A METHODOLOGY FOR QUANTIFYING
THE THERMAL AND MECHANICAL CONDITIONS FOR
WELD METAL SOLIDIFICATION CRACKING

DISSERTATION

Presented in Partial Fulfillment of the Requirement for
the Degree Doctor of Philosophy in the Graduate
School of The Ohio State University

By

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* * * * *

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TO MY PARENTS
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# TABLE OF CONTENTS

ACKNOWLEDGEMENTS ................................................................. iii

VITA ................................................................................ iv

LIST OF TABLES .................................................................................. viii

LIST OF FIGURES ................................................................................ ix

NOMENCLATURE ................................................................................ xiv

CHAPTER PAGE

I. INTRODUCTION ................................................................. 1

   1.1 Background and Research Issues .............................................. 1
   1.2 Statement of Problem and Scope of Research ...................... 11
   1.3 Anticipated Benefits .............................................................. 13

II. DISSERTATION OBJECTIVES AND ORGANIZATION ............... 14

III. WELD METAL SOLIDIFICATION CRACKING PHENOMENON ....... 18

   3.1 The Nature of Solidification Cracking .................................. 19
   3.2 Measurement of Material's Resistance to Weld Metal Solidification
       Cracking .................................................................................. 47
   3.3 Strain Development during Welding ......................................... 62
   3.4 Closure .................................................................................. 71

IV. DEVELOPMENT OF COMPUTATIONAL MODELS ..................... 73

   4.1 Basic Considerations ............................................................. 76
   4.2 Heat Transfer Analysis ........................................................... 79
4.2.1 Relevant Factors and Their Treatment in the Heat Transfer Analysis
4.2.2 Mathematical Formulation
4.2.3 Finite Element Implementation
4.3 Stress/Strain Analysis
4.3.1 The Effects of Solidification Process
4.3.2 Material Properties
4.3.3 Finite Element Implementation
4.4 Construction of the Mechanical Strain Curves

V. VERIFICATION OF COMPUTATIONAL MODELS

5.1 Deformation Characteristics of Weld Edges
5.1.1 Comparison with Matsuda et al.’s Measurements
5.1.2 Comparison with Chihoski’s Measurements
5.2 Strain Fields in the Vicinity of Weld Pool
5.3 Closure

VI. ASSESSMENT OF MECHANICAL DRIVING FORCE FOR WELD METAL SOLIDIFICATION CRACKING

6.1 The Approach: Hot Strain versus Threshold Strain
6.2 Hot Strain Assessment
6.2.1 Model Description
6.2.2 Results and Discussion

VII. CONCLUSIONS AND RECOMMENDATIONS

7.1 Summary and Conclusions
7.2 Recommendations for Future Work

LIST OF REFERENCES
# LIST OF TABLES

<table>
<thead>
<tr>
<th>TABLE</th>
<th>PAGE</th>
</tr>
</thead>
<tbody>
<tr>
<td>4.1</td>
<td>Nominal chemical composition of Al-2024 and Al-5052 (weight percentage)</td>
</tr>
<tr>
<td>4.2</td>
<td>Some thermophysical properties used in this work</td>
</tr>
<tr>
<td>5.1</td>
<td>Similarity comparison between the Johnson’s measurement and corresponding finite element model</td>
</tr>
</tbody>
</table>
# LIST OF FIGURES

<table>
<thead>
<tr>
<th>FIGURE</th>
<th>PAGE</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.1</td>
<td>4</td>
</tr>
<tr>
<td>1.2</td>
<td>6</td>
</tr>
<tr>
<td>3.1</td>
<td>21</td>
</tr>
<tr>
<td>3.2</td>
<td>31</td>
</tr>
<tr>
<td>3.3</td>
<td>31</td>
</tr>
<tr>
<td>3.4</td>
<td>34</td>
</tr>
<tr>
<td>3.5</td>
<td>34</td>
</tr>
<tr>
<td>3.6</td>
<td>40</td>
</tr>
<tr>
<td>3.7</td>
<td>42</td>
</tr>
<tr>
<td>3.8</td>
<td>45</td>
</tr>
</tbody>
</table>
3.9 Modified generalized theory ..................................... 45
3.10 Simplified sketch of the operation of the Trans-Varestraint testing .. 53
3.11 Solidification brittleness range for AISI 304, 321, 316 and 310 and aluminum alloys .................................................. 54
3.12 Cracking threshold strain curves in the weld metals of 2017 and 2219 Al-Cu alloys. ................................................................. 57
3.13 Effects of composition on the CST for the weld metals of Al-Mg and Al-Cu binary alloys. ......................................................... 57
3.14 Determination of local strain in MISO technique. ....................... 60
3.15 Ductility curves of different austenitic stainless steels and Inconel alloy. .......................................................... 60
3.16 Strain fields near the weld arc in an area of 18x42 mm ............... 65
3.17 Typical moving characteristics of weld edges during welding. .... 69
3.18 Comparison of deformation characteristics at the crack initiation point and a crack propagation point .............................. 70
4.1 Geometry of the workpiece and coordinate system used in developing the finite element analysis models ............................ 75
4.2 Schematic drawing of (a) the GTA welding process and (b) the weld pool showing the region of interest ............................... 80
4.3 Models of the liquid-solid region ........................................ 88
4.4 Constitutional supercooling in alloy solidification ..................... 89
4.5 Relationship between the degree of constitutional supercooling and the interface morphology during solidification of an alloy ........ 89
4.6 Simplification of the commonly observed columnar dendritic grains by cellular ones for the purpose of simulating the effect of latent heat release in the heat conduction analysis. ................. 96
4.7 The pseudo phase diagram for Al-2024. .................................. 98

4.8 Temperature dependent thermophysical properties of Al-2024 and Al-5052. (a) Thermal conductivity (b) Specific heat ...................... 104

4.9 A weldment is divided into two regions for change of initial temperature ...................................................... 119

4.10 Linear thermal expansion coefficients for 2024-T4 and 5052-O ........ 130

4.11 Temperature dependence of Young's modulus ....................... 132

4.12 Uniaxial stress versus true plastic strain curves. (a) 2024-T4 (b) 5052-O .............................................................. 133

4.13 General feature of the finite element mesh for thermal stressstrain analysis ........................................................ 137

4.14 The heat transfer analysis mesh corresponding to the thermal stressstrain analysis mesh shown in the previous figure .................... 138

4.15 The construction of a mechanical strain curve in the solidification temperature range. (a) Temperature history (b) Mechanical strain history (c) The mechanical strain curve ......................... 146

5.1 Dimensions of the specimens used by Matsuda et al. ............... 152

5.2 Locations and arrangement of the indentation marks for displacement measurement according to Matsuda et al. .............. 153

5.3 Finite element mesh used in the heat transfer analysis of 100 mm wide plate ......................................................... 158

5.4 Finite element mesh for the stress analysis of 50 mm wide plate as welding starts at x=30 mm. All elements are shown .............. 159

5.5 Finite element mesh for the stress analysis of 50 mm wide plate as welding starts at x=30 mm. Only the elements that are active at the instant of t=4.6 second are shown ......................... 160

5.6 An overview of the temperature field as the welding arc just passed by the indentation marks (t=5.0 second) .......................... 164
5.7 Temperature distribution around the weld pool (t=5.0 second) .... 165

5.8 Effect of different treatments of latent heat on the temperature
distribution in the weld pool and adjacent areas (t=5.0 second) .... 166

5.9 Comparison between the finite element results and the experiment
measurements. W=50 mm ............................................. 168

5.10 Surface of the weld pool at the moment when welding arc is
passing the connecting line between marks A and B ............... 169

5.11 Comparison of the “adjusted” finite element results with Matsuda
et al.’s results ......................................................... 171

5.12 Experimental set-up in Chihoski’s study. (a) edge weld; (b) butt
weld ......................................................................... 173

5.13 Chihoski’s deformation patterns for edge welds .............. 175

5.14 Chihoski’s deformation patterns for the butt welds .......... 176

5.15 The location of the 18x42 mm measuring area in relation to the
weld bead, weld pool and moving torch ............................... 178

5.16 Comparison of the maximum shear strains .................... 182

5.17 Comparison of the transverse strain fields ..................... 183

6.1 The hot strain versus threshold strain approach for quantitative
solidification cracking assessment .................................... 187

6.2 Transient temperature distributions in the Sigmajig test. 15V, 93A,
and 5 mm/second. Top: at 0.5 second; Bottom: at 1.0 second ...... 200

6.3 Transient temperature distributions in the Sigmajig test. 15V, 93A,
and 5 mm/second. Top: at 2.0 seconds; Bottom: at 3.0 seconds .... 201

6.4 Transient temperature distributions in the Sigmajig test. 15V, 93A,
and 5 mm/second. Top: at 4.0 seconds; Bottom: at 5.0 seconds .... 202

6.5 Transverse stress fields at 5 seconds. Top: no restraint; Bottom: top
6.6 Transient transverse hot strain distribution at 5 seconds. Top: no restraint; Bottom: top edge of the plate is fixed ................... 204

6.7 Transverse hot strain distribution at 5 seconds. Element rebirth technique is not used to properly model the solidification effect. The top edge of the plate is fixed. ................... 205

6.8 Effect of restraint on the hot strain distribution at 5 seconds. The necessity of element rebirth technique for properly modeling the effect of solidification process is also clearly demonstrated. ........... 207

6.9 Hot strain curves for 4 locations on the weld centerline. They have different distance, X, from the left edge of the plate where the weld starts. No restraint, TL=911K, Tc=865K, Ts=775K. ................... 210

6.10 Effect of the weld starting position on the hot strain curves. No restraint, TL=911K, Tc=865K, Ts=775K. ................... 212
NOMENCLATURE

BTR  Brittle Temperature Range
FEA  Finite element analysis
HAZ  Heat-affected-zone
MISO Measurement by means of In-Situ Observation
SEM  Scanning electron microscope

\( C_o \)  Alloy composition (wt%)  
\( C_p \)  Specific heat (J/k-g)  
\( C_s \)  Solute concentration of solid at the solid-liquid interface (wt%)  
\( D_L \)  Solute diffusion coefficient in liquid (mm\(^2\)/s)  
\( E \)  Elastic modulus (Young’s modulus) (GPa)  
\( G \)  Shear modulus (GPa)  
\( H \)  Specimen thickness (mm)  
\( I \)  Welding current (A)  
\( L \)  Volumetric latent heat (J/mm\(^3\))  
\( Q \)  Total internal heat generation rate (J/mm\(^3\))  
\( R \)  Grain growth velocity in molten weld pool (mm/s), or radius curvature of the bending block in Varestraint test (1/mm)  
\( S_{ij} \)  Deviatoric stress tensor (MPa)  
\( T \)  Temperature (K)  
\( T_i \)  Initial temperature (K)  
\( T_L \)  Liquidus temperature of an alloy (K)  
\( T_o \)  Reference temperature for thermal expansion coefficient definition (K)  
\( T_S \)  Solidus temperature of an alloy (K)  
\( T_{\infty} \)  Ambient temperature (K)  
\( V \)  Welding arc voltage (V)  
\( c \)  Solidification contraction (shrinkage)  
\( d \)  Crack width (mm)  
\( e_{ij} \)  Elastic deviatoric strain tensor  
\( f_E \)  Volume fraction of terminal eutectic product (%)  
\( f_s \)  Solid fraction in mushy zone (%)
$h$ Coefficient of heat convection (W/K-mm²)
$k$ Partition ratio
$l$ Crack length (mm)
$m_L$ Slope of liquidus temperature in a phase diagram (K/wt%)
$m_s$ Slope of solidus temperature in a phase diagram (K/wt%)
$p$ Hydrostatic pressure (MPa)
$q_{arc}$ Welding heat input (J/mm²)
$q_c$ Surface convection heat loss (J/mm²)
$q_i$ Internal heat generation rate due to solidification (J/mm³)
$q_r$ Surface radiation heat loss (J/mm²)
$r_b$ Characteristic radius of the welding arc column (mm)
$t$ Time (s)
$v$ Welding arc travel speed (mm/s)

$\alpha$ Thermal expansion coefficient
$\alpha_{TL}$ Thermal expansion coefficient for the reference temperature being $T_L$
$\varepsilon$ Emissivity of radiation
$\varepsilon_{th}$ Thermal Strain
$\varepsilon_a$ Augmented strain in Varestraint test
$\varepsilon_c$ Threshold strain
$\varepsilon_h$ Hot strain
$\varepsilon_L$ Longitudinal strain
$\varepsilon_T$ Transverse strain
$\varepsilon_v$ Volumetric mechanical strain
$\gamma_{SL}$ Liquid-solid interfacial tension (J/mm²)
$\gamma_{SS}$ Grain boundary energy (J/mm²)
$\gamma_{xy}$ Shear strain
$\eta$ Welding arc efficiency, viscosity of molten liquid (poise)
$\kappa$ Thermal conductivity (W/K-mm)
$\theta$ Dihedral angle
$\rho$ Density (g/mm³)
$\sigma$ Boltzmann constant
$\sigma_{ij}$ Stress tensor (MPa)
$\nu$ Poisson’s ratio
CHAPTER I
INTRODUCTION

1.1 Background and Research Issues

Welding is one of the most important and widely used joining and fabrication technologies in modern industry. Welded structures are superior in many aspects to riveted structures, castings, and forgings. It is for this reason that welding is widely used in the fabrication of buildings, bridges, ships, oil-drilling rigs, pipelines, spaceships, nuclear reactors, and pressure vessels. Today in the United States, over 50% of the gross national product is associated with the production of welded products.

Welding as a fabrication technique, on the other hand, presents a number of difficult problems to the design and manufacturing community. In a welding process, a local heat source is applied to the joint area. The joint area is first heated, melted, solidified and finally cooled to form a weld. The concentration of the heat source in welding results in a transient, nonuniform temperature field in the welded structure.

While welding joins the components of a structure together, the complex thermal cycles experienced in the vicinity of the weld often change the material
properties in this region. The thermal cycles also result in the build-up of thermal stresses and strains in the weldments. Last, but not least, weld metal solidification cracking may occur. The last problem, weld metal solidification cracking, is the subject of this research.

Since the early 1960s, there has been an increasing demand for stiffer, stronger, yet lighter materials in a diversity of fields including space, aeronautics, energy, and civil construction. Many materials, such as stainless steel, nickel based superalloys and aluminum alloys, have been developed with the emphasis on performance and reliability. Those materials usually involve complex phase transformations to achieve superior mechanical properties but are seldom selected for their weldability. This often leads to situations in which a material with superb properties can not be welded and fabricated with ease and satisfaction. A recent incident of such problems was reported by the U.S. Navy on August 1, 1991. Weld cracks in the hull and internal structures of the partially completed first Seawolf SSN-21 attack submarine will necessitate disassembling and rebuilding the $2 billion vessel. Defense Department officials said the problem is expected to delay delivery by as much as a year and cost tens of millions of dollars.

Weld metal solidification cracking has been a major, persistent problem in a variety of engineering alloys. It has been listed as one of the ten Grand Challenges for the 21st century by the welding community [1], even though it has been the subject of extensive studies over the past 50 years.
In practice, weld metal solidification cracking can vary from small discontinuities to major centerline cracks extending the full length of seam welds. According to their location in the weld bead, solidification cracks can be classified as center-line crack, shoulder or flare crack, radial crack, and crater crack [4]. Most of the laboratory weldability tests deal with center-line cracks.

It has been well recognized that weld metal solidification cracking has the same root as cracking found in ingot casting, continuous casting and near-net shape manufacturing. In fact, many of the early studies on weld metal solidification cracking were conducted by researchers who also investigated solidification cracking in other manufacturing processes. Solidification cracking has been referred to as “hot tearing” [3] or “hot-shortness”[4] in foundry terminology.

Factors affecting weld metal solidification cracking can be considered as either metallurgical or mechanical in origin [5]. The metallurgical factors relate to conditions of solidification of the metal, grain size, presence of low-melting eutectic films, etc.. The mechanical factors relate to conditions of stress-strain build-up in the weld metal during solidification as the result of transient nonuniform heating and cooling at various weld and near-weld positions. Over the past 50 years, it has been confirmed that the formation of cracks results from competition between the material resistance to cracking and the mechanical driving force during the course of solidification of weld metal, as shown in Figure 1.1. Essentially,
weld metal solidification cracking, like many other fracture problems, occurs when the mechanical driving force exceeds the material resistance.

Although the precise mechanisms responsible for weld metal solidification cracking are still not fully understood, studies have revealed the nature of the solidification cracking process. Because of the rapid transient heating and cooling characteristics of welding processes, the solidification of molten weld metal is a non-equilibrium process that leads to the formation of dendritic microstructure under the condition of constitutional supercooling. In the latter stage of

Figure 1.1 — General understanding of weld metal solidification cracking phenomenon
solidification, low melting-point constituents segregate between the dendrites and form liquid films or droplets over the interface of dendrites. This latter liquid and solid co-existence stage is often referred to as the "brittle temperature range" (BTR), in which the mechanical strength and ductility are very low. Mechanical tensile strains are developed at the same time due to weld metal solidification shrinkage, thermal contraction of the parent metal and external restraint of the welded structure. Cracking occurs when the mechanical driving force exceeds the material resistance to cracking. This concept is illustrated in Figure 1.2.

The prediction and prevention of solidification cracking would be conceptually straightforward, based upon Figure 1.2, provided that the material resistance could be properly and quantitatively measured in a laboratory test and that the mechanical driving force in a particular welded structure could be quantitatively evaluated. In fact, this very quantitative driving force versus resistance philosophy has been widely and successfully practiced in many other disciplines. For example, in linear elastic fracture mechanics, fracture toughness is regarded as the material resistance to brittle fracture and is obtained from some simple laboratory tests, whereas stress intensity factor is considered as the mechanical driving force and can be quantitatively evaluated for the concerned structures. The same can also be said for the conventional structural design in which stresses are quantitatively compared with material's yielding strengths.
Figure 1.2 — Schematic explanation for the occurrence of solidification cracking from a viewpoint of the relationship between ductility of alloy and deformation during welding
As for weld metal solidification cracking, the fundamental parameter representing material resistance is the ductility curve in the BTR, and the strains accumulated in the BTR as weld metal solidifies from the liquidus temperature are considered as the mechanical driving force [6,7,8,9,10].

Laboratory weldability tests have been a major means in the evaluation of weld metal solidification cracking susceptibility of materials. A variety of weldability tests has been devised and utilized to study the characteristics of the ductility curve in the brittle temperature range for various engineering alloys. Pan [6] and Prokhorov [7] were among the first to obtain the ductility curves in the BTR for aluminum alloys using a rapid tensile type test. They also observed the effect of deformation rate on the ductility curves. With the Trans-Varestraint test, in which the augmented strain was used as the indication of the strain experienced in weld metal, Senda et al. [8] and Arata et al. [9] systematically investigated the solidification cracking susceptibility of carbon steels, stainless steels, aluminum alloys, etc. Most of their results were presented in terms of the ductility curve characteristics. In the 1980’s, Matsuda et al. [10] published a series of basic research articles on weld metal solidification cracking phenomena based on a new technique, Measurement by means of In-Situ Observation (MISO). With the assistance of the MISO technique, Matsuda et al. were able to directly measure the ductility curves around the solidification crack tip on a very local scale (about 1mm in gauge length).
It is worth noting that, when a material was tested under similar welding conditions, the ductilities in the BTR measured by the MISO technique were often an order of magnitude higher than these measured by the augmented strain in the Trans-Varestain test, whereas the BTR itself often showed little change. In fact, the minimum ductilities obtained by the MISO technique were often quite measurable for alloys that are generally regarded as highly susceptible to weld metal solidification cracking (about 2% for stainless steel AISI 301S, for example).

All of these studies and other considerable evidence have confirmed that the existence of the BTR and the ductility curve in the BTR are intrinsic features of the solidification cracking phenomena. In other words, as long as there is a solidification process involved in a fusion welding process, there will exist a temperature range in which ductility is low. On the other hand, there is a small, but quite measurable ductility in the BTR for most engineering alloys. This latter fact ensures resistance to solidification cracking, enables one to measure the resistance to cracking, and suggests that the solidification cracking is preventable.

Since the 1950's, more than 150 separate and distinct weldability tests aiming at assessing the solidification and liquation cracking susceptibility have been developed, and completely new or modified tests are under continued development. However, the design and fabrication of solidification crack-free structures have not been completely successful despite tremendous effort. A critical
problem remains the lack of adequate systematic techniques to quantify the strain variations during the solidification process. There are many problems in appropriately quantifying the assessment of laboratory weldability testing results, which can be seen by comparing the results of Arata et al. [9] and Matsuda et al. [10]. In fact, this is one of the motivations that so many weldability tests have been devised. Also, it is extremely difficult, if not impossible, to reliably apply laboratory weldability testing results to "real" fabrication problems, since the mechanical driving force under real fabrication conditions has rarely been quantitatively determined. Therefore, today's knowledge about weld metal solidification cracking phenomena only allows us to relatively rank and compare the cracking susceptibility of different alloys under laboratory conditions. Yet such relative ranks of different alloys by different testing methods can be inconsistent. More importantly, laboratory weldability has had a varying degree of correlation with actual "field" weldability. It is almost impossible, by relying on the laboratory weldability test results, to confidently judge whether a particular material will definitely develop solidification cracking under a particular welding condition -- a very practical and important question. The weld metal solidification cracking problem has never been addressed from a design perspective.

Analyzing the thermal and mechanical responses of a weldment to the welding process has always been a challenging task. This is particularly true for the thermal and mechanical conditions in a solidifying weld pool and the
surrounding areas, owing to the complex solidification process involved. The challenges first come from the proper description of the complex physical nature of the fusion welding process. Modeling of welding related thermal and mechanical responses involves topics such as arc physics, fluid flow in the molten weld pool, nonlinear heat transfer, visco-elasto-plastic deformation, and transformation plasticity due to solid state phase transformation and solidification processes. Satisfactory solutions to the problem hence may not be obtainable from simple and analytical studies. Any successful description of the thermal cycles and stress-strain evolutions around a molten weld pool will thus be numerically and computationally intensive in nature. Despite of the fact that a great deal of progress in computational modeling of welding process has been accomplished in recent years by the explosive growth in computer capability and by the equally rapid development in numerical methods, the computational modeling must make some inevitable simplifications and assumptions pertaining to the welding problems. Most computational modeling has been focused on the residual stress and distortion of welded structures in which assumptions and simplifications on the material behaviors at the elevated temperature ranges were often made. Such treatments are capable of providing results accurate enough for the residual stress and distortion analysis, but obviously are not sufficiently accurate for computing the transient thermal stress/strain fields in the vicinity of a weld pool. While the residual stress and distortion computation can be easily
compared with the experimental results, the stress/strain patterns in the vicinity of a weld pool have rarely been studied by either theoretical/numerical or experimental approaches. So far, the only available literature on the deformation near the molten pool pertaining to the weld metal solidification cracking problem has been from a few experimental measurements.

1.2 Statement of Problem and Scope of Research

Today's major obstacle for the further advancement in solving the weld metal solidification cracking problem is related to the lack of quantitative description of the thermal and mechanical aspect of the problem. Also, the computational modeling techniques currently used in the thermal and mechanical analyses of welded structures are inadequate to describe the thermal and mechanical conditions experienced during the course of weld metal solidification cracking.

This dissertation aims at addressing the mechanical aspect of the weld metal solidification cracking problem by developing a systematic methodology that is capable of adequately quantifying the mechanical driving force for solidification cracking and the underlying finite element analysis (FEA) procedure. A commercial finite element analysis code, ABAQUS, is utilized for this purpose.

Based on an extensive literature review of the weld metal solidification cracking problem, this dissertation proposes a hypothesis to provide a quantitative
measure of the mechanical driving force and the material resistance to cracking, and the strain information pertaining to center line cracking is examined and obtained from the analyses of finite element models. Also, in this research the microscopic dendritic grain growth model is incorporated into the thermal model for a better prediction of the temperature fields in the vicinity of the weld pool. Element rebirth technique is used for appropriately modeling the effect of solidification process on the strain fields behind the molten weld pool. The computational modeling procedure and the strain calculation results are justified based upon the experimental data obtained from the published literature. Experimental work is not within the scope of this study.

The welding process simulated in this dissertation is autogenous gas tungsten arc welding (GTA), since most of the weldability tests are conducted with the GTA process. The computational models are developed for full penetration bead-on-plate welds on thin plate to simplify the study to two-dimensional transient nonlinear numerical problems. This reduces the computational time, yet two commonly used weldability tests – the Sagmajig test and the Houldcroft test – could still be adequately simulated. The methodology from these models should extend readily to three-dimensional problems as well.

Aluminum alloys 2024-T4 and 5052-O are used in this study, based on the considerations of experimental measurement results available to the present work, the alloys' tendency to solidification cracking, and the vast information
regarding the thermophysical and mechanical properties available for the alloys.

1.3 Anticipated Benefits

To date, the mechanical aspect of weld metal solidification cracking has seldom been studied. No quantitative methods are available for assessing the cracking phenomenon. This dissertation addresses this mechanical aspect of the problem by developing a quantitative assessment methodology. The anticipated benefits from this research include: (1) better understanding of the cracking mechanism; (2) design of new weldability tests that are able to truly and quantitatively reveal the material resistance to cracking; and (3) prevention of cracking by selecting appropriate materials and welding procedures in the design stage.

Solidification cracking also occurs in other metal processing related areas such as ingot casting, continuous casting, and near-net shape manufacturing. This research should also benefit these industries as well as welding.

Finally, the new finite element procedure developed in this dissertation should also be useful in modeling other welding-related thermal and mechanical problems. In particular, the element rebirth technique should improve the prediction of welding distortion problems whose accuracy is still quite inadequate.
Quantitative assessment of the mechanical driving force is vital to the further advancement of weld metal solidification cracking research and prevention. Such information not only provides much needed information about the stress-strain conditions for cracking in a laboratory weldability test but also as a bridge links the results of laboratory weldability tests and real welding fabrication situations. The primary objectives of this dissertation therefore aim at developing a methodology that is capable of quantitatively assessing the mechanical driving forces during the course of solidification cracking, and shedding new light on the prediction and prevention of the cracking problem. Effort will be devoted to two areas: to establish the finite element analysis (FEA) modeling procedures that are capable of obtaining the stress-strain fields in the vicinity of the weld pool; and to extract the appropriate strain information from the FEA models pertaining to the weld metal solidification cracking problem.

As for the finite element analysis, the modeling techniques and procedures commonly used for today’s welding simulation are inadequate for assessing the stress and strain evolutions in the solidification temperature range, primarily
because they can not appropriately deal with the effect of solidification process. Therefore, the objective is to examine the old modeling techniques and to develop new ones. In this regard, two new modeling techniques are proposed and explicated. The first new technique is to incorporate the microscopic solidification kinetics into the heat transfer analysis for improved modeling of the effect of latent heat of fusion. The second one is to develop an element rebirth scheme for the stress-strain analysis so that the effects of solidification process on the mechanical strain build-up in the solidified weld metal can be properly modeled.

The second effort of this dissertation is to obtain, from the results of FEA models, the mechanical driving force for the weld metal solidification cracking – the mechanical strain evolution in the BTR during the course of solidification. It involves construction of the mechanical strain evolution curves in the BTR, for the locations of interest, as they cool from the liquidus temperature. Finally, the quantitative assessment of weld metal solidification cracking is presented.

This dissertation consists of seven chapters. The first chapter provides essential information on weld metal solidification cracking phenomenon pertaining to the scope of this work. Introduced and discussed in detail in Chapter I are two fundamental concepts – the ductility curve and the mechanical strain curve in the BTR, and one fundamental hypothesis – the mechanical driving force versus the material resistance to cracking, of the weld metal solidification cracking phenomenon. They form the base of this dissertation. Also introduced in Chapter I are the
critical issues regarding today's research on the solidification cracking problem. This leads to Chapter II (this chapter), the objectives of this dissertation.

