and Raju, 1983), then equation (A.7) is obtained.

\[ \left( \frac{1}{(m-2)/2} - \frac{1}{a_f (m-2)/2} \right) = \frac{2-m}{2} \pi^{\gamma/2} (UF)^m C(\Delta S)^m (N_f - N_i) \]  \hspace{1cm} (A.7)

Based on the plasticity model of a discrete surface of tensile yielding or slip ahead of a crack, it was predicted that the Paris exponent \( m \) is 4 (Suresh, 1998); and from experiments on most aluminum alloys, \( m \) approaches a value of 4.0 (Basner et al., 2001). The crack length at failure \( (a_f) \) is much larger than crack-initiation site size \( (a_o) \). so

\[ \int_{a_0}^{(m-2)/2} \] is neglected. Thus, Eq. (A.7) simplified to

\[ \left( \frac{1}{a_0^{(m-2)/2}} \right) \approx \left( \frac{2-m}{2} \pi^{\gamma/2} (UF)^m (Q)^{-m/2} C(\Delta S)^m (N_f - N_i) \right) \]  \hspace{1cm} (A.8)

\[ -1 = a_0^{(\gamma/2-1)} \left( \frac{2-m}{2} \pi^{\gamma/2} (UF)^m (Q)^{-m/2} C(\Delta S)^m (N_f - N_i) \right) \]  \hspace{1cm} (A.9)

\[ 1 \approx \frac{m-2}{2} (N_f - N_i) a_0^{-1} C(UF\Delta S / \sqrt{\pi a_0 / Q})^m \]  \hspace{1cm} (A.10)

\[ \Delta K_0 = UF\Delta S / \sqrt{\pi a_0 / Q} \]  \hspace{1cm} (A.11)

Then substituting equation (A.11) into equation (A.10), it is obtained:
\[ \frac{(N_f - N_i)}{a_0} = \left( \frac{2}{(m - 2)C} \right) \Delta K_0^{-m} \]  

(A.12)

As presented previously when the SDAS is greater than ~ 28 \( \mu \)m, cracks initiate from both pores and oxides. Replica experiments (discussed later) show that the crack initiation life \( (N_i) \) is a very small part of the total fatigue life for cracks that initiate at pores or oxides, then Eq. (A.12) is further simplified:

\[ \frac{N_f}{a_0} \approx D \Delta K_0^{-m} \]  

(A.13)

where the constant \( D \) is

\[ D = \left( \frac{2}{(m - 2)C} \right) \]

Eqs. (A.7), (A.12), and (A.13) correspond to Eqs. (4.2), (4.3) and (4.4) in Chapter 4, respectively.
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