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Cyclic deformation behavior of an automotive material

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CYCLIC DEFORMATION BEHAVIOR OF AN AUTOMOTIVE MATERIAL

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A project presented to Ryerson University

in partial fulfillment of the

requirements for the degree of

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in the program of

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Toronto, Ontario, Canada, 2008

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CYCLIC DEFORMATION BEHAVIOR OF AN
ALUMINUM MATRIX COMPOSITE

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ABSTRACT

CYCLIC DEFORMATION BEHAVIOR OF AN AUTOMOTIVE MATERIAL

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The A356 Al-Si-Mg cast alloy is being used in the automotive industry to replace some heavy components due to its fabrication flexibility and high strength-to-weight ratio. This study was aimed at identifying cyclic deformation characteristics and fracture mechanisms of the A356 alloy in different material conditions. The microstructure consisted of primary α -Al matrix and eutectic regions containing Si particles of acicular (T5) and spherical (T6 and ModT6) morphologies. The ModT6 sample had a higher yield strength (YS) and ultimate tensile strength (UTS) but a lower strain hardening exponent than the T6 sample, while the T5 sample had a lower YS and UTS but a higher initial strain hardening than the T6 sample. The T6 sample exhibited a higher cyclic hardening capacity and longer fatigue life. Crack initiation in both tensile and fatigue tests occurred at the sub-surface voids. Quasi-cleavage fracture characteristics in the T5 condition and dimple-like fracture features in the T6 and ModT6 were observed.

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To my wife, Elham Loghmani

CHAPTER 1: INTRODUCTION	1
1.1 Background	1
1.2 Problem Statement	2
1.3 Research Objectives	3
1.4 Scope of the Study	4
1.5 Significance of the Study	5
1.6 Thesis Organization	6
CHAPTER 2: LITERATURE REVIEW	7
2.1 Al-Si Alloys	7
2.2 Microstructure and Hardness of A356 Alloys	8
2.3 Tensile Properties of A356 Alloys	10
2.4 Fatigue Properties of A356 Alloys	13
2.4.1 Fatigue crack initiation	16
2.4.2 Effect of pore location and size	20
CHAPTER 3: MATERIAL AND EXPERIMENTAL PROCEDURE	22
3.1 Materials and Composition	22
3.2 Sample Preparation and Microstructure Examination	23
3.3 Mechanical Testing	24
3.3.1 Tensile Test	24
3.3.2 Fatigue Test	25
3.3.3 Fracture Surface Analysis	26
3.4 Data Analysis	27
3.5 Statistical Analysis	28
CHAPTER 4: RESULTS AND DISCUSSION	29
4.1 Microstructure	29
4.2 Tensile Properties	30
4.3 Fatigue Properties	31
4.3.1 Fatigue crack initiation	31
4.3.2 Effect of pore location and size	32
4.4 Fracture Surface Analysis	33
4.5 Statistical Analysis	34
CHAPTER 5: CONCLUSIONS AND RECOMMENDATIONS	35
5.1 Conclusions	35
5.2 Recommendations	36
REFERENCES	37
APPENDICES	38
APPENDIX A: AUTHOR'S DECLARATION	38
APPENDIX B: BORROWER'S PAGE	39
APPENDIX C: ABSTRACT	40
APPENDIX D: ACKNOWLEDGEMENTS	41
APPENDIX E: TABLE OF CONTENTS	42
APPENDIX F: LIST OF FIGURES	43
APPENDIX G: LIST OF TABLES	44
APPENDIX H: GLOSSARY	45
APPENDIX I: NOMENCLATURE	46
APPENDIX J: SCOPE FOR FUTURE RESEARCH	47

TABLE OF CONTENTS

AUTHOR'S DECLARATION.....	ii
BORROWER'S PAGE.....	iii
ABSTRACT.....	iv
ACKNOWLEDGEMENTS.....	v
TABLE OF CONTENTS.....	vii
LIST OF TABLES.....	ix
LIST OF FIGURES.....	x
NOMENCLATURE.....	xvii
CHAPTER 1: INTRODUCTION.....	1
CHAPTER 2: LITERATURE REVIEW.....	3
2.1 Al-Si Alloys.....	3
2.2 Microstructure and Hardness of A356 Alloys.....	5
2.3 Tensile Properties of A356 Alloys.....	10
2.4 Fatigue Properties of A356 Alloys.....	13
2.4.1 Fatigue crack initiation.....	16
2.4.2 Effect of pore location and size.....	20
CHAPTER 3: MATERIAL AND EXPERIMENTAL PROCEDURE.....	22
3.1 Materials and Composition.....	22
3.2 Sample Preparation and Microstructure Examination.....	22

3.3	Tensile and Low Cycle Fatigue Tests.....	23
CHAPTER 4: EXPERIMENTAL RESULTS AND DISCUSSION		25
4.1	Microstructure of A356 in T5, T6 and ModT6 Conditions	25
4.2	Hardness and Tensile Properties of A356 in T5, T6 and ModT6 Conditions.....	30
4.3	Low Cycle Fatigue Properties of A356 in T5, T6 and ModT6 Conditions.....	33
4.4	Fractography.....	40
CHAPTER 5: CONCLUSIONS AND RECOMMENDATIONS FOR FUTURE RESEARCH.....		48
5.1	Conclusions.....	48
5.2	Scope for Future Research.....	50
REFERENCES.....		51

LIST OF TABLES

TABLE 1: Equivalent diameter of the Si particles measured for different times of solution treatment at 540°C [3].....	8
TABLE 2: Mechanical properties of A356 aluminum alloy obtained by tensile tests in different treatment conditions [3].....	10
TABLE 3: Chemical composition of cast A356 alloys [2].....	14
TABLE 4: Rockwell hardness of A356 alloy in the T5, T6 and ModT6 conditions.....	30
TABLE 5: Tensile properties of cast A356 aluminum alloy in different conditions, tested at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$, where the available typical tensile properties of A356 alloy [14] are included for comparison.....	31
TABLE 6: Low cycle fatigue parameters for the A356 cast aluminum alloy in different conditions.....	40

