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COMPUTATIONAL-ANALYSIS ASSISTED INTRODUCTION OF FRICTION STIR WELDING INTO DEVELOPMENT OF LIGHT-WEIGHT HIGH-SURVIVABILITY MILITARY VEHICLES

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COMPUTATIONAL-ANALYSIS ASSISTED INTRODUCTION OF FRICITION STIR WELDING INTO DEVELOPMENT OF LIGHT-WEIGHT HIGH-SURVIVABILITY MILITARY VEHICLES

A Dissertation
Presented to
the Graduate School of
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In Partial Fulfillment
of the Requirements for the Degree
Doctor of Philosophy
Mechanical Engineering

by
Guruprasad Arakere
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Accepted by:
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Dr. Rajendra Singh
ABSTRACT

High strength aluminum alloys with superior blast/ballistic resistance against armor piercing (AP) threats and with high vehicle light-weighing potential are being increasingly used as military-vehicle armor. Due to the complex structure of these vehicles, they are commonly constructed through joining (mainly welding) of the individual components. Unfortunately, these alloys are not very amenable to conventional fusion based welding technologies (e.g. Gas Metal Arc Welding (GMAW)) and in-order to obtain high-quality welds, solid-state joining technologies such as Friction Stir Welding (FSW) have to be employed. However, since FSW is a relatively new and fairly complex joining technology, its introduction into advanced military vehicle underbody structures is not straight forward and entails a comprehensive multi-prong approach which addresses concurrently and interactively all the aspects associated with the components/vehicle-underbody design, fabrication and testing. One such approach is developed and applied in the present work. The approach consists of a number of well-defined steps taking place concurrently and relies on two-way interactions between various steps. In the present work, two of these steps are analyzed in great detail: (a) Friction Stir Welding process modeling; and (b) Development and parameterization of material models for the different weld-zones.

Within the FSW process modeling, interactions between the rotating and advancing pin-shaped tool (terminated at one end with a circular-cylindrical shoulder) with the clamped welding-plates and the associated material and heat transport are studied computationally using a fully-coupled thermo-mechanical finite-element analysis. To surmount potential numerical problems associated with extensive mesh distortions/entanglement, an Arbitrary Lagrangian Eulerian (ALE) formulation was used which enabled adaptive re-meshing (to ensure the continuing presence of a high-quality mesh) while allowing full tracking of the material free surfaces/interfaces. To demonstrate the utility of the present computational approach, the
analysis is applied to the aluminum-alloy grades, AA5083 (a solid-solution strengthened and strain-hardened/stabilized Al-Mg alloy) and AA2139 (a precipitation hardened quaternary Al-Cu-Mg-Ag alloy). Both of these alloys are currently being used in military-vehicle hull structural and armor systems. In the case of non-age-hardenable AA5083, the dominant microstructure evolution processes taking place during FSW are extensive plastic deformation and dynamic recrystallization of highly-deformed material subjected to elevated temperatures approaching the melting temperature. In the case of AA2139, in addition to plastic deformation and dynamic recrystallization, precipitates coarsening, over-aging, dissolution and re-precipitation had to be also considered. To account for the competition between plastic-deformation controlled strengthening and dynamic-recrystallization induced softening phenomena during the FSW process, the original Johnson-Cook strain- and strain-rate hardening and temperature-softening material strength model is modified using the available recrystallization-kinetics experimental data. Lastly, the computational results obtained in the present work are compared with their experimental counterparts available in the open literature. This comparison revealed that general trends regarding spatial distribution and temporal evolutions of various material-state quantities and their dependence on the FSW process parameters are reasonably well predicted by the present computational approach.

The introduction of newer joining technologies like the so-called Friction Stir Welding (FSW) into automotive engineering entails the knowledge of the joint-material microstructure and properties. Since, the development of vehicles (including military vehicles capable of surviving blast and ballistic impacts) nowadays involves extensive use of the computational engineering analyses (CEA), robust high-fidelity material models are needed for the FSW joints. A two-level material-homogenization procedure is proposed and utilized in the present work in-order to help manage computational cost and computer storage requirements for such CEAs. The method
utilizes experimental (microstructure, micro-hardness, tensile testing and X-ray diffraction) data to construct: (a) the material model for each weld zone; and (b) the material model for the entire weld. The procedure is validated by comparing its predictions with the available experimental results and with the predictions of more-detailed but more costly computational analyses.

**Keywords:** Friction Stir Welding; Process Development; AA2139; AA5083; Blast-survivable and Ballistic Threat-resistant Military Vehicles; Material Model; Finite Element Analysis.
DEDICATION

This thesis would have been impossible without the unwavering love and support from my father, Mr. Udayashankar Arakere, my mother, Mrs. Suma Shankar, my brother Mr. Ajay Prasad Arakere and my wife Mrs. Shreya Prakash, to whom this thesis is dedicated.
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CHAPTER 1

INTRODUCTION

1.1. Background

In order to respond to the new enemy threats and warfare tactics, military systems, in particular those supporting the U.S. ground forces, are being continuously transformed to become faster, more agile, and more mobile so that they can be quickly transported to operations conducted throughout the world. Consequently, an increased emphasis is being placed on the development of improved lightweight vehicle-armor systems as well as on the development of new high-performance armor materials/structures. Therefore, a number of research and development programs are under way to engineer light-weight, highly mobile, transportable and lethal battlefield vehicles with a target weight under 20 tons. To attain these goals, significant advances are needed in the areas of light-weight structural- and armor-materials development (including aluminum-based structural/armor-grade materials). Due to the complex structure of the military battle-field and tactical vehicle underbodies, the use of aluminum alloy components generally requires component joining by welding. Unfortunately, the high-performance aluminum grades used in vehicle-armor applications are normally not very amenable to conventional fusion-based welding technologies. The problems associated with fusion welding of the advanced high-strength aluminum alloys used in military-vehicle applications can be overcome through the use of solid-state joining technologies such as Friction Stir Welding (FSW).

However, FSW is a relatively new, fairly complex and expensive joining technology. Its introduction into design/development of advanced military vehicle structures is not straightforward and involves very expensive and time consuming build and test experimental approaches, significantly increasing the vehicle design lead time and resulting in vehicles with deficient
blast/ballistic survivability. However, advancements in computer hardware and software could provide computational modeling as a viable tool for the introduction of the FSW process into the design/development of advanced military vehicle structures in a time- and cost-efficient manner. Consequently, the focus of the present work is the development of a new fully-integrated approach for the concurrent design, FSW-based manufacturing and testing of high-survivability military vehicle-underbodies, using both computational and experimental techniques. The present approach addresses concurrently and interactively aspects associated with the vehicle design, manufacturing and blast-survivability performance testing. Since blast-survivability and ballistic resistance (destructive) testing of full-size military vehicle-underbodies is quite costly and time consuming, it is commonly replaced with the corresponding fabrication/testing of sub-scale (look-alike) test structures. Consequently, within the present work attention will be given to the fabrication and testing of such sub-scale structures and not to the full-scale vehicle-underbodies. This approach contains a number of discrete steps, these steps are carried out concurrently and multiple iterations/interactions between different steps are encountered.

1.2. Literature Review

The relevant literature survey for each of the sub-topics covered in the present work is provided in Chapters 2-5.