In Chapter III, prior research on weld metal solidification cracking is reviewed and discussed. The purpose of this extensive review is to provide a clear understanding of the cracking phenomenon so that the approach and methodology being developed in this dissertation can be fully appreciated. The review is also helpful in providing necessary information for the verification and explanation of the numerical analysis results. The nature of weld metal solidification cracking is reviewed first, followed by reviews of the parameters that measure the susceptibility to the cracking, and finally, the mechanical conditions involved in the cracking phenomenon.

In Chapter IV, the development of the finite element models is presented and discussed, after reviewing some relevant computational issues. The procedures for quantifying the strain field behind the weld pool and constructing the mechanical strain curve are also presented in the chapter.

The finite element modeling procedure developed in Chapter IV is verified in Chapter V, based on the quantitative comparisons of the predicted results from the finite element analyses with the experimental results reported in the literature. Since the objective of this work is to reveal stress-strain evolution during solidification, the comparisons are focused on the regions surrounding the molten weld pool. However, the comparisons are on a macroscopic level;
discussions on the validity of the strain calculation in the solidification temperature range is deliberately postponed to the next chapter.

Chapter VI first proposes the hot strain versus threshold strain hypothesis in order to provide a quantitative and computable measure for the mechanical driving force for solidification cracking. This hypothesis and the hot strain calculation are then debated and justified based on their ability to explain experimentally observed solidification cracking tendencies reported by other researchers. The magnitude of the hot strain curves is checked with experimentally measured local ductilities. The necessity of element rebirth is also demonstrated in this chapter. Conclusions and recommendations for future research are presented in the last chapter, Chapter VII.
CHAPTER III

WELD METAL SOLIDIFICATION CRACKING PHENOMENON

This chapter reviews the fundamentals of the weld metal solidification cracking phenomena. Instead of covering every aspect of the solidification cracking problem (certainly an almost impossible task), it tries to lay out the physical foundation upon which the methodology of this dissertation can be developed. The purpose of the review is to identify the key physical phenomena behind the solidification cracking so that the computational models can be properly established, and the results of computational analysis can be properly interpreted and appreciated. The literature survey is particularly important to this study: the validity of the computational models developed in this study will be verified with experimental and other results obtained from the literature.

Studies on weld metal solidification cracking have focused on three major topics: (1) the nature of the cracking phenomenon, the characteristics that distinguish it from other cracking problems, and the physical processes that provide resistance to cracking; (2) the measurement of cracking susceptibility, or the measurement of quantitative indices (parameters) that represent the material resistance to solidification cracking and distinguish one alloy or welding condition...
from the others in terms of the tendency to solidification cracking; (3) the determination of the mechanical conditions that cause solidification cracking. Of course, these three topics have been closely interrelated in the literature.

3.1 The Nature of Solidification Cracking

The study of solidification cracking began in the late 1940s. Early work was often related to solidification cracking in castings. Many fundamental concepts were proposed before the 1960s.

Pumphrey [4] proposed the Shrinkage Britteness Theory in 1948 based on studies of solidification cracking phenomena in castings and welds in aluminum alloys [11,12,13,14]. This theory proposed several important concepts regarding the nature of solidification cracking.

The first concept is the brittle temperature range (BTR). For alloys possessing a freezing range, the primary dendrites begin forming at the liquidus temperature and grow at the expense of the liquid during subsequent cooling. At a later stage in the cooling, the dendrites interlock and form a coherent network with the remaining liquid occupying the interstices. Once the network forms, the material is incapable of appreciable extension, and fracture takes place by the formation of a rift in the network. The strength and the ductility at cracking are not the same as those of the individual solid dendrites, since the dendrites are separated by a film of liquid. Therefore, there are very little plastic deformation (brittleness)
and strength. From the failure resistance viewpoint, the alloy is susceptible to cracking only while it is at a temperature in the BTR, i.e., at a temperature between that at which it gains coherence and the solidus.

Medovar [15] used binary phase diagrams to relate the constituent to the BTR (Figure 3.1), though he referred to the BTR a little differently as the "effective interval". For a eutectic system, solidification is divided into two regions. Under the equilibrium condition, the region of liquid/solid state (Region II) and the region of solid/liquid state (Region I) are separated by line b-δ' which is the initial temperature of formation of a continuous crystal network (the coherent temperature). Cracking occurs in Region I. The high solidification rate in the weld pool results in the suppression of diffusional processes and in a sharp intensification of dendritic micro-segregation. Medovar emphasized that this consequently shifts the cracking curve to the left, i.e., toward the region of lower concentrations of the element capable of giving a eutectic reaction in the molten pool (Figure 3.1, broken curves bα" and bδ"). The "effective interval" during welding is then considerably larger than that for equilibrium conditions. Medovar also pointed out, under conditions of welding, the tensile stress/strain could reach relatively higher levels than for the cases of casting, since the nonuniform heating and cooling of the weld pieces, i.e., the existence of a "rigid fixture", provides an additional significant contraction source. Both circumstances, the intensification of
Figure 3.1 — Correlation among hot cracking (hot shortness), quantity of eutectic and effective interval of solidification (BTR) (From Medovar [15])

- b-$\delta'$: Initial temperature of formation of continuous dendrite network under equilibrium conditions
- b-$\delta''$: Initial temperature under actual welding conditions
- b-$\alpha'$: Equilibrium solidus
- b-$\alpha''$: Non-equilibrium solidus
dendritic segregation and the considerably higher strains, accounts for the greater susceptibility of the weld to hot cracking.

The second concept of the shrinkage brittleness theory is the shrinkage brittleness, a measure of the inherent tendency to cracking of an alloy system. Cracking may be brought about by handling a casting or weld while part of it is in the brittle temperature range, but in general it is a result of the restraint of free contraction. The degree of cracking is determined by the amount of contraction which occurs in a particular part while it is at a temperature in the brittle range, and the greater the contraction the more severe the cracking. Therefore, the absolute tendency of a certain alloy to cracking can be described by the inherent tendency to cracking which is independent of all external factors except those (if any) which alter the extent of the brittle range. However, the actual severity of cracking in particular circumstances also depends on many other factors such as the temperature gradient in the surrounding solid metal, external constraints as well as the method of expressing the result of a test. It is thus apparent why a certain alloy may exhibit widely differing degrees of cracking in different conditions. In the first instance, the extent of the brittle range is likely to be influenced by any factor which modifies the course of freezing, and this will, as a result, alter the inherent tendency to cracking. Secondly, the actual degree of cracking, being related to the inherent cracking tendency, is susceptible to the influence of external factors peculiar to the practical conditions. According to Pumphrey, the
inherent tendency to cracking is referred to as the **shrinkage-brittleness**, whereas the actual severity of cracking in a test is referred to as the **rimosity**. The shrinkage-brittleness is given by:

\[
SB = 3\alpha \text{ BTR} 
\]

(3.1)

where \(\alpha\) is the average coefficient of linear contraction over the solidification temperature range. It is obvious that the inherent tendency to cracking depends upon both the average coefficient of linear contraction and the extent of the brittle temperature range. However, the latter is more likely to be influenced by various factors and is more practically important for the metallurgist to improve the resistance to cracking.

The third concept in the shrinkage brittleness theory is related to the material resistance to cracking. Pumphrey attributed the resistance to **healing** and **accommodation** processes. Accommodation is a process in which, by means of deforming plastically, dendrites move relatively to each other to accommodate the contraction strain without the formation of cracks \([11]\). Verö \([16]\), by studying castings of aluminum-silicon alloys, proposed the concept of healing, whereby incipient cracks are filled with liquid and their ill-effects thus overcome.

Physically, the shrinkage-brittleness is the amount of free volumetric contraction per unit volume during cooling through the brittle range. From the viewpoint of mechanical driving force versus material resistance, hence Pumphrey’s
shrinkage-brittleness should be regarded as a part of the driving force for cracking rather than the material resistance to cracking.

However, in the shrinkage-brittleness theory, the philosophy of driving force versus the resistance is not well conceived. In one instance, Pumphrey stated that the healing and accommodation processes can not alter the shrinkage-brittleness; they are regarded as a separate effect on the tendency to cracking, and have the result of modifying the relationship between shrinkage-brittleness and the actual severity of cracking in particular circumstances (the rimosity). He then continued, this argument is justified in that the shrinkage-brittleness is the strain applied to the dendrite network, and the healing and accommodation indicate the capability of the network to sustain the applied strain.

According to Pumphrey, the extent of the brittle temperature range as well as the healing and accommodation processes are influenced by any factor which modifies the course of solidification. They are primarily determined by the constitution of the alloy, but they are also dependent upon the amount and the mode of distribution of the liquid phase. The temperature at which a coherent network forms during the solidification of an alloy is dependent on the shape of the dendrites. If the dendrite arms are long and slender, they will make contact when the proportion of solid is lower than when they are short and thick. In other words, if a dendrite grows in such a way as to entrap liquid within it, it will come into contact with adjacent dendrites at an earlier stage in the solidification
process than when there is no entrapment, and the brittle range will be thus larger. The shape of the dendrites is dependent principally upon the temperature gradients within the cooling mass, which are in turn determined by the rate of cooling. The cracking tendency will be a minimum when the cooling rate is very low, which agrees with experimental observation.

Medovar [15,17,18,19] and Bochvar [20,21] extensively studied the effect of the eutectic upon the healing and dendritic morphology that affect the cracking susceptibility. Based on their observation and the others [22,23,24], the resistance of welds to cracking is determined not only by the magnitude of the BTR but also by the quantity of eutectic. In many cases, the initial increase in the composition of elements that form eutectic liquid leads to increased cracking tendency; but further increase will reverse the effect. The favorable effect of an increased quantity of eutectic upon the resistance of welds to cracking is explained by its dual action – not only by the healing of inter-crystalline interstices, but also by the refinement of the primary structure. The quantity of low-melting eutectic liquid depends upon the concentration of the eutectic forming elements as well as upon the solidification rate. For an alloy whose solidification is terminated with a eutectic reaction, the microstructure of the alloy then consists of two components – the pre-eutectic primary phase and the eutectic. If the concentration of the element capable of giving a eutectic reaction in the molten weld pool is not great, the crystals of the primary phase may grow anisotropically without obstacles, which
leads the formation of a coarse, columnar structure in the weld. As the temperature reaches the solidus, an insignificant quantity of eutectic liquid forced out by the front of growing columnar crystals becomes located only along the crystal boundaries instead of at the interstices of dendrites and can cause the appearance of hot cracks.

In case the concentration of the element capable of forming a eutectic is sufficiently high, the secondary and tertiary dendrite branches of the primary crystals cannot meet completely, and the interstices remaining between them are filled with eutectic liquid. The manifestation of the anisotropic crystal-growth is rendered difficult by the shortage of "building material". The filling of the interdendritic cavities is promoted by constituent segregation, as well as by the continuous supply of liquid metal, which is forced by the welding arc into the rear part of the weld pool, where the solidification process take place. As a result, instead of an coarse structure, there is formed a refined structure. Such a fine structure, owing to the considerable development of the interlock network of the crystals, is known to endow the weld with increased resistance to the action of tensile stresses/strains, i.e. to give it maximum resistance to hot-crack formation.

Although it is possible to increase the composition of eutectic forming elements to prevent cracking in welds, this approach often has a side effect: the decrease in the ductility of the weld metal caused by an excessive increase in the
quantity of brittle eutectic. According to Medovar [15], weld metal solidification cracking could also be affected by the use of preheat or heat generated during welding. This actually results from the changes in the quantity of eutectic and the effective interval of solidification.

Medovar also argued that the specific effect of increasing the quantity of eutectic upon the resistance of welds to solidification crack formation may not be discoverable by methods providing for the application of an external load to the specimen being welded. Under actual welding conditions, the eutectic liquid has enough time to fill all interdendritic interstices and eliminate all cracks. Under conditions of considerable external deformation, this may not occur, and instead of the elimination of the cracks, the latter may even be intensified as a result of the fact that the separation of the crystals may become too great. This question, however, can be answered only by investigations.

Based on the studies of hot tearing of copper-aluminum alloy and carbon steels, Pellini [25,3] proposed the Strain Theory of Hot Tearing. The investigation was conducted by simultaneous radiography detection of cracking and thermal analysis of solidification of castings. The time for hot tearing to occur was thus related to the solidification conditions. In addition, the development of strength and ductility through the period of solidification was determined by tensile type tests at various stages of solidification. For the carbon steels, the tearing over the whole range of carbon contents begins in the vicinity of the solidus temperature
of the Fe-C phase diagram. Pellini then proposed the "film stage fracture" concept. This concept actually originated from the solidus determination experiment where, when slowly heating to the solidus temperature, a thin but continuous liquid film is believed to be formed, causing the strength and ductility of the sample to be extremely low [26]. Pellini believes that this film stage is a fundamental phenomenon for an alloy system and uses it to explain the solidification cracking. The liquid film provides the metal condition which permits solidification cracking; however, the actual occurrence of the cracking is determined by mechanical factors inherent to the rate of extension per unit time imposed in liquid film regions.

According to the strain theory, the primary requirement for solidification cracking is the development of a hot spot which must extend to compensate for the contraction of surrounding solid regions. The extension forced on the hot spot results in solidification ruptures when the critical film stage condition is reached because of the severe concentration of the overall extension into these zones to produce high unit strains. The nature of strain distributions resulting from the extension of the hot zone at various stages of solidification are as shown schematically in Figure 3.2. While the metal is in the mushy condition which exists at temperatures considerably above the solidus (Figure 3.2A), the extension of the hot zone is distributed relatively uniformly since the interdendritic liquid areas are relatively wide and mass flow of the pasty mass results. Furthermore it
is possible that at this stage of solidification any tendency to separation can be countered by flow of feed metal (healing effect). As the hot zone approaches the film stage the extension in the hot zone necessarily is forced into the liquid films which have no appreciable strength compared to the adjoining solid dendrites. Figure 3.2B illustrates the high value of unit strains developed in the liquid films at this time. If separation does not occur at the film stage of solidification, solidification cracking is no longer possible (Figure 3.2C), because the strains in the hot spot area are distributed in a relatively uniform fashion through the coherent and ductile solid metal. Therefore, solidification cracking is most likely to occur during the film stage. It should be noted that the film stage in the strain theory is essentially similar to the BTR of the shrinkage brittleness theory. A major difference is that the term "mushy stage" used by Pellini actually means a macroscopic "mushy state" particular to the castings and is different from the microscopic "mushy zone" of the dendritic growth.

The strain theory emphasizes the effect of mechanical factors. The cause for cracking is mechanical, because the metal developing a condition of essentially continuous liquid films is fundamental to alloy systems and accordingly is not subject to control for castings; whereas the magnitude of the mechanical strain imposed on the liquid film could be altered.

Pellini also explicitly expressed that it is the strain rather than the stress that controls the occurrence of solidification cracking. The exact time during film
stage that the strain reaches a critical value determines the level of stress required to develop the strain. This can be understood by studying Figure 3.3 as deduced from the tensile type tests [3]. As the metal approaches complete solidification, the value of load required to develop separation increases; however, the critical extension required decreases. If cracking were stress controlled, it would have then developed at a higher temperature where the stress is minimum. However, the cracking begins at temperatures close to the solidus.

Borland proposed the Generalized Theory of Cracking [27,28] in 1960. This theory includes relevant ideas from both the shrinkage brittleness theory and the strain theory; it also suggests that for cracking to occur it is not a sufficient condition that a wide freezing range exists, liquid should also be present over a relatively wide temperature interval in a form that allows high stresses to be built up between grains. In other words, the tendency to cracking depends not only upon the quantity but also upon the distribution of the liquid. The latter can be characterized by the dihedral angle which is dependent upon the ratio of the solid-liquid interfacial energy to the grain boundary energy. A measure of crack susceptibility is the force (or stress) required to initiate or halt a propagating crack [29].

In agreement with others [30], Borland believes that the solidification process is divided into four stages, as sketched in Figure 3.4: (1) Primary dendrite formation; the solid phase is dispersed while the liquid is continuous. Both solid and liquid phases are capable of relative movement. Cracking is not possible in this
Figure 3.2 — Illustrating nature of strain distributions which exist during various stages of solidification in a casting system. (From Pellini [25])

Figure 3.3 — Mechanism of solidification cracking as deduced from casting studies. (From Pellini [25])
stage. (2) Dendrite interlocking: the material is cooled below the coherent temperature, both liquid and solid phases are continuous, but only the liquid is capable of relative movement and is able to circulate freely between the interlocking dendrites. Cracking is possible in this stage; but may be refilled and healed by a flow of liquid into the void so formed. Accommodation is possible to avoid cracking but the healing is dominant. (3) Critical solidification range; the temperature is cooled below the Critical Temperature so the solid crystals are in an advanced stage of development and the semi-continuous network restricts the free circulation of liquid between the dendrites. Relative movement of the two phases is impossible. Accommodation is the only way to prevent cracking, since healing cannot happen in this stage. Furthermore, the dihedral angle becomes important in this stage; a low angle will give a small solid-liquid contact area and will allow high stresses to build up on the solid bridges between grains. (4) Complete solidification; no solidification cracking occurs.

Borland [31] argued that cracking can occur in three different ways: (1) Necking of liquid films open to external (free) surfaces and subsequent void (crack) formation; (2) Separation of highly stressed (strained) thin liquid films separating adjoining grains. No solid-solid bonding occurs. Thick liquid films prevent cracking due to ability of liquid circulation to accommodate strains; (3) Breaking of solid-solid bonds in regions where the liquid coverage of grain surfaces is sufficiently extensive to allow breaking stresses to be imposed on the solid-solid
bridges. Extensive solid-solid bridging prevents cracking. It has been established that in most practical situations cracking takes place after a small amount of solid-solid bridging has occurred. If this is the case then it might be possible to stop cracking by adding, to the crack susceptible alloy, a very small amount(s) of element(s) which increases the solid-solid bonding in the critical solidification range. In the generalized theory of cracking, this increase in solid-solid bonding can be achieved by increasing the dihedral angle of contact between the liquid and the solid grains, as shown in Figure 3.5. When the dihedral angle is very small, the grain faces and edges are almost completely covered (wetted) by the liquid phase. If this condition exists in the weld pool over a relatively large temperature interval then it can be reasoned that cracks could form under adverse strain conditions, because only small areas of the adjacent grains will then be joined. The liquid phase will progressively occupy less of the grain faces as the dihedral angle increases. The liquid can exist only as a continuous network along the grain edges at $\theta = 60$ degrees. Further increase in the dihedral angle will cause the liquid phase to exist only around the corners of the grains. Therefore, the tendency of solidification cracking decreases as the dihedral angle increases. Arata et al. [32] measured the dihedral angles of the eutectic products in the weld metals of aluminum alloys and concluded that the resistance to cracking, as indicated by the minimum augmented strain required to cause cracking in
Figure 3.4 — Effect of constitutional features on cracking susceptibility in binary systems. (*From Borland [27]*)

Figure 3.5 — Effect of dihedral angle on distribution of liquid phase on grain corners, edges, and faces. (*From Borland [27]*
the Slow Bending Trans-Varestraint tests, especially for slow straining rate, is significantly increased as the dihedral angle increases.

It is well known [33,34,35,36] that if three phases meet along a common edge, the boundaries will reach local equilibrium at the angles required by vectorial equilibrium of the interfacial tensions. Assuming isotropic surface tension, the dihedral angle \( \theta \) for the case where two of the phases are the same (solid) is expressed as

\[
\frac{\gamma_{ss}}{\gamma_{sl}} = 2\cos\left(\frac{\theta}{2}\right)
\]

(3.2)

where \( \gamma_{sl} \) is the solid-liquid interface energy; \( \gamma_{ss} \) is the grain boundary energy. Under the equilibrium condition, the shape of the grains is greatly influenced by the dihedral angle. However, the quantitative relationship between the interfacial energy and the dihedral angle under the dynamic, non-equilibrium weld solidification condition still remains for further investigation. The dendritic growth in the weld pool indicates that the shape of the dendrite structure, at least in the early stage of the growth, is not solely governed by the interfacial energy but by the interplay of interfacial energy, diffusion and interface kinetics [36]. Besides, in some alloy systems, the interfacial energy is anisotropic [36]. Nevertheless, it is quite possible that the interfacial energy plays an important role in the final stage of the solidification where the cracking most likely to occur, according to the evidence collected by Borland [27].
Not only could the interfacial tension influence the dihedral angle and hence the shape of the last solidified liquid, but also it could affect the filling and healing effect of the last solidified liquid. In line with the shrinkage brittleness theory, Pumphrey [4] developed a simple model to explain the effects of viscosity and interfacial tension of the liquid on the formation of crack and healing. The basic assumption is that solidification contraction and the contraction of the surrounding solid metal will create a gap (crack) if the supply of liquid metal is not sufficient to compensate for the increase in volume of the gap. Hence, for cracking not to occur during welding, the liquid metal must be able to flow between the solid bridges of the network. For a gap of constant width, \( d \), and length, \( l \), which may correspond to the distance from the point of coherent temperature to the effective solidus in the weld puddle, cracking will take place when the pressure required to force the necessary quantity of liquid along the gap is greater than can be supported by the surface film, if the supply of liquid is driven by capillary flow. The condition for the absence of cracking is:

\[
\frac{dl^2}{d^2} \leq \frac{\gamma_{SL}}{4 \eta(1 + c)}
\]

(3.3)

where \( \gamma_{SL} \) is the interfacial tension, \( c \) the solidification contraction, \( \eta \) the viscosity of the liquid, and \( \dot{\gamma} \) is the rate of increase in width of the gap.

Although the gap used in Pumphrey’s model is a much simplified representation of the cracks in real solidification situations, the model does indicate that
increase in interfacial tension will decrease the possibility of cracking, which agrees with other cracking theories. Decrease in the viscosity of the liquid and the solidification contraction will have the same effect. Since the length $l$ is proportional to the brittle temperature range, the likelihood of cracking is proportional to the square of the brittle temperature range. Obviously, experimental observations support the effects predicted by the model.

Bochvar et al. [37] summarized the basic understandings of Russian researchers to the solidification problem, based on the work prior to 1960. Among other arguments, the following are worthy of mention in particular. (1) The theory for capacity to resist solidification cracking during welding and casting can be developed only on the basis of comparing the deformation processes with the processes of change in ductility (deformability). (2) In the effective crystallization temperature range, there is a well-defined dip in the ductility of alloys (the BTR). In the BTR, an alloy in the solid-liquid state has a small, but quite measurable ductility resulting from the plastic deformation due to the mutual displacement of grains and the deformation of grains themselves in the solid-liquid state. This fact ensures resistance to solidification cracking during welding and casting, and enables one to measure their resistance to cracking. Hydrostatic pressure and capillary forces may fill in (heal) cracks with liquid metal. (3) The formation of cracks depends on the correlation of three characteristics: the brittleness temperature range, the ductility in that range, and the unit of increase in the
elastic-plastic deformation with unit temperature decreases. This latter characteristic is referred to as the deformation/temperature ratio. Attempts at making a unilateral quantitative assessment of the resistance to solidification cracking, allowing for only one of the three characteristics indicated above, cause gross errors. In other words, the resistance to solidification cracking cannot be assessed quantitatively by taking these factors separately, although the severity of cracking can be indicated, to a first approximation, by the variation in one of the three characteristics in cases where the other two remain practically constant.

(4) In the general case, the capacity of a metal for resisting crack initiation, and its capacity for resisting the propagation of these cracks, are of different natures, since cracks are initiated in the effective crystallization temperature range, while their propagation may take place in the range or at lower temperatures. The length and width of cracks cannot be used as a measure of resistance to their formation.

Using the scanning electron microscope (SEM), Arata et al. [38,9], Matsuda et al. [39] and Borland [29] studied the fractographs of weld metal solidification cracks of stainless steels and aluminum alloys. The cracks were produced with the traditional Trans-Varestraint test, the Slow-Bending Trans-Varestraint test and the Houldcroft test. Under the high strain rate imposed by the traditional Trans-Varestraint test, SEM observation reveals that the appearance of the solidification crack surface is gradually changing as the temperature of crack front
decreases for both stainless steels and aluminum alloys. A typical solidification crack surface is shown in Figure 3.6. The surface can be roughly characterized into three types. The crack surface in the higher temperature zone near the end of the molten pool consists of many fine protuberances (Figure 3.6-b) which can be considered as the primary and secondary arms of columnar dendrites. In addition, tear-like protuberances were occasionally observed near the liquidus temperature as shown in Figure 3.6-c, corresponding to the liquid on the secondary dendrite arms. These regions are named as Type D, for dendritic surface. In the region at the lowest temperature, the crack surface is very flat and intergranular with a small number of holes which were in the liquid state at the moment of cracking. This is called the Type F as shown in Figure 3.6-e. The remaining liquid in type F region does not fully cover the grain boundary, but disperses and becomes more and more sporadic with the decrease in temperature. The type F is not the so-called liquid film stage, but liquid lake or liquid droplet stage. In the region at the medium temperature, the protuberances of secondary arms gradually become obscure and only grooves between primary dendrite arms parallel to the growing direction of the columnar crystal are noticeable (Figure 3.6-d). This region is named as Type D-F, considering it as a transition from type D to type F. The type D-F can be considered as the film stage.

Furthermore, metallographic studies [39] revealed that, in type F region of fully austenitic stainless steels, the grain boundaries gradually migrate, with the
Figure 3.6 — Feature of solidification crack surface in the weld metal of AISI 310S type stainless steels. (a) general appearance; (b) Type D; (c) Type D1; (d) Type D-F; (e) Type F (From Matsuda et al. [39])
gradual decrease in the amount of liquid phase at the original columnar solidification grain boundary. The crack passes through the migrating grain boundaries in the region of type F and through the solidification grain boundaries in the regions of type D and type D-F. The detrimental effect of phosphorus in fully austenitic stainless steels could be explained from the fact that P suppresses the lower temperature limits of types D, D-F, and F; the phosphide is not solidified in the type F region. On the other hand, increase in sulphur content in fully austenitic stainless steel hardly influences the ductility in the brittleness temperature range, since the sulphide is completely solidified in the lower temperature range of type F.

When the strain rate is low, the fracture surface may not have all three distinctive features as in the case of high strain rate of the Trans-Varestraint test. A study by Arata et al. [38] using Slow-Bending Trans-Varestraint test shows that the fracture surface morphology at the crack initiation point is dependent upon the temperature at which the crack initiates, independent of the augmented-strain and strain rate. Cracking occurs when the actual strain reaches the cracking threshold for the particular strain rate imposed as shown in Figure 3.7. The cracking threshold curve is obtained from the Slow-Bending Trans-Varestraint test whereas the crack surface morphology is based on the Trans-Varestraint test results. For example, the fracture surface is type D for a strain rate of 2.5% per second, and type D-F for 1.0% per second. Furthermore, when the solidification
crack continuously propagates forward to follow the welding arc as in the case of weld center line cracking of a "real weld", the fracture surface in the Slow-Bending Trans-Varestraint test indicates that the morphology during the forward propagation is the same as the morphology at the location of crack initiation which is the function of strain rate. This means that the solidification crack propagates at about the same temperature as that for crack initiation under the constant strain rate imposed during the Slow-Bending Trans-Varestraint test.

Figure 3.7 — Effect of strain rate on the separation temperature of cracking and crack surface morphology (From Arata et al. [38])
Borland [29] examined the fracture surface of solidification cracks in Al-Sn Houldcroft test specimens using SEM. For most of the crack propagation stage, the morphology of the fractured surface resembles that of the type D-F as mentioned above. In agreement with Arata and Matsuda, Borland concluded that solid-solid bridging did not take place or only occurred as isolated events during crack propagation. Therefore the mechanics of propagation in the Houldcroft test mainly involves the separation of liquid films at the rear of the weld pool or in the worst situation the breaking of small areas of solid-solid bonding. To the contrary, however, Borland observed more solid-solid bridging in the regions where a crack was initiated and halted.