LIST OF FIGURES

Fig.1: Microstructure of the as-thixo A356 aluminum alloy showing the semisolid structure of the material [3].....	5
Fig. 2: SEM image of the as-thixo A356 aluminum alloy showing the different phases analyzed via EDS [3].....	6
Fig. 3: SEM image of the as-solutioned (540°C, 1 h) A356 aluminum alloy showing the different phases analyzed using EDS [3].....	7
Fig.4: Plot of HRF and electrical conductivity vs. time in the solution state at 540°C [3].....	8
Fig. 5: Plot of HRF and electrical conductivity vs. time of ageing at 160°C in the T5 & T6 conditions [3].....	9
Fig. 6: Plot of HRF and electrical conductivity vs. time of ageing at 200°C in the T5 & T6 conditions [3].....	9
Fig. 7: Fracture around the globules in the eutectic region for the A356 alloy in the as-received and T5 conditions [3].....	11

Fig. 8: Fracture surfaces around the globules in the eutectic region for the A356 alloy in the as-received and T5 conditions [3].....	12
Fig 9: Dimple formation on the fracture surface of the T6 treated samples showing high ductility levels [3].....	13
Fig.10: Plate-like Fe-rich intermetallic particles, (a) Low-Fe casting, (b) High-Fe casting [2].....	15
Fig. 11: S–N curves for low-Fe and high-Fe castings for (a) S _t and (b) S _b specimens[2].....	16
Fig. 12: Overview of a fracture surface in a fatigue fractured low-Fe St specimen ($\sigma_a = 140$ MPa, $R = -1$). Three regions were distinguished, region I initiating site, region II-steady crack propagation region, and region III-fast crack propagation region [2].....	17

Fig.13: Fracture surface near the initiating site in a low-Fe St specimen

($\sigma_a = 140$ MPa, $R = -1$). (a) The enlarged image of the initiating site showing the crack initiating from the micro-cell(s). A is the α -Al matrix in the microcell(s), and B is the eutectic region. (b) Plastic deformation α -Al in the eutectic region leading to Si-particle (labeled C) cracking or debonding in low-Fe samples [2].....18

Fig.14: (a) Fracture surface near the initiating site in a high-Fe St specimen

($\sigma_a = 140$ MPa, $R = -1$) showing the crack initiating from a micro-cell.(A: α -Al matrix; B: eutectic region). (b) De-bonded Si-particles (C) and the edge of a cracked or de-bonded plate-like Fe-rich intermetallic (D) observed within the eutectic region of the crack initiating micro-cell [2].....18

Fig. 15: (a) Optical micrograph showing micro-cracks (A) initiated at de-

bonded Si-particles, (b) Optical micrograph showing micro-cracks

(B) initiating from a large Fe-rich intermetallic particle, (c)

Formation of a dominant crack by micro cracks linking [2].....19

Fig. 16: Crack initiation life calculated from the initiation model [2].....20

Fig.17: Fracture surfaces and crack initiating pores (a) SEM image of a typical fracture surface highlighting the crack initiation pore (A), (b) and (c) two typical initiation pores and the detailed fracture surfaces around them [1].....	21
Fig. 18: The dependence of the fatigue life on the equivalent diameter of the failure dominant pore [1].....	21
Fig. 19: <i>Geometry and dimensions of sub-sized samples for tensile and fatigue tests</i>	24
Fig.20: Low-magnification optical images showing the microstructure of A356 alloy in different heat treatment conditions. (a) T5, (b) T6, and (c) ModT6.....	26
Fig.21: Typical SEM images showing the microstructure of A356 alloy in different heat treatment conditions. (a) T5 SEM micrograph, (b) T6 SEM micrograph, (c) ModT6 SEM micrograph.....	28
Fig.22: (a) Typical T6 EDS spectra showing aluminum, silicon, magnesium and iron distribution, (b) a magnified EDS spectrum showing the iron content in the bright plate-like particles seen in Fig.22(a).....	29
Fig.23: Typical true stress-strain curves for A356-T5, T6 and ModT6, obtained at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$	32

Fig. 24: Stress amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T5 heat treatment conditions.....	34
Fig.25: Stress amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T6 heat treatment conditions.....	34
Fig.26: Stress amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in ModT6 heat treatment conditions.....	35
Fig.27: Plastic strain amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T5 heat treatment conditions.....	36
Fig.28: Plastic strain amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T6 heat treatment conditions.....	37
Fig.29: Plastic strain amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in ModT6 heat treatment conditions.....	37
Fig.30: Effect of the total strain amplitude applied on the fatigue life of A356 cast aluminum alloy in T5 heat treatment conditions.....	38

Fig.31: Effect of the total strain amplitude applied on the fatigue life of A356	
cast aluminum alloy in T6 heat treatment conditions.....	39
Fig.32: Effect of the total strain amplitude applied on the fatigue life of A356	
cast aluminum alloy in ModT6 heat treatment conditions.....	39
Fig. 33: Typical SEM images showing the crack initiation sites of tensile	
fracture surfaces of A356 cast aluminum alloy. (a) T5, (b) T6, and	
(c) ModT6 tested at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$	41
Fig.34: Typical SEM images showing the quasi-cleavage and dimple-like	
tensile fracture surfaces of A356 cast aluminum alloy. (a) T5, (b) T6,	
and (c) ModT6, tested at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$	42
Fig.35: Typical SEM images showing fatigue crack initiation sites of A356 cast	
aluminum alloy: (a) T5 fatigued at total strain amplitude of 0.2%, (b) T6	
fatigued at total strain amplitude of 0.4%, and (c) ModT6 fatigued at total	
strain amplitude of 0.6%.....	44
Fig. 36: SEM micrographs showing the typical fatigue crack propagation features	
of A356 cast aluminum alloy in the T5 condition. (a) quasi-cleavage	
cracking, and (b) fatigue striations., tested at total strain amplitude of 0.6%.....	45

Fig.37: SEM micrographs showing the typical fatigue crack propagation features of A356 cast aluminum alloy in the T6 condition. (a) dimple-like characteristics, and (b) fatigue striations, tested at total strain amplitude of 0.4%.....	46
--	----

Fig.38: SEM micrographs showing the typical fatigue crack propagation features of A356 cast aluminum alloy in the ModT6 condition. (a) dimple-like characteristics, and (b) fatigue striations, tested at total strain amplitude of 0.6%.....	47
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NOMENCLATURE

Acronym	Definition
ASTM	American Society for Testing and Materials
EDS	Energy dispersive x-ray spectroscopy
HRF	Rockwell hardness (F scale)
LCF	Low cycle fatigue
SEM	Scanning electron microscope
UTS	Ultimate tensile strength
YS	Yield strength