1.3. Thesis Objective and Outline

The overall objective of the present work was to develop a comprehensive multi-prong computer aided engineering (CAE) based approach which addresses concurrently and interactively all the aspects associated with the FSWed sub-scale military-vehicle underbody design, fabrication and testing. Toward that end, a combined computational and experimental approach consisting of a number of well-defined steps taking place concurrently and relying on
two-way interactions between various steps is developed. Application of the developed approach could enable the low-cost, short lead-time development of blast-resistant vehicle underbody structures. The organization of the present work is as follows:

In Chapter 2, a detail fully-coupled thermo-mechanical finite-element computational investigation of the effect of various FSW-process parameters on the heat and mass transport of the material and on the microstructure evolution for the case of the AA5083 wrought aluminum alloy was conducted. An effort was made to more accurately account for the competition between strain-hardening and dynamic-recrystallization processes in this alloy during the FSW process. While previous investigations recognized the effect of plastic strain, strain rate and temperature on the material strength, only reversible effects of the temperature were accounted for. This shortcoming was rectified in the present work by recognizing that, via dynamic-recrystallization, exposure of the material to high temperatures may result in permanent microstructure/property changes.

In Chapter 3, the fully-coupled thermo-mechanical finite-element FSW process model developed in Chapter 2, was combined with the basic physical metallurgy of two wrought aluminum alloys, AA5083-H131 (a solid-solution strengthened and strain-hardened/stabilized Al-Mg alloy) and AA2139 (a precipitation hardened quaternary Al-Cu-Mg-Ag alloy), to predict/assess their FSW behaviors. The operation and interaction of various microstructure-evolution processes taking place during FSW (e.g. extensive plastic deformation, dynamic recrystallization, precipitates coarsening, over-aging, dissolution and re-precipitation) was considered to predict the material microstructure/properties in the various FSW zones of the two alloys.

In Chapter 4, a comprehensive multi-prong approach which addresses concurrently and interactively all the aspects associated with the FSWed components/vehicle underbody design,
FSW-based manufacturing and testing is developed and applied in the present work. The approach consists of a number of well-defined steps taking place concurrently and relies on two-way interactions between various steps. The computational models developed in the earlier chapters are used within the developed approach to predict the FSW behavior of the material under consideration. The developed approach is critically assessed using a SWOT (Strengths, Weaknesses, Opportunities and Threats) analysis.

In Chapter 5, a new two-level weld-zone homogenization procedure is developed and implemented in order to reduce the memory/storage requirements and increase the computational speed of computer-aided transient non-linear dynamics engineering analyses. Within the first level of homogenization, homogenized effective mechanical properties are determined for each FSW zone. Within the second level of homogenization, homogenized properties of the entire FSW-joint local cross-section are computed. The procedure is validated against the results of the computational analyses in which weld zones are accounted for explicitly and against the available experimental results.
CHAPTER 2
NUMERICAL INVESTIGATION OF THE FRICTION-STIR WELDING OF
AA5083 AND THE MATERIAL EVOLUTION PROCESS

2.1. Abstract

Computational based investigation of the interaction between the rotating and advancing sides of a pin-shaped tool with the clamped welding-plates and the resulting material and heat transport during the Friction Stir Welding (FSW) process using a fully-coupled thermo-mechanical finite-element analysis is conducted. The numerical problems associated with the excessive material deformation due to the finite-element mesh distortions is overcome through the use of an Arbitrary Lagrangian Eulerian (ALE) formulation which enables the adaptive re-meshing (to ensure the presence of a high quality mesh) while allowing the material free surfaces to be tracked. The utility of the developed computational procedure is demonstrated for the case of FSW of AA5083 (a solid-solution strengthened and strain-hardened Al-Mg wrought alloy). The competition between the plastic-deformation induced material strengthening and the dynamic-recrystallization induced material softening taking place during the FSW process is accounted for through a modified Johnson-Cook (accounts for the large-strain, large strain-rate hardening and temperature-softening) material strength model using the available recrystallization-kinetics experimental data. Finally, the obtained numerical results are compared with their experimental counterparts available in the open literature. The developed computational approach well predicted the overall general trends in the spatial distribution and temporal evolutions of the different material-state quantities and their dependence on the FSW process parameters.
2.2. Introduction

Friction-stir welding (FSW) is a solid-state material joining process employed for metallic and polymer-based materials in applications in which the original material microstructure/properties must remain unchanged as much as possible after joining [2.1–2.3]. The FSW process consists of a rotating tool moving along the contacting surfaces of two rigidly butt-clamped plates, as displayed in Figure 2-1(a). As displayed in the figure, the FSW tool consists of two main parts, a threaded cylindrical pin, at one end, and equipped with a shoulder, at the other. The work-pieces to be welded are firmly clamped, as well as placed on a rigid backing support. Meanwhile, the FSW-tool shoulder is makes a firm contact with the top-most surface of the work-piece. As the rotating tool translates along the butting work piece surfaces, the tool-shoulder generates heat at the shoulder/work-piece interface, to a lesser extent, at the pin/work-piece contact surfaces, as a result of the frictional-energy dissipation. This causes an increase in work piece material temperature resulting in the softening of the material adjacent to the work piece/tool interface. The subsequent translation of the tool along the butting surfaces causes the thermally-softened material in front of the tool to be transferred (i.e. extruded around the tool) to the wake of the tool and compacted/forged at the wake to form a joint/weld. Other than butt joint welding, the FSW process is often used for lap- as well as T- joints.

The FSW process has been the preferred welding technique for aluminum components and its applications in the welding of other difficult-to-weld metals is slowly expanding. Several industrial sectors such as shipbuilding and marine, aerospace, railway, land transportation, etc. are currently employing the FSW process.

The FSW process offers a number of advantages when compared to the commonly used fusion welding technologies, such as: (a) excellent as-welded mechanical properties of the weldment; (b) improved safety, since, toxic fumes or the molten material spatter is absent; (c)
absence of welding consumables such as the gas shield or filler metal; (d) ease of process automation; (e) capability of the process to operate in several positions (e.g. horizontal, vertical, overhead, orbital, etc.), due to the absence of a weld pool; (f) reduces the need for expensive post-weld machining activities due to minimal weld-thickness under/over-matching; and (g) low environmental impact. In spite of the several advantages, the FSW process has some disadvantages such as: (a) presence of an exit hole in the work piece left upon tool withdrawn; (b) large tool-vertical and plates-clamping forces are required; (c) welding of variable-thickness and non-linear welds is difficult; and (d) the process involves lower welding rates compared to conventional fusion-welding technologies.

The FSW process involves the presence an advancing side of the weld (i.e. a side in the work piece whose circumferential velocity of the rotating tool is in the same direction as the tool traverse direction) and the retreating side (i.e. the side in the work piece on which the two velocities are in opposite directions). The presence of the advancing and retreating sides in the work piece results in an asymmetry in the heat transfer, material flow and weld microstructure-properties [2.4].

The FSW process involves extremely complex relationships and contest between the associated thermo-mechanical processes involving frictional energy dissipation, plastic deformation, heat dissipation, material flow, dynamic recrystallization, etc. [2.5-2.8]. Examinations of the weld region generally reveal the presence of the following four zones, Figure 2-1(b): (a) a base-metal/un-effected zone far away from the weld where no material microstructure/property changes take place; (b) the heat-affected zone (HAZ) where material microstructure/properties are influenced by the thermal effects associated with the FSW process. The HAZ zone is mostly found in the case of fusion-welds. However, the material microstructural changes in the case of FSW is quite different due to the presence of lower
temperatures and a diffuse heat source; (c) the thermo-mechanically affected zone (TMAZ) which is located closer to the butting surfaces compared to the HAZ. The material microstructure/properties in the TMAZ are affected by the thermal as well as the mechanical aspects of the FSW process. Generally, the original grains within this weld-zone undergo severe plastic deformation; and (d) the weld-nugget which is the inner-most zone of an FSW joint. The weld-nugget contains the so-called “onion-ring” features which are a result of the way material is transported from the regions ahead of the tool to the wake regions behind the tool. The work piece material within the weld-nugget contains a very-fine dynamically-recrystallized (i.e. equiaxed grains) microstructure, since this zone is subjected to the most extreme conditions of plastic deformation and high temperature exposure.