With the aid of direct observation of solidification and solidification cracking behaviors during welding, Matsuda et al. [40] proposed the Modified Generalized Theory of weld metal solidification cracking in 1982. A high speed movie camera mounted on an optical microscope was used to characterize the initiation and propagation of the solidification cracking in a tensile hot cracking test under very high strain rate. The strain rate was about 130% per second near the center of the weld metal just behind trailing edge of the weld pool [10]. The BTR was converted from the crack length and also considered as the true solidification range. For all alloys studied under such a high pulling rate, the crack initiated at a position about 1/3 of the total crack length from the crack end that was close to the weld pool. The crack then propagated toward both lower and higher
temperature sides; the crack tip at the higher temperature side almost reached the solidification front and stopped first. Sometimes inflow of liquid metal into crack was observed, indicating the occurrence of healing. Examination of the fractured surfaces revealed the existence of types D, D-F and F regions with the crack initiation site located in the D-F region. Using a modified Scheil's equation of Brody et al. [41], Matsuda et al. showed that much of the solidification was achieved at the early stage of the solidification under the testing conditions. In fact, the solid fraction was more than 0.6 for a temperature drop of about 20°C from the initial liquidus temperature for most alloys studied. This and other studies of casting solidification [42] certainly showed the rapid development of columnar dendrites in early stage of solidification and the early attainment of the liquid film stage in which most of the grain boundaries are covered by a thin film of liquid and a certain degree of solid-solid bridging is formed due to the rapid growth of the secondary dendrite arms. The solidification and cracking features in the weld metal are illustrated in Figure 3.8 by Matsuda et al. and the Modified Generalized Theory was accordingly proposed as shown in Figure 3.9.

The stages 1 and 2 in Borland's Generalized Theory are shifted to much higher temperatures. These two stages together called the "liquid mass stage" in the Modified Generalized Theory. The coherent temperature separates the stage 1 and stage 2 whereas the stage 3 is separated from stage 2 by the critical temperature. The stage 3 is further divided into the "liquid film stage" and the "liquid
Figure 3.8 — Features of solidification, crack initiation and propagation during welding. (From Matsuda et al. [40])

Figure 3.9 — Modified generalized theory. (From Matsuda et al. [40])
droplet stage" which are represented by stage 3(h) and stage 3(l) in Figure 3.9, respectively. A solidification crack initiates in the liquid film stage and propagates toward both the liquid mass stage and liquid droplet stage. The cracking can not initiate in the droplet stage. The propagation in the liquid mass stage is somewhat slower than in the liquid droplet stage because of "healing" effect.

It should be noted that the Modified Generalized Theory was proposed based on the studies under rapid strain rate conditions. Cracking in practical welding situations may differ in some aspects. For example, the crack once formed in the practical case will propagate at the most susceptible stage, namely at the liquid film stage to follow the movement of welding arc. It is for this reason that most of the solidification crack fractographs show some dendritic features and that the flat feature (type F) corresponding to the liquid droplet stage is observed only at the location where crack initiates or arrests as mentioned by Borland.

Another difference between high speed test and practical welding situation where the deformation speed is moderate, is the contribution of grain boundary sliding to the total strain of weld metal within BTR. It is well accepted, in the fields of hot workability and creep below solidus temperature, that flow stress and ductility increase with the increase in strain rate, and that grain boundary sliding is one of major reasons for this strain rate dependence. Studies by Matsuda et al. [43] indicated that such strain rate dependence was also valid
within the weld metal in the solidification brittleness temperature range (BTR).
Matsuda \cite{44} also showed that, when the strain rate was as low as that under real
welding fabrication conditions – about 0.4 - 1.2\% per second – solidification grain
boundary sliding within the BTR was easily observed in the weld metals of AISI
310S austenitic stainless steel and Inconel alloy 600 for both tensile hot cracking
and the Slow-Bending Trans-Varestraint tests. The sliding phenomenon was not
observed at high strain rates of 21-24\% per second.

3.2 Measurement of Material's Resistance to Weld Metal Solidification
Cracking

This is no doubt that a great number of studies on weld metal solidification crack-
ing have been concerned with the measurement of cracking susceptibility of vari-
ous materials. So far, over 150 separate and distinct weldability testing
techniques have been developed primarily for this purpose \cite{45}. Consequently,
many indices and parameters have been utilized to indicate the cracking suscepti-
bility. These weldability tests vary widely in their approach and utility, but can
generally be classified as representative (self-restraint) and simulative (external-
restraint) test techniques \cite{46}.

Representative test techniques, such as the circular patch test, seek to repro-
duce the actual welding conditions as closely as possible in an effort to accurately
"represent" the situation of interest. These tests depend on self-restraint induced
by the specimen design and/or fixture. In most cases, these test techniques only provide a simple "go or no-go" solution for a specific material/process/restraint combination. They are ineffective in quantifying weldability among different materials because of the difficulties in isolating the material factor from the test results.

Simulative test techniques, such as the Varestraint test, attempt to simulate some aspect of the thermo-mechanical response of the material to the welding process. These tests normally involve the application of an external augmented strain or stress whose magnitude can be easily quantified as the "quantitative" cracking index. This approach allows the metallurgical and compositional factors associated with cracking to be isolated from the mechanical factors and permits their effects to be studied and quantified. As a result, simulative tests have been successful in providing a comprehensive order ranking of families of alloys or heats of a given alloy. However, the test conditions, especially the combination of thermal and mechanical history, are inherently different from actual welding conditions. As well, the assessment of these differences for essentially all the testing techniques have been very primitive, if it has been done at all. The lack of this assessment has provoked difficulties in standardizing the test procedures and resulted in poor reproduction and correlation. Discrepancies between test results and field experience are not uncommon.
It is difficult to list each and every one of the indices used in all the weldability tests. Nor it is necessary, for quite number of the indices can not truly represent the material resistance to cracking. They are often an overall indication of particular combinations of composition, process and mechanical conditions.

For example, in many weldability tests, the crack length and the number of cracks appearing in a specific sample is commonly used as a measure of crack susceptibility. Borland [29] pointed out that these cannot be used to quantitatively assess the crack susceptibility. Instead, Borland used the force required to initiate or halt a propagating crack as a measure of crack susceptibility. He experimentally measured this force and the elongation required for rupture using a weld simulated tensile test [31], and the results showed the existence of a strength and ductility dip in the brittle temperature range.

Pan [6] and some Russian researchers [7,47,48] were the first to propose and experimentally illustrate the existence of the ductility curves within the BTR, primarily by means of tensile tests during solidification of TIG arc spot welds. Figure 1.2 depicts the low ductility curve in the BTR during solidification according to Prokhorov [7] and Pan [6]. The properties of this ductility curve can be characterized by: (1) the magnitude of the BTR, (2) the minimum ductility ($e_{min}$) and (3) the shape of the curve. The deformation in this range is very localized by the circulation of liquid layers around the solid grains. The latter do not undergo
substantial changes in shape, moving relative to one another, due to the fact that the liquid and solid phases have very different resistances to shear deformation. This deformation is equivalent to the sum of the change in the shape of the work and the free contraction deformation (shrinkage). The plastic deformation accumulates throughout the entire BTR, starting at its upper boundary. The ductility curve is represented by the solid line whereas the actual deformation during solidification is shown by the dashed straight lines. In order to judge the susceptibility to solidification cracking of a alloy, the development of the plastic deformations must be compared with the deformability (ductility) of the alloy in the entire welding cycle. Solidification cracks will form if the magnitude of deformation of the solidifying metal exceeds the ductility curve in the BTR, that is, for the cases of straight lines 1 and 2. Prokhorov also emphasized the strain rate dependence of the ductility curve and, more importantly, the difference between the resistance of metal to solidification cracking and the strength of a component subjected to external forces. Failure to take the latter into account could result in fundamental mistakes. For example, the most widespread mistakes are the measurement of internal deformation (or the deformation at the location of cracking) simply by means of tensile strain gauges and the ignorance of the thermal/kinetic phenomena when assessing the resistance of metals. In other words, caution should be taken when experimentally measuring the resistance (ductility curve); the measuring parameters reflecting the resistance should be clearly defined first.
Since the 1970s, Japanese researchers have done a series of fundamental studies on the weld metal solidification cracking phenomenon. Using the Trans-Varestraint Test modified from the original Varestraint Test \cite{49}, Senda et al. \cite{8} first obtained the ductility curves in the BTR for various constructional materials, including carbon steels, low alloy high strength steels, austenitic stainless steels, aluminum alloys and brass. They agreed in principle to the concept of Prokhorov \cite{7} and Bochvar et al. \cite{37} for the conditions for the occurrence of solidification cracking, that is, the solidification crack will form when the deformation of the solidifying metal exceeds the ductility in the BTR. The essential detail of the Trans-Varestraint testing apparatus is shown in Figure 3.10. Welding begins at point A and stops at point C. When the welding arc passes through the point B, a pneumatically-actuated loading yoke bends the specimen downward suddenly to conform to the radius of curvature of the top surface of the bending block. The strain on the top surface of the specimen is represented by the augmented-strain calculated as follows:

\[
\varepsilon_a = \frac{H}{2R} \times 100\% \tag{3.4}
\]

where: $H =$ specimen thickness; $R =$ radius of curvature of the bending block. In general, the longest crack is seen along the weld center-line right behind the instantaneous solid-liquid interface at the moment of loading. The maximum crack length initially increases with the increase in the augmented strain and saturates with further increase in augmented strain. Because the strain rate applied to the
specimen is extremely high, about 60%/sec in the case of augmented strain of 2.5% [50], the strain can be considered as to be applied simultaneously. This allows easy measurement of temperature distribution in the vicinity of the molten weld pool at the instance of solidification cracking. Senda et al. argued that, when the actual deformation in the solid-liquid region exceeds the ductility in the BTR, the solidification crack will initiate and propagate, and the applied augmented strain reflects the ductility at the location where the solidification crack arrests. Since the temperature at such location is experimentally measured by means of thermocouples, the relationship between the ductility and temperature can then be obtained for a particular combination of alloy and welding condition, if a set of augmented strain is used to obtain various locations of the crack tips. Figure 3.11 is an example of the BTR for some stainless steels and aluminum alloys.

Assuming the testing plate perfectly conforms to the contour of the bending block of constant radius and simple beam bending theory applies, the average total strain at the top surface, where the solidification crack is most likely to form, will then be uniform and equal to the augmented strain calculated from Equation (2.4). In reality, however, the total strain may not be equal to the augmented strain and not uniform as revealed from the direct observation of solidification cracking by Matsuda et al. [43]. More importantly, it is not the total strain (the sum of mechanical strain and thermal strain) but the mechanical strain that causes the solidification cracking. Indeed, Senda et al. seemed to avoid the use of
Figure 3.10 — Simplified sketch of the operation of the Trans-Varestraint testing. (From Senda et al. [8])
Figure 3.11 — Solidification brittleness range for AISI 304, 321, 316 and 310 and aluminum alloys (broken line for each steel shows the critical strain rate for temperature drop. (From Senda et al. [8])
ductility when presenting their results. However, other researchers [39] later did refer to the ductility in terms of the augmented strain. There are other problems with the original Trans-Varestraint test. One of them is the strain rate used. It is too high in comparison with the average strain rates across the weld bead during normal arc welding which are less than 8% per second [51].

Arata et al. [32,50] later investigated the effect of straining rate on solidification cracking threshold strain of aluminum alloy and austenitic stainless steel weld metals, using the Slow-Bending Trans-Varestraint test. The resistance to cracking was represented by the threshold strain for cracking which is the minimum augmented strain required to cause the solidification cracking to initiate under a given strain rate. The threshold strain for cracking is different from the minimum augmented strain of the Trans-Varestraint test in that the former is a measure of crack initiation whereas the latter is for crack arrest. Furthermore, the average strain rate and average strain crossing the top surface of the weld metal in the Slow-Bending Trans-Varestraint test were directly measured with strain gauges and verified by means of a camera method adopted from Chernavski [52]. The transverse strain imposed across the weld metal in the Slow-Bending Trans-Varestraint test increases quite linearly with time but not with temperature drop. Solidification cracking always initiated at a particular location in the solid and liquid coexisting region. Arata et al. assumed that this location, at the instant the strain started to be applied, was located at the upper limit of the
BTR which was set to the liquidus temperature by Arata et al.. The threshold strain increases as the strain rate decreases for most aluminum alloys they studied. On the other hand, the austenitic stainless steels have very low threshold strain regardless of the strain rate imposed. In fact, this very low strain was close to the low limit of the test machine. The threshold strains as a function of temperature (strain rate) are shown in Figure 3.7 and Figure 3.12 for stainless steel 310 and aluminum alloys 2017, 2219 respectively.

Based on the results of Slow-Bending Trans-Varestraint tests, the solidification cracking susceptibility can be evaluated with a new index: the critical strain rate for temperature drop (CST). It is defined as the slope of the tangential line to the cracking threshold curve from the upper limit temperature of the BTR (the effective liquidus temperature according to Arata et al.) as illustrated as bold dashed lines in Figure 3.12. The concept of CST was first proposed by Senda et al. [8] for the ductility curves of Trans-Varestraint tests. It is more reasonable to use the cracking threshold curves obtained from the Slow-Bending Trans-Varestraint tests. Figure 3.13 shows the effect of alloy composition on the CST of Al-Mg and Al-Cu binary alloy weld metals. The tendencies for cracking in this figure remarkably resembles the well known qualitative tendencies for these alloys [14,53].

At the location where the temperature is the coherent temperature at the moment the augmented-strain begins to be applied (most likely the solidification
Figure 3.12 — Cracking threshold strain curves in the weld metals of 2017 and 2219 Al-Cu alloys. (*From Arata et al. [32]*)

Figure 3.13 — Effects of composition on the CST for the weld metals of Al-Mg and Al-Cu binary alloys. (*From Arata et al. [32]*)
crack initiation site), the strain rate is ideally equal to the applied augmented strain rate. However, at the locations in front of this location, the strain accumulated would be less since the strain cannot be accumulated in the liquid phase. At the locations behind of this location, the strain history is not the same as that of the applied augmented strain because the strain history at a particular point should be tracked from the moment that point passes through the coherent temperature. This may be one of the reasons concerning the observation by Arata et al. [32,38] that the crack length behind the crack initiation point was shorter than predicted from the cracking threshold curves for both aluminum alloys and austenitic stainless steels.

Matsuda et al. [10,43,54] developed the Measurement by Means of In-Situ Observation technique (MISO) to study the local ductility in the vicinity of the solidification cracking tips during crack initiation and propagation. With optical microscope and high speed movie camera, they recorded the dynamic movement of reference points consisting of ruggedness or structure on the surface of weld metal. The strain was calculated from the relative distance change of two reference marks that were 0.9 to 1.7 mm apart from each other as shown in Figure 3.14 by picture analysis method. These two marks were chosen so that they faced each other across the crack initiation site and parallel to the loading direction. The gauge length in the MISO technique was much shorter than those used in other tests that measure the ductility or strain variation in weld metal. It
therefore characterized the local ductility in the vicinity of the crack tip. In fact, further decrease in gauge length would scatter the measured data, due to nonuniform strain distribution at grain boundaries [10]. Examples of the ductility curve for some stainless steels are shown in Figure 3.15 [54]. They were obtained using tensile hot cracking test under high strain rate (130 % per second). Solid marks indicate both the crack initiation temperature and the minimum ductility required to cause cracking. The BTRs are clearly revealed. The upper limit of the BTR was set to the liquidus temperature measured by the thermal analysis in an electric furnace. The ductility at a given temperature was the strain when the crack tip propagated to the position of that temperature obtained using the method of Senda et al. [8]. In comparison with the results of Trans-Varestraint tests (Figure 3.11), the values of the minimum ductility measured with the MISO technique were generally about an order of magnitude higher. Matsuda et al. also showed that the minimum ductility measured by MISO technique combined with the Trans-Varestraint test gave nearly the same values as that measured with the MISO technique combined with the tensile hot cracking test. This suggests that the augmented-strain of the Trans-Varestraint test can not be used for the local strain in the vicinity of the weld pool, although it does represent the average strain across the top surface of the whole weld bead as indicated by Arata et al. [32].
Figure 3.14 — Determination of local strain in MISO technique. (From Matsuda et al. [10])

Figure 3.15 — Ductility curves of different austenitic stainless steels and Inconel alloy. (From Matsuda et al. [54])
Matsuda et al. also studied the effect of strain rate on the ductility for crack initiation of plain carbon steels, austenitic stainless steels and Inconel alloy by the MISO technique combined with the tensile hot cracking test and the Houldcroft-type cracking test. Three crosshead speeds were used in the tensile hot cracking test for each alloy. The first two speeds were 20 and 2 mm/sec whereas the third one was set to a value between 0.06 to 0.45 mm/sec to reveal the critical strain rate below which the solidification cracking will not occur, and the critical minimum ductility at this critical strain rate. The study showed that the minimum ductility for crack initiation had a dependency on strain rate irrespective of materials used. Contrary to the results of the Slow-Bending Trans-Varestraint test, the minimum ductility decreased with decrease in strain rate. The minimum ductilities under the highest strain rate reflected well the commonly accepted order of crack susceptibilities for the materials used and the effect of harmful elements such as S and P. However, under the critical strain rate, the minimum ductility was in the range of 1 to 2%, showing little variation to the composition changes of the alloys. The critical strain rate instead corresponded well with the common ranking of the cracking susceptibilities of alloys. In fact, low critical strain rate under constant minimum ductility means large BTR. More importantly, the critical strain rates from the tensile hot cracking test were very close to those from the Houldcroft-type cracking test and other self-restrained cracking test when measured with the MISO technique. Since the Houldcroft and other
self-restraint type cracking tests are generally considered to behave like the real welding fabrication conditions, it seems more reasonable to use the critical strain rate for comparing the susceptibility of solidification cracking under weld fabrication conditions. Matsuda et al. [44] attributed such a strain rate dependence to the effect of grain boundary sliding on total strain accumulation. However, there is another deformation mechanism at elevated temperature, i.e., the recovery process. Under certain circumstances, the recovery mechanism may be the controlling process over the grain boundary sliding mechanism [55].

3.3 Strain Development during Welding

In welding the processes of deformation and solidification take place simultaneously. Knowledge about the characteristics of the deformation at high temperature is essential to solidification cracking problem. The deformation provides the "driving force" for cracking, and in order to better comprehend the cracking mechanisms, it is necessary to know the deformation patterns under the "practical" as well as under the laboratory weldability test conditions. However, research on the deformation in the vicinity of the weld puddle has been difficult in nature and the success was limited.

In the past both theoretical/numerical and experimental research has been done to determine the magnitude and distribution of welding stresses and strains. The theoretical/numerical work has often been unsatisfactory because of
the complexity of the situation. Several factors contribute to this complexity: (1) the physical and mechanical properties of most materials at high temperatures, especially in the region of the solidus, are not well known, (2) these properties are often highly temperature dependent, and (3) heat flow theory is least accurate for the region nearest the weld, the most critical region for weld metal solidification cracking to occur. The majority of the numerical analysis has been on the residual stress and distortion of welded structures in which assumptions and simplifications on the material behaviors at elevated temperature ranges were made. Such treatments are capable of providing results accurate enough for the residual stress and distortion analysis, but obviously not enough for computing the transient thermal stress/strain in the vicinity of weld pool. The residual stress and distortion computation can also be relatively easily compared with the experimental results. On the other hand, the stress/strain patterns in the vicinity of weld puddle during welding have rarely been studied by either theoretical/numerical or experimental approaches. So far, the only available literature on the deformation near the molten pool pertaining to the weld metal solidification cracking problem is from a few experimental measurements and will be reviewed here.

Experimental measurements of the transient strain or deformation fields around weld pool have been a difficult task. Useful techniques have been limited to moiré fringe technique \cite{56,57,58} and indentation film recording technique
Holography also has the potential to measure the welding strains. The indentation film technique is simple in device and can measure the strain changes at locations very close to the solidification crack tip when combined with the MISO technique [10], but it only provides information at a small number of points. Moiré fringe technique, on the other hand, is more accurate in theory and is able to obtain the distribution of welding strains over a large surface area around the welding arc. However, moiré fringe technique can not measure the strains within the weld bead.

As an example of moiré fringe measurement, Johnson obtained the strain field around a moving arc for bead-on-plate welds of Al-Mg alloy using the projected-grating grid-analyzer technique [56,57]. High-frequency gratings (500 line/in) were etched onto the bottom of each coupon for strain measurement. High speed movie camera (16 frames/second) was used to record the moiré fringe patterns during welding. Assuming no out-of-plane deformation, small strain and quasi-stationary stress/strain state around the weld pool, the error in strain was about 0.003 (0.3%). Figure 3.16 shows typical distributions of the transverse normal strain ($\varepsilon_x$) and the in-plane shear strain ($\gamma_{xy}$). The transverse strains for points away from the weld centerline are positive (in expansion), but near the centerline a substantial amount of compressive strain occurs. This agrees with the results from Dantu [59] and Daytlov and Sidoruk [60]. Such a compressive effect becomes greater at high welding speeds. The longitudinal strain ($\varepsilon_y$) is
Figure 3.16 — Strain fields near the weld arc in an area of 18x42 mm (From Johnson [56])
generally quite uniform and slightly positive. The shear strains are substantial around the weld pool. The planes of maximum shear strain are oriented almost perpendicular and parallel to the weld bead. This indicates that the 2-dimensional models, either plane strain or generalized plane strain assumption, used in some theoretical/numerical analyses, are incapable of predicting the existence of such shear strains. Johnson attributed the formation of the shear strain to the asymmetry of temperature distribution about the X axis which is normal to the weld bead so that when plastic flow occurs, there is a net shifting of metal rearward. This shearing displaces the metal near the weld rearward.

Chihoski [58,61] also used moiré fringe method to study the distortion of edge welds and butt welds. Parallel lines of 50 lines/inch were scraped on black anodized 0.1 inch thick 2014-T6 aluminum alloy plates. Chihoski’s approach did not obtain the displacements of each point, so strains could not be calculated. It instead gave the distorted shapes of the weld coupons. Perhaps the most important part of Chihoski’s work [61,62] is that it provided a framework for rationalizing stress and strain in welds. By comparing the distorted shapes of an edge weld and a butt weld under identical welding conditions, the strain, and to some extent stress, in a butt joint can be rationalized and many physical phenomena can be explained. For example, Chihoski demonstrated that, when the welding speed is low, there exists a tensile transverse stress at the location just behind the solidified rear edge of the weld puddle, indicating the possibility of solidification
cracking. When the welding speed increases, such tensile stress zone becomes compressive so solidification cracking could be prevented. However, Chihsoki's work was not based on rigorous continuum mechanics and did not present stresses and strains quantitatively.

Using the indentation film technique, Matsuda [51] studied the moving characteristics of weld edges of aluminum alloys with or without solidification cracking. The measuring points were marked on the surface of bead-on-plate specimen with the indentation of Vickers hardness test before welding. The transverse movement was measured by the relative distance of two indentations which were about 1 mm away from the fusion lines, separated by and normal to the weld bead. The longitudinal displacement was measured by two indentations 5 mm apart from each other on the same side of the bead. The location variations of the indentations were recorded every 0.2 second by a camera with an accuracy of 0.01 mm. It should be noted that this method is not capable of measuring the shear strain due to the symmetric arrangement of the indentations with respect to the weld bead. Typical moving characteristics in both transverse and longitudinal directions are shown in Figure 3.17. Circle and square marks represent deformations in transverse and longitudinal directions respectively. The location of the weld puddle is also shown in the figure. Similar to Johnson's moiré fringe measurement[56,57], Figure 3.17 shows the longitudinal deformations are positive and almost constant for both cracked and crack-free beads, though it
is a little higher in the crack-free bead; the transverse contraction gradually develops as the arc moves away. On the other hand, when solidification crack forms, substantial transverse expansion is observed. In general, this is in agreement with Chihoski’s work \cite{58,61} for edge weld noticing that the cracked weld to some extent resembles the edge weld. However, there is a major discrepancy between Chihoski’s work and Matsuda’s. Chihoski’s results first showed a compression zone, even for the edge weld; then the compression gradually changed to the tension. Matsuda also revealed that increasing welding speed or decreasing coupon width, thickness or material hardness increases the magnitude but maintains the sign of deformation.

By comparing the deformation characteristics at the middle of cracked coupon that represents the crack propagation site and at the starting edge of the coupon representing the initiation site of solidification crack (Figure 3.18), Matsuda concluded that the amount of deformation and the deformation rate are higher at the initiation site than at the propagating crack tip under the quasi-stationary state; and the minimum ductility required for cracking is also higher at the initiation site.

It should be noted that the above measurements only give the total strains or displacements, rather than mechanical strains or stresses which are expected to be more significant. One of the questions that is essential to the solidification cracking problem yet remains unsolved is how to evaluate the mechanical strains
Figure 3.17 — Typical moving characteristics of weld edges during welding. (From Matsuda et al. [51])
Figure 3.18 — Comparison of deformation characteristics at the crack initiation point and a crack propagation point. (From Matsuda et al. [51])
or thermal stresses in the vicinity of the molten weld pool from the measured total strain values.

3.4 Closure

The weld metal solidification cracking phenomena have received considerable attention during the last four decades and remain an active area for both applied and fundamental research. The nature of the phenomena is such that they encompass a broad spectrum of engineering and scientific disciplines, including heat and mass transfer, solidification and crystallization processes, fluid mechanics, and solid mechanics.

The complex interdisciplinary nature of the weld metal solidification cracking phenomena has, unfortunately, led to a segmentation of reported results. While many theoretical and experimental investigations have been reported which deal with specific aspects of the phenomena, few have succeeded in providing a consistent framework which integrate all relevant descriptions of the phenomena for the quantitative prediction of cracking behaviors under either laboratory or manufacturing conditions. Nevertheless, the fundamental mechanism for solidification cracking and the existence of the BTR have been clearly revealed.
It is worth noting that most previous research has been aimed at elucidating the metallurgical processes involved. Although the findings from such research have often been quite useful, the fact remains that the metallurgical processes occur while the surrounding plate is undergoing rapid heating and cooling and a phase change (solidification) is taking place. These factors result in a stress-strain pattern which directly affects cracking in and near the weld. This pattern could be expected to vary according to the set of welding parameters used and the mechanical properties of the alloy. Research which does not take into account this mechanical situation is therefore bound to be of limited usefulness.

However, research on the deformation in the vicinity of the weld puddle has been difficult in nature and the success has been limited.
CHAPTER IV
DEVELOPMENT OF COMPUTATIONAL MODELS

This chapter presents the methodology that is used to obtain the mechanical strain curves in the solidification temperature range. The methodology is based on finite element analysis of the stress and strain fields in the vicinity of a weld pool. A commercial general-purpose finite element code, ABAQUS version 4.9.1, is utilized to conduct the intended analyses. The development of heat transfer models is discussed first, followed by the stress-strain analysis models. The chapter is closed with a discussion on the construction of the mechanical strain curves in the solidification temperature range, based on the finite element analysis of the stress-strain fields.

Since the general-purpose finite element analysis code ABAQUS is utilized in this study, it is not necessary to recite the detailed mathematical formulation and computer implementation of the finite element method itself, particularly those pertaining to ABAQUS. The reader is referred to ABAQUS Theory Manual [63]. Instead, this chapter will focus on another important aspect of the finite element analysis: the appropriate identification and description of the welding problems of the study in the form which is solvable by ABAQUS.
Autogenous gas tungsten arc welding (GTA) process is simulated since it has been used in most of the weldability tests. The finite element models are developed for full penetration bead-on-plate weld on thin plate, effectively reducing the modeling effort to two dimensional transient nonlinear problems. This would greatly reduce the computational cost, yet the two of the commonly used weldability tests – the Sigmajig test and the Houldcroft test – could still be properly simulated. The methodology resulting from these models should be able to readily deal with three dimensional problems as well.

Without losing generality, the coordinate system is fixed on the workpiece, the x-axis is along the centerline of the plate, and the y-axis along the edge of the plate, as shown in Figure 4.1. The welding arc starts at \( x=x_0 (\text{at } t=0) \) and moves in the positive direction along the x-axis.

Aluminum alloys 2024-T4 and 5052-O are used in this study, based on the considerations of experimental results available to the this work, the alloy's tendency to solidification cracking, and the information regarding to the availability of thermophysical and mechanical properties of the alloy systems. The nominal compositions of Al-2024 and Al-5052 are presented in Table 4.1 [64].