Symbol	Definition
b	Fatigue strength exponent
c	Fatigue ductility exponent
K	Strength coefficient as per Hollomon equation
K'	Cyclic strength coefficient
n	Strain hardening exponent as per Hollomon equation
n'	Cyclic strain hardening exponent
ε	True strain
ε'_f	Fatigue ductility coefficient

$$\frac{\Delta \varepsilon_p}{2}$$

Plastic strain amplitude

σ

True stress

σ'_f

Fatigue strength coefficient

CHAPTER 1

INTRODUCTION

Implementation of lightweight materials in the automotive and aerospace industries is one of the most effective methods to reduce fuel consumption and greenhouse gas emissions. Because of the effective weight reduction and subsequent improvement in fuel economy of vehicles, as well as flexibility in fabrication and high strength-to-weight ratio, cast A356 Al-Si-Mg alloy is being used to replace some relatively heavy cast iron and steel components in the automotive and aerospace industries [1]. The microstructure of these alloys consists of primary α aluminum phase and a eutectic mixture of aluminum, silicon, magnesium and iron [2, 3]. While being highly productive, the casting process normally introduces defects, such as porosity and entrapped oxide films. It has been demonstrated that, when the defects are present at or near the specimen surface, they can cause rapid crack initiation, and crack growth which then dominate the tensile and fatigue behavior. Other microstructural features, such as Si-particles and Fe-rich intermetallic particles, emerge as important determinants of tensile and fatigue performance [1, 2]. Mechanical properties of Al-Si alloys are related to the morphology of silicon particles (size, shape and distribution), grain size, and dendrite parameters, which can be changed by rapid solidification, chemical modification, and heat treatment in the solid state [4]. This investigation is mainly focused on the effect of heat treatment on the strain-controlled low cycle fatigue behavior.

A lot of work has been done on the stress-controlled S-N curves and fatigue crack propagation behavior of the A356 alloy in the T6 (solution, quenching and age hardening) condition [1, 2, 5-9]; however, only limited work has been reported on the strain-controlled low cycle fatigue (LCF) of the A356-T6 alloy [10, 11]. To further lower the manufacturing cost and conserve energy associated with solution treatment, T5 heat treatment involving only artificial aging directly after fabrication without any solution treatment [12, 13] is considered in the automotive industry. To the authors' knowledge, while very limited reports on the mechanical properties of the A356 alloy in the T5 condition have been seen in the literature [3, 12, 14], no low cycle fatigue behavior of this alloy in the T5 condition has been reported in the literature. The objective of this study was, therefore, to identify cyclic deformation characteristics, low cycle fatigue resistance, and fracture mechanisms of A356 in the T5, T6 and modified T6 conditions in relation to their microstructures, including porosities.

CHAPTER 2

LITERATURE REVIEW

2.1 Al-Si Alloys

The alloys from the Al-Si system are most widely used in the foundry industry thanks to their good castability and high strength-to-weight ratio; the microstructure of these materials consists of a primary phase, aluminum or silicon, and a eutectic mixture of these two elements. These materials contain many impurities: alloying elements partly go into solid solution in the matrix and partly form intermetallic particles during the solidification process. Among cast aluminum alloys, A356 ranks as one of the commercially important alloys used in automotive applications, due to its excellent casting characteristics and good mechanical properties [3].

However, the casting process inevitably introduces defects, such as porosity and entrapped oxide films [2]. Studies have demonstrated that, whenever defects are present at or near the specimen surface, these defects can cause crack initiation, and crack growth will then dominate the fatigue behavior [2]. Continuous improvement in casting techniques has led to a decrease in the size and density of defects. Other microstructural features, such as silicon and Fe-rich intermetallic particles, emerge as important determinants of fatigue performance. Iron is picked up during the molten processing of aluminum and, hence, inevitably exists in commercial cast aluminum-silicon alloys [2]. For most foundry alloys, the presence of iron is detrimental due to the formation of plate-

like Fe-rich intermetallic compounds, such as $\beta\text{-Al}_5\text{FeSi}$. It has been indicated that such particles can bring about high stress concentrations, thereby enhancing crack initiation and decreasing the ductility of the castings [2].

Mechanical properties of Al-Si alloys are related to the morphology of silicon particles (size, shape and distribution), Al grain size, and shape and dendrite parameters [4]. Modification used to change the silicon morphology can be achieved by rapid solidification, and chemical and thermal modifications in the solid state. Na, Sr and Sb are commonly used for chemical modification. Addition of small amounts of these elements to the melt changes the coarse and large needles of silicon into a fine and spherical form. Heat treatment also changes silicon morphology and the kinetics of spheroidization are affected by the as-cast microstructure [4].

The strength of Al-Si cast binary alloys can be improved by the addition of small amounts of magnesium (about 0.35%) [15]. The important alloy of this type is A356, which contains 7% Si for castability and 0.35% Mg to make the alloy heat-treatable. The magnesium silicide (Mg_2Si) content of the alloy is in the range of 0.5 to 0.6 % and the precipitation strengthening is attributed to the metastable phase of Mg_2Si . By reducing the iron level of A356 alloys to about 0.1%, considerable improvement in the strength of this alloy can be attained [15].

2.2 Microstructure and Hardness of A356 Alloys

A356 is one of the candidate alloys to perform the shaping of aluminum alloys in the semisolid state or thixo-casting that has become a widely accepted industrial process in recent years [3]. Fig. 1 shows the SEM microstructure of the as-cast A356 aluminum alloy sample. The microstructure of the electromagnetically stirred bars is composed of solid matrix globules (solid part) surrounded by the eutectic region (liquid part). The eutectic of the studied system contains Si, Fe and Mg particles embedded in Al [3].

Fig.1: Microstructure of the as-thixo A356 aluminium alloy showing the semisolid structure of the material [3].

In the as-thixo state, the EDS reveals that the α globules, as indicated by arrow 1 in Fig.2, are constituted by Al and Mg in small quantities. The eutectic phase (arrow 2 in Fig.2) is constituted by phases rich in Fe and Mg, where Fe is combined with the other elements to form different intermetallic phases, such as α -AlFeSi or lamellar particles $\text{Fe}_2\text{Si}_2\text{Al}_9$ or FeAl_3 , and Mg is combined with aluminium to form Mg_2Al_3 or Mg_2Si [3].