An important feature in the FSW process is that heat transfer takes place through thermal conduction as well as transport of the work-piece material from the region in front of the tool to the region behind the translating tool. The work-piece material properties, FSW tool geometry and the FSW process parameters significantly influence the heat and the mass transfer process. The material transport in the FSW process is accompanied by severe plastic deformation and dynamic recrystallization of the transported material. The material strain rates involved the FSW process [2.13, 2.14] is generally around 30 s$^{-1}$.

The weld quality and process efficiency are influenced the following key FSW process parameters: (a) FSW-tool rotation and traverse velocities; (b) tool plunge depth; (c) tool tilt-angle; and (d) tool-design and material. It is critical to achieve a delicate balance between the FSW-tool rotation and traverse speeds, since, low work piece temperatures results in insufficient material softening causing material flaws due to low ductility of the material. On the other hand, high work piece temperatures results in significant material microstructure/property changes as well as incipient-melting flaws. The FSW-tool plunge depth (defined as the depth of the lowest
point of the shoulder) is an important process parameter which ensures that the required level of shoulder/work-piece contact pressure is attained so that the tool completely penetrates the weld. Insufficient tool-plunge depths typically results in low-quality welds due to inadequate material forging at the rear of the tool, while excessive tool-plunge depths usually leads to weld undermatching i.e. the weld thickness lower compared to that the base material thickness. It has been found experimentally, that rearward tilting of the tool by about 2-4 degrees improves the effect of the forging process.

The past two decades have seen considerable experimental research efforts towards gaining a better understanding of the FSW joining mechanisms and the evolution of the welded-materials microstructure/properties [2.15-2.18] as well as to explain the effect of various FSW process parameters on the weld quality/integrity [2.19-2.23]. Although the experimental efforts were able to correlate the welded-materials properties/microstructure with the FSW process parameters they provide very little real time understanding of the physics of heat/mass transfer and microstructure-evolution processes. Therefore, it is hoped that a good level of understanding of the underlying mechanisms can be gained by carrying out a detailed computational investigation of the FSW process. This chapter attempts to provide one such example of the application of the developed computational model.

A detailed review of the available literature revealed a number of prior research efforts dealing with the computational investigations of the FSW process. The work by Zhang and others involved the development of a semi-coupled thermo-mechanical finite-element investigation of the FSW process [2.29-2.31]. A number of computational solid mechanics and computational fluid dynamics based efforts were reported in the literature whose main objective was to investigate the effect of various FSW process parameters on the resulting heat/mass transport [e.g. 2.29-2.33].
Development of a detail finite-element based computational procedure to investigate the effect of various FSW-process parameters on the heat/mass transport of the work piece material and on the associated microstructure/property evolution for the case of the AA5083 wrought aluminum alloy is presented in Chapter 2. While similar investigations have been carried out by other researchers [2.29-2.31], an effort is made in the present work to more accurately account for the competition between material strain-hardening and dynamic-recrystallization processes present in this alloy during the FSW process. Though the prior investigations correctly accounted for the effect of plastic strain, strain rate and temperature on the material strength, the effects of the temperature on the strength were assumed to be reversible process. The present work rectifies the above mentioned shortcoming by recognizing that, due to the presence of the dynamic-recrystallization phenomena and exposure of the weld-material to high temperatures may result in permanent microstructure/property changes.

The organization of this chapter is as follows: In several subsections of Section 2.3, details are provided regarding the formulation of the problem, fully-coupled thermo-mechanical finite-element analysis and its integration, work-piece material models, tool/work-piece contact algorithm and the arbitrary Lagrangian Eulerian adaptive-meshing method. The key results obtained in the present work are presented and discussed in Section 2.4, while the key conclusions resulting from the present study are summarized in Section 2.5.
Figure 2-1. (a) A typical set-up of the Friction Stir Welding (FSW) process; and (b) The common microstructural zones associated with the typical FSW joint.
2.3. Computational Approach

2.3.1. Problem Definition

As mentioned earlier, the primary objective within this chapter is to develop a detail finite-element computational investigation of the FSW process. Relatively simple work-piece and tool geometries are employed, since, the main purpose of the investigation is to enable establishment of the basic relations between the main FSW process parameters and the work-piece materials flow pattern. Since the FSW process is often employed for joining aluminum alloys, a prototypical aluminum alloy AA5083 in a H131-temper condition is used as the work-piece material. An overview of this alloy and its H131-temper condition are provided later in the section.

2.3.2. Computational Models

Figures 2-2(a)-(b) displays the dimensioned geometrical models for the work-piece and the FSW-tool. In order to simplify the computational model, the work-piece is assumed to be a single part with “a perfect clamping” condition. The work-piece radius and thickness are 40.0mm and 3.0mm, respectively. The circular work-piece displayed in Figure 2-2(a) represents a circular region surrounding the FSW-tool, in an otherwise infinitely long FSW work-piece due to the use of an Arbitrary Lagrangian Eulerian (ALE) formulation. The plate is modeled with a concentric circular through-the-thickness hole of radius 3.0mm. The FSW tool consisting of two parts is modeled as a 3.0mm-radius cylindrical pin on the lower section and a 9.0mm-radius circular-disc shaped upper shoulder section. An inclination angle of 80.5 degrees (with respect to the vertical axis of the tool) is present at the bottom surface of the shoulder.

The finite-element model of the work-piece consists of c.a. 9,000 first-order eight-node reduced-integration hexahedral thermo-mechanically coupled solid elements, while the tool was
meshed using c.a. 2,000 first-order four node reduced-integration rigid-shell elements. A single-node heat-capacity element was employed to model the thermal properties of the tool. Figures 2-3(a)-(b) shows the finite-element meshed models for the work-piece and the tool, respectively.

Figure 2-2. Dimensioned geometrical models for the: (a) FSW-tool; and (b) FSW work-piece.
Figure 2-3. Finite-element meshed models for: (a) FSW tool; and (b) FSW work-piece
2.3.3. Thermo-mechanical Finite Element Computational Analysis

A coupled thermo-mechanical finite-element analysis was developed and employed to investigate the FSW process. A thermo-mechanical analysis involves both the nodal velocities and nodal temperatures as part of the nodal degrees of freedom. Within such an analysis, the solid-mechanics and heat-transfer parts of the analysis are fully-coupled. That is, the work of plastic deformation and that associated with frictional sliding are considered as sources of heat generation within the thermal problem, while, the effect of local temperature on the mechanical aspect of the analysis is accounted for through the use of temperature-dependant work-piece material properties.

The computational model employs the following initial conditions at the beginning of the analysis: (a) fixed rotational speed for the tool in the range of 200-400rpm; (b) zero translational velocity for the tool, while the work-piece is assumed to be stationary; and (c) the tool and the work-piece are at an initial ambient temperature of 298K.

The computational model utilizes the following boundary conditions throughout the analysis: (a) the work-piece material at bottom surface is constrained in the through-the-thickness direction; (b) the rotational speed of the tool is held at the same initial angular velocity; (c) a contact pressure of 70MPa is applied over the tool-shoulder/work-piece contact interface; and (d) the work-piece material is not translated along the weld-line during the first 2s. After the initial phase, the effect of tool translation along the weld-line is obtained by applying a constant material-flow velocity in the weld-line direction over the (in-flow) and (out-flow) boundaries of the work-piece. The external regions of the work-piece which are not in contact with the tool are specified with heat-convection thermal boundary conditions. The work-piece/air and the work-piece/backing-plate interfaces are assigned typical values for the heat transfer coefficient.
A relaxed hourglass-stiffness method was used in order to deal with the potential hourglassing problem due to the use of reduced integration elements as well as the incompressible nature of plastic deformation in the work-piece material.

The tool and the work-piece are allowed to interact over their contact surfaces. Specifically, contacts between the bottom surface of the tool-shoulder and the top surface of the work-piece as well as those between the outer surface of the pin and the work-piece hole were considered. Further details of the contact algorithm used are given in sub-section 2.3.5.