<table>
<thead>
<tr>
<th></th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
<th>Cr</th>
<th>Zn</th>
<th>Ti</th>
</tr>
</thead>
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<td>≤0.5</td>
<td>3.8-4.9</td>
<td>0.3-0.9</td>
<td>1.2-1.8</td>
<td>≤0.1</td>
<td>≤0.25</td>
<td>≤0.15</td>
</tr>
<tr>
<td>Al-5052-O</td>
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<td>≤0.4</td>
<td>≤0.1</td>
<td>≤0.1</td>
<td>2.2-2.8</td>
<td>0.15-0.35</td>
<td>≤0.1</td>
<td>-</td>
</tr>
</tbody>
</table>
Figure 4.1 — Geometry of the workpiece and coordinate system used in developing the finite element analysis models
4.1 Basic Considerations

The centerpiece of the methodology proposed in this dissertation is the finite element analysis (FEA) of the stress and strain fields in the vicinity of weld pool. Such stress-strain problems are in general within the confines of the thermal stress problem that is commonly solved based on the continuum solid mechanics.

Finite element analysis is essentially a tool to carry out the intended numerical modeling. The basic procedure for FEA modeling includes (1) identifying the governing physical phenomena or mechanisms of the problem, (2) describing these physical phenomena in mathematical (partial differential equation) forms, and (3) solving the partial differential equations, in accordance with the imposed boundary and initial conditions, by using the finite element techniques and procedures to obtain the approximate solutions to the problem.

It must be recognized that, for problems as complicated as the ones studied in this dissertation, the most important decisions are made before one writes down the first equation, namely, what to consider important and what to neglect if necessary. Not only are approximations made at the solution stage (as for the case of finite element method), but also judgements at the problem-setting stage.

The complex nature of welding poses many difficulties for successful modeling of welding related thermal and mechanical problems. From the numerical solution technique perspective, the severe material and geometrical
nonlinearities involved cause a great deal of computational cost in order to obtain reasonably accurate results. Although the cost of computing has been considered a deterrent in the past, the rapid evolution of powerful, low cost computer is changing this view. In reality, the greatest limitation has always been the knowledge of the physics behind the welding problems. For example, despite the fact that the energy transfer from the welding arc to the weld and the molten metal flow in the weld pool have shown significant influence on the shape of weld pool and the temperature distribution in the vicinity of the welding pool, it is impossible to treat them on a rigorous scientific basis.

It is hence not surprising that some simplifications and idealizations have been made, some factors been ignored, and some engineering judgements been used in the past when dealing with welding related thermal and mechanical problems. Undoubtedly, such simplifications and idealizations are required due to the limited knowledge of the physics, mathematics and computational modeling techniques. However, it must be pointed out that such simplifications and idealizations are case dependent. Suitable simplifications and idealizations to one specific problem may not be applicable to another. A typical example is the modeling of residual stresses versus the modeling of distortion of the same welded structure. It has been well documented \(^{65,66}\) that the simplifications made for the residual stresses analysis are often not good ones for the distortion prediction, and vice versa.
Since this dissertation deals with the temperature and stress-strain distributions in the solidification temperature range, a task that has rarely been studied in the past, it is necessary to reexamine the simplifications, idealizations and methodologies used in prior computational modeling studies, and to develop new ones to meet the requirements of this study. Specifically, two new modeling techniques are proposed here. The first one is to incorporate the microscopic solidification kinetics into the heat transfer analysis for improved modeling of the effect of latent heat of fusion on the temperature distribution in the vicinity of the weld pool. The second one is to develop an element rebirth scheme for the stress-strain analysis so that the solidification effect on the strain build-up in the resolidified weld metal can be adequately simulated.

Generally speaking, modeling of the thermal and mechanical fields during a welding process is a coupled problem in a sense that simultaneous solution is required. Because of thermal expansion and contraction, temperature changes always create the thermal stress and consequently the residual stresses and deformations in a welded structure. On the other hand, changes in stress-strain field will not play a noticeable role on the temperature field, unless one of the following two conditions hold: (1) there are significant dimensional changes, and (2) the heat generation due to the mechanical work is significant in comparison with other thermal energy input. As Hibbitt [67], Argyris et al. [68] and Mahin et al. [69] have pointed out, neither of these two conditions is true for arc welding, a
process concerned in this work. The coupling effect is basically one-way. Therefore, the heat transfer analysis and the stress-strain analysis are uncoupled in this study. The heat transfer analysis is performed first, independent of the stress analysis. The temperature history obtained from the heat transfer analysis is then used as the thermal loading condition in the stress simulation.

4.2 Heat Transfer Analysis

The objective of the heat transfer analysis is to obtain the transient temperature field during the course of welding.

4.2.1 Relevant Factors and Their Treatment in the Heat Transfer Analysis

Figure 4.2(a) depicts the gas tungsten arc welding process. The schematic drawing of the weld pool, indicating the region of interest, is given in Figure 4.2(b). In a GTA welding process, the welding arc transfers its energy to the surface of the workpiece beneath it, to the electrode by thermal conduction, and to the surroundings by radiation. As a result of energy transfer between the arc and the workpiece, a molten pool is formed and subsequently grows to a steady shape that follows the traveling welding arc. The thermal energy is further conducted away from the molten pool to the heat affected zone (HAZ) and the base metal. Any attempt to simulate this transient temperature field around the welding arc should therefore address at least the following relevant factors that govern the
Figure 4.2 — Schematic drawing of (a) the GTA welding process and (b) the weld pool showing the region of interest
development of this temperature field: (1) energy transfer from the arc to the surface of the workpiece, (2) convection and radiation heat transfer between the specimen and the atmosphere, (3) fluid flow and coupled conduction-convection heat transfer in the weld pool, (4) phase transformation, particularly the melting and solidification process at the solid/liquid interface, (5) fusion zone geometry, (6) realistic weld geometry, and (7) accurate thermophysical properties for the workpiece.

Of these factors, the following three need to have further detailed discussion here: the arc energy input, the fluid flow in the weld pool, and the solidification process. The thermophysical properties will be discussed in Section 4.2.3, Finite Element Implementation for heat transfer analysis.

**Arc Energy Input**

The formation of a welding arc itself presents a number of physical phenomena which, despite intensive investigation over several decades, have defied anything like complete understanding \cite{70}. From the heat transfer point of view, the welding arc in a GTA process is a gaseous electric discharge of extremely high temperature and flow velocity which conducts electric current and generates heat between the electrode and the workpiece. The heat generated in the arc is transferred to the workpiece in at least the following forms: radiation due to the extremely high temperature (as high as 20,000K for GTA process) of the arc, and
convection by the plasma jet flow [71]. Modeling of the welding arc itself has been a on-going research topic in the past few years. However, few attempts have been made to directly incorporate the arc physics models into the heat transfer analysis of the workpiece.

In theory, the heat generated by welding arc should be regarded as radiative and convective boundary conditions for the heat transfer analysis of the workpiece. It is, however, impractical to do so because of the lack of mathematical models that can adequately describe the complex phenomena involved in such a heat generation process. Instead, researchers in the past have used some simplified forms to treat the welding arc when the heat transfer analysis of the workpiece is concerned. Essentially, the welding arc is considered as a heat flux applied on the surface of the workpiece or in a volume internal to the weldment. One of them was the point source approach [72,73], in which the arc energy input was treated as an infinitesimally concentrated point heat flux. This treatment, together with the assumption of constant material properties and quasi-stationary state, makes it possible to obtain analytical solutions that are capable of describing with reasonable accuracy the temperature distributions in simple weldments, such as bead-on-plate welding. The accuracy of these analytical solutions is reasonably high in dealing with temperature changes in areas not too close to the welding arc. The accuracy generally drops considerably for the temperature distributions in the heat-affected zone and weld metal.
Better simulations of the temperature distribution in the vicinity of the welding arc have been obtained when the heat flux from the arc was modeled as a finite heat source and the temperature dependence of thermophysical properties was considered. However, such treatments necessitate the use of computational numerical analysis methods. For GTA process, almost all finite heat source approaches have assumed the welding arc to be a two-dimensional heat flux disk with a radially symmetric Gaussian distribution that is applied on the surface of the workpiece [74,75,76,77,78,79]. If the electrode is moving along the x-axis and in the positive direction of the x-axis (Figure 4.2), then the Gaussian distribution of the arc heat input takes the following form:

\[
q_{ar}(x, y, t) = q_0 e^{-\frac{3r^2}{r_b^2}} = \frac{3\eta V I}{\pi r_b^2} \exp\left\{-\frac{3\left[(x - vt - x_o)^2 + y^2\right]}{r_b^2}\right\}
\]

where \( V, I, v \) are the arc voltage, current and travel speed respectively, \( \eta \) the arc efficiency, \( r \) the distance from the center of the welding arc, \( r_b \) the characteristic radial dimensional distribution parameter that defines the region in which 95 percent (1 - \( e^{-3} \)) of the heat flux is deposited.

Experiments [80,81] have shown that, at least for stationary GTA welding process, the Gaussian distribution is indeed a quite good approximation of the
energy density distribution. However, the Gaussian distribution is basically an empirical curve fit to the experimental results. Variables that have been known to have strong influence on the energy density distribution of the welding arc, such as electrode diameter, electrode tip shape, the distance between the electrode tip to the workpiece, and so on, are not explicitly expressed in Equation (4.1). In fact the only quantity in the Gaussian distribution that could reflect their effects is the characteristic radius \( r_b \). But there are no existing models that relate \( r_b \) to these variables.

Nevertheless, for the commonly used welding parameters, literature have shown that the use of the characteristic radius is capable of providing good approximations to the arc energy density distributions for the heat transfer analysis of workpieces. Therefore, this dissertation will also use the Gaussian distribution to model the energy transfer from the welding arc to the surface of workpiece. Detailed modeling techniques will be presented in Section 4.2.2, Mathematical Formulation of the heat transfer model.

Fluid Flow in the Weld Pool

Another major concern in modeling the temperature distribution in the vicinity of a weld pool is the fluid flow in the weld pool. Research in modeling heat and fluid flow in weld pool has been very active and much progress has been made in the past decade [82,83]. However, many difficulties still remain for
modeling a moving weld pool in the GTA process, since most of research has been focused on the stationary weld pool and on the establishment of proper modeling techniques. It is therefore beyond the scope of this dissertation to conduct a complete simulation of heat and fluid flow in the weld pool, accounting for conduction, convection and electro-magnetic body forces in the molten weld pool; the role of surface tension active elements in the fluid flow pattern development; radiative, convective and evaporative losses at the weld pool surface; and heat conduction into the solid surrounding the weld pool. Instead of being explicitly simulated in this study, the effect of the fluid flow in the weld pool on the overall temperature distribution of the workpiece will be considered in the heat transfer models by means of effective thermal properties such as the effective thermal conductivity. Such an approach has been in principle used by many investigators in the past and the effectiveness of the approach has been well demonstrated [69,77,79].

**Latent Heat of Fusion**

The third major phenomenon that affects the accuracy of the temperature distribution prediction in the vicinity of weld pool is the solidification process. From the viewpoint of macroscopic heat transfer analysis, it relates to modeling the release and absorption of the latent heat of fusion during the melting and solidification of the weld pool. Obviously, such release and absorption are
related to the microscopic solidification process that determines the evolution and morphology of the solidified crystal grains. Previous studies \cite{84,85} have indicated that the latent heat of fusion highly influences the weld pool shape and size, as well as the temperature distribution in the vicinity of weld pool. Therefore the latent heat effect must be considered in this study.

There have been two approaches in the past to include the effect of latent heat of fusion in welding heat transfer analysis. The first one essentially assumes that solidification occurs at one temperature with zero solidification range \cite{86,76}. The release or absorption of the latent heat is modeled as a singularity in specific heat of the metal. While such a treatment is fine for the solidification process in which a microscopically planar solid-liquid interface exists, it fails to recognize the fact that the solidification of an alloy under most of practical circumstances, such as welding and casting, takes place over a temperature range encompassed by the liquidus and solidus temperatures of the alloy. In the second approach \cite{69,78}, the solidification temperature range is taken into account by distributing the latent heat as an appropriate step increase in specific heat over the entire solidification range. In other words, the latent heat is treated as a additive term to the specific heat over the range bounded by the equilibrium solidus and liquidus. This implies that the solidification rate with respect to temperature is constant. In fact, this is the standard treatment of latent heat in ABAQUS \cite{87}. 
Both of these two models deviate from the nature of solidification of alloys. Figure 4.3 illustrates the basic concept of microscopic solidification process in accordance with the solidification theory [36,88,89]. First of all, alloys begin the solidification process at the liquidus temperature and complete it at the solidus temperature, that is, there is a region (the mushy zone) over which solid and liquid coexist at a given time during solidification and the fraction of solid in the mushy zone gradually increases from zero at the liquidus to 100 percent at the solidus. Secondly, the solidification takes place in the mushy zone with temperature dependent solidifying rate, so the release rate of latent heat of fusion is not constant.

The existence of the mushy zone often results from the constitutional supercooling phenomenon (Figure 4.4). It determines the microstructural characteristics of the solidified metals. In general, constitutional supercooling, if only slight, gives rise to cellular perturbation of the crystal surface – but larger supercooling leads to fully developed dendrites (Figure 4.5).

Doherty [90] pointed out that the dendrite, by virtue of its extended surface, has a considerable increased surface energy compared to the equilibrium shape of the crystal and is therefore thermodynamically unstable compared with this equilibrium shape. The origin of the dendrite must therefore result from the growth is now universally recognized as arising in thermal and/or solute diffusion controlled growth process since the sharp leading tip of a dendrite allows
Figure 4.3 — Models of the liquid-solid region. The equilibrium melting range, $\Delta T_0$, does not, except for the lever rule case, correspond to the range, $\Delta T'$, over which the mushy zone develops. The dendrite tips need a certain undercooling which is determined by the stability of the tip. The dendrite roots will usually have much higher concentrations than $C_0/k$, due to nonequilibrium solidification. This often leads to interdendritic precipitation of eutectic phases of volume fraction, $f_e$, even if the composition is not on the eutectic tie-line. The volume fraction of solid, $f_s$, in the mushy zone will follow an S-shaped curve like that in the uppermost diagram. (From Kurz and Fisher[88])
Figure 4.4 — Constitutional supercooling in alloy solidification. (a) Phase diagram; (b) Solute-enriched boundary layer in front of a solidifying planar interface for a given growth rate; (c) Stable planar interface; (d) Unstable interface and the constitutional supercooling region. (From Kurz and Fisher [88])

Figure 4.5 — Relationship between the degree of constitutional supercooling and the interface morphology during solidification of an alloy. S, L, and M denote solid, liquid and mushy zones. (From Kou [81])
the loss by diffusion of either latent heat or rejected solute. Whether the dendrite will grow or shrink depends upon the interaction of the solute and thermal fields, upon liquid-solid surface energy, and upon interface kinetics.

Flemings [91] suggests that the transition from cellular structure to dendritic structure mainly depends upon the growth rate. When regular cells form and grow at relatively low rates, they grow perpendicular to the liquid-solid interface regardless of crystal orientation. In other words, the growth direction is controlled by the heat flow direction so that cells will grow in the direction of the maximum thermal gradient. When growth rate increases, however, crystallographic effects begin to exert an influence. The cell-growth direction deviates toward the preferred crystallographic growth direction (\(<100>\) for cubic metals); the cross section of the cell generally begins to deviate from its previously circular geometry to form the flange feature even the secondary arms. The driving forces leading to preferred dendritic directions are anisotropic surface energy, and interface kinetics.

The conditions required to initiate breakdown of a plane front in metals has been well established by the constitutional supercooling theory [92] and by the interface stability theory [93,94,95,96]. It is also known that the greater the degree of constitutional supercooling, the greater the tendency for a given material to switch from the cellular to the dendritic mode of solidification. However, for the complex solidification in an arc welding process, no theory is available that can
predict quantitatively the transition from the cellular to the columnar dendritic mode of solidification or the transition from the columnar dendritic to the equiaxed dendritic mode; nor there are any theories or models that can quantitatively relate the evolution of the dendritic crystals and their morphologies to the constitutional supercoolings and other related factors.

Since the present work concerns the stress and strain evolution within the solidification range, it is important to have a more realistic model for the release rate of latent heat of fusion in order to better simulate the temperature field in the solidification range. In this study, microscopic solidification kinetics is intended to be utilized for this purpose. The concept of incorporating the microscopic solidification kinetics is based on the work of Stefanescu [97] and Flemings [36]. Because no theories or models exist that can quantitatively describe the morphology of the dendrites and its relationship to the constitutional supercooling and other factors, simplified microscopic solidification kinetics for simple cellular morphology will be used. Detailed mathematical treatment will be presented in the following section.

4.2.2 Mathematical Formulation

Based on the discussions in the last section, the basic heat transfer models to be developed in this dissertation are assumed to be two dimensional heat conduction models with prescribed heat flux moving along the weld to simulate the
welding arc, and convection and radiation heat loss from the top and bottom surfaces of the workpiece. In addition, the heat loss from the edges of the workpiece is ignored, since the workpiece is thin.

In this section, the general mathematical approach is presented. Particular values of the variables will not be given here, unless they are necessary to explain the approach.

For the aforementioned two dimensional heat conduction models, all of the boundary conditions, i.e., the surface heat loss and the heat flux from the welding arc, can be included in the governing heat diffusion equation in terms of internal heat generation or loss (the body heat flux). In accordance with the coordinate system given in Figure 4.1, the governing equation can be then written as:

$$\frac{\partial}{\partial x} (\kappa \frac{\partial T}{\partial x}) + \frac{\partial}{\partial y} (\kappa \frac{\partial T}{\partial y}) + Q = \rho C_p \frac{\partial T}{\partial t}$$ (4.2)

$$Q = -\frac{q_c}{H} - \frac{q_r}{H} + \frac{q_{arc}}{H} + q_i$$ (4.3)

$$= -\frac{2h(T - T_\infty)}{H} - \frac{2\sigma \epsilon(T^4 - T_{\infty}^4)}{H} + \frac{q_{arc}}{H} + q_i$$

where $H$ is the thickness of the workpiece, and $Q$ is the rate of total internal heat generation which consists of four terms: surface convection heat loss, $q_c$, surface radiation heat loss, $q_r$, heat flux input from the arc as described in Equation (4.1),
\( q_{arc} \) and heat generation due to solidification (latent heat of fusion), \( q_i \). The explanations for other variables are provided in the NOMENCLATURE.

Therefore, the mathematical model for the problem in hand becomes a heat conduction model with insulated boundary conditions but with a internal heat generation \( Q \), which is capable of accounting for all the actual surface boundary conditions. Such a treatment of the surface boundary conditions is necessary because the two-dimensional element library in ABAQUS does not provide the corresponding surface boundary options.

As mentioned before, the latent heat of fusion highly influences the shape and size of the weld pool, as well as the temperature distributions in the vicinity of the weld pool. In this work, the effect of latent heat is incorporated into the governing heat conduction equation as a part of the internal energy generation. A simple microscopic solidification kinetics model is utilized for this purpose.

The concept of incorporating the microscopic solidification kinetics proposed in this dissertation is based on the work of Stefanescu et al. [97,98] and Desbiolles et al. [99] for heat transfer modeling of equiaxed dendritic microstructures in castings, and Flemings [100] for models of local solute-redistribution of columnar grains.

In agreement with Stefanescu and Desbiolles, the rate of latent heat release is assumed to be proportional to the rate of solid fraction change in the mushy zone and can be written as
\[ q_i = L \frac{df_s}{dt} \] 

where \( f_s \) is the fraction of solid and \( L \) is a proportional constant. \( f_s = 0 \) for \( T > T_L \) and \( f_s = 1 \) for \( T < T_S \). By definition, the proportional constant \( L \) is the total energy released in solidification per unit volume (volumetric latent heat), which is the product of the density and the latent heat. Obviously, the release rate of latent heat, \( q_i \), is negative during melting and positive during freezing.

As Stefanescu \(^{[97]}\) pointed out, in conventional heat transfer modeling, the fraction of solid, \( f_s \), can not be calculated from first principles for lack of available direct expressions, and the problem is circumvented, for example, by assuming that \( f_s \) (and, thus, the enthalpy) changes as a linear function of temperature within a temperature range (implicit enthalpy method and specific heat method). In Stefanescu's \(^{[97]}\) and Desbiolles's \(^{[99]}\) work, the fraction of solid, \( f_s \), was calculated by considering the solidification kinetics and incorporated into the macroscopic heat transfer modeling that is described by the foregoing heat conduction equation. Stefanescu proposed two possible approaches in calculation of \( f_s \) for casting processes: (1) from cooling curve data using computer-aided cooling curve analysis (CA-CCA), which is valid only for the situation of uniform temperature field in the casting, and (2) from nucleation and growth laws, which are based on the well known Johnson-Mehl relation, but can only deal with equiaxial grain growth in the casting.
However, the microscopic models for solidification kinetics used in casting simulations do not apply to the solidification processes during welding because of the epitaxial and competitive nature of crystal growth in the weld pool [101]. For the solidification process in a weld pool, there is no nucleation and the equiaxed dendritic (or equiaxed eutectic) growth is rare. In fact, columnar grain growth is the dominant process. Thus new approaches are needed to obtain the solid fraction $f_s$ during welding.

As an ideal simplification, one may first assume that the grains in the fusion zone (commonly in various columnar dendritic morphologies) could be represented by simple cellular cells, as shown in Figure 4.6. In this connection, Fleming’s local solute-redistribution equation [100] can be employed. It relates the $f_s$ to the liquid composition at the liquid-solid interface of the columnar dendritic grain and cooling rate, interface velocity, and so on.

The local solute-redistribution equation takes the following form:

$$C_s' = kC_0\left[\frac{a}{k-1} + \left(1 - \frac{ak}{k-1}\right)(1 - f_s)^{(k-1)}\right]$$  \hspace{1cm} (4.5)

where $a = -D_LG/m_LRC_0$; $C_0$ is the composition of the alloy, $k$ the partition ratio; $R$ is the growth velocity, $G$ the temperature gradient at the liquid-solid interface, and $D_L$ the solute diffusion coefficient in liquid.
In addition, Flemings\textsuperscript{[100]} showed "near zero constitutional supercooling" in the inter-cell liquid and thus implied that $f_s$ can be related to the temperature at that location through the liquid composition, i.e., the temperature at the location concerned can be assumed to be the local liquidus temperature corresponding to the liquid concentration at that position. The relationship between the solid-liquid interface composition $C_{s*}$ and temperature can be determined in the light of phase diagram, assuming local equilibrium exists. Thus the rate of solidification is expressed as:

Figure 4.6 — Simplification of the commonly observed columnar dendritic grains by cellular ones for the purpose of simulating the effect of latent heat release in the heat conduction analysis.
\[ \frac{\partial f_s}{\partial t} = \frac{\partial f_s}{\partial C_s} \frac{\partial C_s}{\partial T} \frac{\partial T}{\partial t} \] (4.6)

Now let us take aluminum alloy Al-2024 as an example to further the discussion. Al-2024 is essentially an Al-Cu alloy \cite{64} with additions of Mg and Mn. It has the nominal composition of 4.4% Cu, 0.6% Mn and 1.5% Mg. It has a terminal eutectic reaction under most solidification conditions. Treat Al-2024 as an Al-Cu binary alloy with the following properties: (1) having 4.5% Cu, (2) having the same eutectic composition \( C_E \) and maximum solid solubility \( C_{SE} \) as a true Al-Cu binary alloy indicated by the true Al-Cu phase diagram \cite{102}, but (3) keeping its actual liquidus and solidus \( (T_L = 911 \text{ K}, \text{ and } T_s = 775 \text{ K}) \), (4) having constant \( k \), \( m_L \) and \( m_s \), (in fact, those values for binary alloys systems Al-Cu, Al-Mg and Al-Mn are quite constant, as shown in their respective phase diagrams \cite{102}). Then construct a “pseudo” phase diagram, as shown in Figure 4.7. The values of \( k, m_L \) in Equation (4.5) thus can be obtained as \( k = 0.17018, m_L = -4.74 \text{ K/Wt\%} \).

The diffusion coefficient in liquid \( D_L \) of Al-Cu binary alloy is about \( 3 \times 10^{-9} \text{ m}^2/\text{s} \) \cite{88}. The growth rate in a weld pool should be no greater than the arc travel speed\cite{101}, which is between 5 to 12.5 mm/sec. Assume that the upper limit of the temperature gradient \( G \) in GTA aluminum weld metal is about 500 K/cm (the actual maximum temperature gradient calculated in this work is about 180 K/cm). Then \( a \) in Equation (4.5) is about \( 1 \times 10^{-3} \), negligible in comparison with unity. The local solute-redistribution equation (Equation 4.5) is simplified to the form of the Scheil equation:
\[ C'_i = kC_0(1 - f_s)^{(k-1)} \]  
(4.7)

And

\[ \frac{\partial f_i}{\partial C'_i} = \frac{1}{kC_0(1 - k)} \frac{1}{(1 - f_s)^{2-k}} \]  
(4.8)

\[ = \frac{(kC_0)^{1-k}}{1 - k} C_s^{2-k} \]

From the pseudo phase diagram, we have

\[ C_s' = \frac{1}{m_s}(T - T_s) + C_{SE}; \]  
(4.9)

\[ \frac{\partial T}{\partial C'_s} = m_s \]

Figure 4.7 — The pseudo phase diagram for Al-2024.
The release rate of latent heat of fusion in the solidification range is finally expressed as

\[ q_i = L \frac{\partial f_s}{\partial t} = L \frac{\partial f_s}{\partial C_s} \frac{\partial C_s}{\partial T} \frac{\partial T}{\partial t} = \frac{L(kC_0)^{1-k}}{m_s(1-k)} \left[ \frac{T - T_s}{m_s} + C_{se} \right]^{2-k} \frac{\partial T}{\partial t} \] (4.10)

Therefore the governing heat diffusion equation used in this study can be written as

\[ \rho C_p \frac{\partial T}{\partial t} = \frac{\partial}{\partial x} \left( \kappa \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( \kappa \frac{\partial T}{\partial y} \right) - \frac{2h(T - T_\infty)}{H} - \frac{2\sigma e(T^4 - T_{\infty}^4)}{H} + \frac{3\eta V}{H\pi r_b^2} \exp \left\{ -\frac{3[(x-vt)^2+y^2]}{r_b^2} \right\} + \left[ \rho C_p - \frac{L(kC_0)^{1-k}}{m_s(1-k)} \left[ \frac{T - T_s}{m_s} + C_{se} \right]^{2-k} \right] \frac{\partial T}{\partial t} \] (4.11)

Moving the latent heat term \( q_i \) to the right hand side of Equation (4.2), we obtain the heat diffusion equation that could be implemented in ABAQUS

\[ \frac{\partial}{\partial x} \left( \kappa \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( \kappa \frac{\partial T}{\partial y} \right) - \frac{2h(T - T_\infty)}{H} - \frac{2\sigma e(T^4 - T_{\infty}^4)}{H} + q_{arc} \] (4.12)

\[ = \left[ \rho C_p - L \frac{\partial f_s}{\partial T} \right] \frac{\partial T}{\partial t} = \left[ \rho C_p - \frac{L(kC_0)^{1-k}}{m_s(1-k)} \left[ \frac{T - T_s}{m_s} + C_{se} \right]^{2-k} \right] \frac{\partial T}{\partial t} \]
Or

\[ \frac{\partial}{\partial x} \left( \kappa \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( \kappa \frac{\partial T}{\partial y} \right) - \frac{2k(T - T_m)}{H} - \frac{2\sigma \varepsilon (T^4 - T_m^4)}{H} + \frac{q_{arc}}{H} = \rho C_p(T) \frac{\partial T}{\partial t} \]

\[ C_p'(T) = C_p - \frac{L(kC_0)^{1-k}}{\rho m_s (1-k)} \left[ \frac{T - T_s}{m_s} + \frac{2}{k-1} C_{SE} \right]^{\frac{2-k}{k-1}} \]

(4.13)

It should be noted that, when integrating Equation (4.10) over the entire solidification range, the enthalpy obtained only accounted for 91% of the latent heat of fusion. The reason is that neither the local solute-redistribution equation (4.5) nor the Scheil equation (4.7) can describe a complete solidification process unless the solidification range goes to infinity. It is often expected that some liquid will remain until an invariant temperature (e.g., eutectic) is reached and the remainder of the liquid then solidifies eutectically so that the Scheil equation can only be used before this terminal reaction.