Fig.4: Plot of HRF and electrical conductivity vs. time in the solution state at 540 °C [3].

The mean equivalent diameter of the Si particles was measured for all the solution times (Table 1) indicating that the spheroidization of such phases is accompanied with the softening of the material [3].

TABLE 1: Equivalent diameter of the Si particles measured at different times of solution treatment at 540 °C [3].

Time (h)	Equivalent diameter (μm)
1	1.9 ± 0.8
2	2.3 ± 1.2
4	2.8 ± 1.2
8	3.1 ± 1.9
16	3.2 ± 1.5

In Figs. 5 and 6, the aging curves are plotted at 160 °C and 200°C for the material in the T5 and T6 conditions. The curves show higher levels of hardness at the same time and temperature of treatment for the material in the T6 condition. It can be seen that, at

Fig.3: SEM image of the as-solutioned (540°C, 1 h) A356 aluminium alloy showing the different phases analyzed using EDS [3].

The hardness and electrical conductivity of the material were plotted versus time during the solution treatment at a temperature of 540°C for 1,2,4,8 and 16 hours (Fig.4). The hardness shows an increase for the shorter times followed by a decrease at the larger ones; after 8 h, the hardness levels off, indicating that the aluminium matrix is saturated. The slight increase in hardness is because of the enhanced solution hardening mechanisms of Fe and Mg due to their high solubility at 540°C; after 8 h, the coarsening of the Si phase results in the drop to the minimum hardness [3].

2.3 Tensile Properties of A356 Alloys

Tensile tests were performed on solution treated-artificially aged A356 at 160°C and 200°C in both T5 and T6 conditions. The results show that the YS and UTS increase with increasing time in both materials conditions; in the T5 condition, the UTS is comparable with that measured for the T6 one, although the YS is somewhat lower in the T5 condition than in the T6 condition. Table 2 shows a summary of the YS and UTS values [3].

TABLE 2: Mechanical properties of A356 aluminium alloy obtained by tensile tests in different treatment conditions [3].

Specimens condition	Yield stress (MPa)	UTS (MPa)
As-thixo	104	241
Solution 1 h	106	231
T6 at 160°C, 0.5 h	115	243
T6 at 160°C, 4 h	127	245
T6 at 160°C, 25 h	153	264
T6 at 200°C, 0.5 h	111	236
T6 at 200°C, 4 h	154	257
T6 at 200°C, 25 h	247	297
T5 at 160°C, 0.5 h	103	239
T5 at 160°C, 4 h	107	236
T5 at 160°C, 25 h	129	244
T5 at 200°C, 0.5 h	104	238
T5 at 200°C, 2h	150	252
T5 at 200°C, 25 h	201	280

200°C, over aging starts after 2 h; while, at 160°C, hardness increases continuously to a maximum reached after 1 day in both T5 and T6 conditions [3].

Fig.5: Plot of HRF and electrical conductivity vs. time of ageing at 160°C in the T5 & T6 conditions [3].

Fig.6: Plot of HRF and electrical conductivity vs. time of ageing at 200°C in the T5 & T6 conditions [3].

The fracture surfaces of the as-thixo and T5 specimens revealed a brittle characteristic, as clearly shown in Fig.8, while the specimens in the T6 condition revealed a ductile feature demonstrated by the presence of dimples around the bigger particles visible in the microstructure (Fig.9) [3].

Fig.8: Fracture surfaces around the globules in the eutectic region for the A356 alloy in the as-received and T5 conditions [3].

The scanning electron microscopy (SEM) of the longitudinal sections of tensile tested samples reveals that the fracture propagates in the eutectic region around the globules, as shown in Fig.7 [3].

Fig.7: Fracture around the globules in the eutectic region for the A356 alloy in the as-received and T5 conditions [3].

under resonant vibration. However, by increasing the Fe-content (0.97 wt.%), the fracture resistance decreased due to the β -Al₅FeSi phase formation. Fatigue cracks were found to initiate and grow along the eutectic silicon particles and the Fe-rich intermetallic phase [16]. Yi et al. [2] investigated the effect of Fe-content on fatigue crack initiation and propagation in a cast aluminum–silicon alloy (A356-T6). The chemical compositions of the A356-T6 aluminum-silicon alloys with a low Fe-content (<0.1 wt.% Fe, designated as ‘Low-Fe’) and high Fe-content (>0.55 wt.% Fe, designated as ‘High-Fe’) are shown in Table 3 [2].

TABLE 3: The chemical composition of cast A356 alloys [2]

Casting	Si	Fe	Mg	Cu	Mn	Cr	Ni	Zn	Ti	B
Low-Fe	7.25	0.06	0.32	<0.01	<0.01	<0.01	<0.01	0.02	<0.01	<0.01
High-Fe	6.59	0.57	0.39	<0.01	<0.01	<0.01	<0.01	0.02	0.02	<0.01

Plate-like Fe-rich intermetallic particles (possible β -phase, Al₅FeSi) were observed in the castings and their size was proportional to the Fe-content. For the low-Fe casting, only small needle-like Fe-rich intermetallic particles were observed in the eutectic region (Fig. 10(a)). For the high-Fe casting, large plate-like Fe-rich intermetallic particles were extensively observed in the eutectic region, as shown in Fig. 10(b) [2].

Fig.9: Dimple formation on the fracture surface of the T6 treated samples showing high ductility levels [3].

2.4 Fatigue Properties of A356 Alloys

Hornig *et al.* [16] investigated the fracture behavior of A356 alloys by varying the Fe-content within the range of 0.14-0.97 wt.%, and their results indicated that, when the Fe content is below 0.57 wt.%, the Fe-rich intermetallic particles are mainly π - $\text{Al}_8\text{Mg}_3\text{FeSi}_6$ and α - $\text{Al}_{15}\text{Fe}_3\text{Si}_2$ phases, and there is no obvious difference in the fracture resistance

Fig.11: S-N curves for low-Fe and high-Fe castings for (a) S_t and (b) S_b specimens [2].

2.4.1 Fatigue crack initiation

As shown in Fig.12, the features responsible for crack initiation can be distinguished as (I) crack initiation; (II) steady crack growth; and (III) fast crack growth. Closer examination of region-I at higher magnification indicates that the cracks initiated from microcells at the specimen surface [2]. Details of the initiation site observed in a low-Fe

Fig.10: Plate-like Fe-rich intermetallic particles, (a) Low-Fe casting, (b) High-Fe casting [2].