The numerical calculations are carried out using the ABAQUS/Explicit [2.34] finite-element program.

2.3.4. Material Models

No mechanical properties (except for the density) are specified for the tool material, since, the tool is considered to be a rigid body. While, its thermal capacity had to be specified, since, the tool was acquiring a portion of the heat generated as a result of tool-work-piece interfacial slip during the FSW process. Considering the fact that the tool is often made of hot-worked tool steel such as AISI H13, temperature-invariant thermal properties and density of this material were used to compute the thermal capacity of the tool [2.35].

As mentioned earlier, the aluminum alloy whose FSW behavior is analyzed is AA5083-H131. Often, age-hardened Al-alloys (e.g. AA6061-T6) are friction-stir welded, the material microstructure changes in these alloys is substantially more complex due to the unstable nature of its precipitates (i.e. precipitates can undergo partial or complete dissolution during alloy exposure to high temperature and can reappear upon cooling in different morphologies and number densities, and even precipitates with different crystal structures may appear). The alloy considered in the present work, AA5083 (nominal chemical composition: 4.5 wt.% Mg, 0.25 wt.% Cr, 0.75 wt.% Mn), is a Mg/Mn based solid-solution hardened alloy, which, in its H131
temper state is cold-work hardened and stabilized (to obtain a needed level of ageing/over-ageing resistance). Though, Al₆Mn precipitates are present in this alloy, due to the aforementioned stabilizing heat-treatment, they are relatively resistant to both dissolution and coarsening so that precipitate-portion of the material microstructure can be taken as mainly unchanged during the FSW process.

Both the thermal and mechanical properties had to be specified for the work-piece material. The work-piece made of an aluminum alloy, AA5083, consisted of the following temperature-invariant and microstructure-invariant thermal properties: (a) thermal conductivity, \(k=120\text{W/m} \cdot \text{K}\); (b) specific heat, \(c_p=880\text{J/kg} \cdot \text{K}\); and (c) density, \(\rho= 2700\text{kg/m}^3\).

The work-piece is modeled as an isotropic linear-elastic material and the materials plastic response is assumed to be strain-rate sensitive, strain-hardenable and (reversibly) thermally-softenable. With these types of materials, the mechanical response is represented by the following three relations: (a) a yield criterion, which is a mathematical relation which defines the condition which must be satisfied for the onset (and continuation) of plastic deformation; (b) a flow rule, which is a relation which describes the rate of change of different plastic-strain components during plastic deformation; and (c) a constitutive law, which is a relation which describes how the material-strength changes as a function of the extent of plastic deformation, the rate of plastic deformation and temperature. Further details of the mechanical model relations mentioned above is given below:

*Yield Condition*

The criterion for the onset of plastic deformation/yielding is given by the von Mises yield condition, according to which the equivalent stress \(\bar{\sigma}\) in the material must be equal to the material yield strength, \(\sigma_y\), i.e.:
\[ f(\sigma'_{ij}, \sigma_y) = \bar{\sigma} - \sigma_y = \sqrt{(3/2)\sigma'_{ij} \varepsilon'_{ij} - \sigma_y} \geq 0 \]  

(1)

where, \( f \) is the yield function, \( \sigma'_{ij} \) and \( \varepsilon'_{ij} \) are the stress and strain components, superscript ‘ is used to denote the deviatoric quantities.

**Plastic Flow Rule**

The flow rule used within the present work is associative, according to the rule, the plastic flow takes place in the direction of the stress-gradient of the yield surface as:

\[ \dot{\varepsilon}^{pl} = \frac{df}{d\sigma'_{ij}} \]  

(2)

where, superscript \( pl \) is used to denote a plasticity-related quantity, a raised dot denotes a time derivative and \( \lambda \) is a proportionality constant.

**Material Constitutive Law:**

Within the present work, the material yield strength was assumed to be controlled by strain, strain-rate-hardening as well as the thermally-activated slip-controlled thermal-softening effect which is assumed to be reversible process. Accordingly, the Johnson-Cook strength model [2.36] was employed as the constitutive law for the work-piece material considered. The computational modeling of the FSW process involves large material strain, high material deformation rate and high-temperature conditions which are well represented by the constitutive model considered in the present work. The yield strength according to the Johnson-Cook model is:

\[ \sigma_y = A + B(\bar{\varepsilon}^{pl})^n \left[ 1 + C_1 \log(\dot{\varepsilon}^{pl} / \dot{\varepsilon}^{pl}_0) \right] \left[ 1 - T^m_H \right] \]  

(3)
where, $\bar{\varepsilon}^{pl}$ is the equivalent plastic strain, $\dot{\bar{\varepsilon}}^{pl}$ the equivalent plastic strain rate, $\dot{\varepsilon}^{pl}_r$ a reference equivalent plastic strain rate, $A$ the zero-plastic-strain, unit-plastic-strain-rate, room-temperature yield strength, $B$ the strain-hardening constant, $n$ the strain-hardening exponent, $C_1$ the strain-rate constant, $m$ the thermal-softening exponent and $T_H=(T-T_{room})/(T_{melt}-T_{room})$ a room-temperature ($T_{room}$) dependent homologous temperature, while, $T_{melt}$ is the melting temperature. The temperatures are given in Kelvin. Table 2-1 provides a summary of the Johnson-Cook strength model parameters for AA5083.

The equivalent plastic strain evolution in the original Johnson-Cook strength model is assumed to be controlled completely by the plastic-deformation process. In the present work, the equivalent plastic-strain evolution is assumed to be controlled by competition between the plastic yielding of the material and dynamic-recrystallization. Since this represents one of the key contributions of the present work, details of the proposed modifications to the Johnson-Cook strength model are provided in the results and discussion section, Section III.

**Table 2-1. Johnson-Cook Strength Model Parameters for AA5083-H131 alloy**

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Symbol</th>
<th>Units</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Reference Strength</td>
<td>$A$</td>
<td>MPa</td>
<td>167.0</td>
</tr>
<tr>
<td>Strain-hardening Parameter</td>
<td>$B$</td>
<td>MPa</td>
<td>596.0</td>
</tr>
<tr>
<td>Strain-hardening Exponent</td>
<td>$n$</td>
<td>N/A</td>
<td>0.551</td>
</tr>
<tr>
<td>Strain-rate Coefficient</td>
<td>$C$</td>
<td>N/A</td>
<td>0.001</td>
</tr>
<tr>
<td>Ambient Temperature</td>
<td>$T_{room}$</td>
<td>K</td>
<td>293</td>
</tr>
<tr>
<td>Melting Temperature</td>
<td>$T_{melt}$</td>
<td>K</td>
<td>893.0</td>
</tr>
<tr>
<td>Temperature Exponent</td>
<td>$m$</td>
<td>N/A</td>
<td>1.0</td>
</tr>
</tbody>
</table>
Material Stress State Integration

During material loading, the stress is updated by integrating the rate-form of Hooke’s law:

\[ \Delta \sigma_j = C_{ijkl} \Delta \varepsilon_{kl} = C_{ijkl} \Delta \varepsilon_{kl} - C_{ijkl} \Delta \varepsilon_{kl} \]  \hspace{1cm} (4)

where, \( C_{ijkl} \) is the elastic-stiffness tensor of forth-order, and the total strain rate \( \dot{\varepsilon} \) is assumed to be comprised of its elastic, \( \dot{\varepsilon}^e \), and plastic, \( \dot{\varepsilon}^p \), components. At the end of each step during the loading process, the total strain rate is known (computed from the known velocity gradient).

Plugging in Eq. (2) into Eq. (4), Eqs. (1)-(4) now constitute a set of eight equations with eight unknowns (\( \sigma_j, \dot{\varepsilon}, \sigma, \)). The eight equations can be readily solved/integrated using one of the numerical integration techniques.