In reality, there are indeed two solidification stages for Al-2024 [64]: (1) the pre-eutectic stage of columnar dendritic grain growth, and (2) the subsequent eutectic solidification stage that occurs at temperatures close to the solidus.

Now let us examine how well the latent heat release model can be applied to Al-2024. First of all, the eutectic solidification will start when the remaining
liquid reaches the eutectic composition, $C_E$, (the concentration of the solid at the liquid-solid interface reaches the corresponding composition, $C_{SE}$). Obviously, the volume fraction of the eutectic product, $f_E$, is expected to be the same as the volume fraction of this remaining liquid, $f_L$. Therefore, the volume fraction of eutectic, $f_E$, can be obtained from the Scheil equation:

$$C_{SE} = k C_0 f_E^{k-1}$$  \hspace{1cm} (4.14)

Together with the pseudo phase diagram for Al-2024, we have

$$f_E = \left(\frac{C_{SE}}{k C_0}\right)^{\frac{1}{k-1}} = 0.089$$  \hspace{1cm} (4.15)

Therefore, 8.9 percent of the latent heat of fusion is released during the eutectic stage of solidification, in agreement with the integrated value of 91 percent for the pre-eutectic solidification. In addition, since the amount of the eutectic product is small, the majority of the latent heat is released in the cellular grain growth stage. This suggests that the foregoing cellular dendritic growth model should be a good approximation to the solidification process in the weld pool of Al-2024 alloy as long as the release of latent heat is concerned.

Secondly, it is worth noting that the 8.9 percent of eutectic product predicted in present model is very close to the experimentally measured values of the eutectic fractions in the weld metals of Al-2017 ($f_E = 5.8\%$) and Al-2219 ($f_E = $...
7%) by Arata et al. [32] – two aluminum alloys with similar compositions to Al-2024. In another study, Chen [103] also measured the eutectic fraction of Al-2219. He obtained slightly higher values (7-10%). This justifies the treatment of the microscopic solidification process proposed in the present study.

Noting that Al-2024 is not really a binary alloy, the eutectic solidification in Al-2024 probably does not occur at the solidus, but most likely within a unknown temperature range just above the solidus. In this study, the enthalpy released by the eutectic solidification is then assumed to be uniformly distributed in a temperature range of 30 K right above the solidus and added to the enthalpy released during the dendritic growth stage. Such a more realistic treatment of the eutectic enthalpy release is also beneficial from the numerical computation point by avoiding a steep nonlinearity. Nevertheless, the effect of the eutectic solidification was small.

The above treatment worked equally successful when applying to the other aluminum alloy, Al-5052, used in this dissertation. The calculated eutectic volume fraction in this work was 1.1 percent whereas the experimental measurement by Arata et al. [32] showed about 1.8 percent in the weld metal of Al-5052.
4.2.3  **Finite Element Implementation**

The finite element implementation of the mathematical formulations discussed in
the last section is divided into three parts: geometrical discretization of the work-
piece, consideration of thermophysical properties and imposition of boundary
conditions. The detailed mathematical formulations of finite element method for
heat transfer analysis pertaining to the ABAQUS package is given in the
ABAQUS Theory Manual [63] and is not recited here.

Table 4.2 and Figure 4.8 provide the thermophysical properties for Al-2024-
T4 and Al-5052-O alloys used in this work. The basic thermophysical properties
required in heat transfer analysis are density, specific heat and thermal conductiv-
ity. In order to model the effect of latent heat of fusion, the solidus and liquidus
temperature of an alloy are also needed.

In this study, the materials are assumed isotropic and homogeneous, but
with temperature dependent properties. For welding analysis, the material prop-
erties are needed for a temperature range from room temperature to melting and
above. Such data are scanty and vary from source to source. Most of the data

<table>
<thead>
<tr>
<th></th>
<th>Density (g/mm³)</th>
<th>Latent Heat (J/g)</th>
<th>Solidus (K)</th>
<th>Liquidus (K)</th>
<th>Coherent (K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-2024</td>
<td>2.78x10⁻³</td>
<td>395.4</td>
<td>775</td>
<td>911</td>
<td>865</td>
</tr>
<tr>
<td>Al-5052</td>
<td>2.68x10⁻³</td>
<td>395.4</td>
<td>848</td>
<td>923*</td>
<td>900</td>
</tr>
</tbody>
</table>

Figure 4.8 — Temperature dependent thermophysical properties of Al-2024 and Al-5052. (a) Thermal conductivity; it is artificially increased in the liquid state to simulate the fluid flow effect. (b) Specific heat; the effect of latent heat is also shown. It is clear that the latent heat dramatically increases the effective specific heat in the solidification temperature range. Compared with the uniform latent heat distribution in the solidification range, the cellular grain growth model predicts that a major part of the latent heat is released in the earlier stage of the solidification. This is in fact in agreement with some experimental observations. (See Matsuda’s Modified Generalized Theory of hot cracking in Chapter III for references on the experimental observations.)
used here come from several publications by the Thermophysical Properties Research Center at Purdue University.

In this study, the density is considered to be constant over the entire temperature range concerned, as required by ABAQUS. Such an assumption does not depart from the reality very much, since changes in density with temperature are very small. The largest change, which occurs during the transformation from solid to liquid, is less than 6 percent \(^{(104)}\). The densities for Al-2024 and Al-5052 are taken as 2.78x10\(^{-3}\) and 2.68x10\(^{-3}\) g/mm\(^3\) \(^{(64,105)}\), respectively.

The temperature dependent thermal conductivity of Al-2024 up to its solidus temperature (775 K) is taken directly from the literature \(^{(106)}\). Temperature dependent thermal conductivity for Al-5052 was not found in literature. Its temperature dependency is then assumed according to its ambient temperature values \(^{(64,105)}\) and the temperature dependencies for Al-5%Mg and Al-7% Mg alloys \(^{(107)}\).

In order to account for the increased convection caused by the stirring effect of the arc in a weld pool, artificially high thermal conductivity for the molten metal in the weld pool has often been reported; many suggested values 2 to 5 times the conductivity at the solidus \(^{(77,86,69,108)}\). In this study, the approach by Mahin \textit{et al}. \(^{(69)}\) was adopted; the thermal conductivity is enhanced linearly from the solidus to the liquidus and remained constant above the liquidus. The conductivity value for the liquid is twice that at the solidus temperature. In comparison
with the otherwise sudden increase in conductivity at the solidus (or liquidus) temperature, this gradual change seems more appropriate in modeling the mushy zone and, at the same time, avoids the numerical difficulties associated with the sudden changes of material properties. The choice of low bound values for the effective thermal conductivity is due to the relatively high solid-state thermal conductivity of aluminum alloys.

Figure 4.8(b) shows the specific heat as a function of temperature for both Al-2024 and Al-5052. Although the temperature dependent specific heat for Al-2024 up to its solidus temperature is readily available [106], once again it could not be found for Al-5052. Judging from the fact that content of alloying elements has a weak influence on the specific heat [107,106] (for similar alloy system, of course), the specific heat of Al-2024 and Al-5052 are assumed to be the same.

The specific heat for liquid aluminum above the liquidus temperature is assumed as a constant value of 1.09 J/g-K according to Flemings [110] and Kurz [104]. It was then linearly interpolated between the solidus and liquidus for both Al-2024 and Al-5052.

The latent heat of fusion for pure aluminum is in the range of 378 to 415 J/g [110,104,111], and 388 J/g for Al-Cu and Al-Zn alloys [106]. In this work, a value of 395.4 J/g is used. The release rate of latent heat (the latent heat distribution over the solidification range) can be treated in three ways in ABAQUS, based on the mathematical formulations developed in the last section.
The first way is obviously to consider the latent heat as internal heat generation in the solidification temperature range as proposed in the last section (Equation 4.11). A user subroutine (HETVAL) in ABAQUS can be used for this purpose \cite{87}. However, the rate of temperature change needs to be calculated in the subroutine. Although it can be done by using the state variable in ABAQUS, it tends to be cumbersome. Besides, in order to improve the efficiency of its finite element solution procedure, ABAQUS \cite{63} assumes that the prescribed body fluxes (internal heat generation rate) have a weak temperature dependency so that the higher order terms in the Jacobian matrix contributed by the body fluxes can be omitted. Since the rate of internal heat generation due to the solidification strongly depends upon temperature, it is then better not to use this first approach to avoid unnecessary numerical errors.

The second way is to consider the latent heat as part of the specific heat in the solidification temperature range (see Equation 4.13). In other words, an effective specific heat is defined as \( C_p'(T) \) inside the solidification temperature range but \( C_p \) outside. Figure 4. 8(b) presents the latent heat of fusion in terms of effective specific heat, \( C_p' \), (Equation 4.13). The effective specific heat is utilized in ABAQUS input file via the command, *SPECIFIC HEAT. Such a treatment, however, causes a dramatic change in specific heat in the solidification temperature range, resulting in numerical instabilities and slow convergence rate during an analysis \cite{63}.
The third way is to utilize the *LATENT HEAT command, which evokes a special solution procedure in ABAQUS for dealing with the severe latent heat effect. It is basically to modify the corresponding Jacobian contribution of the internal energy (specific heat) to a secant term during the early iterations of the solution to a time step [63]. Therefore this third approach is used in this work.

In order to use the *LATENT HEAT command to model the cellular grain growth in the weld pool, the solidification temperature range is divided into many temperature intervals. Within each of the temperature interval, the energy released due to solidification (the enthalpy) is calculated by integrating the terms on the right hand side of equation (4.13) over the interval. The energy released during the final eutectic solidification (8.9 percent of the latent heat for Al-2024 and 1 percent for Al-5052) is added to the temperature interval of the last 30 K above the solidus to account for the energy release during the eutectic solidification. The calculated enthalpy is then input to ABAQUS using the *LATENT HEAT command.

Four-node bilinear isoparametric quadrilateral elements (element type: DC2D4) are used to discretize the workpiece. Since the heat transfer during arc welding is a moving boundary problem and the finite element mesh is fixed on the workpiece, the solidification front, in association with severe latent heat release, moves through the mesh with time and can cause discontinuous spatial temperature gradient within the elements. Simple finite elements, such as the
linear and quadratic elements used in ABAQUS, do not allow temperature gradient discontinuities within the element, although they do allow such discontinuities between elements, in the direction of the normal to their sides. The best we can do is to use a fine mesh of lowest order elements, thus providing a high number of gradient discontinuity surfaces. In ABAQUS, the lowest order two dimensional element for heat conduction is element type DC2D4, and we therefore choose this type of element. Smaller elements are used near the weld to account for the possible severe temperature gradient in those regions.

The treatment of boundary conditions is fairly simple; the arc energy input and surface heat losses due to convection and radiation are converted as a single internal heat generation term in this work, as discussed in the last section. In this connection, a user subroutine, DFLUX, is used to incorporate them into the ABAQUS program in forms of body heat flux. An arc efficiency, $\eta$, of 60 percent, typical for the GTA process, is used for arc energy input. The characteristic arc radius, $r_b$, is assumed the same as the diameter of the electrode (1.6 mm).

Because of the shielding gas flow in a GTA welding process, the convection heat transfer (film) coefficient $h$ is in theory a very complex function of the welding process and the distance from the electrode. Forced convection exists around weld pool, whereas natural convection takes place in the regions away from weld pool. In this study, the treatment of film coefficient $h$ is simplified. It takes a constant value of 84 W/m$^2$-K for the entire plate surface, which is quite high
for forced air flow on flat plate \cite{112}. Similarly, the radiative emissivity $\varepsilon$ is assumed to be 0.3, according to Incropera \cite{113}, for the surface conditions prepared by normal pre-welding cleaning procedures used for aluminum alloys. The simplified treatment for the film coefficient and radiative emissivity is based on the fact that the intense heat input from the electrode dominates the heat exchange between the workpiece and its surroundings. The effects of natural convection and radiation should be of secondary in nature, at least for the region surrounding the weld pool. Therefore, the use of true film coefficients, which is expected to vary between 84 W/m²-K and 5 W/m²-K (the values for natural convection), will not alter noticeably the temperature fields calculated based on the constant values. This has been confirmed in the preliminary studies. For the workpieces simulated in this study, the temperature fields for the models without convection and radiation heat losses were higher than those of otherwise the same but with the specified values for the convection ($h = 84$ W/m²-K) and radiation ($\varepsilon = 0.3$) heat losses. However, the differences were less than 4K.

4.3 Stress/Strain Analysis

As mentioned earlier, the welding thermal stress/strain problems in general have been treated within the confines of the continuum mechanics of solid. The thermal stresses and deformations are driven by the nonlinear distribution of
temperature field in weldment. In this study, they are treated as a transient static thermal stress problem, which neglects the inertia effect. The solutions to the welding thermal stress/strain problems thus need to satisfy, in addition to the initial and boundary conditions, the following within the solution domain (the welded structures): (1) the static equilibrium equation; (2) the kinematic equation which relates the displacements to the strains of a deformed body; and (3) the constitutive equation which defines the relationship between stresses and strains.

ABAQUS is a displacement based finite element analysis package (the displacements are the basic solution variables), in which the solutions are approximated by replacing the static equilibrium equation with a weak requirement – the equilibrium must be maintained in an average sense over a finite number of divisions of the volume of the body. ABAQUS uses the virtual work principle to obtain this weak form.

This study uses the nonlinear geometrical formulation option in ABAQUS to define the kinematic relations in order to deal with possible large strains and finite displacements that may occur in the vicinity of the weld pool. Such kinematic relations are mere geometrical relations, straightforward, and well defined. On the other hand, it is very difficult, if not impossible, to precisely describe the constitutive relationship suitable for welding stress-strain problems due to the strong temperature dependence and the wide temperature range which some regions of the weldment (heat affected zone and weld metal) experience. In this
study, the material constitutive behaviors are assumed as temperature-dependent but time (rate)-independent elasto-plastic ones, owing to the short time the material is exposed at the elevated temperature and the lack of appropriate elasto-viscoplastic models and their relevant experimental data. Although it seems that the assumption of time-independent plasticity is rather artificial, comparisons of the computational results obtained in this study with experimental measurements show that such an assumption is adequate for the problems concerned here. In addition, time-independent plasticity has been frequently documented for the computational analyses of stresses and distortions of welded structures; and the analyses with viscoplastic constitutive relations do not necessarily ensure more accurate results than those with the time-independent plastic relations if the viscoplastic models and their parameters can not be accurately specified (which is true for most engineering alloys under the conditions during welding).

The selection of appropriate constitutive models for wide temperature range is only part of the story that must be dealt with for welding related stress and distortion analysis. The other part relates to the effects of solidification process in the weld pool that does not need to be considered for many of the thermal stress problems based on solid continuum mechanics. This study will emphasize three of the phenomena related to the weld pool solidification process that have not been adequately recognized and appropriately reflected in the finite element
analysis models in the past, but are deemed to have significant influence on the stress/strain distributions in the weldment: the deformation in the weld pool, the solidification shrinkage, and the change of initial temperature in the weld metal.

4.3.1 The Effects of Solidification Process

Deformation in The Molten Weld Pool

The first phenomenon is that, owing to the nature of recrystallization, the resolidified weld metal, at the instant of transforming from liquid to solid, restores to its virgin strain-free state. In other words, all the possible deformation (strain) accumulated in the molten weld pool, caused by the fluid flow, should be "annealed" at the liquid-solid interface upon the recrystallization. In addition, the molten liquid in the weld pool should only exert minimum forces to the surrounding solid, thus having little effect on the stress-strain distribution in the surrounding areas. Therefore, except for the effect of the fluid flow in the weld pool on the shape of the weld pool, which could be adequately dealt with in the heat transfer analysis, the fluid flow behavior in the weld pool can be omitted in a finite element modeling of stresses and distortions in welding; the best way to deal with the molten weld pool in such an analysis is to simply exclude it from the solution domain.

Noting that the position of the weld pool continuously changes in order to keep up with the traveling electrode, the exclusion of the weld pool from the
solution domain has to be done in a dynamic and continuous fashion during the entire course of welding, a non-trivial task. This is perhaps the compelling reason that the majority of the documented finite element investigations on the stress and distortion of welds have considered the molten weld pool to be part of the solution domain and treated the entire solution domain – the base metal, heat-affected-zone, fusion zone, and the molten weld pool – as one piece of solid in accordance with the solid continuum theory (mostly temperature-dependent but rate-independent elasto-plasticity, or rate dependent elasto-viscoplasticity).

However, if the molten weld pool is considered to exhibit solid behavior the same way as the rest of the model, it is then inevitable for a continuum solid mechanics based finite element analysis procedure to allow some plastic strains being calculated inside the weld pool. Not only will these plastic strains then be accumulated in the weld metal as the molten weld pool solidifies and cools to the ambient temperature, but also the thermal expansion/contraction of the weld pool will cause further deformation in the vicinity of the weld pool, since as a pseudo-solid, the weld pool must deform compatibly with the rest of model. Therefore, the deformations in the weld pool have to be eliminated one way or another in a finite element analysis, in order to obtain a realistic solution to the stresses and distortions in a weld.

There have been some efforts in the past to minimize the accumulation of plastic strains in the molten weld pool. Leung et al. [114], for example, also using
ABAQUS to model residual stresses in a cross section of a single-pass-weld plate, used hybrid elements which treat the hydrostatic stress as an independently interpolated state variable with an assumed constant hydrostatic stress. They also limited the temperature histories of the molten material to the melting temperature in an attempt to limit the plastic deformations in the weld pool. Obviously, such an attempt could not produce a strain free weld pool, but rather would reduce the extent of deformation inside the weld pool.

In their attempt to analyze stress in a continuous casting process, Tszeng and Kobayashi [115] described another special procedure for the liquid regions. The concept is that hydrostatic pressure does not cause plastic deformation in metallic materials in which the plastic deformation process is primarily based on the dislocation gliding mechanism. In this connection, a Poisson’s ratio $\nu$ of very close to 0.5 is assigned artificially for the temperatures above liquidus temperature, $T_L$. This makes the stress state in the liquid region very close to the hydrostatic state (being incompressible for mechanical loading). This should exclude, in accordance with the incremental (visco-) plasticity, the onset of plastic deformation which must never appear in the liquid region. In order to avoid singularity in forming the stiffness matrix, the Young’s modulus $E$ is set to a very small number, instead of being exactly zero, for the temperature above $T_L$. It can be shown that, for an elastoplastic body (using the tensor notation),
\[ \sigma_{ij} = S_{ij} + p \delta_{ij} \]

\[ S_{ij} = 2Ge_{ij} \]

\[ p = 3K\varepsilon_v - \beta(T - T_i) \]

\[ G = \frac{E}{2(1 + \nu)}; K = \frac{E}{3(1 - 2\nu)}; \beta = \frac{E\alpha}{3(1 - 2\nu)} \]

where \( \sigma_{ij} \) is the stress tensor; \( S_{ij} \) the deviatoric stress tensor; \( e_{ij}^e \) the elastic deviatoric strain tensor; \( p \) the hydrostatic pressure; \( \varepsilon_v \) the volumetric mechanical strain, namely, the sum of elastic and (visco-) plastic strains; \( T_i \) the initial temperature at which the thermal strain vanishes. Therefore, when the Poisson’s ratio, \( \nu \), is set close to 0.5 and Young’s modulus, \( E \), is very small, the deviatoric stress \( S_{ij} \) can be suppressed while keeping the hydrostatic pressure finite. This treatment of Poisson’s ratio has also been reported for welding stress analysis [116].

The above treatment of the liquid region has a serious shortcoming for the cases involving the weld pool. Since the liquid region is assumed almost incompressible (\( \nu \approx 0.5 \)), i.e., the volumetric mechanical strain \( \varepsilon_v \) in the liquid region is very close to zero. The hydrostatic pressure inside the liquid region is then proportional to the thermal expansion/shrinkage (which is nothing but volumetric change), according to the third equation of Equations (4.16). This could produce a much higher hydrostatic pressure than it would physically exhibit in the liquid region, which is particularly true for the Tseng’s case in which the values of the Poisson’s ratio and Young’s modulus are set such that the bulk modulus \( K \) in the liquid region is kept approximately equal to that of the room temperature. It
should be noted that the high hydrostatic pressure assumption might be reasonable for the continuous casting process where high head pressure of the molten metal exists; it is however definitely inappropriate when a molten weld pool is concerned, for the actual hydrostatic pressure in the weld pool is usually small. Although such high hydrostatic pressure will not cause plastic deformation inside the liquid region as intended, it will certainly cause some unrealistic large plastic deformations in the surrounding solid region where the elastic incompressibility can not be assumed.

Neutron diffraction measurements were recently used by Mahin et al. [117], to provide experimental data for a detailed verification of a plane stress analysis of a traveling GTA weld. In that study, a parallel-sided full penetration weld was generated in a 4.7 mm thick 304L stainless steel plate in an effort to establish a quasi-two-dimensional plane stress condition. Comparison of the predicted residual elastic strain distributions with the neutron diffraction measurements indicated that the quasi-2D conditions assumed in the analysis were achieved only in the regions of the heat-affected zone greater than two weld widths from the weld fusion boundary. They attributed the discrepancies in the weld region to excessive drop-through of the weld and neglecting to reinitialize the strain history in the molten weld pool region at the time of resolidification.

The approach used in this study is to exclude the moving molten weld pool from the solution domain; therefore, the solution domain is composed only of the
continually changing solid part of the welds. The reason behind this decision is not only to completely eliminate the deformation in the weld pool, but also to appropriately model another solidification related phenomenon: the change of initial temperature for weld metal which will be discussed next.

The approach is based on an adaptive mesh generation scheme by which the elements whose temperature is above the coherent temperature (so they are inside the traveling weld pool) are simply removed from the finite element model and added back into the model after their temperatures drop below the coherent temperature. ABAQUS provides the capability of removing elements from the model and adding them back into the model by the *MODEL CHANGE input option. When elements are added back into the model, ABAQUS assumes that they are “annealed” – that is, to have zero strain, plastic strain, creep strain, etc., before they are added to the model. Since the weld pool is following the moving welding arc, such element removal/inclusion is dynamic, namely, following the location of the weld pool. In this dissertation, this procedure is referred as to the element rebirth technique.

Initial Temperature Change of Weld Metal

The second phenomenon caused by solidification process is the change of initial temperature for the weld metal of autogenous weld. When dealing with the thermal strain calculation, it is necessary to specify an initial temperature at
which the material is in a stress (strain) free state. Now divide the entire element meshes into two regions, as shown in Figure 4.9, based on whether the peak temperature exceeds the melting temperature. The first region (Region I) in the figure consists of the elements representing the base metal and the heat-affected zone where the peak temperatures are below the melting temperature; and the second region (Region II) encompasses those with peak temperatures above the melting temperature, which eventually become the weld metal upon the

![Figure 4.9](image)

Figure 4.9 — A weldment is divided into two regions for change of initial temperature. I: Base metal and HAZ where the peak temperature is below the melting temperature, the initial temperature is set to ambient temperature throughout the analysis. II: Weld metal where change of initial temperature is necessary: ahead the weld pool (IIa), the initial temperature is the same as in region I; behind the weld pool (IIb), the initial temperature is changed to the melting temperature. Note that region IIb continuously expands at the expense of region IIa as welding proceeds.
completion of welding. For the elements in the base metal and heat-affected-
zone, specifying the initial temperature is no problem: it is the ambient tempera-
ture before welding starts so that the elements in this region will subject to ther-
mal expansions when their temperatures are higher than the ambient tempera-
ture. On the other hand, two initial temperature values are needed for the ele-
ments in the second region for a GTA welding process: the first value is set to the
ambient temperature if an element is ahead of the weld pool (now belongs to
region IIa) so that it will expand as the welding arc approaches toward it. The
second value is set to the melting temperature for the same element after it enters
the weld pool (in region IIb), since, as mentioned earlier, the solidification pro-
cess introduces a stress-free state in the resolidified metal at the instant of
liquid-to-solid transformation. With its initial temperature being changed to the
melting temperature, the element, when representing the resolidified weld metal,
would be able to appropriately simulate the thermal contraction the weld metal
experiences when it cools down from the melting temperature. In other words,
for an element in the weld metal region (Region II), its initial temperature needs
to be changed from the ambient temperature to the melting temperature of the
alloy after it enters the weld pool so that it will expand before it experiences the
liquid-to-solid transformation (solidification) process, and contract afterwards. It
should be emphasized that regions IIa and IIb are continuously changing their
territories: region IIb grows at the expense of region IIa as welding proceeds.
The change of initial temperature is accomplished in this study by an element rebirth technique, in which two sets of nodes and elements with different initial temperatures are used to represent the weld metal. The first set is used for Region Ila and the other set for Region IIb of Figure 4.9. The element rebirth technique is discussed in detail in Section 4.3.3, Finite Element Implementation of the stress-strain analysis models.

**Solidification Shrinkage**

Solidification shrinkage is the third solidification related phenomenon that can have significant impact on the stress and strain fields of a weld. It is well known that most metals and alloys contract on solidifying; the volume change results from the liquid-solid contraction, which is in the range of about 3 to 6 percent for metals and much higher for refractory oxides \(^{110}\). For aluminum, this volumetric solidification shrinkage is 6.6 percent, which is equivalent to 2.2 percent of linear contraction. It should be emphasized that this solidification shrinkage is different from the thermal contraction caused by a decrease in temperature: solidification shrinkage still exists even for the solidification of a pure metal which has a zero freezing temperature range.

Process metallurgists have long since noticed the effect of solidification shrinkage on the soundness and surface shape of casting products; extra reservoirs of metal, termed *risers*, are used on castings and ingots to supply this feed
metal for compensating the solidification shrinkage. What is surprising is that many of the finite element analysis investigations, particularly those for welding processes, have neglected this shrinkage effect on the thermal stresses during the course of solidification.

The solidification shrinkage is not trivial at all; the thermal contraction due to solidification shrinkage is 150 percent of the total thermal contraction of aluminum as it cools from the melting temperature to room temperature (about 1.5 percent). It is therefore expected that the solidification shrinkage would play an important role in the stress/strain distribution, particularly in the vicinity of the weld pool, and the distortion of welded structures.

Insofar as the thermal stress modeling of an alloy is concerned, the effect of solidification shrinkage is phenomenologically equivalent to the thermal expansion/contraction due to temperature change. It causes mechanical deformations (often plastic deformation) inside the liquid region as well as in the other parts of the weldment, owing to the compatibility requirement of the continuum solid mechanics. In this study, the solidification shrinkage is thus assumed to be linearly distributed in the solidification temperature range and therefore can be effectively treated as an additional thermal expansion/contraction term caused by temperature changes.

In summary, an integrated approach to model systematically the effects of weld pool solidification process on the stress-strain distribution of a welded
structure is proposed and implemented in this dissertation. These effects are considered unique in that, in terms of computational procedure, they are rarely encountered in other thermal stress problems, even for the solidification stresses of castings and ingots. Although there are other ways to model these effects, this study decides to use an element rebirth technique, combined with modifying the thermal expansion coefficient in the solidification temperature range, to circumvent the modeling difficulties due to the solidification process in weld pool. Even though recent studies have documented that the element rebirth techniques are a natural way for modeling the metal deposition of gas metal arc welding and other welding processes involving metal transfer from the electrode to the weld pool [118,119,120], to the author’s knowledge, this study is the first to integrate all three aspects of the solidification process into the modeling of stresses and distortions of welds. In addition, the use of element rebirth technique in simulating the autogenous GTA welding process has not been documented. Furthermore, only one element mesh has been used in the past for the entire heating and cooling cycle of the weld metal, making it impossible to change the initial temperature for the solidified weld metal.