S-N curves were generated for the low-Fe and high-Fe castings for both S_t and S_b specimens (Fig. 11), where the S_t specimen was taken from the top part of the wedge ingot and the S_b specimen was taken from the bottom part of the wedge ingot. The effect of Fe-content on the fatigue life is dependent on the applied stress level. It seems that a critical stress level exists, below which high Fe-content reduces the observed fatigue life. However, if the applied stress is above the critical level, the high Fe-content increases the fatigue life slightly. From the S-N curves, this critical stress amplitude is approximately 130-140 MPa for these castings [2].

Fig.13: Fracture surface near the initiating site in a low-Fe S_t specimen ($\sigma_a = 140$ MPa, $R = -1$). (a) The enlarged image of the initiating site showing the crack initiating from the micro-cell(s). A is the α -Al matrix in the microcell(s), and B is the eutectic region. (b) Plastic deformation α -Al in the eutectic region leading to Si-particle (labeled C) cracking or debonding in low-Fe samples [2].

Fig.14: (a) Fracture surface near the initiating site in a high-Fe S_t specimen ($\sigma_a = 140$ MPa, $R = -1$) showing the crack initiating from a micro-cell. (A: α -Al matrix; B: eutectic region). (b) De-bonded Si-particles (C) and the edge of a cracked or de-bonded plate-like Fe-rich intermetallic (D) observed within the eutectic region of the crack initiating micro-cell [2].

Therefore, it can be concluded that, in the absence of other defects (such as large pores or oxide films), fatigue crack initiation in both low-Fe and high-Fe specimens will be dominated by the local plastic deformation accumulation within the micro-cells

casting are shown in Fig. 13(a). The α -Al matrix (region A) and the surrounding eutectic (region B) are clearly distinguished at the initiation site. Fig. 13(b) shows that there is significant plastic deformation of the α -Al in the eutectic region leading to either cracking of the Si-particles or de-cohesion of the particle-matrix interface [2]. Similar observations were also made for high-Fe specimens (see Fig.14). The only difference from the low-Fe casting is the existence of large plate-like Fe-rich intermetallic particles, which were also found to crack or de-bond from the matrix in the initiating micro-cell (see Fig. 14(b)) [2].

Fig.12: Overview of a fracture surface in a fatigue fractured low-Fe S_i specimen ($\sigma_a = 140$ MPa, $R = -1$). Three regions were distinguished, region I initiating site, region II-steady crack propagation region, and region III-fast crack propagation region [2].

One possible reason is that the stress concentration developed around the particles (either silicon or Fe-rich intermetallic) in the eutectic region is strongly dependent on their size and morphology [2]. Basically, the large plate-like Fe-rich intermetallic particles are more effective in blocking the movement of dislocations from one micro-cell to another. Accordingly, more dislocations would accumulate between the interface of the particle and the matrix, where a high strain concentration would occur [2]. When the stress exceeds a critical level, a fatigue crack formed either by cracking the particles or along the interface between the particles and the matrix, depending on the strength of the particles and the interface [2].

Fig. 16: Crack initiation life calculated from the initiation model [2].

2.4.2 Effect of pore location and size

SEM shows that most of the fatigue cracks initiate from pores, usually, from large pores located at, or close to, the specimen surface [1]. These are termed the failure-dominant pores.

intersecting the specimen surface [2]. Figure 15 shows that micro-cracks initially developed from the eutectic region by de-bonding and cracking of Si-particles or large Fe-rich intermetallic particles (Fig. 15(a) and (b)), due to the local accumulation of plastic deformation within the cells in the preferentially oriented grains and intersecting the surface [2]. These micro-cracks were linked by the localized plastic deformation in the form of shear bands to form the dominant small-crack, as shown in Fig. 15(c) [2].

Fig. 15: (a) Optical micrograph showing micro-cracks (A) initiated at de-bonded Si-particles, (b) Optical micrograph showing micro-cracks (B) initiating from a large Fe-rich intermetallic particle, (c) Formation of a dominant crack by micro cracks linking [2].

As shown in Fig. 16, the presence of large plate-like Fe-rich intermetallic particles in the high-Fe casting reduces the crack initiation life for all stress levels. According to Fig.16, the initiation life for high-Fe castings is approximately one order of magnitude lower than that for the low-Fe castings in the lifetime regime of 10^4 - 10^7 cycles [2].

CHAPTER 3

MATERIAL AND EXPERIMENTAL PROCEDURE

3.1 Materials and Composition

This study was conducted using a cast aluminum alloy A356 with a nominal composition (wt%) of 7.0 Si and 0.35 Mg [14] in T5, T6 and modified T6 (ModT6) conditions. The heat treatment procedure for various aluminum alloys has been described in [13]. The T5 condition was obtained via artificial aging only at 182°C for 8 hours; the T6 condition was achieved with a solution heat treatment at 549°C for 4 hours, quenching into room-temperature water, and artificial aging at 182°C for 8 hours, while the ModT6 condition was done with a solution heat treatment at 549°C for 20 minutes in a SiC fluidized bed, quenching to 21~27°C in the SiC fluidized bed, and artificial aging at 182°C for 8 hours in the SiC fluidized bed as well.

3.2 Sample Preparation and Microstructure Examination

In order not to change potentially the microstructure of the alloy, the metallographic samples were cold mounted using LECOSSET 7007. The sample surfaces were prepared with utmost care. After hand-grinding with 0/0, 1/0, 2/0, 3/0 and 4/0 emery papers, the specimen surface was further ground using 1200 grit silicon carbide paper. Then the specimens were polished using 1 micron diamond paste with diamond extender as a lubricant. Chemomet cloths were finally used to finely polish the sample. Samples were etched using Kellers etchant (5 ml HNO₃, 3 ml HCl, 2 ml HF, and 190 ml water) for

A typical fatigue fracture surface highlighting the pore as the crack initiator is shown in Fig. 17(a), while Fig. 17 (b) and (c) give details around the failure-dominant pores. The dependence of the fatigue life on the equivalent diameter of the failure-dominant pore is presented in Fig. 18 [1].

Fig.17: Fracture surfaces and crack initiating pores (a) SEM image of a typical fracture surface highlighting the crack initiation pore (A), (b) and (c) two typical initiation pores and the detailed fracture surfaces around them [1].