2.3.5. Interactions between the Tool and Work-piece

A penalty contact algorithm was employed to determine the normal interactions between the tool and the work-piece. Within the penalty contact algorithm, the interpenetration of the contact surfaces is resisted by linear spring forces/contact-pressures whose values are directly proportional to the depth of contact penetration. These forces, hence, tend to pull the surfaces into an equilibrium position with no penetration. Unless, the nodes on the “slave surface” contact the “master surface”, the contact pressures between the bodies are not transmitted. It should be noted that when the surfaces are in contact there is no limit to the magnitude of the contact pressure that can be transmitted. On the other hand, the shear stresses transmitted across the contacting interfaces are defined through the use of a static and a kinetic friction coefficient. Also, a stick/slip critical shear-stress level (i.e. a maximum value of shear stress which can be
transmitted across a contact interface before the contacting surfaces begin to slide) is also employed. The static and the kinetic friction coefficients are set to a value of 0.3. A modified coulomb friction model was employed to define the stick/slip behavior. According to this model, there is an upper limit for the shear-stress which can be transmitted across the contacting interfaces. This is equal to the shear strength of the softer of the two contacting materials. Thus at a given level of applied contact pressure, the stick/slip critical shear-stress level is defined as the smaller of the following two shear-stress values: (a) the softer-material shear strength; and (b) a product of the friction coefficient and the contact pressure. The contact pressure and the contact shear stress are calculated as part of the complete FSW boundary-value problem.

As mentioned earlier, the frictional energy dissipation at the tool/work-piece contact interfaces due to the frictional-slip/sliding is considered as a potential heat source. The heat generated per unit contact surface area per unit time, $q$, is taken to scale with the magnitude of the tangential/interface-shear stress, $\tau$, and the slip rate, $ds/dt$ as:

$$ q = \eta \tau (ds/dt) $$

(5)

where $\eta$ denotes the fraction of the frictional-slip energy which is converted to heat. The heat flux, $q$, is then divided between the tool and the work-piece. Within the present work, it was assumed that the heat generated at the tool/work-piece interface is equally partitioned between the tool and the work-piece.

2.3.6. Explicit Formulation of the FSW Thermo-Mechanical Problem

As mentioned earlier, the computational procedure employed for modeling the FSW process involved a fully coupled thermo-mechanical finite-element analysis. In this type of analysis, the thermal-energy conservation equation is in the form:
\[ \rho c_p \frac{T}{T} = \nabla (k \nabla T) + \eta \sigma_{ij}^{\prime} \delta_{ij} \]  \hspace{1cm} (6)

and an equation for the dynamic mechanical-equilibrium which is given by:

\[
\frac{\partial \sigma_{ij}^\prime}{\partial x_j} + f_i = \rho \frac{\partial^2 u_i}{\partial t^2} \quad (i,j=1,2,3) \]  \hspace{1cm} (7)

are solved, where \(\nabla\) stands for a gradient/divergence operator, \(x\) is the spatial coordinate and \(f\) is the applied body force.

Within the ABAQUS/Explicit program, an explicit forward-difference integration scheme is used to integrate Eq. (7):

\[ T_{i+1} = T_i + \Delta t_{i+1} T_i \]  \hspace{1cm} (8)

where, subscript \(i\) refers to the time-step increment number.

At the end of each time increment \(i\), the rate of change of the temperature given by the temperature-rate vector, \(T_i\), is calculated as follows:

\[ T_i = C^{-1} (P_i - F_i) \]  \hspace{1cm} (9)

where \(C^{-1}\) is the (inverse) lumped thermal capacity matrix, \(P\) the applied source vector while \(F_i\) is the internal thermal-flux vector.

The dynamic mechanical-equilibrium equation is solved using a central-difference integration procedure as follows:

\[ \delta \xi_{r1/2} = \delta \xi_{r-1/2} + \frac{\Delta t_{i+1} + \Delta t_i}{2} \delta \xi \]  \hspace{1cm} (10)

\[ \delta \xi_{r1} = u_i + \Delta t_{i+1} \delta \xi_{r1/2} \]  \hspace{1cm} (11)
At the end of each time increment $i$, the acceleration-vector is calculated as follows:

$$ \mathbf{a}_i = M^{-1}(L_i - I_i) $$

(12)

where $M^{-1}$ is the (inverse) mass matrix, $L_i$ the applied-load vector while $I_i$ is the internal-force vector.

### 2.3.7. Computational Cost

Generally, computational analysis of most manufacturing/fabrication processes such as the FSW process are very costly/time-consuming. This problem of computational cost can be a major issue with the use of an explicit finite-element method which are conditionally stable (i.e. the time increment employed should be smaller than a critical time increment often referred to as the stable time increment). The mechanical and the thermal problems are associated with their respective stable time increments, within the fully-coupled thermo-mechanical analysis employed in the present work. Hence, the overall stable time increment is the smaller of two time increments.

The mechanical stable time increment is defined by the criterion that, within a given time increment, the stress/deformation wave must not propagate a distance longer than the minimal dimension of any finite-element in the mesh. Hence, the mechanical stable time increment is defined as $\Delta t_{\text{max,mech}} = \frac{l_{\text{min}}}{c_d}$, where $l_{\text{min}}$ is the smallest-element edge length, while $c_d$ is the dilatational wave propagation velocity (sound speed) which is defined as $c_d = \sqrt{E/\rho}$ where $E$ is the Young’s modulus. For the alloy considered in the present work, the sound speed, $c_d$, is ca. 5,100m/s. and the smallest work-piece element size used is ~0.6mm, hence, the stable time increment $\Delta t_{\text{max,mech}}$ is ~1.0·10^{-7}s. An explicit finite-element computational procedure for the FSW process with a simulation time of 20s, would use ~2·10^8 time increments. With the
available computational facilities, this would require an estimated wall time of 96hrs per analysis. Since the computational cost for the FSW computational analysis was extremely high, a mass-scaling algorithm was employed. Within the mass-scaling algorithm, the work-piece material density is artificially increased in order to increase the stable time increment. The increase in the material density does not affect the amount of heat generated by the dissipation of plastic-deformation work and frictional-slip and the thermal stable time-increment. In order to ensure that the mechanical part of the solution is not significantly altered by the mass-scaling algorithm, care was taken to ensure that a kinetic-energy over internal-energy ratio is less than 10%.

In a similar fashion, the thermal stable time increment is generally defined by the condition that, within a given time increment, the thermal wave must not propagate a distance longer than the minimal dimension of any finite-element in the mesh. Hence, the thermal stable time increment is defined as $\Delta t_{\text{max,therm}} = \frac{l_{\text{min}}^2}{2\alpha}$, where $l_{\text{min}}$ is the smallest-element edge length, while $\alpha$ is the thermal diffusivity. Using the previously mentioned values for the thermal-property for AA5083, the thermal stable time increment has been computed as $\Delta t_{\text{max,therm}}$ is $\approx 1.0 \cdot 10^{-3}$s. As mentioned earlier, the mass-scaling algorithm does not affect the thermal stable time increment and that the modified mechanical stable time increment does not exceed the thermal stable time increment. Hence, it should be noted that the mass-scaling does not affect the thermal portion of the fully-coupled thermo-mechanical FSW problem.

2.3.8. Arbitrary Lagrangian-Eulerian (ALE) Method

As discussed earlier, large amounts of plastic deformation as well as large-scale movement/extrusion of the work-piece material from the regions in front of the tool to the region behind the tool are encountered during the FSW process. Under such material processing conditions, the traditional pure-Lagrangian based formulation, in which the computational
domain/mesh is attached to the material and moves/deforms with the material, can encounter extreme numerical difficulties. To overcome this numerical problem, a finite-element analysis procedure based on the Arbitrary Lagrangian-Eulerian (ALE) formulation was employed for the FSW process modeling. The key feature of the ALE procedure, is the ability of procedure to adaptively re-mesh the computational mesh (during a computational run) in order to ensure that the mesh remains of a high quality. The lagrangian character of the ALE mesh enables tracking of the material surfaces, thereby preventing the formation of partially-filled elements.