4.3.2 Material Properties

Like the thermophysical properties, the mechanical properties are assumed to be dependent on temperature. The mechanical properties are in general closely
related to the microstructures of the alloy, which in turn are influenced by the thermal history experienced. It is therefore quite possible for an alloy to exhibit different mechanical behaviors, depending upon whether it is during the heating or cooling periods of welding, as noted by Karlsson and Josefson \textsuperscript{[118]}, and Josefson \textsuperscript{[121]}. This also lead to the spatial dependence of material properties in which the weld metal, HAZ and base metal have different properties.

However, the aluminum alloys selected for this investigation, 2024-T4 and 5052-O, are simplified to behave both homogeneously and insensitively to the thermal cycles during welding, based on the following considerations. The homogeneity is due to the fact that the weld metal and base metal have the same chemical composition because of the autogenous GTA welding process used in this study. 5052-O is a non-heat-treatable alloy in its annealed state (no cold work hardening present); thus the thermal cycles of welding are expected to have minimum influence. 2024-T4, though a heat-treatable alloy whose mechanical strength can be enhanced by the precipitation process, is under the solution heat-treated and naturally aged to a substantially stable condition (the T4 process) before welding. The thermal cycle in welding can be fairly assumed to be similar to, and thus to produce similar mechanical properties to the T4 process. Such a simplification leads to an important advantage in the development of the finite element models: the standard material library in ABAQUS can be utilized.
The thermo-mechanical properties which must be specified are the thermal expansion coefficient, and the constitutive relations between stresses and strains. As mentioned earlier in this section, we assume that the constitutive relations can be described by temperature dependent but rate-independent elastoplasticity. In particular, the isotropic work-hardening incremental plasticity based on the Huber-von Mises yield surface and the associated flow rule is selected as the constitutive models used in the study. In this connection, the parameters necessary to describe the constitutive models are the elastic modulus (Young’s modulus), $E$; Poisson’s Ratio, $v$; virgin yielding stress; and true stress versus true plastic strain curves for the definition of the flow rule.

Finally, it is worth pointing out that ABAQUS uses linear interpolation to obtain material property information from the values that are provided in the input file, if the interpolating point is between two entries; the program assumes constant properties if the point is outside the range provided.

**Thermal Expansion Coefficient**

In ABAQUS, the thermal expansion coefficient is defined on an average basis, which is compatible with the definition used in most experimental documentation $^{107,122}$. According to ABAQUS, the apparent linear thermal expansion coefficient $\alpha$, with its reference temperature being at $T_0$ is defined as
\[ \alpha_{T_0}(T) = \frac{l - l_o}{l_o (T - T_o)} = \frac{\Delta l}{l_o \Delta T} \]  

(4.17)

where \( l_o \) is the reference length at \( T_o \) and \( \Delta l \) is the linear thermal expansion. Note that the notation \( \alpha_{T_0}(T) \) indicates that the linear thermal expansion coefficient, besides being a function of temperature, depends upon the choice of the reference temperature. For this reason, experimental measurements are often expressed in terms of the linear thermal expansion, \( l - l_o \), rather than presenting the thermal expansion coefficient directly. The thermal expansion coefficient is then calculated in accordance with the selection of the reference temperature.

From the thermal stress analysis point of view, the stresses in the solution domain are caused by temperature changes from a initial stress-free state (at which the temperature is referred as the initial temperature \( T_i \).) ABAQUS uses the following definition for the thermal strain at the current temperature \( T \):

\[
\varepsilon^th = \frac{l - l_i}{l_i} \equiv \frac{(l - l_o) - (l_i - l_o)}{l_o}
\]

\[ = \frac{l - l_o}{l_o} - \frac{l_i - l_o}{l_o} = \alpha_{T_0}(T - T_o) - \alpha_{T_i}(T_i - T_o) \]

\[
(4.18)
\]

where \( l_i \) is the initial length at the initial temperature \( T_i \). This equation indicates that ABAQUS only considers the change of thermal strains from the initial state to the current state.
With the above notion, it is obvious that the selection of the reference temperature (hence the reference length) is quite arbitrary and independent upon the initial temperature in a particular thermal stress analysis, making the selection of the reference temperature a matter of convenience in ABAQUS.

Based on the experimental data collected by Touloukian et al. [122], the linear thermal expansion is a very weak function of alloy's composition: 2024-T4 and 5052-O have almost the identical linear thermal expansion curves, which are also very close to that of pure aluminum. Therefore, this study assumes a single thermal expansion curve for both of the alloys up to their respective solidus temperatures (775K for 2024-T4 and 848K for 5052-O, see Table 4.2). Above the solidus temperatures, separate curves are used to include the solidification shrinkages.

As discussed in the previous section, the effect of solidification shrinkage can be incorporated into the finite element models by modifying the thermal expansion coefficient in the solidification temperature range. This study uses the following procedure for this purpose. First of all, the solidification rate is assumed to be constant within the solidification temperature range (the mushy zone). Therefore, the solidification shrinkage, which should be proportional to the solid fraction in the mushy zone, is linearly increased from zero at the liquidus temperature to 2.2 percent (6.6 percent of volumetric shrinkage) upon the completion of solidification (the solidus temperature). Secondly, it is
apparent that the thermal contraction of liquid metal does not cause any mechanical stress/strain in the solid portion of the model, provided that the liquid metal can flow freely. On the other hand, it does contribute to the development of the mechanical stress/strain once a solid network of dendrites is formed to prohibit the free flow of the liquid metal. Therefore, the solidification shrinkage can be effectively neglected in the upper portion of the solidification temperature range where a solid network has not formed. Based on these considerations, only the solidification shrinkage and thermal contraction due to temperature change below the coherent temperature (the temperature at which solid network of dendrites is formed) are included in the thermal stress models. According to Matsuda’s study [40] on the GTA weld pool solidification of plain carbon and stainless steels, as well as some other studies on aluminum alloy castings quoted by Matsuda, the solid fraction at which dendrites begin to form network is 0.31, the solid fraction above which the interdendritic liquid can not flow is 0.67, and the strength during solidification originates at a solid fraction of 0.67. Therefore, this study defines the coherent temperature as such that it separates the solidification temperature range at one third (1/3) from the liquidus temperature. The coherent temperature is assumed to be 865K for Al-2024, and 900K for Al-5052.

Based on the above discussion, we now use the effective linear thermal expansion to calculate the thermal expansion coefficient that is needed in the finite element analysis. The effective linear thermal expansion curves for both 2024-T4
and 5052-O are shown in Figure 4.10(a). Note that the reference temperature is set to 273K in the figure, following the convention used by Goldsmith et al. [107] and Touloukian et al. [122]. The curves can be divided into three parts: (1) for temperatures below the solidus temperature, they are simply equal to the linear thermal expansion due to temperature change; (2) between the solidus and coherent temperature, the sum of solidification shrinkage and thermal contraction due to temperature change; and (3) no change in thermal expansion/contraction occurs above the coherent temperature.

For mathematical convenience, we choose the liquidus temperature as the reference temperature for calculating the thermal expansion coefficient. The advantage is that the thermal expansion coefficient for temperatures above the coherent temperature will be zero, resulting in absolutely no thermal expansion or contraction being calculated in the finite element analysis. Hence,

\[
\alpha_{T_l}(T) = \frac{l - l_i}{l_i(T - T_l)}
\]  

(4.19)

and

\[
\varepsilon^{th} = \frac{l - l_i}{l_i} = \alpha_{T_l}(T - T_l) - \alpha_{T_i}(T_i - T_l)
\]  

(4.20)

The thermal expansion coefficients are shown in Figure 4.10(b).
Figure 4.10 — Linear thermal expansion coefficients for 2024-T4 and 5052-O. (a) Effective linear thermal expansion as the reference temperature being set at 273K; note that the solidification shrinkage is added in the solidification range, and the expansion levels off above the coherent temperatures (b) Thermal expansion coefficient as the reference temperature being set at the liquidus temperatures; the values of the coefficient above the coherent temperature are zero.
Poisson’s Ratio

The Poisson’s ratio has a weak dependence upon temperature \cite{116}. It is thus assumed to be a constant throughout the entire temperature range for both alloys. This study adopts the value of 0.34 as provided by Dally et al. \cite{123}.

Young’s Modulus

The Young’s modulus strongly depends upon the temperature, but not so upon the variations in alloying elements \cite{105,124,125}. Thus this study assumes both alloys having the same Young’s modulus up to about 650K as documented in the literature \cite{105,125} (Figure 4.11). No data are found above this temperature. The Young’s modulus is thus assumed to decrease linearly to a very low value (0.5 GPa) at the solidus temperature, further decrease linearly to an even lower value (0.05 GPa) at the coherent temperature, and to hold constant throughout the temperature range above the coherent temperature, to approximate the near zero strength of the molten weld pool.

Stress versus Plastic Strain Curves

To characterize the strain hardening behavior represented by the flow rule of the plasticity, uniaxial stress versus plastic strain relationships are needed for various temperatures between the room temperature and the liquidus temperature. The stress versus strain curves are obtained for 2024 from several sources
[105,126,127], and re-plotted in Figure 4.12(a) in terms of the stress versus true plastic strain as required in ABAQUS. Since no stress-strain curve data are available for temperatures above 650K, they are linearly interpolated based on the data [105,125] of the yielding strength, ultimate strength, and ductility (reduction in area) up to 775K (the solidus temperature of 2024-T4).

No stress-strain curves are found for 5052-O. A linear strain-hardening model is used based on the information available [105,125] for the yielding strength, ultimate strength and ductility for various temperatures from the room temperature to solidus temperature (Figure 4.12(b)).

Figure 4.11 — Temperature dependence of Young’s modulus
Figure 4.12 — Uniaxial stress versus true plastic strain curves. (a) 2024-T4; (b) 5052-O. The curves for the liquidus temperatures are not shown.
For both 2024-T4 and 5052-O, the stress versus plastic strain curves above the coherent temperature are assumed to be temperature independent and the flow stresses are assumed very small, about 10 percent of those at their respective solidus temperatures.

Note that the strain-hardening effects are very small for both 2024-T4 and 5052-O for temperatures above 600K, the temperature range in which the plastic deformation is mostly likely to be confined during welding, indicating that the kinematic strain-hardening (Bauschinger effect) is not a strong factor for these two alloys.

4.3.3 Finite Element Implementation

Solution Strategy

In this study, the heat transfer (heat conduction) analysis and thermal stress-strain analysis are uncoupled as discussed earlier. For such an uncoupled thermal stress-strain analysis, ABAQUS first solves the heat conduction problem independently from the stress problem to obtain the temperature history. The temperature history at the nodes is then written to a file using the *NODE FILE output option during the heat transfer analysis. ABAQUS provides a very simple interface which uses this file as input to define the temperature field at different
times as a transient loading condition in the stress analysis: the FILE parameter on the *TEMPERATURE option in the stress analysis is used to read the temperature file back into the stress model. This mode of transferring the temperatures is based on node numbers – the temperature at node N on the *NODE FILE output from the heat transfer analysis is applied at node N in the stress mesh.

The advantage of uncoupling the nonlinear heat transfer and stress analysis is to reduce the computational cost: they generally have different convergence rates, sometimes the difference can be of several orders of magnitude. In the uncoupled calculations where the convergence can be satisfied separately, larger time increment can be used for the one with faster convergent rate to reduce its computation time. In a coupled approach, however, both the heat transfer and stress analysis have to use the same small time increment determined by the slower convergence rate. The computation for long simulation times can become prohibitively expensive. In addition, noting that the molten weld pool is required to be included in the heat transfer analysis but excluded from the stress analysis, the uncoupled approach makes it possible to apply the element rebirth technique to the stress analysis only.

Since we are considering the thermal stress-strain problem pertaining to thin plates, plane stress condition is assumed: the top and bottom surfaces of the plate are assumed to be stress free. Two kinds of plane stress elements are used: eight-node second order (quadratic) isoparametric quadrilateral elements with
reduced integration scheme (element type: CPS8R) are used for the regions away from the weld metal because they are the more cost effective elements provided in ABAQUS [63] and provide the most accurate prediction of strains at the Gauss integration points (the Barlow points) if the elements are well shaped [128]; whereas six-node second order triangular elements (CPS6) are used for the weld metal and the near heat-affected-zone, since the use of these elements, in conjunction with the element rebirth technique, provide a closer approximation of the boundary of the weld pool than using the quadrilateral elements. Figure 4.13 illustrates the general features of the finite element meshes used for the thermal stress-strain analysis; Figure 4.14 shows the corresponding heat transfer meshes. Note that, because of the use of first order quadrilateral elements, finer meshes are used in the heat transfer models to provide the temperature history for each of the nodes in the thermal stress-strain model.

Various boundary and loading conditions are modeled in this study. Their descriptions are more appropriate if discussed together with the individual requirements for each of the models which will appear in the following chapters, and thus are not provided here.

For transient, nonlinear stress problems such as the problems concerned in this study, ABAQUS first divides the total simulation time into many time increments so that the solution follows the path of history. Within each time increment, ABAQUS uses an iteration scheme based on Newton's method (or
Figure 4.13 — General feature of the finite element mesh for thermal stress-strain analysis. The weld starts at the left end of the plate. Only half of the plate is modeled due to symmetry about the center line of the weld metal. Note the orientation change for the reborn elements that represent the resolidified weld metal. The fusion line is approximated between the elements in the model. The reborn elements are connected with the rest of the model along this fusion line.
Figure 4.14 — The heat transfer analysis mesh corresponding to the thermal stress-strain analysis mesh shown in the previous figure. $X_0$ denotes the distance between the welding start point and the left edge of the plate. No element rebirth is necessary for the heat transfer analysis.
modified Newton’s method if the Jacobian matrix cannot be computed exactly) to obtain an approximate solution which satisfies the basic equilibrium equations within the user prescribed force tolerance. This study follows the general guideline recommended by ABAQUS for choosing the force tolerance: a small fraction (in the range of 1% to 0.1%) of the magnitude of typical significant forces at nodes which can be estimated as the yield stress multiplied by a typical element area.

**Element Rebirth Technique**

As discussed in the previous section, an element rebirth scheme is adopted in this study to model the effect of solidification process on the thermal stress-strain fields. The general guideline for such element rebirth procedure is discussed here.

The basic concept is to partition the entire course of the transient thermo-mechanical responses during welding into a series of staggered solution and element rebirth alternation. Ideally, in order to obtain accurate modeling of the movement of weld pool, the thermal stress analysis should be carried out for an infinitesimally small length of time, and then the new elements (which will be infinitesimally small in size) corresponding to the fresh resolidified weld metal should be added to the mesh and those representing the area that just enters the weld pool should be removed before carrying out the analysis for another
infinitesimally small period of time. This alternate process of analysis and redefining the mesh, along with new boundary conditions, will continue until the end of the welding. However, from a practical point, this is impossible, and we have to opt for an approximate solution by adopting the same scheme of alternating between the analysis and the redefinition of the mesh on the level of finite time and finite size.

The first step in the element rebirth scheme is to decide the element size in the weld metal. Since the objective is to obtain the strain evolution curve in the solidification temperature range, it is necessary to have enough nodes within it. It was found that at least 10 nodes should be inside this range along the centerline of the weld, in order to have reasonable results. Smaller element size and more elements would generally provide better results, but consume more computational time and memory. Therefore, a few preliminary heat transfer runs are needed to obtain the size of the solidification temperature range before the mesh sizes for the thermal stress-strain analysis are finally decided. Noting that the temperature distribution during welding is a function of material property and welding parameters, the appropriate mesh sizes vary accordingly.

Having chosen the appropriate element size, the second step is to generate the finite element meshes for both the heat transfer and thermal stress analyses. First of all, the fusion zone (weld metal) needs to be identified with the help of the preliminary heat transfer runs. Secondly, noting that ABAQUS specifies the
initial temperature on the nodal basis, two sets of elements (and two sets of nodes) are constructed for the weld metal in the thermal stress-strain analysis, in order to model the change in initial temperature of the weld metal: each set of the nodes and elements covers the entire weld metal (region II of Figure 4.9) so that these two sets are overlapping each other; one set takes the ambient temperature as its initial temperature and the other takes the liquidus temperature. By using the element rebirth/removal technique, the active elements for a particular step only include those in the first element set (whose initial temperature is the ambient temperature) that are ahead of the weld pool and those behind the weld pool in the second element set at that particular step. This is illustrated in Figure 4.13.

Now one may imagine that all of the nodes and elements in the model be divided into two groups based on their initial temperatures. The first group consists of those with the ambient temperature as their initial temperature so that it covers the entire physical geometry of welded structure (the solution domain, region I plus region II). The second group takes the liquidus temperature as the initial temperature and covers only the weld metal (region II). Next, the nodes and elements in the first group are generated. After that, the second group of nodes and elements is then generated with the physical positions (not the number) of the nodes coinciding with those of the nodes in the first group that also cover the weld metal. As such, the nodes and elements in the second group are
not connected and hence cannot deform compatibly with those in the first group. To overcome this problem, the last step in the mesh generation is to use the *MPC command to link the corresponding nodes on the fusion line of the weld metal (they should be located on the outer boundary of the second group) of the two groups as shown in Figure 4.13.

Since it is not necessary to have element rebirth in the heat transfer analysis, only one set of elements is needed accordingly. However, two sets of nodes are still needed for the heat transfer analysis – the second set of nodes is to provide the temperature information for the second set of nodes (hence the second set of elements) of the stress analysis. The element generation procedure is the same in principle as that for the heat transfer analysis without considering the element rebirth technique; the only extra step here is to have each of the nodes in the second set linked to the corresponding nodes in the first set by using the *EQUATION command in ABAQUS.

The last step in element rebirth scheme is to decide an element removal/rebirth sequence. For this purpose, both of the heat transfer and thermal stress analyses are divided into the same number of steps. The element removal and rebirth are accomplished in a step by step fashion. While the ideal (but impossible) way is to add back or remove from the model an element (or better a part of an element) as it passes through the coherent temperature, the practical goal is set to limit the addition of the elements to those whose temperatures fall between
the liquidus and the coherent temperature since the last step (thus representing the fresh resolidified weld metal), and to limit the removal of elements to those whose temperatures rise beyond the liquidus temperature since last step (representing the metal just melted). The number of steps is proportional to the total welding time, so a better controlling parameter is the time period of each of the steps. A suitable upper limit to the time period of a step is such that no more than two elements along the center line of the weld are added back to the model, if the aforementioned guideline for choosing element size is followed. When the time period is beyond this limit, some of the elements being added back to the model will have temperatures too far away from the coherent temperature at the moment of element rebirth.

4.4 Construction of the Mechanical Strain Curves

Having obtained the stress and strain information in the vicinity of the weld pool in the finite element analysis, the question now is what is the controlling parameter representing the mechanical driving force for the solidification cracking. In the past, force (stress) and total strain have been proposed for this purpose. In this study, the mechanical strain, which is the sum of the elastic strain and the plastic strain, is proposed to represent the mechanical driving force. The mechanical strain has obvious advantages over both the stress and the total strain. The total strain by definition consists of elastic strain, plastic strain and thermal
strain. However, the thermal strain by itself does not directly cause cracking, if it
does not cause any elastic and/or plastic deformations (hence stresses) in the
body. This can be seen by considering an idealized case in which an infinitesimally
small body is slowly, uniformly and freely cooled from the liquid state to the
room temperature so that neither stress nor mechanical strain is created in the
body. Obviously, cracking does not occur in this case, even though the thermal
strain generated in the cooling process can be up to several percent.

The stress in principle should provide physically sound indication to the
solidification cracking problem, since the stress and the mechanical strain are
related to each other by the constitutive equations. However, the stress is in prac-
tice very insensitive to the changes in various loading and boundary conditions,
owing to the fact that the material exhibits very low yield strength and near perfect-
plastic deformation behaviors at the elevated temperatures.

However, the experimental measurement of mechanical strain is difficult.
The reason is that most of the modern strain measurement techniques, such as
electric resistance strain gage, moiré fringe, or X-ray diffraction, cannot differenti-
ate the mechanical strains from the thermal strains – what they measure is the
sum of the thermal strain and mechanical (or elastic for X-ray diffraction) strain.
The mechanical strain has to be deduced from the experimental measurement
results with the assistance of temperature information.
This study is interested in the centerline solidification cracking phenomenon. In agreement with many other documented investigations, this study assumes that the cracking is caused by the strain which acts perpendicularly to the crack. Therefore, the transverse mechanical strain acting on the center line of the weld is used for quantifying the mechanical driving force of the centerline solidification cracking.

The task of constructing the mechanical strain curve in the solidification temperature range itself is quite simple. After choosing a particular point on the centerline of the weld under the concern, the temperature history and elastic and plastic strain history for this point are readily available from the ABAQUS result files, which can be presented in the forms shown in Figure 4.15 (a) and (b). The mechanical strain curve (Figure 4.15(c)) is then constructed according to the principle shown in the figure.
Figure 4.15 — The construction of a mechanical strain curve in the solidification temperature range. (a) Temperature history; (b) Mechanical strain (sum of elastic and plastic strains) history; (c) The mechanical strain curve wanted. Note that all these curves are for a particular physical point.
CHAPTER V
VERIFICATION OF COMPUTATIONAL MODELS

This chapter presents the comparisons of the predictions from the finite element models with some experimental results reported in the literature. The validity of the finite element models is then justified based on these comparisons. It is worth noting that the objective of the verification is not for a particular finite element model of a particular welding situation, but for the modeling approach proposed in this dissertation which has been discussed in Chapter IV.

Since the objective of this study is to understand the strain evolutions in the temperature range in which solidification cracking occurs, it is most desirable to have direct comparisons within this range. Unfortunately, experimental measurements of stresses and/or strains inside the fusion zone at the temperature concerned are extremely difficult to conduct. Suitable quantitative results could not be found. The quantitative comparisons in this chapter are then based on experimental measurements in the regions outside the fusion zone but very close to the molten weld pool. An argument for such approach is that, quite similar to those used in the linear elastic fracture mechanics for the concept of “K-dominant” [129], the localized deformation in the solidification temperature
range is controlled by the areas that surround it. Furthermore, as one of the primary subjects of the following chapter, the finite element predictions of the strain evolution in the solidification temperature range will be qualitatively checked with common knowledge regarding the likelihood of solidification cracking under various circumstances. This combined approach should be able to verify the validity of the finite element modeling procedure as well as the methodology for quantifying the mechanical driving force for solidification cracking that are proposed in this dissertation.

As for the heat transfer models, results from various previous finite element analyses (for example, Tsai et al. [120] and Goldak et al. [84]) have demonstrated that the predictions of thermal history for points relatively far from the heat source are usually very accurate. Problems often rise for those in the vicinity of the weld pool. Hence, whilst in-depth deformation comparisons will be made, comparison of the heat transfer model will be kept brief and focused on the predictions of the width of the fusion zone size, and, whenever possible, the weld pool size.

Most of the FEA calculations were performed on the Cray Y-MP8/864 supercomputer at the Ohio Supercomputer Center. The computation was quite intensive. For the cases studied in this dissertation, the average CPU time for a heat transfer analysis was about 150 minutes, and about 240 minutes for a stress analysis, in addition to producing large data files. Using larger element sizes should
reduce the CPU and memory consumption, but it is prohibited by the requirement of the element rebirth scheme.

Based on the past experience [84, 114, 117, 116, 118, 120], the element sizes used in this study should be small enough to ensure the convergence of the finite element calculation results. In fact, the element rebirth scheme requires smaller elements than the convergency does. Nevertheless, some preliminary studies were conducted to examine the convergence of the finite element models used in this work. Changes in temperature calculations were less than about 3K if the elements were twice as big as the ones used in the "formal" heat transfer analyses that will be presented next. Since the strain build-ups in the resolidified weld metal (the reborn elements) strongly depends on the size of the elements, and larger elements than those used in this study would violate the restrictions by the element rebirth scheme, the convergence of the thermal stress analysis was instead examined based on the FEA models without the element rebirth procedure. It was found that doubling the element sizes almost does not change the stress and strain fields in the models used in this and the next chapter chapters.
5.1 Deformation Characteristics of Weld Edges

5.1.1 Comparison with Matsuda et al.'s Measurements

The first comparison to be made is with Matsuda et al.'s [51] measurements of the moving characteristics of weld edges of aluminum alloys. In that work, detailed information about the welding parameters and the dimensions of the weld specimen were given, making it possible to develop the matching finite element models so that they can be quantitatively compared. For the purpose of appreciation of the experimental results and the finite element models, this section first reviews some relevant experimental setup of Matsuda et al.'s work, then describes the corresponding finite element models. Finally comparisons are made and discussed.

Review of Experiment Procedure

The aim of Matsuda’s work was to investigate the effects of various welding conditions on the macroscopic moving characteristics of the fusion boundary near the trailing edge of the molten weld pool as solidification cracking takes place. Bead-on-plate welds were made on rectangular plates with GTA welding process for various plate thicknesses, traveling speeds, and alloy compositions. The measurements were conducted on two Al-Mg alloys (5052 and 5083) and one Al-Cu alloy (2017).
The basic testing procedure involved (1) making some indentation marks at four prescribed locations on the surface of the specimens with Vickers hardness test as the measuring points; (2) capturing the movements of the indentation marks as the welding arc passes by with a 35 mm camera at a speed of 5 frames per second; and (3) measuring the mutual movements of the indentation marks on the films developed after welding under a microscope at X20 magnification. Matsuda claimed that the accuracy in determining the relative distance changes is within 0.01 mm.

Figure 5.1 shows the dimensions of the specimen. All of the specimens were 150 mm long, but they varied in width and thickness to simulate the effect of the rigidity of workpiece. As shown in the figure, the width of the plate is designated by \( W \). For the cases in which \( W \) was less than 100 mm, parallel slots were cut on 150x100 mm plates. For those with \( W \) larger than 100 mm, no slots were made and \( W \) was the actual width of the plate.

For 2 mm thick plates, the welding current was varied for different welding travel speeds so as to obtain full-penetration welds and keep a constant weld width of about 7.5 mm.

Welding was started at two different locations during the test as shown in Figure 5.1. Generally speaking, no solidification crack was observed in those plates in which the weld was started at 30 mm from the edge of the plate (point A). When the weld started at the edge of the plate (point A'), the crack occurred
Figure 5.1 — Dimensions of the specimens used by Matsuda et al.
along the center line of the weld bead and followed the weld pool as welding proceeded. However, when a plate had a width over 150 mm, no cracking was reported by Matsuda et al..

Four indentation marks forming a rectangle shape were located at point B in Figure 5.1, about 90 mm from the edge where the weld started. Steady-state weld bead was maintained at these locations. Detailed arrangement of the indentation marks is illustrated in Figure 5.2. They were about 1.25 mm from the fusion boundary of the weld bead. The transverse deformation was measured as the distance change of the indentation marks which were separated by and normal to the weld bead. The longitudinal deformation was measured with the indentations on the same side of the weld bead. The location changes were recorded

Figure 5.2 — Locations and arrangement of the indentation marks for displacement measurement according to Matsuda et al.
every 0.2 second by a camera with an accuracy of 0.01 mm. It should be noted that this method is not capable of measuring the shear strain due to the symmetric arrangement of the indentations with respect to the weld bead.

**Description of Finite Element Models**

Only a subset of Matsuda’s measurements are modeled and compared in this study. It includes a total of five cases. All of them are based on one welding condition (16 volt, 145 amp and 750 mm/min), and one material (Al-5052). Only one plate thickness (2 mm) is used in the finite element models since it gives full penetration weld bead, as indicated in Matsuda et al.’s work. Three plate widths (50, 100, and 250 mm) are used when modeling the crack-free plates. However, only two are modeled for the cracked plates (W=50 and 100 mm), as the solidification cracking was not reported for the plates over 150 mm wide.

The general finite element modeling procedure and the relevant material properties are presented in Chapter IV.