Fig. 18: The dependence of the fatigue life on the equivalent diameter of the failure dominant pore [1].

Fig.19: Geometry and dimensions of sub-sized samples for tensile and fatigue tests

about 6 seconds. An optical microscope equipped with an image analysis system (CLEMEX software) was used to observe the microstructure of the etched samples. Rockwell hardness of the test samples was measured by a digital Buehler hardness tester in the F scale of the Wilson Conversion Chart (1/16" ball and 60 kg load). All tests were conducted in the mechanical and industrial engineering department of Ryerson University.

3.3 Tensile and Low Cycle Fatigue Tests

The sub-sized samples for both tensile and fatigue tests (prepared by Magna COSMA International in accordance with ASTM E8 standards) had a gauge length of 25 mm and a width of 6 mm. Geometry and dimensions of the samples are shown in Fig.19. The tensile tests were carried out using a computerized United tensile testing machine at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$. Two specimens were tensile tested for each material condition. The strain-controlled, push-pull type low cycle fatigue (LCF) tests were carried out at zero mean strain ($R_\epsilon = -1$, completely reversed strain cycle), at a room temperature of 25°C at total strain amplitudes of 0.1%, 0.2%, 0.3%, 0.4%, 0.5% and 0.6%. Five specimens were tested at each level of the strain amplitudes using a computerized Instron fatigue testing system. Triangular waveform loading at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$ was applied during the LCF tests. The strain-controlled tests at lower strain amplitude levels were continued up to 10,000 cycles, then the tests were changed to load control tests with a frequency of 50 Hz. The fracture surfaces of the specimens after tensile and fatigue tests were examined using SEM/EDS to identify fatigue crack initiation sites and propagation mechanisms.

Fig.20: Low-magnification optical images showing the microstructure of A356 alloy in different heat treatment conditions. (a) T5, (b) T6, and (c) ModT6.

CHAPTER 4

EXPERIMENTAL RESULTS AND DISCUSSION

4.1 Microstructure of A356 in T5, T6 and ModT6 Conditions

Figure 20 and Figure 21 show the microstructure of the cast A356 aluminum alloy in the T5, T6 and ModT6 conditions, taken with an optical microscope at a lower magnification and SEM at a higher magnification, respectively. It is seen that the microstructure of A356 alloy was composed of primary α aluminum surrounded by eutectic regions with a cluster of eutectic silicon particles (grey phases). EDS analysis revealed that the eutectic region contained elements of Al, Si, Mg and Fe, as shown in Fig.21 (d) and (e). Silicon particles were small but plate-shaped or acicular in the T5 condition (Fig. 21(a)), while they became bigger but had a spherical shape in the T6 and in ModT6 conditions (Fig. 21(b) and (c)). The plate-like Si particles were thermally very unstable and instantly disintegrated and spheroidized at sufficiently high temperatures during the solution treatment [12, 17]. The morphology of Si particles was important because large and elongated Si particles were reported to fracture more frequently than spherical ones [12, 18]. The addition of silicon in this alloy resulted in excellent castability, good fluidity, and resistance to hot cracking and thermal expansion. Another important alloying element was magnesium, which led to an improvement in mechanical properties, such as the strength-to-weight ratio and the yield stress, arising from the precipitation mechanism. Mg combined with Si to form the aging-hardening compound Mg_2Si , which precipitated from a solid solution during artificial aging [12].

Fig.21: Typical SEM images showing the microstructure of A356 alloy in different heat treatment conditions: (a) T5 SEM micrograph, (b) T6 SEM micrograph, (c) ModT6 SEM micrograph.

The bright plate-like particles in the microstructure were Fe-rich intermetallic particles, as determined by EDS analysis shown in Fig.22 (a) and (b). Iron is easily picked up during the molten processing of aluminum and hence inevitably existent in commercial cast aluminum-silicon alloys [2]. For most cast alloys, the presence of iron was detrimental due to the formation of plate-like Fe-rich intermetallic compounds, such as β -Al₅FeSi. It has been indicated that such particles could bring about high stress concentrations, thereby enhancing crack initiation and decreasing the ductility of the castings [2]. According to the quantitative image analysis, the eutectic region in T5 samples was obtained to be about $66 \pm 4\%$ of the total area, while, in T6 and ModT6, it became about $60 \pm 4\%$ and $52 \pm 2\%$, respectively.

4.2 Hardness and Tensile Properties of A356 in T5, T6 and ModT6 Conditions

The Rockwell hardness of the samples in different conditions was measured in F scale and the obtained results are shown in Table 4. It is seen that the T6 sample has a higher hardness than the T5 sample due to the effect of solution treatment. A similar result was also reported by Cavaliere *et al.* [3] for T5 and T6 material conditions. The ModT6 sample was observed to have nearly the same hardness as the T6 sample if the experimental scatter was considered.

Table 4: Rockwell hardness of A356 alloy in the T5, T6 and ModT6 conditions.

A356 samples	T5	T6	ModT6
HRF	75 ± 3	89 ± 2	85 ± 5

Table 5 shows the tensile properties of the A356 alloy in different material conditions, obtained at a strain rate of 10^{-2} s^{-1} . Due to the spheroidization of Si particles in the T6 condition, this material exhibited higher ductility than the T5 and ModT6 materials. Similar results were also reported by Cavaliere *et al.* [3] and Kumai *et al.* [19]. However, the average yield strength (YS) and ultimate tensile strength (UTS) of T6 specimens lay in-between those of T5 and ModT6 specimens. For comparison, the available typical tensile properties of A356 alloy in the literature [14] are also listed in Table 5. It is seen that while the YS of the present T6 alloy was essentially the same, both the UTS and ductility were higher than the reported typical data [14]. For the T5 condition in the present study, the average value of all tensile properties (YS, UTS, and ductility) was higher than the reported typical data [14].

Fig.22: (a) Typical T6 EDS spectra showing aluminum, silicon, magnesium and iron distribution, (b) a magnified EDS spectrum showing the iron content in the bright plate-like particles seen in Fig.22(a).

condition (solution treatment at 550°C for 8 hours and artificial aging at 150°C for 24 hours [21]) was somewhat different from the T6 condition in the present study.

Fig.23: Typical true stress-strain curves for A356-T5, T6 and ModT6, obtained at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$.

Table 5: Tensile properties of cast A356 aluminum alloy in different conditions, tested at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$, where the available typical tensile properties of A356 alloy [14] are included for comparison.