The main aspects of the ALE model/formulation used in the present work are the following: (a) In the circumferential direction, the work-piece mesh is assumed to be stationary (i.e. of the Eulerian character), while in the radial and in the through-the-thickness directions, the same mesh is allowed to follow the material (i.e. of the Lagrangian character); (b) The rim surfaces (i.e. on the inflow and outflow boundaries) of the work-piece plate are treated as being pure Eulerian and are thus stationary; and (c) The top and bottom surfaces of the work-piece are of “sliding” type, i.e. the mesh is allowed to follow the material in the direction normal to the surface but is not attached to the material in the other two orthogonal directions.

2.4. Results and Discussion

2.4.1. Modification of the Material Model to Include the Effects of Dynamic Recrystallization

As mentioned earlier, a modified Johnson-Cook model which includes the effect of strain-hardening, strain-rate sensitivity and temperature-softening on the material yield-strength is employed for the work-piece material, AA5083-H321. As discussed earlier, the original Johnson-Cook model, given in Eq. (3), provides only a reversible effect of the temperature in promoting plastic deformation through the thermal activation of dislocation glide and climb. That is higher temperatures during the FSW process advance plastic yielding of the material but, per se, is not considered to (irreversibly) change the material microstructure/properties. As described
earlier, during the FSW process, the weld zone material becomes heavily plastically deformed, hence, the weld zones are generally subjected to temperatures very near, yet lower than, the material melting temperature. The material, under such processing conditions, tends to undergo annealing while it is being deformed plastically. The FSW processing conditions result in the stir/nugget region to undergo dynamic recrystallization, due to which the weld-zones material strength/hardness (at high welding temperatures, as well as, at the room temperature) is much lower compared to the base (H131 temper condition) material. The above mentioned effect is not accounted for in the original Johnson-Cook model. Rather, only the effect of high temperatures on promoting plastic deformation via thermal activation is taken into account.

In the present work, a modification to the differential equation governing the evolution of the equivalent plastic strain is proposed, in order to overcome the aforementioned deficiency of the original Johnson-Cook model. The evolution of the equivalent plastic strain within the original Johnson-Cook model is governed by simultaneously satisfying the Hooke’s law, yield criterion and flow rule relations, Section 2.3.4. Hence, within the current material model, only the effect of material strain-hardening due to an increase in the dislocation density and the resulting increase in the dislocation-motion resistance imposed by the neighboring dislocations is considered. The effects of dynamic recrystallization are accounted for through a simple phenomenological-based relation for the additional (negative) component in the equivalent plastic strain rate. This equation is based on the following physics-based arguments: (a) Dynamic recrystallization is a thermally activated process and consequently the correction term in the equivalent plastic strain evolution equation must contain a Boltzmann probability term in the form exp(-Q/RT) where Q is an activation energy, while R is the universal gas-constant. In other words, an arrhenius-type function is utilized to represent the dynamic-recrystallization correction to the Johnson-Cook strength model; (b) It is convenient to replace the Q/RT term in the
Boltzmann probability relation with $q/T_h$ (where $q$ is a dimensionless activation energy), since the rate of recrystallization across various alloy systems appear to scale with the dimensionless absolute-zero based homologous temperature, $T_h$; and (c) Since the rate at which the weld-material tends to recrystallize increases as the amount of cold work is increased, $q$ should be a decreasing function of the equivalent plastic strain $\varepsilon_{pl}$.

From the arguments mentioned above, the contribution of the dynamic-recrystallization towards the evolution of the equivalent plastic strain during the FSW process can be expressed as:

$$
\dot{\varepsilon}_{pl,\text{dyn., rec}} = \dot{\varepsilon}_{o,pl,\text{dyn., rec}} e^{-q(\varepsilon_{pl})/T_h}
$$

(13)

where, $\dot{\varepsilon}_{o,pl,\text{dyn., rec}}$ is a dynamic-recrystallization frequency/pre-exponential term. An analysis of the available experimental data pertaining to the kinetics of recrystallization of AA5083 [2.37] showed that $q$ scales inversely with $\varepsilon_{pl}$ raised to a power of 2.9. Based on this finding and using the curve-fitting results for the experimental recrystallization-kinetics data reported in Ref. [2.37], it is found that Eq. (13) can be rewritten as:

$$
\dot{\varepsilon}_{pl,\text{dyn., rec}} = 21.5 e^{-1/(\varepsilon_{pl}^{0.3}T_h)}
$$

(14)

Figures 2-4(a)-(c) shows the effect of Eq. (14) on modifying the behavior of AA5083 under simple uniaxial tensile conditions. When the $T_h$ values are relatively low ($T_h = 0.3$), it is seen in Figure 2-4(a) that the effects of dynamic recrystallization are very small hence the material strain hardens. On the other hand, when $T_h$ is relatively high ($T_h = 0.9$), the effect of dynamic recrystallization is significant despite extensive plastic deformation, hence, the material undergoes pronounced strain softening, Figure 2-4(b). In Figure 2-4(c), it is seen that when the effects of strain hardening and dynamic recrystallization are comparable, at the intermediate
values of \( T_h \) (\( T_h = 0.5 \)), no significant change in material strength takes place during plastic deformation.

From Figure 2-4(c), it is evident that the oscillating behavior of the material strength is the result of the competition and the interaction between strain-hardening and dynamic recrystallization induced softening processes. That is, softer material tends to harden at a high rate and, when the amount of plastic strain in the work-piece becomes sufficiently large, the rate of dynamic-recrystallization becomes high enough to bring the strength down. This type of oscillating-strength behavior is often a signature of the undergoing dynamic-recrystallization process.

The recrystallization kinetics of the material are generally described using the so-called Johnson-Mehl-Avrami equation e.g. [2.38]. This equation defines the relationship between the volume fraction of the work-piece material recrystallized and the time taken. It is typically given by a characteristic S-shaped curve starting from a non-zero annealing time (the incubation period), increases with a higher and higher slope and, ultimately, the slope decreases as the volume fraction of the recrystallized material approaches unity, Figure 2-5. A major part of the range (80-90%) of the recrystallized-material volume-fraction is generally covered by the inner steepest part of this curve. Taking this fact into account, the simple model proposed here assumes that the entire recrystallized-material volume-fraction vs. time curve can be represented by its inner part and that this portion can be linearized. The slope of this new linear function, on the other hand, is taken to be a function of the temperature and the equivalent plastic strain. Eq. (14) is then obtained by assuming that \( \dot{\varepsilon}_{pl,\text{dyn-rec}} \) scales linearly with the rate of recrystallization.

The overall effect of dynamic recrystallization of the work-piece material on the material evolution during FSW is accounted for through a modified Johnson-Cook material model which is implemented into a user-material subroutine VUMAT.for and linked with ABAQUS/Explicit.
finite-element solver. Several FSW cases were analyzed in the present work in order to validate the implementation of the material model. It is found that when the effects of dynamic recrystallization are suppressed, the results (not shown for brevity), based on the user-material model and the Johnson-Cook model are essentially identical.
Figure 2-4. The strength vs. equivalent plastic strain curves are for the original and the modified Johnson-cook strength models are compared. The results obtained are for under a uniaxial strain-rate of 0.001 s\(^{-1}\) for three different homologous temperatures:

(a) \(\theta = 0.3\); (b) \(\theta = 0.9\); (c) \(\theta = 0.5\).
Figure 2-4. Continued
2.4.2. Typical Results of the Computational Procedure

In this section, a detailed discussion of the typical results obtained as part of the fully-coupled finite element computational analysis of the FSW process is presented. The temporal evolution and spatial distribution of several material-related quantities such as: equivalent plastic strain, stress and strain components, temperature, material velocity, local material strength, tracer particle analysis which provide the locations of the material particles as they enter/pass through the circular region surrounding the rotating pin tool, etc are obtained as part of the developed computational procedure. The results obtained allowed an analysis of the effect of all the key FSW process parameters on the work-piece material state. This section provides only a few
representative and unique results, since similar results were shown and discussed in a series of papers by Zhang et al. [2.29-2.31].