**Heat Transfer Models**

The heat transfer is modeled as a two-dimensional heat conduction problem, since full-penetration weld was easily produced for the 2 mm plate as reported by Matsuda et al.. A symmetric half of the plates is discretized with 4-noded linear isoparametric quadrilateral elements. The same mesh is used if the
plates have the same width, regardless of the positions at which welding starts. Smaller elements are used in the weld metal and adjacent areas. The element size varied from 0.625x0.625 mm for weld metal to 2.5x5 mm for regions away from the weld. Figure 5.3 shows the finite element mesh for the 100 mm wide plate, which consists of 6960 elements.

The convective and radiative heat loss from the top and bottom surfaces of the plate is considered in the heat transfer analysis, whereas the heat loss from the four sides of the plate is ignored. The film coefficient for convection is assumed to be 84 W/m²K. The radiative emissivity is assumed to be 0.3, according to Incropera [113], for the surface conditions prepared by normal pre-welding cleaning procedure for aluminum alloy. As discussed in Chapter IV, the heat flux from the moving welding arc is assumed to have a spatial distribution of radially symmetric Gaussian profile in the plane of the plate, but uniformly distributed over the thickness direction. The diameter and other dimensional parameters of the tungsten electrode were not provided by Matsuda et al.. This study then assumes that the electrode has a diameter of 1.6 mm, as commonly used in the Sigmajig test and the Houldcroft test [103] for thin plate weldability tests. The arc beam radius necessary to define the Gaussian distribution in equation (4.1) is taken as the same as the diameter of the electrode (1.6 mm). An arc efficiency of 60% is used for the GTA process. According to the discussion in Chapter IV, all the boundary conditions and the heat input from the arc can be consolidated as
the single body heat generation term, \( Q \), in the analysis (see equation 4.3). This body heat generation term is implemented in ABAQUS using a user subroutine DFLUX. With aforementioned values, the body heat generation term for the cases concerned here can be written as

\[
Q(W \text{ m}^{-3}) = 259.62 \exp\left\{-1.1719\left[(x - 12.5t - x_0)^2 + y^2\right]\right\} - 8.4 \times 10^{-5}(T - T_\infty) - 1.7 \times 10^{-14}(T^4 - T_\infty^4)
\]

(5.1)

where \( T_\infty \) (K) is the ambient temperature of the surroundings, which is taken as the same as initial temperature of the plate (300 K). \( x_0 \) (mm) is the x-coordinate of the location where the welding starts, either at 0 mm (point A of Figure 5.1) or at 30 mm (point A'). The welding time is \( t \) (second).

For each plate width, two heat transfer analyses are performed for the two different welding starting locations.

**Stress-Strain Models**

The mechanical responses of the plates during welding are modeled as the plane stress problem: the top and bottom surfaces of the plate are stress-free. Eight-noded second order isoparametric quadrilateral elements with reduced integration scheme (CPS8R) are used for the areas away from the fusion zone, whereas six-noded second order triangular elements (CPS6) are used for the weld metal and the adjacent regions. Unlike in the heat transfer analysis, two
meshes are generated in the stress-strain analysis for the two different weld starting positions of the same plate width. The finite element mesh for the $W=50$ mm wide plate with weld starting at $x_0=30$ mm is shown in Figure 5.4. Note that the two sets of triangular elements representing the weld metal and its neighboring areas actually overlap each other in the figure; but only some of the elements are active at a specific time. Figure 5.5 shows those active elements for a peculiar time ($t=4.6$ second) for the same plate.

All the plates modeled in this study are assumed to be free of any constraint during welding. Therefore, no displacement boundary conditions are imposed on the finite element model, except for the symmetry requirement along the centerline of the weld metal. In order to prevent the free rigid body movement of the plate (displacement singularity), the displacement of a single node located at $x=150$ mm, $y=0$ mm is also set to zero in both $x$ and $y$ directions.

For the cases in which solidification cracking takes place along the centerline of the weld metal (i.e., the welding starts at $x=0$ mm), the symmetry requirement for the reborn elements is released to appropriately simulate the free movement of the cracked surface. The symmetry requirement is always imposed on the elements ahead of the moving welding arc. As for the cases where solidification cracking occurs, no relaxation in the symmetry requirement is needed.

Following the guideline discussed in the previous chapter, the time period for a step is 0.2 second. The element rebirth/removal is done at the beginning of
Figure 5.3 — Finite element mesh used in the heat transfer analysis of 100 mm wide plate.
Figure 5.4 — Finite element mesh for the stress analysis of 50 mm wide plate as welding starts at x=30 mm. All elements are shown.
Figure 5.5 — Finite element mesh for the stress analysis of 50 mm wide plate as welding starts at x=30 mm. Only the elements that are active at the moment of t=4.6 second are shown.
each step. In other words, elements are removed from and added to the model every 0.2 second. In addition, the time period in a step is further divided into many time increments during the numerical computation in order to guarantee that the convergence of the solution will be satisfied within the user prescribed force tolerance. The maximum time increment in a step is limited to no more than 0.01 second, which is small enough that the converged solutions are usually achieved in one iteration after the temperature field in the model reaches the quasi-stationary state. Smaller time increments are often required before this quasi-stationary state, i.e., in the initial transient period in order to satisfy the prescribed force tolerance. This is accomplished by implementing the automatic time increment scheme in ABAQUS which automatically select a smaller time increment when necessary.

Comparisons and Discussions

Temperature Fields

As mentioned at the beginning of this chapter, the verification of the heat transfer analysis is focused on the comparisons of the temperature fields around the weld pool, particularly on the comparisons of the width of the weld bead and the weld pool size. Figure 5.6 shows an overview of the quasi-stationary temperature field as the welding arc just passed by the indentation marks (t=5.0 second). The plate is 100 mm wide and the weld starts at x=30 mm. The weld
pool and the surrounding areas are enlarged in Figure 5.7, so that the finite element predictions of the weld pool size can be more accurately obtained from the figure. However, before we can proceed to do so, it is fundamental to know the temperature that corresponds to the experimentally observed weld pool boundary. In this study, we assume that the boundary of the weld pool corresponds to a particular constant temperature within the solidification temperature range (between the solidus and liquidus temperature); and we further assume that this constant temperature is the coherent temperature of the alloy which is just below the liquidus temperature. The choice of coherent temperature seems arguable, but not without any reasons. For one, Lin [130], using the thermocouple technique, determined experimentally that the peak temperature experienced at the fusion line is 1659K for S310 stainless steel whose liquidus temperature is 1673K according to the thermal analysis measurement by Matsuda et al. [54].

As such, the heat transfer model predicts that the width of the weld bead, which is also the width of the weld pool, and the length of the weld pool are 7.35 and 12.8 mm respectively, when the weld pool is under the quasi-stationary state. This is in very good agreement with the data provided by Matsuda et al.: the weld bead was about 7.5 mm wide; and though not directly given, a measurement of the weld pool on one of the figures in Matsuda et al.’s article revealed that the length of the weld pool was about 13.5 mm. Therefore, the heat transfer models developed in this work is justified.
Figure 5.8 presents the effect of different treatments of the latent heat during welding. Presented here are three latent heat release models based on: (1) the consideration of the new micro-dendritic grain growth model proposed in this study, (2) the common assumption of uniform latent heat distribution in the solidification range, and (3) no latent heat assumption. It is clear that, in comparison with the uniform latent heat distribution, the micro-solidification model does not alter very much the temperature distribution in the region away from the molten weld pool; it also has little influence on the heating (melting) process ahead of the weld arc; however, it indeed significantly changes the temperature field within the solidification temperature range behind the weld arc where it matters most insofar as the solidification cracking is concerned. The new model pushes the coherent temperature away from the welding arc, narrows the region between the coherent temperature and the solidus temperature. If the latent heat were not considered during welding, the solidification temperature range behind the weld pool would be much narrower but the weld bead would be wider.
Figure 5.6 — An overview of the temperature field as the welding arc just passed by the indentation marks (t=5.0 second). The weld starts at x=30 mm. W=100 mm, Al-5052, T_L=923K, T_C=900K, T_S=848K.
Figure 5.7 — Temperature distribution around the weld pool. The coherent temperature is used to define the width of weld metal and the shape of weld pool. The weld starts at \( x = 30 \text{ mm}, t = 5.0 \text{ second}, W = 100 \text{ mm}, \text{Al-5052, } T_L = 923\text{K}, T_C = 900\text{K}, T_S = 848\text{K}.\)
Figure 5.8 — Effect of different treatments of latent heat on the temperature distribution in the weld pool and adjacent areas. The weld starts at x=30 mm, t=5.0 second, W=100 mm, Al-5052, T_L=923K, T_C=900K, T_S=848K.
Moving Characteristics of the Indentation Marks

Matsuda et al. tried to quantify the moving characteristics of the weld edge by a series of plots showing the relative distance change of indentation marks as a function of time. The following quote from Matsuda et al.'s article describes how they constructed these plots: “Zero on the abscissa means the instance when the tip of the welding electrode has passed on the connecting line between the first two marks and the deformation $\Delta l$ in the vertical shows the difference between the instantaneous and the original distance as positive for expansion and negative for contraction.” The first two marks mentioned in the quote should be referred to as the marks A and B in Figure 5.2 of this work in relation to the welding travel direction. This interpretation is also clearly indicated by Matsuda et al. in one of their $\Delta l$ versus time plots. Following this instruction, Figure 5.9 presents the comparison of the deformation for 50 mm wide plate. The finite element results are plotted as either continuous or dashed lines, whereas the experimental measurements are shown as discrete points. Deformations for both the cracked and crack-free plates are shown. It is clear that the quantitative comparison is not as desirable, although the finite element analysis is able to predict all the deformation features observed by Matsuda et al.: the cracked plate exhibits large transverse expansion rapidly whereas transverse contraction occurs for the crack-free plate; as for the longitudinal deformation, positive and almost constant deformation occurs in both cases with the crack-free one showing a little larger expansion.
The quantitative discrepancy between the experimental and finite element results can be reduced by carefully examining the choice of the origin for time axis (the choice of the zero time) in Matsuda et al.'s work. Figure 5.10 duplicates some photographs in Matsuda et al.'s article which were used to measure the relative movement of the indentation marks.

The shape of the molten weld pool, the width of the weld bead, the indentation marks, and the location of the crack tip can be simultaneously observed in the same photos. In the figure, the welding torch is moving upwards and is shown in the first photo (photo (a)). Therefore, the indentation marks used for

![Graph showing comparison between FEM and Experiment results](image-url)

**Figure 5.9** — Comparison between the finite element results and the experiment measurements. W=50 mm
Figure 5.10 — Surface of the weld pool at the moment when welding arc is passing the connecting line between marks A and B. Travel speed: 750 mm/min. (From Matsuoka et al. [51])
transverse deformation measurement should be the two on the lower part of the photo, which are marks A and B in Figure 5.2. If the photo (a) were representing the instant of \( t=0 \), then the electrode has clearly passed the connecting line of the marks A and B. In fact, the electrode has just passed the other two marks (C and D). Therefore it seems that Matsuda et al. actually used a different reference time for setting up the zero point on time axis. This explanation is further evidenced by the consequently "modified" comparison with Matsuda et al.'s work, and the comparison with another experimental measurement of near fusion zone displacements based on moiré fringe technique by Chihoski [58], which will be presented next.

With this explanation, the reference time (zero on abscissa) should be the moment when the electrode is at the location about 5 mm passing the connecting line between the marks A and B, which takes 0.4 second for a 750 mm/min traveling speed. Considering also that the photographs were not taken continuously – 0.2 second elapsed between the consecutive films, we then need to shift the reference time for the finite element result accordingly by 0.5 to 0.6 second. Figure 5.11 presents the comparison for the adjusted finite element results. The transverse deformations of all the five cases studied in this dissertation are compared in the figure: two cracked ones (50 and 100 mm wide) and three crack-free ones (50, 100, 250 mm wide).
Clearly, the predicted transverse deformations for the cracked cases show much improved quantitative agreement with the experimental data. However, the predictions for both 100 mm and 250 mm wide crack-free plates still do not quantitatively agree with the Matsuda et al.'s measurement. They are essentially parallel to the experimental data, but with a net negative shift in displacement.

Now let us turn out the attention to a striking prediction regarding the transverse deformation in the neighborhood of the molten weld pool: the fusion line usually experiences contraction in this region for both the cracked and crack-free situations. As shown in Figure 5.11, this phenomenon is particularly noticeable

![Figure 5.11](image)

Figure 5.11 — Comparison of the “adjusted” finite element results with Matsuda et al.’s results. Only the transverse deformations are shown.
for the plates with width over 100 mm. But this is a considerable departure from the accepted and apparent view of immediate post-puddle tension caused by the clear decrease in temperature of the centerline after the weld pool, and is also not expected from Matsuda et al. observation. For crack-free plates, this negative deformation continues and gradually saturates. On the other hand, this contraction stops only after the welding arc has passed by and, as observed by Matsuda et al., expansion ensues.

However, this discrepancy between the finite element prediction and experimental observation (or the common accepted concept) does not necessarily mean that the finite element models developed in this study cannot reliably predict the deformation behaviors in the vicinity of the weld pool, since the report of Matsuda et al. actually did not cover the time period in which the predicted contractive deformation takes place. Quite to the contrary, this predicted phenomenon has been reported in another experimental investigation of the deformation behaviors in the vicinity of the weld pool, thus demonstrating the capability of the finite element models.

5.1.2 Comparison with Chihoski's Measurements

Based on his moiré fringe photographic studies of aluminum during autogenous GTA welding, Chihoski [58] proposed that the lagging temperature rise (in the areas on the sides of a fast moving aluminum weld pool) has the effect of
applying a transverse compression on the post-puddle weld bead. Therefore the weld fusion line exhibits contraction right after the welding arc. It is also worth noting that Chihoski’s article provided an excellent explanation for such a phenomenon by using a simple one-dimensional expansion/contraction model that consists of a series of strips normal to the weld bead.

Chihoski performed the deformation measurement on 2.54 mm thick 2014-T6 plate for two kinds of welds: edge welds and butt welds as shown in Figure 5.12. The edge weld was actually not a weld but a 2.54 mm meltdown on the edge of a 102x610 mm plate, corresponding to a 5.8 mm butt weld which joins

Figure 5.12 — Experimental set-up in Chihoski’s study. (a) edge weld; (b) butt weld (From Chihoski [58])
two of 102x610 mm plates that were tack welded at both ends before welding. Therefore, the edge weld resembles the cracked plate in Matsuda et al.'s work whereas the butt weld resembles the crack-free plate.

Chihoski presented the experimental measurement in the form of the distorted shapes of a series of lines that are originally parallel to the edge of the plate that was to be welded. These lines were 5.08 mm (0.2 in.) apart with the first line very close to the fusion line. They were labeled for reference: the first line is A, the rest in order are B, C, D, E, F, G, H, and I. Three welding speeds were used in Chihoski's study: 152, 330 and 508 mm/min (6, 13 and 20 ipm) for both edge and butt welds.

Unfortunately, Chihoski did not provide complete welding conditions. For example, the welding voltage was not given. Furthermore, as will be seen next, the scales used in Chihoski's deformation plots were not well explained so that the matching finite element models could not be established in the study for quantitative verification. Nevertheless, it is found that some of Chihoski's experimental setup are quite close to those used in Matsuda et al.'s work. Therefore, Chihoski's results should provide a qualitative comparison with the finite element analysis results based on Matsuda's work.

Figures 5.13 and 5.14 show the experimental results for both edge and butt welds, respectively. The negative deformation feature around the weld pool in the finite element calculation (Figure 5.11) is well illustrated in these two figures.
Figure 5.13 — Chihoski's deformation patterns for edge welds. The edge weld resembles the cracked plate in Matsuda's work. (From Chihoski [58])
Figure 5.14 — Chihoski’s deformation patterns for the butt welds. The butt weld resembles the crack-free situation in Matsuda’s work. (From Chihoski [58])
5.2 Strain Fields in the Vicinity of Weld Pool

The last experiment used for the verification of the finite element model developed in this study is Johnson's measurement of the principal and shear strain fields in the neighborhood of an autogenous GTA weld pool in 3 mm thick plate based on a projected-grating moiré fringe strain analysis technique [56, 57]. Dynamic strain patterns were recorded using a 16 mm movie camera with a speed of 16 frames per second. Johnson estimated that the accuracy of the strain measurement was within 0.3 percent, primarily based on the repeatability of the measurement results.

The material used in Johnson's studies was an Al-Mg alloy similar to 5056 or 5256, which has similar thermophysical and mechanical properties, particularly at the elevated temperatures, to those of the 5052 alloy [64,122,125] that was used in Matsuda et al.'s work [51]. The specimen was rectangular and measured 300x800 mm. The welds were made along the centerline of the long direction, starting from the edge of the plate; and no solidification cracking took place in Johnson's studies. Three welding currents were chosen for each of the three travel speeds – 200, 500, and 800 mm/min – to give weld beads ranging from 50 percent to full penetration.

The moiré fringe measurement was conducted on an 18 x 42 mm rectangle area located around the center of the plate (Figure 5.15). The side that is close to the weld bead is 5 mm from the centerline of the weld. From this figure, the
width of the weld bead is estimated between 8 to 8.5 mm. Johnson assumed that the temperature distribution reached the quasi-steady state when the electrode moved to the measurement area. In addition, he also ignored the end effect on the strain pattern in the measurement area because the plate was of constant width and sufficiently long. This led Johnson to assume a quasi-stationary strain pattern around the electrode.

Unfortunately, Johnson did not provide complete welding conditions; for example, the welding voltage was not given so that the matching finite element models could not be established in this study. Nevertheless, if we compare the material, geometry and welding conditions used by Johnson with those used by Matsuda et al., it is then clear that, as shown in Table 5.3, the conditions of one

![Diagram](image)

Figure 5.15 — The location of the 18x42 mm measuring area in relation to the weld bead, weld pool and moving torch (From Johnson [57])
particular case in Johnson’s work – full penetration weld made with 800 mm/min travel speed – essentially resemble those used in the finite element verification model which has been developed for the 250 mm wide plate with 750 mm/min welding speed in Matsuda et al.’s work. Hence, this finite element model is used without modification for comparison with the moiré fringe measurement results of that particular case. And this comparison should be able to provide at least a semi-quantitative verification of the finite element model, even though it is developed for Matsuda et al.’s work. The term “semi-quantitative” used here means that not only should the general characteristics of the strain fields in the neighborhood of the weld pool be favorably compared, but also the magnitude of the strain fields should be similar, if the finite element model is valid.

Table 5.1 – Similarity comparison between the Johnson’s measurement and corresponding finite element model

<table>
<thead>
<tr>
<th></th>
<th>Johnson’s Experiment</th>
<th>FEM Model</th>
</tr>
</thead>
<tbody>
<tr>
<td>Alloy</td>
<td>5056</td>
<td>5052</td>
</tr>
<tr>
<td>Weld type</td>
<td>full penetration</td>
<td>full penetration</td>
</tr>
<tr>
<td>Travel speed</td>
<td>800 mm/sec</td>
<td>750 mm/sec</td>
</tr>
<tr>
<td>Current</td>
<td>250A</td>
<td>145A</td>
</tr>
<tr>
<td>Voltage</td>
<td>n/a</td>
<td>16V</td>
</tr>
<tr>
<td>Weld bead width</td>
<td>8 - 8.5 mm</td>
<td>about 7.5 mm</td>
</tr>
<tr>
<td>Plate width</td>
<td>300mm</td>
<td>250mm</td>
</tr>
<tr>
<td>Plate thickness</td>
<td>3mm</td>
<td>2mm</td>
</tr>
</tbody>
</table>
Figures 5.16 and 5.17 present the comparisons of the maximum shear strain fields and the transverse strain fields respectively. The strain values are plotted as three-dimensional graphs to show clearly how strains develop around the weld pool. Moiré fringe measurement results and the finite element results are shown as dashed lines and solid lines respectively. The measurement area in these two figures in relation to the weld pool is shown in Figure 5.15. The strains at the nodal points instead of at the normally more accurate integration points are used in order to make the comparison with Johnson’s plots more legible. For the same reason, the figures only show the strain values of the nodal points which are located on the same 3x3 mm grid intersection points as used in Johnson’s plots. In addition, comparisons are made only for a 15x42 mm area instead of the 18x42 mm area shown in Figure 5.15. The 3 mm wide area within the X=20 mm line and X=23 mm line is not compared because there are no nodal points on the X=23 mm line. The normal strain in the X-direction (the transverse strain, $\varepsilon_T$) has been plotted with negative, or compressive strain upward in order to better show the straining behavior near the weld.

It is clear that the predicted magnitude of maximum shear strain field is in very good agreement with Johnson’s moiré fringe measurement results. The difference between these two strain fields are generally within the range of the experimental errors. In addition, the predicted orientations of the maximum shear strain are also consistent with the Johnson’s results: the planes of
maximum shear strain are oriented in almost the same direction as the X and Y axes for the region just behind the weld pool. The finite element analysis shows that the angle between the maximum shear strain and the weld centerline (Y direction in Figure 5.15) is generally within 2 to 5 degrees along the X=5mm line. Johnson did not provide values for this angle, but according to one of the figures in his articles, it should be also within 5 degrees. The predicted transverse strain field is in general consistent with the experimental results, even though not as good as in the maximum strain case. The major difference in the transverse strain distribution occurs along the X=5mm line. The moiré fringe measurement results reveals that, as the electrode passes by, a point on the X=5mm line begins to contract from a tensile state (positive strain) and eventually experiences compression (negative strain). The finite element analysis shows the same trend but hardly produces any negative strain. A possible cause for this discrepancy may be the difference in the weld bead widths. The finite element model assumes a little smaller weld bead (7.5 mm) than that in Johnson’s test (8 - 8.5 mm). If everything else is the same, this bead width difference means that the X=5 mm line of the finite element model is equivalent to the Y=6 mm line in the Johnson’s test where the strain values are considerably lower than those on the X=5 mm line of the same plot.
Figure 5.16 — Comparison of the maximum shear strain fields. Dashed lines: Johnson’s moiré measurements; Solid lines: FEA results. The electrode is located at the intersection of X and Y axes.
Figure 5.17 — Comparison of the transverse strain fields. Dashed lines: Johnson’s moiré measurements; Solid lines: FEA results. The electrode is located at the intersection of X and Y axes.
Nevertheless, the finite element model adequately predicts the most striking feature of the transverse strain field that was emphasized and observed in the entire series of welding results by Johnson: the sharp negative curvature in the strain pattern.

5.3 Closure

In this chapter, predictions from the finite element models developed in this dissertation have been compared quantitatively with the experimental measurement results from three separate sources using different testing techniques. It has been found that the finite element predictions regarding to the deformation of the weld edge, the strain fields in the neighborhood of the weld pool, and the weld pool size are in general in very good agreement with the experimental results. Thus the first objective in verifying the finite element modeling approach proposed in this dissertation, namely, verifying quantitatively the strain fields and deformation patterns in the regions outside the fusion zone, but very close to the molten weld pool, has been achieved.
CHAPTER VI

ASSESSMENT OF MECHANICAL DRIVING FORCE FOR WELD METAL SOLIDIFICATION CRACKING

In this chapter, the hot strain versus threshold strain approach – a methodology for quantitative solidification cracking assessment – is first proposed. It is based on the quantitative determination of strain evolution in the solidification temperature range using the finite element modeling procedure developed in Chapter IV. Next, the mechanical driving forces (the hot strains) for weld metal solidification cracking under several welding conditions are assessed quantitatively based on the finite element analysis results. The objective is to demonstrate the usefulness of the proposed finite element modeling procedure and the ensuing methodology for quantifying the mechanical driving force, which are the core of this dissertation. Although no experimental measurement data are available to directly and quantitatively verify the predicted results, validity of the proposed methodology is established as it can correctly explain various experimental observations regarding the tendency of solidification cracking. Once again, the validity of the entire approach of the dissertation should be judged upon the discussions of this chapter and the previous chapter as a whole.
6.1 The Approach: Hot Strain versus Threshold Strain

As discussed in Chapters I and III, numerous investigations since the 1940’s have established that weld metal solidification cracking takes place in a brittle temperature range (BTR) during the solidification process in which liquid and solid phases coexist. The material resistance to cracking in this BTR is substantially lower than that in the adjacent temperature ranges. The likelihood for an alloy to crack depends upon the competition between the recovery of the material resistance and the build-up of the mechanical driving force in the BTR: unless the material can sustain the mechanical driving force in the BTR, cracking will occur.

The ductility in the BTR, though small, is still quite measurable for most of the engineering alloys. This fact ensures resistance to solidification cracking, enables one to measure the resistance to cracking, and suggests that the solidification cracking is preventable.

The development of the quantitative cracking assessment methodology proposed in this dissertation is based on these well established concepts (see Figure 1.2). These concepts are illustrated in Figure 6.1 in terms of hot strain versus threshold strain which will be discussed next. This fundamental concept.

Based on the rationale discussed in Chapter IV, this dissertation chooses to use the term **hot strain** as a quantitative measure of the mechanical driving force for solidification cracking. A further assumption is that the hot strain is the mechanical strain – the sum of the (visco)plastic strain and elastic strain – in the
solidification temperature range, which can be determined by using the finite element analysis procedure developed in this work.

On the other hand, the material resistance is represented by the ductility in the BTR. In this study, the ductility is measured by the **threshold strain** whose value is that of the hot strain experienced at the crack initiation location at the instant of crack initiation. As shown in Figure 6.1, the threshold strain in the BTR is not constant but a function of the temperature at which the cracking occurs. Solidification cracking will not occur only if the hot strain, which is also a function of temperature, does not exceed the threshold strain in the entire brittle

![Diagram of solidification temperature range and threshold strain curve](image)

**Figure 6.1** — The hot strain versus threshold strain approach for quantitative solidification cracking assessment.
temperature range. Therefore, the threshold strain curve in the BTR represents the material resistance to solidification cracking.

By proposing the hot strain as a measure of the mechanical driving force and the threshold strain curve in the BTR for the material resistance to solidification cracking, the criterion for incipient crack growth can be expressed as

\[ \varepsilon_h(T) = \varepsilon_c(T) \] (6.1)

and a solidification crack-free weld can be made only if the following conditions is satisfied in the entire solidification temperature range

\[ \varepsilon_h(T) < \varepsilon_c(T) \] (6.2)

We emphasize that both the hot strain, \( \varepsilon_h \), and the threshold strain, \( \varepsilon_c \), are functions of temperature, \( T \), and likely depend on the strain rate as well.

One of the fundamental differences between the cracking assessment approach proposed in this dissertation (the hot strain versus threshold strain curve hypothesis) and the others is that the methodology developed in this work enables the use of the same hot strain calculation procedure for both weldability tests and actual fabrication situations. In other words, Figure 6.1 (representing the approach in this work) differs from Figure 1.2 in the former is based on the quantitative evaluation of the hot strain. Thus it is not only a schematic explanation for the occurrence of solidification cracking in view of the competition of
material resistance versus mechanical driving force, but it is also possible to predict, in the design stage, whether solidification cracking will occur under an actual fabrication situation, since the hot strains experienced in a fabrication process can be compared directly and quantitatively with the threshold strain of the material that is measured in a weldability test. None of the currently available weldability testing procedures are capable of achieving this, primarily because the procedures in these weldability tests for quantifying the mechanical driving force—which will be equal in value to the material resistance at the moment of cracking—cannot be used to evaluate the mechanical driving force in real fabrication situations. The Varestraint test, for example, is perhaps one of the most commonly used weldability tests for solidification cracking. It uses the "augmented strain" as the measure of material resistance to solidification cracking. However, since there is no way to obtain the augmented strains under actual welding construction situations, the augmented strain can only, at its best, relatively rank the order of solidification cracking susceptibility of different alloys, and cannot predict whether or not a specific alloy will develop solidification cracking under a given fabrication situation.

The methodology proposed in this dissertation is perhaps the first one to evaluate quantitatively the hot strains for solidification cracking. Thus, the major objective of this dissertation is to establish the validity of the methodology and its underlying finite element models. It is for this reason that the entire previous
chapter (Chapter V) has been devoted to verify the underlying finite element models. The foundation of the hot strain versus threshold strain approach is contingent on the computation of the hot strain evolution during solidification process, since the threshold strain is based on the hot strain computation in the specially designed weldability test. Therefore, the usefulness of the approach of this work depends on the validity of the hot strain computation, and so a major portion of this chapter will also be devoted to establish the legitimacy of the hot strain assessment in the solidification temperature range. This dissertation is less concerned about the measurement of threshold strains for various alloys under various welding situations or the hot strains in a peculiar welding fabrication condition – they would require more extensive resources to achieve than practical for this dissertation.