Material condition	YS, MPa	UTS, MPa	Elongation, %
T5	177	233	9
T6	200	298	12
ModT6	258	315	10
T51, sand cast [14]	125	180	3
T51, permanent mold [14]	140	200	5
T6, sand cast [14]	205	275	6
T6, permanent mold [14]	205	285	12

Fig.23 shows typical true stress-true strain curves obtained at a strain rate of 10^{-2} s^{-1} . Again, the true stress of the T6 sample lay in-between that of T5 and ModT6 samples at a given true strain. However, the T6 sample exhibited a higher overall hardening rate (i.e., the steeper slope on the true stress-true strain curve in Fig. 22), while the T5 sample had a slightly higher initial hardening rate. The following Hollomon equation,

$$\sigma = K\varepsilon^n, \quad (1)$$

has been suggested in ASTM E646 to describe approximately the flow curve in the uniform deformation stage after yielding up to the point corresponding to the UTS, where σ is the true stress, ε is the true strain, K is the strength coefficient and n is the strain-hardening exponent. Based on eq. (1) the evaluated hardening exponent for T5, T6 and ModT6 materials was 0.177 ± 0.003 , 0.166 ± 0.004 and 0.089 ± 0.005 , respectively. Similar results were also reported by Rometsch and Schaffer [20] where the n values ranging between slightly higher than 0.17 and slightly lower than 0.08 were given for the A356 cast alloy, depending on the aging time-temperature combinations. Also, a range of n values between 0.05 and 0.15 was given by Lee [21] for the A356-T6 alloy, depending upon the variation of microporosity. It should be noted that the T6 heat treatment

Fig. 24: Stress amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T5 heat treatment conditions.

Fig. 25: Stress amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T6 heat treatment conditions.

4.3 Low Cycle Fatigue Properties of A356 in T5, T6 and ModT6 Conditions

Figures 24 to 26 show the evolution of stress amplitude as a function of the number of cycles at different strain amplitudes on a semi-log scale for A356 T5, T6 and ModT6 samples. The obtained results showed that, at the higher total strain amplitudes applied (0.4-0.6%), like the tensile test results (Table 5), the magnitude of the initial stress amplitudes of the T6 sample lay in-between the T5 and ModT6 samples, and the initial cyclic strain hardening occurred in all three material conditions. After a few cycles (about 5~10 cycles), nearly constant stress amplitudes were observed until failure for the T5 and ModT6 samples (Figs.24 and 26), but the cyclic strain hardening continued until failure for the T6 sample (Fig.25). At lower strain amplitudes (0.1-0.3%), the stress amplitude remained almost constant in all three material conditions. As also seen from Figs.24 to 26, the stress amplitude increased and the fatigue life decreased with increasing total strain amplitude in all three material conditions. Similar results on the cyclic hardening characteristics and fatigue life with respect to the applied total strain amplitudes were also reported by Begum *et al.* [22] and Lin and Chen [23] on the cyclic deformation characteristics of extruded magnesium alloy.

(Figs.27 and 29). The plastic strain amplitude at lower total strain amplitudes (0.1-0.2%) remained basically constant especially for T6 and ModT6 conditions.

Fig.27: Plastic strain amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T5 heat treatment conditions.

Fig.26: Stress amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in ModT6 heat treatment conditions.

A physical quantity, plastic strain amplitude $\left(\frac{\Delta\epsilon_p}{2}\right)$, has generally been considered to govern the fatigue life through initiating several damaging processes and influencing the internal microstructure, which eventually affects the strain resistance and finally the fatigue life [18]. The plastic strain amplitude as a function of the number of cycles is shown in Fig.27, Fig.28 and Fig.29 for the T5, T6 and ModT6 conditions, respectively. As the total strain amplitude increased, the curve was shifted higher for all three material conditions. It is seen from Figs.27 to 29 that, at a given higher total strain amplitude applied (0.3-0.6%), the magnitude of the initial plastic strain amplitude became smaller in the sequence of T5, T6, and ModT6, although the plastic strain amplitude in the T6 condition (Fig.28) decreased more rapidly than that in the T5 and ModT6 conditions

The obtained fatigue life versus the applied total strain amplitude is shown in Figs.30-32 for the three materials conditions (T5, T6 and ModT6). The fatigue life, which increased with decreasing total strain amplitude, was basically longer in the T6 condition than in the ModT6 and T5 conditions due to the stronger hardening capacity (Fig. 25) and the resultant decrease in the plastic strain amplitude (Fig. 28). This could also be seen in Table 6, where the fatigue parameters are listed, including the cyclic strain-hardening exponent n' , cyclic strength coefficient K' , fatigue strength exponent b , fatigue strength coefficient σ'_f , fatigue ductility exponent c , and fatigue ductility coefficient ϵ'_f . It is seen from Table 6, that while the fatigue ductility exponent of the T6 condition lay in-between that of the T5 and ModT6 conditions, both the cyclic strain hardening exponent and fatigue strength exponent (absolute value) of the T6 condition were higher than those of the T5 and ModT6 conditions.

Fig.30: Effect of the total strain amplitude applied on the fatigue life of A356 cast aluminum alloy in T5 heat treatment conditions.

Fig.28: Plastic strain amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in T6 heat treatment conditions.

Fig.29: Plastic strain amplitude versus the number of cycles during cyclic deformation of A356 cast aluminum alloy in ModT6 heat treatment conditions.

Table 6: Low cycle fatigue parameters for the A356 cast aluminum alloy in different conditions.

Low cycle fatigue parameters	T5	T6	ModT6
Cyclic strain hardening exponent, n'	0.15	0.24	0.21
Cyclic strength coefficient, K' , MPa	502	1628	1105
Fatigue strength coefficient, σ'_f , MPa	345	800	584
Fatigue strength exponent, b	-0.10	-0.15	-0.12
Fatigue ductility coefficient, ϵ'_f , %	5.08	2.60	2.90
Fatigue ductility exponent, c	-0.57	-0.55	-0.52

4.4 Fractography

Fracture surfaces of tensile specimens were examined using SEM. Figure 33 show some typical initiation sites on the tensile fractured surfaces for the A356-T5, T6 and ModT6 conditions. It is seen that cracks initiated basically from voids/defects at or near the specimen's surface in all the three material conditions. "Quasi-cleavage" facets were observed on the fracture surface of the T5 samples (Fig.34 (a)). This could be attributed to the coarse grain and acicular silicon particles of the material, as seen in Figure 21(a). Since the acicular silicon particles were more brittle and fracture faster than rounded particles [12, 21], the material showed semi-brittle behavior. The dimple-like fracture surfaces of T6 and ModT6 samples (Fig. 34(b) and (c)) revealed the ductile behavior of these samples during the tensile tests. This could be attributed to the finer microstructure and rounded silicon particles, as shown in Fig.21 (b) and (c).