2.4.2.1. Nodal Velocity Results

The spatial variation of the work-piece nodal velocities on the boundary regions of the work-piece at times of 0.0s and 0.5s are displayed in Figures 2-6(a)-(b). The FSW tool is not displayed in Figures 2-6(a)-(b), as well as in other figures presented in the remainder of this chapter in order to maintain clarity. As seen in the figures, the initial conditions assigned in the form of an unidirectional velocity field in the welding direction, transforms into a very complex velocity field in the region right below the tool shoulder (within which the material is stirred around the pin) and the remainder of the field (within which the material flows around the stir region). As shown in Figures 2-6(a)-(b), the initially unfilled region underneath the tool shoulder becomes filled as the FSW process proceeds (an increase in the work-piece hole upper-rim altitude is seen). Thereafter, the region underneath the tool shoulder remains completely filled throughout the FSW process. As the FSW tool traverses along the welding direction, the work-piece material under the tool is continuously refreshed. Figure 2-6(c) displays a close up of the stir region under the tool shoulder to better reveal the character of the nodal-velocity field. In addition to the above mentioned results, a transverse section of the work-piece is displayed in Figure 2-6(d) which reveals the tool-induced material-stirring effect through the work-piece thickness.
Figure 2-6. Results of the nodal-velocity field associated with friction stir welding: (a) the initial velocity state; (b) the fully developed state; (c) a close-up of the work-piece (b); and (d) a transverse section of part (b).
Figure 2-6. continued
2.4.2.2. Trajectories of the Material Particles

The spatial distribution and temporal evolution of the nodal velocities are shown in Figures 2-6(a)-(b). The finite-element analysis employed in the present work is based on the ALE method in which the motion of the computational mesh is not completely tied to the motion of the material. Hence, the nodal velocity results displayed in Figures 2-6(a)-(b) are associated with the velocity of the material points passing through the nodes at that instant. It should be noted that different material points are associated with the same nodes. Material-particle trajectories are employed within the present work to observe the material extrusion around the tool pin and its subsequent forging at the tool wake. This is accomplished in the ABAQUS/Explicit program through the use of tracer particles which are attached to the material points and not to the nodes in the mesh.

The results obtained through the use of the tracer particles on the retreating-side and advancing-side are displayed in Figures 2-7(a)-(b), respectively. The initial location of the tracer particles shown in these figures is halfway between the top and bottom surfaces of the work-piece. The tracer-particle trajectories are shown in color for clarity. The following key aspects of the FSW process are observed from the results displayed in Figures 2-7(a)-(b): (a) The retreating side of the work-piece material, as shown by the yellow and green tracer-particle trajectories, Figure 2-7(a)), for the most part, does not enter the stir zone under the tool-shoulder and flows around it; (b) The work-piece material on the advancing side (as represented by the white and cyan tracer-particle trajectories, Figure 2-7(b)), is extruded to the retreating side and is stirred along with the retreating side material to form the welded joint; and (c) The advancing-side material further away from the initial butting surfaces remains on the advancing side and either enters the stir region on the advancing side or flows around it.
Figure 2-7. Results of the tracer-particles for the work-piece material on the: (a) Retreating-side; and (b) advancing-side.
2.4.2.3. Temperature Field

Figures 2-8(a)-(b) show the spatial distribution of the temperature in the work-piece during FSW process. The results displayed in these figures correspond to the temperature distributions over the medial longitudinal and medial transverse sections. Examination of the results displayed in these figures and of the results obtained in the present work (but not shown for brevity) reveals that: (a) Work-piece temperatures in a range between 350°C and 450°C are observed for the current FSW process conditions such as tool contact pressure, tool rotational and translational speeds; (b) The temperature differences between the top and bottom surfaces of the work piece are significantly reduced as the tool rotational speed and contact pressure are increased; (c) The peak temperatures were found to be in the work-piece material below the tool shoulder and temperatures gradually decreased from this region as a function of the distance in the radial and through-the-thickness directions; and (d) The plastic deformation of the material contributed around 30% towards the overall heat generation, while, the remaining heat generated was associated with the frictional dissipation at the tool/work piece contact surfaces and the plastic deformation contribution increases slowly with an increase in the translational velocity of the tool.
Figure 2-8. Temperature distribution over half of the work-piece obtained by cutting along: (a) the longitudinal; and (b)-(c) transverse directions: Maximum (red) = 400°C; Minimum (blue) = 25°C.
2.4.2.4. Equivalent Plastic Strain Field

The spatial distribution of the equivalent plastic strain within the work-piece during FSW is displayed in Figures 2-9(a)-(b). The equivalent plastic strain distribution results over the medial longitudinal and medial transverse sections are displayed in Figures 2-9(a)-(b), respectively. An examination of the results shown in these figures and those obtained in the present work (but not shown for brevity) reveals that: (a) The equivalent plastic strains in a range between 30 and 50 are observed, depending on the FSW process conditions such as tool contact pressure, tool rotational and translational speeds; (b) The distribution of the equivalent plastic strain showed a high level of asymmetry relative to the initial location of the butting surfaces. The key reasons for the observed asymmetry was due to the differences in the material transport (at the advancing and the retreating sides of the weld) from the region ahead of the tool to the region behind the tool; (c) The equivalent plastic strains gradually decreased from the region below the tool shoulder as a function of the distance in the radial and through-the-thickness directions with the highest equivalent plastic strains found in the work-piece material right below
the tool shoulder; and (d) The differences in the equivalent plastic strains between the top and bottom surfaces of the work piece are reduced as the tool translational speed is decreased and the tool/work-piece contact pressure is increased. These results suggest that under the FSW process conditions used, the extent of material stirring/mixing (which plays a critical role in weld quality/joint-strength) is increased.

2.4.2.5. Residual Stress Field

It is well established that weldments fabricated using the FSW process contain significant level of residual stresses both in the longitudinal (in the welding direction) and in the transverse (normal to the direction of welding) directions. The presence of residual stresses in the weldment is due to the non-uniform distributions in the extent of plastic deformations and in temperature in different regions within the weld joint. It is well known that the residual stresses in the weldment can adversely affect the structural and environmental resistance/durability of welded joints. Hence, it is important that the residual stresses be quantified and their magnitudes and spatial distributions be correlated with various FSW process parameters. Within the present work, an effort is made to develop capabilities for computational investigations of the residual stress distribution. This is accomplished by importing the results of the explicit FSW simulation into the implicit finite-element program ABAQUS/Standard and carrying out a quasi-static fully coupled thermo-mechanical analysis. The extremely long computational times required by the ABAQUS/Explicit program for this type of investigation makes its use inappropriate. The FSW tool is removed and the work-piece boundary conditions are eliminated while the temperature is gradually decreased down to room temperature, within the implicit quasi-static thermo-mechanical analysis.