As mentioned in the beginning of this chapter, there is no experimental measurement data yet available to directly confirm the finite element calculations of the hot strains (mechanical strains) in the solidification temperature range under the welding conditions. In theory, the validity of the hot strain calculation is not confined to the solidification cracking problem. But in practice, we are limited to the calculation of the hot strains for various situations in which the solidification cracking tendency has been well established. If the hot strain versus threshold strain hypothesis proposed in this study is valid, it would then be able to correctly predict the tendency of the cracking in these situations.
Before this qualitative justification of hot strain calculation can proceed, it should realize that the hot strains need to be calculated first prior to generating the threshold strain curves for any materials. Therefore, until its validity can be established otherwise, the methodology of hot strain assessment in this work cannot be justified based on the comparison of the threshold strain curves for various alloys with their known relative cracking tendencies, although this kind of comparison has been widely used in the development of various weldability tests.

Fortunately, previous investigations have revealed the factors that can affect the tendency of solidification cracking (the material resistance to cracking). In general, the intrinsic cracking process is governed by variables such as: the interfacial tension that determines the energy needed to separate the interdendritic bonding, the fluidity of the molten metal that determines how fast the liquid metal can fill the interdendritic region, and the dendritic morphology that determines the bridging of the dendrites which serves as the concentrator for the strain imposed on the dendrites. Any other factors that can influence these intrinsic variables will also be capable of influencing the material resistance to solidification cracking (the threshold strain curve in this work). Among the factors that strongly affect the material resistance to solidification cracking are the alloy composition, the welding parameters (current, voltage, and travel speed), and the welding process and procedure. On the other hand, the material resistance to
cracking is less likely to be dependent upon the size of the workpiece and the restraint imposed on the workpiece by fixtures.

With these in mind, the appropriate candidates for the purpose of judging the hot strain calculation – hence the hot strain versus threshold strain hypothesis – would be the ones with the same threshold strain curve but each of them with different hot strains. The higher the hot strain is under such these situations, the more likely the solidification cracking will take place. We should be able to judge the validity of the methodology and the underlying finite element modeling procedure by comparing the predicted likelihood of solidification cracking with the observation reported in the literature regarding the tendency of cracking under these circumstances.

Obviously, the most logical and perhaps the easiest way to maintain a constant threshold strain curve is to restrict the finite element modeling effort to one alloy, one welding process and one set of welding parameters. The changes in the hot strains under various conditions can be brought in by varying the rigidity of the workpiece or by changing the restraint conditions imposed on the workpiece.

Finally, the discussion of this chapter will be limited to the centerline weld metal solidification cracking of two-dimensional plates so that the finite element modeling techniques developed in Chapter IV can be applied.
6.2 Hot Strain Assessment

6.2.1 Model Description

Based on the rationale discussed in the last section, the finite element modeling effort should be restricted to one alloy. Instead of using aluminum alloy 5052-O (see Chapter V), this chapter chooses aluminum alloy 2024-T4. The reason behind this decision is that alloy 2024 is generally more prone to solidification cracking than alloy 5052 and calls for more attention to solidification cracking in practice.

The geometry of the finite element models essentially resembles that used in the Sigmajig test [131]. Full penetration bead-on-plate weld is deposited along the centerline of the plate using autogenous GTA welding process. Some of the reasons to choose the Sigmajig test are: (1) this makes it possible to use some of the cracking phenomena observed in the Sigmajig test to justify the hot strain versus threshold strain approach; (2) the Sigmajig test specimen is usually very thin (usually less than 2 mm), making the two-dimensional assumption of the finite element models fairly close to the real situation; and (3) the Sigmajig test has the smallest specimen dimensions among the most commonly used solidification cracking weldability tests, (i.e., a 50x50 mm square plate, which is also much smaller than the plates used in Chapter V for the verification of the underlying finite element modeling procedure). The choice of the Sigmajig testing plate substantially reduces the requirements on computational resources. The plate
thickness used in the finite element models is kept the same as that used in Chapter V (2mm). The welding current, voltage and speed are 15V, 93A, and 300 mm/min (5 mm/second), respectively. The electrode is assumed to have a diameter of 1.6 mm. According to Chen [103], these welding parameters should be able to produce a full penetration weld bead of about 10 mm in width.

However, unlike in the previous chapter for Matsuda et al.'s work [51], the exact boundary conditions and the loading procedure in the Sigmajig test will not be followed. The first reason is that the boundary conditions in the Sigmajig test are quite complex; their inappropriate treatment in the finite element models will deteriorate the credibility of the models. Secondly, the Sigmajig test has been exclusively used for studying the solidification cracking problems of ferrous alloys [132,133], especially for stainless steels, and no report on aluminum alloys has been found. Thus developing the exact matching models is not necessary.

Instead of simulating the boundary conditions and loading procedure in the Sigmajig test, two simple boundary conditions are analyzed in the finite element models. The first boundary condition is that the nodal displacements along the two sides of the plate parallel to the weld are set to zero in both X and Y directions: this is more or less similar to the situation in the Sigmajig test where the fixture is used to restrain the in-plane movement of the two sides of the plate parallel to the weld but without applying any stress prior to welding. The second
boundary condition is to simulate the situation in which no fixture is used to restrain the plate at all. Although there is no corresponding loading condition in the Sigmajig test for this case, many documented studies on solidification cracking have chosen this kind of setup, for example, in the Houldcroft test [134] and in Matsuda et al.'s article [51].

The finite element modeling procedure developed in Chapter IV is followed to set up the finite element models. The treatment of thermophysical and mechanical properties of alloy 2024 pertaining to the finite element models has also been discussed in Chapter IV and will not be repeated here. Due to symmetry about the weld centerline, only half of the plate needs to be analyzed. For the heat transfer analysis, the half of the plate is discretized with 5000 four-noded linear isoparametric quadrilateral elements. Convective and radiative heat loss from the top and bottom surfaces of the plate are considered in the model. The heat input of the arc is modeled as a transient moving body heat flux with Gaussian distribution. Figure 4.14 shows the finite element mesh for the heat transfer analysis. The welding arc moves from left to the right in the figure, and the weld starts either at the left edge of the plate \(x_0=0\) mm or 12.5 mm \((1/4\) of the plate length) from the edge \(x_0=12.5\) mm.

Figure 4.13 shows the corresponding finite element mesh for the stress-strain analysis for the case of \(x_0=0\) mm. In order to properly model the weld pool shape with the element rebirth technique, six-noded second order triangular
plane stress elements (CPS6) are used for the fusion zone and the near heat affected zone. The rest of the model is discretized with eight-noded plane stress quadratic isoparametric quadrilateral element with a reduced integration scheme (CPS8R).

In the stress-strain analysis, the welding time is divided into many steps, with each step corresponding to a time period of 0.5 second (2.5 mm of weld). The fusion line and the weld pool shape are determined from the heat transfer analysis. For each time step, the old elements (representing the area ahead of the weld pool that has entered the weld pool since the last step) are removed from and the new elements corresponding to the resolidified weld metal since the last step are added back to the analysis model by using the *MODEL CHANGE option available in ABAQUS.

All the finite element models are assumed to be free of solidification cracking. By no means does this suggest that solidification cracking does not occur under the conditions simulated by these finite element models. To the contrary, experience indicates that centerline solidification cracking will almost certainly take place in some of the situations. Although the crack-free treatment is partly due to the fact that the values of the threshold strain are not known for alloy 2024 under the concerned welding conditions, the primary reason is related to the single threshold strain curve assumption for all the cases being studied here – the
crack-free treatment makes it possible to compare the hot strains in different cases, and to reveal which case is more prone to solidification cracking.

Since only half of the plate is modeled in finite element analysis, the crack-free assumption requires that the symmetric boundary condition be applied along the centerline of the weld, in addition to the other boundary conditions discussed earlier in this section.

6.2.2 Results and Discussion

Temperature and Stress-Strain Fields

Figures 6.2 to 6.4 show the transient temperature distributions at different times. The liquidus, coherent and solidus temperatures of 2024 alloy are 911, 865 and 775 K respectively. The coherent temperature contours are used to define the boundary of the weld pool and the fusion line. During the initial transient stage, the weld pool and the fusion zone grow as time elapses. The initial transient stage ends at about 3 seconds after welding starts. From then on, the weld pool and the fusion zone essentially stop expanding, though the contours of other lower temperatures still expand with time. The lower the temperature of a contour, the longer the time required for that contour to stop growing.

Since we are concerned with the centerline solidification cracking, the transverse stress and strain fields normal to the weld bead are more important than
the other orientations. Figures 6.5 and 6.6 show the transverse stress and strain fields at 5 seconds, respectively. At this instant, the electrode is located at the center of the plate. The weld pool is removed from the finite element model so it is not shown in the figure. The effects of boundary conditions (fixture) are also shown in these two figures. It is apparent from Figure 6.5 that the areas around the weld pool are subject to minimal stresses. In addition, in comparison with the subtle differences of the transverse strain distributions (Figure 6.6), the stress distributions along the weld centerline behind the weld pool are not very sensitive to the boundary condition changes. This can be explained by the fact that the high temperatures experienced in these areas causes the yield strength to be very low in comparison with rest of the plate where the temperature is low. Although the stresses in these areas are quite close to the corresponding yield strengths, they are still substantially lower than the rest of the areas. This fact also suggests that the use of stress as the parameter for expressing the solidification cracking criteria appears to be impractical since stress is insensitive to the deformation process experienced in the solidification temperature range. As reasoned in Chapter IV, the hot strain (mechanical strain) is a much better choice for this purpose. Therefore, the rest of this chapter will focus on the discussion of the hot strain and the stress fields will not be further discussed.

In order to demonstrate the subtle influence of the solidification process on the strain field in the solidified weld metal, Figure 6.7 presents the finite element
analysis results of transverse hot strain distribution for a model in which element rebirth procedure is not used. In other words, no solidification process is considered in the model so that the entire finite element model is treated as a thermal stress problem based on continuum solid mechanics. The model was fixed along the top edge of the plate. Therefore, except for the element rebirth procedure, this model is identical to the bottom one in Figure 6.5. It is clear that there is substantial compression plastic deformations inside the weld pool, and they are not annealed during the resolidification process. This results in substantial compressive plastic strains in the resolidified weld metal – the magnitudes of this plastic strain are more than 30 times the yielding strain.

So far, the attention has been on the overall temperature, stress and strain fields in the plates. We will next focus on the hot strain distribution along the weld centerline, which matters the most for the centerline solidification cracking.
Figure 6.2 — Transient temperature distributions in the Sigmajig test. 15V, 93A, and 5 mm/second. Top: at 0.5 second; Bottom: at 1.0 second.
Figure 6.3 — Transient temperature distributions in the Sigmajig test. 15V, 93A, and 5 mm/second. Top: at 2.0 seconds; Bottom: at 3.0 seconds.
Figure 6.4 — Transient temperature distributions in the Sigmajig test. 15V, 93A, and 5 mm/second. Top: at 4.0 seconds; Bottom: at 5.0 seconds.
Figure 6.5 — Transverse stress fields at 5 seconds. Top: no restraint; Bottom: the edges parallel to the weld bead (the top edge in the model) is fixed.
Figure 6.6 — Transient transverse hot strain distribution at 5 seconds. Top: no restraint; Bottom: top edge of the plate is fixed.
Figure 6.7 — Transverse hot strain distribution at 5 seconds. Element rebirth technique is not used to properly model the solidification effect. The top edge of the plate is fixed.
Hot Strains in the Solidification Temperature Range

Although the general definition of hot strain has been made in Section 6.1, it becomes necessary to clarify the specific definition of the hot strain pertaining to the centerline solidification cracking, which is to be discussed here in details. For the rest of this chapter, we will adopt a fundamental assumption that seems universally accepted, either explicitly or implicitly, regarding the mechanical driving force for the centerline solidification cracking: the transverse strain (or stress) causes the centerline solidification cracking. Thus, unless otherwise stated, the hot strain for the centerline solidification cracking is specifically referred to as the transverse mechanical strain experienced along the centerline of the weld metal in the solidification temperature range.

Figure 6.8 shows the effect of restraint on the hot strain distribution along the weld centerline at 5 seconds. In the first case (labeled as Free BC in the figure), the plate was set to be free of any displacement restraint along the four edges of the plate. In the other case, the two edges of the plate parallel to the weld were restrained from any displacement and labeled as Fixed BC. Note that these curves are equivalent to the hot strain curves at a given position along the centerline of the weld that is needed in Figure 6.1, provided that the quasi-stationary state for both the temperature and the stress-strain fields are established in the plate. The effect of the solidification process is also apparent in the figure by comparing the hot strain distributions under the same restraint condition. The
strain-free state due to resolidification cannot be appropriately modeled if the element rebirth technique is not used – the hot strains at the moment of resolidification are about -3.0% and -8.5% for the two restraint conditions that were modeled in Figure 6.8.

The striking feature shown in Figure 6.8 is that the predicted hot strain is substantially lower when the plate is constrained 25 mm away from the weld centerline, than the case without the constraint. At first glance, this prediction seems quite contradictory to the commonly accepted view of the effect of

Figure 6.8 — Effect of restraint on the hot strain distribution at 5 seconds. The necessity of element rebirth technique for properly modeling the effect of solidification process is also clearly demonstrated.
restraint on the solidification cracking tendency. However, this prediction can be well accounted for by the thermal expansion of the base metal. First of all, the temperature of the base metal and the heat-affect-zone is still continuously rising when the welding arc has just passed by, as evidenced in Figure 6.4. Hence the base metal and the heat-affected-zone are still in the expansion state. Secondly, when no restraint is applied to the plate, the thermal expansion of the base metal can be relaxed either at the free boundaries or in the weld metal since it is much softer than the base metal, whereas all the thermal expansion can only be relaxed in the weld metal. In other words, the thermal expansion will push the weld fusion boundary inward to the centerline of the weld metal, as evidenced by the movement of the fusion line discussed in the previous chapter (Chapter V), and this inward movement will be greater if the plate is constrained. Clearly, such an inward movement of base metal will reduce the magnitude of the tensile strains in the weld metal. In fact, Figure 6.8 even shows that, for the restrained plate, a minimal compression – when the weld metal is just resolidified from the liquidus temperature – is gradually changing to tension as the weld metal continuously cools down.

There are documented observations that support the prediction in Figure 6.8. Again, Matsuda et al.’s investigation [51] on the moving characteristics of weld edges is used here as an example of experimental observation. When the weld was started from the edge of the plate, solidification cracking could be
produced easier in a narrow plate than in a wide plate. In fact, Matsuda et al. mentioned that they could not produce solidification cracking if the width of a plate was over 150 mm. Furthermore, Matsuda et al. also showed that the inward transverse displacements of the weld edge were much larger for a plate of 50 mm wide than for a 100 mm or 150 mm wide plate when the weld started 30 mm from the edge of the weld (no solidification cracking was produced in these cases). This indicates that the mechanical driving force, i.e., the hot strains in the solidification temperature range for a narrow plate, is greater than in a wide plate.

Figure 6.9 presents the evolution of hot strains at four positions along the weld centerline of the finite element model whose boundary conditions were set to be free of any restraint from the fixture. Their positions are identified by the distance, x, from the left edge of the plate where the weld starts, as shown in Figure 4.13.

In Figure 6.9, the hot strain curve for a particular point is plotted by monitoring both the temperature drop and the build-up of the hot strain as that point is cooling down from the liquidus temperature. This curve tells how the mechanical driving force for solidification cracking varies during the solidification process, according to the hot strain versus threshold strain hypothesis depicted in Figure 6.1. The procedure for extracting the hot strain curve in the solidification
temperature range from the finite element analysis results has been discussed in Chapter IV and will not be repeated here.

The first observation from Figure 6.9 is quite straightforward: a dendritic grain located on the weld centerline will experience a gradually increasing hot strain as it cools from the liquidus temperature. More importantly, Figure 6.9 suggests that a dendrite located at the left edge of the plate (x=0mm) where welding is initiated will experience the highest hot strain. The greater the distance between a location of interest and the left edge of the plate, the lower the

![Figure 6.9 — Hot strain curves for 4 locations on the weld centerline. They have different distance, X, from the left edge of the plate where the weld starts. No restraint, T_L=911K, T_c=865K, T_s=775K.](image-url)
magnitude of the hot strain will be. This leads to the prediction that solidification cracking is more likely to start at the edge of the plate under the conditions that the plate is free of restraint. This prediction agrees with the common consensus that, if a weld is made from the edge of a plate which has uniform width and is free of any restraint from the fixture, the crack often initiates at the beginning of the weld run (the end cracking [135,136]).

A related, and almost universally observed phenomenon regarding the cracking tendency of a bead-on-plate weld is the influence of the location at which a weld starts. In practice, it is found that cracking is likely to occur if a weld is started from the edges of a plate. To avoid this problem welders often begin the weld an inch or so from the edge, and complete the unwelded portion later. This phenomenon has been utilized in designing the Houldcroft test [134]. Matsuda et al. [51] also used it to measure the moving characteristics of weld edge in cracked and crack-free plates.

According to the hot strain versus threshold strain hypothesis, one should expect that different weld start locations experience different hot strains: the hot strain should be higher when the weld is started from the edge of the plate than from a position away from the edge. Although the hot strain curves shown in Figure 6.9 can be used to explain such a phenomenon, it would be more appropriate to directly compare the hot strains at the different weld start locations. Figure 6.10 presents such a comparison of the hot strain curves obtained from the finite
element analysis results. The two finite element models are identical in all other aspects except for the weld start points: the weld is started from the edge of the plate ($x_o=0$mm) in the first model and from the point that is 12.5 mm (1/4 of the plate length) from the edge ($x_o=12.5$mm) in the second model. No restraint (fixture) is imposed in both models. In Figure 6.10, the continuous lines represent the hot strain build-up at the weld start points ($dx=2.5$mm), whereas the dashed lines show the hot strains 10 mm from the weld start point ($dx=10$mm). A comparison of the continuous lines clearly reveals that the hot strain is much higher when the weld is initiated at the edge of the plate. Once again, another

![Graph showing the effect of weld starting position on hot strain curves.](image)

Figure 6.10 — Effect of the weld starting position on the hot strain curves. No restraint, $T_L=911$K, $T_c=865$K, $T_s=775$K.
commonly documented observation regarding the solidification cracking tendency (the weld start location effect) has been well explained by the differences in hot strains as obtained from the finite element models.

Having demonstrated the consistency between the predictions from the hot strain versus threshold strain approach and the corresponding reported solidification cracking tendencies, we will now show that the magnitudes of the predicted hot strain curves are also credible.

As mentioned in the beginning of this chapter, there has not been any reported experimental measurement of the strain evolutions inside the solidification temperature range. Thus the hot strain curves in Figure 6.9 cannot be directly compared with experimental results. However, it is possible to examine the magnitude of the hot strains in the solidification temperature range, based on Matsuda et al.'s experimental measurement of solidification cracking ductility curves by means of the MISO technique \[^{10,54,43}\]. As reviewed in Chapter III, Matsuda et al. in these reports investigated the local ductility in the vicinity of the solidification crack tips during crack initiation and propagation. With an optical microscope and an high speed movie camera, they were able to measure the relative distance change of two reference marks right behind the solidification front that were 0.9 to 1.7 mm apart from each other, faced each other across the crack initiation site, and were parallel to the loading direction. The gauge length in the MISO technique should be short enough that the strains obtained with such a
gauge length are comparable to those calculated from the finite element analysis. In fact, Matsuda et al. pointed out that further decrease in the gauge length would scatter the measured data, due to nonuniform strain distribution at grain boundaries. Since Matsuda et al. used a relatively rapid load to create the solidification cracking, the experimental procedure was not capable of tracing the “natural” strain build-up at a particular location as it is solidifying from the liquidus temperature. Instead, the ductility curves under various loading conditions were presented for various alloys such as stainless steels, carbon steels, Inconel alloys and aluminum alloys. In general, the ductility values for crack initiation are within 2 to 8 percent, depending upon the composition, welding condition, and loading speed.

If one accepts that the values of the threshold strains are close to the ductility values measured by Matsuda et al., (i.e., in the neighborhood of 2 to 8 percent), then these values provide the ballpark for checking the magnitude of the predicted hot strains. What needs to be done is to construct two finite element models to simulate two weldability tests which have the same threshold strain curve but different hot strain curves. One test will definitely crack because the hot strains experienced exceed the threshold strain, and the other will definitely not crack because the hot strains are less than the threshold strain. Note that the threshold strains are a function of temperature and their exact values are unknown but should be somewhere between 2 to 8 percent.
The requirement for proving the correctness of the hot strain calculation procedure and the magnitude of the predicted hot strains is that, according to equations (6.1) and (6.2), the following conditions need to be satisfied: (1) the hot strains predicted in the first model (the cracked one) must be greater than the threshold strain at least at one temperature in the BTR, and (2) the hot strains predicted from the second model (the crack-free one) must be less than the threshold values for the entire BTR.

The two finite element models used in Figure 6.10 are ideal candidates for checking the magnitude of the predicted hot strains in the solidification temperature range. In the first model, a weld is made along the centerline of the plate and started from the edge of the plate. The centerline solidification cracking would certainly take place in this model, as indicated by the observation of Matsuda et al. [51] on the influence of the plate width and the weld starting location on the centerline solidification cracking tendency in studying the moving characteristics of weld edge. For the same reason, it is expected that the centerline solidification cracking will not occur in the second model in which the weld is started 12.5mm from the edge.

According to Figure 6.10, the predicted hot strains in the first model are actually higher than the upper bound value of the threshold strain (8 percent) before the solidus temperature has been reached, indicating the cracking will definitely take place in the test simulated by the first model. On the other hand,
the predicted hot strains in the second model are so low (lower than but close to the lower bound threshold value) that the centerline solidification cracking should not take place.

Based on the above discussion, it is concluded that the hot strain calculation procedure developed in this dissertation is capable of providing at least a reasonable magnitude for the hot strains experienced during real weld situations.
CHAPTER VII

CONCLUSIONS AND RECOMMENDATIONS

7.1 Summary and Conclusions

In this dissertation, a systematic approach for quantitative assessment of the mechanical driving force of weld metal solidification cracking phenomenon has been proposed, developed and verified.

Weld metal solidification cracking, like many other cracking problems, results from the competition between the mechanical driving force and the material resistance to the cracking. Many investigations in the past have established that the weld metal solidification cracking occurs in the solidification temperature range: the mechanical driving force is the strain accumulated as the weld metal solidifies from the liquidus temperature, and the material resistance is the ductility curve in the brittle temperature range. It is this fundamental mechanical driving force versus material resistance concept that forms the basis of this research.

The extensive literature review (Chapter II) reveals that there has not been a systematic approach to quantitatively evaluate the stress-strain condition for the solidification cracking problems. This lack of quantitative assessment of the
mechanical driving force has become the most critical issue for the further advancement of weld metal solidification cracking research and prevention. Such quantitative assessment not only provides much needed information about the stress-strain conditions for cracking in a laboratory weldability test, but also acts as a bridge to link the results of laboratory weldability tests and real welding fabrication situations.

This research has attempted to address this critical issue by (1) proposing the hot strain versus threshold strain hypothesis to extend the mechanical driving force versus material resistance concept so that the cracking can be quantitatively evaluated; (2) developing a finite element modeling procedure capable of quantitatively determining the hot strain experienced during a welding process.

The hot strain is the mechanical strain accumulated in the solidification temperature range. The threshold strain is a quantitative measure of the material resistance to solidification cracking whose value is equal to that of the hot strain experienced at the crack initiation location at the moment of crack initiation. A solidification crack-free weld can be made only if the following condition is satisfied in the entire solidification temperature range:

\[ \varepsilon_h(T) < \varepsilon_c(T) \]  \hspace{1cm} (7.1)

and the criterion for incipient crack growth can be expressed as

\[ \varepsilon_h(T) = \varepsilon_c(T) \]  \hspace{1cm} (7.2)
The hot strain calculation procedure consists of the finite element modeling of the temperature and the thermal stress-strain fields in the vicinity of a weld pool, and also the construction of hot strain curves in the solidification temperature range based on the finite element analysis results.

Two new modeling techniques have been developed in order to provide more realistic temperature and stress-strain fields in the vicinity of a weld pool. By using a micro-dendritic grain growth model which describes the relationship between the fraction of solid and the temperature in the solidification temperature range, microscopic solidification kinetics has been incorporated into the heat transfer analysis for improved modeling of latent heat of fusion. In the thermal stress-strain analysis, an element rebirth scheme has been developed to appropriately handle the effects of solidification process on the stress-strain distributions in the resolidified weld metal: the deformation in the molten weld pool, the change of the reference temperature for thermal strain calculation in resolidified weld metal, and the solidification shrinkage.

The major effort of this dissertation has been focused on the development and verification of the proposed methodology. In this regard, the hot strain calculation procedure and the underlying finite element models have been applied to two-dimensional problems of full penetration bead-on-plate welds deposited on thin aluminum plates by the autogenous gas tungsten arc welding process. The following conclusions are based on these studies:
On the New Modeling Techniques

(1) The inclusion of the micro-dendritic grain growth model in the heat transfer analysis significantly changes the temperature field within the solidification temperature range behind the welding arc where it matters the most insofar as the solidification cracking is concerned, although it has little influence on the temperature distribution in the rest of the plates.

(2) Element rebirth techniques are needed in order to adequately handle the weld solidification effect on the strain field in the resolidified weld metal. When the element rebirth technique is not used, the conventional finite element thermal stress model based on continuum solid mechanics will lead to a substantial compressive plastic deformation being accumulated in the resolidified weld metal.

On the Verification of the Finite Element Modeling Procedure

(3) The weld pool size and weld bead width predicted from the heat transfer models, when considering the microscopic solidification kinetics, are in excellent agreement with the experimentally measured values.

(4) The predicted strain fields in the neighborhood of the weld pool and the moving characteristics of weld fusion lines are also in very good agreement with the experimental measurement.
On the Determination of Hot Strain

(5) Based on the hot strains determined by the finite element modeling procedure developed in this study, some experimentally observed weld metal solidification cracking tendencies are well explained by the hot strain versus threshold strain hypothesis.

(6) Although no experimental results are available to allow direct verification of the predicted hot strain evolution in the solidification temperature range, this study has managed to demonstrate the credibility of the magnitude of the hot strains obtained from the finite element analysis.

In summary, this dissertation has demonstrated the validity of the hot strain versus threshold strain approach for the prediction of the weld metal solidification cracking and the underlying finite element modeling procedure.

7.2 Recommendations for Future Work

The hot strain versus threshold strain approach and the underlying finite element modeling procedure developed in this dissertation allows for quantitative assessment of solidification cracking problem in the design stage by comparing the hot strains in a structure to be welded with the alloy’s threshold strain (measured under a simple laboratory weldability test). However, much more work must be done before this approach can be successfully used as a
routine design procedure. Among many important subjects, the author suggests the following:

(1) *Design new weldability test(s).* Many of the today's weldability tests really evaluate the arrest, not the initiation, of solidification cracking, so the threshold strain for cracking initiation cannot be measured in these tests. They are also difficult to model in the finite element analysis. Thus, the new weldability test should not only be sensitive to the initiation of cracking and easy to test, but also easy to model.

(2) *Improve the efficiency of the finite element modeling procedure.* The current element rebirth scheme is very cumbersome to implement.

(3) *Apply the element rebirth technique to other fusion welding related stress and distortion problems.* The solidification effects which have been discussed in this research are universal to the fusion welding process, and have been rarely reported in the literature. Perhaps appropriately modeling the solidification effects could explain many discrepancies between the numerical simulation results and experimental measurements (for example the unexplained weld metal stress dip that often appears in the finite element models).
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