Fig.31: Effect of the total strain amplitude applied on the fatigue life of A356 cast aluminum alloy in T6 heat treatment conditions.

Fig.32: Effect of the total strain amplitude applied on the fatigue life of A356 cast aluminum alloy in ModT6 heat treatment conditions.

Fig.34: Typical SEM images showing the quasi-cleavage and dimple-like tensile fracture surfaces of A356 cast aluminum alloy. (a) T5, (b) T6, and (c) ModT6, tested at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$.

Fig.33: Typical SEM images showing the crack initiation sites of tensile fracture surfaces of A356 cast aluminum alloy. (a) T5, (b) T6, and (c) ModT6 tested at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$

Fig.35: Typical SEM images showing fatigue crack initiation sites of A356 cast aluminum alloy: (a) T5 fatigued at total strain amplitude of 0.2%, (b) T6 fatigued at total strain amplitude of 0.4%, and (c) ModT6 fatigued at total strain amplitude of 0.6%.

Fig.35 shows typical SEM images on the initiation sites of fatigue fracture surfaces of A356 alloy in the T5, T6 and ModT6 conditions. Clearly, the fatigue crack initiated from the voids/defects at or close to the surface as well, and propagated into the inside of the samples in all the three conditions. Similar crack initiation sites were also observed in the high cycle fatigue tests on the cast A356-T6 material, due to the high stress/strain concentration induced by larger pores close to the specimen surface [1]. In the crack propagation area of the T5 samples both features of quasi-cleavage and fatigue striations perpendicular to the propagation direction were observed, as seen in Figure 36(a) and (b). Figs 37 and 38 show the observed characteristics in the propagation area on the fatigue fracture surfaces of A356 alloy in the T6 and ModT6 material conditions. Basically, both dimple-like characteristics and fatigue striations were observed for both T6 and ModT6 material conditions. Microvoids and secondary cracks were also seen on the fracture surfaces in all the three material conditions, as shown in Fig.36 (a) and Fig.37 (a). These fracture surface features after tensile and/or fatigue tests indicated that the T6 and Mod T6 materials would have higher ductility than the T5 samples (Table 5). Similar results have also been reported by Cavaliere *et al.* [3].

Fig. 37: SEM micrographs showing the typical fatigue crack propagation features of A356 cast aluminum alloy in the T6 condition. (a) dimple-like characteristics, and (b) fatigue striations, tested at total strain amplitude of 0.4%.

Fig.36: SEM micrographs showing the typical fatigue crack propagation features of A356 cast aluminum alloy in the T5 condition. (a) quasi-cleavage cracking, and (b) fatigue striations., tested at total strain amplitude of 0.6%.

CHAPTER 5

CONCLUSIONS AND RECOMMENDATIONS FOR FUTURE RESEARCH

5.1 Conclusions

The study was conducted using a cast aluminum alloy A356 in the T5, T6 and ModT6 conditions in order to evaluate its cyclic deformation characteristics and fatigue life with emphasis on the effect of the heat treatment conditions. The following results were obtained:

1. The microstructure of the A356 alloy basically consisted of primary α aluminum matrix embedded in the eutectic regions containing clusters of silicon particles. The volume fraction of the eutectic regions became smaller in the T6 and ModT6 conditions with solution treatment. Silicon particles were acicular in the T5 condition and became spherical in the T6 and ModT6 conditions.
2. The hardness in the T6 and ModT6 conditions was essentially the same, but the T5 sample was somewhat lower than the T6 and ModT6 samples.
3. The tensile properties of the alloy in all the three conditions were observed to be equivalent to or higher than the typical values reported in the literature in the corresponding heat treatment conditions. The ModT6 sample had slightly lower ductility but higher yield strength and ultimate tensile strength than the T6 sample, which were higher than the values of the T5 sample.

Fig.38: SEM micrographs showing the typical fatigue crack propagation features of A356 cast aluminum alloy in the ModT6 condition. (a) dimple-like characteristics, and (b) fatigue striations, tested at total strain amplitude of 0.6%.

5.2 Scope for Future Research

As mentioned in the introduction, tensile and fatigue properties of the A356 cast aluminum alloy were basically reported for the T6 heat treatment condition. Except the present study, no fatigue data of the A356 alloy in the T5 condition could be found in the open literature. The present investigation basically involved the strain-controlled low cycle fatigue behavior and the related microstructural characterization and tensile tests. Future work would, therefore, be to evaluate the stress-controlled high cycle fatigue resistance and fatigue crack propagation behavior of T5 material, in conjunction with (1) the changes in the microstructures and porosity (volume fraction, shape and size, etc.), and (2) the effect of chemical composition, such as the content of Fe and Mg.

4. While the T5 sample exhibited an initial high hardening rate, the T6 sample had a high overall hardening rate up to a larger strain value. Due to the initial high yield strength, the hardening rate of the ModT6 sample was relatively small.
5. Cyclic deformation tests showed that the stress amplitude response remained basically constant at lower total strain amplitudes, and cyclic hardening occurred at higher total strain amplitudes. In comparison to the T5 and ModT6 samples, the T6 sample exhibited a higher cyclic hardening capacity, thus leading to a longer fatigue life.
6. SEM examinations of fracture surfaces after both tensile and fatigue tests showed that cracks initiated basically from the voids/defects at or near the specimen surface in all the three material conditions. Due to the presence of acicular silicon particles, the T5 had quasi-cleavage fracture characteristics, while the T6 and ModT6 exhibited dimple-like fracture features. The typical fatigue striations, in conjunction with some secondary cracks and microvoids, were also observed in the fatigue crack propagation regime.

Fig.35. SEM micrographs showing the typical fatigue crack propagation features in A356 cast aluminum alloy in the ModT6 condition (a) dimple-like characteristics and (b) fatigue striations observed at total strain amplitude of 0.6%.

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