Figures 2-10(a)-(b) displays the distribution of the von Mises residual stresses over medial longitudinal and transverse sections of the work-piece, respectively. An examination of
the results displayed in these figures and of the results obtained in the present work (but not shown for brevity) reveals that: (a) Maximum longitudinal residual stresses are generally greater than their maximum transverse counterparts by a factor of roughly two; (b) An increase in the tool rotational and translational velocities result in an increase in the longitudinal and transverse residual stresses; and (c) The residual stresses typically increase in magnitude as the distance from the initial portion of butting surfaces is reduced. However, in the innermost portion of the nugget, they tend to decrease somewhat. This is clearly related to the effect of dynamic recrystallization which is prevalent in this region.
Figure 2-9. Spatial distribution of the equivalent plastic strain over one-half of the work-piece obtained by cutting along: (a) the longitudinal; and (b) transverse directions: Maximum (red) = 120; Minimum (blue) = 0.
Figure 2-10. Spatial distribution of the von Mises residual stress over one-half of the work-piece obtained by cutting along: (a) the longitudinal; and (b) transverse directions:
Maximum (red) = 50MPa; Minimum (blue) = -20MPa.
2.4.3. Comparison between Experimental and Computational Results

A comparison of the computational results presented in the previous section with their experimental counterparts reveals that computational results are in good qualitative agreement with the general experimental observations/findings. However, a good-level of quantitative agreement between the experimental and computational results is essential, if the developed model/procedure is to become an integral part of the FSW process design and guide further development and optimization of the process. A few selected computational results from the present work are compared with their experimental counterparts obtained in the work of Peel et al. [2.26], in order to assess the ability of the present computational procedure to account for the experimentally measured FSW-related results. The experimental investigation conducted by Peel et al [2.26], on the effect of the FSW process on AA5083 (the aluminum alloy investigated in the present work) is quite comprehensive and thorough. The following two types of experimental results obtained from the work of Peel et al. [2.26] could be directly compared with the finite-element based computational results obtained in the present work: (a) variation of the longitudinal and transverse (normal) residual stresses as a function of the distance from the weld center line; and (b) variation of the work-piece material room-temperature strength as a function of the distance from the weld center line.

2.4.3.1. Residual Stress Distribution

The computational results (pertaining to the variation of the longitudinal and transverse residual stresses as a function of the distance from the initial location of the butting surfaces) obtained in the present work are compared with their experimental counterparts reported in Ref. [2.26] as displayed in Figures 2-11(a)-(b). The present computational analysis reasonably well reproduce the residual stress results obtained experimentally, while, some disagreement exists
between the results. Specifically: (a) The residual stresses on the advancing side of the weld (the right-hand side in Figures 2-11(a)-(b)) are generally compressive away from the weld center line; (b) The residual stress magnitude increases and then becomes tensile in nature at a distance of 15-20 mm from the weld center line (at the advancing side); (c) The residual stresses in the innermost portion of the nugget is generally tensile; (d) The stresses gradually decrease toward zero as the distance from the weld center line increases on the retreating side; and (e) The transverse residual stresses are generally lower than their longitudinal counterparts.

2.4.3.2. Material Strength Distribution at Room-temperature

The computational results (pertaining to variation of the room-temperature material strength as a function of the distance from the initial location of the butting surfaces) predicted by the modified Johnson-Cook strength model, are compared with their experimental counterparts reported in Ref. [2.26] as displayed in Figures 2-12. Though the present computational analysis correctly predicts the overall trend, the quantitative agreement between the computed and the experimental results is only fair. The results obtained using the modified Johnson-Cook strength model is quite encouraging since the original Johnson-Cook strength model (in which the effect of dynamic recrystallization is neglected) incorrectly predicts that the highest room-temperature strength levels are located in the innermost region of the nugget zone (where the equivalent plastic strain levels are also the highest).
Figure 2-11. The variation of the residual stresses as a function of the distance from the weld center line along the: (a) longitudinal and (b) transverse. The weld-joint advancing side results are on the right-hand side of the plot.
Figure 2-12. Variation of the room-temperature material strength as a function of the distance from the weld center line. The weld-joint advancing side results are on the right-hand side of the plot.
2.5. Summary and Conclusions

Based on the work presented and discussed, the following main summary and conclusions can be made:

1. A computational-based fully-coupled thermo-mechanical finite-element analysis of the friction stir welding (FSW) process for a prototypical solid-solution strengthened and strain hardened aluminum alloy (AA5083) is performed.

2. The effects of dynamic recrystallization and the associated material softening within the stir zone of the welded joint are accounted for through a modification of the original Johnson-Cook strength model in order to model the microstructure/property evolution during the FSW process.

3. The obtained computational results showed good overall qualitative agreement with the corresponding empirical findings.

4. The validation of the modified Johnson-Cook finite-element procedure was conducted using the limited quantitative experimental results which pertain to the variations of the longitudinal and transverse residual stresses with distance from the weld center line and the associated variations in material strength. A reasonably good agreement is obtained between the computational and experimental results suggesting that the modeling and simulation procedure used are quite adequate.
2.6. References


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CHAPTER 3

NUMERICAL INVESTIGATION OF THE HARDNESS EVOLUTION DURING
FRICION-STIR WELDING OF AA5083 AND AA2139 ALLOYS

3.1. Abstract

The computational model of the Friction Stir Welding (FSW) process developed in the previous chapter is combined with the basic physical metallurgy of two wrought aluminum alloys to assess their FSW behaviors. The alloys selected in the present work are AA5083 (a solid-solution strengthened and strain-hardened/stabilized Al-Mg-Mn alloy) and AA2139 (a precipitation hardened quaternary Al-Cu-Mg-Ag alloy). These aluminum alloys are currently being used in the design of military-vehicle hull structural and armor systems.

The aluminum alloys considered in the present work exhibit very different behaviors during the FSW process. The predominant microstructure evolution processes taking place during the FSW process in the case of the non-age-hardenable AA5083 alloy, are extensive plastic deformation and dynamic recrystallization of the weld zone material which is highly-deformed and subjected to elevated temperatures approaching the melting temperature. In the case of age-hardenable AA2139 alloy, precipitates coarsening, over-aging, dissolution and re-precipitation was also considered in addition to the plastic deformation of the material and dynamic recrystallization within the weld-zone. Within the present work, the spatial variation of the material hardness within the different FSW zones for the two alloys considered is accessed using the data available in the open literature which pertain to the kinetics of the aforementioned microstructure-evolution processes. The computational results obtained are found to be in reasonably good agreement with their experimental counterparts.
3.2. Introduction

Current efforts by the U.S. Army have been chiefly aimed at higher levels of mobility, deploy ability, and sustainability while maintaining or surpassing the current levels of lethality and survivability. Currently, the ability to readily transport and sustain battlefield vehicles is hindered, since, their weights are in excess of 70 tons due to the ever increasing lethality of the ballistic threats. Therefore, a number of research and development programs are under way to engineer light-weight, highly mobile, transportable and lethal battlefield vehicles with a target weight under 20 tons. Towards this end, significant advances in the field of light-weight structural- and armor-materials development (including aluminum-based structural/armor-grade materials) are required in order to meet the goals.

For quite some time, aluminum alloys such as AA5083-H131 have been used for the design of military vehicles such as the M1113 and the M109, in accordance with the MIL-DTL-46027J specification [3.1]. The primary reasons for the use of this alloy in military vehicle design are its light weight, ease of welding using the available techniques, very good performance against threats based on fragmentation, and superior corrosion resistance.

The increased levels of lethal threats, have resulted in the use of higher strength aluminum alloys, such as AA2139 [3.2], AA7039 [3.3], AA2219 [3.4] and AA2519 [3.5] for the design of aluminum-armor based military-vehicle systems. These higher strength alloys provide significantly better protection against ballistic threats posed by armor piercing (AP) threats. The use of these aluminum alloys towards the design of vehicle-hulls is very desirable, since, the increased tensile strengths of these alloys enable significant weight reductions. However, compared to the AA5083-H131 alloys, one of the problems with these alloys are its weldability and corrosion resistance which are inferior by nature.
Within the present work, a computational approach is developed and utilized to predict and compare the welding behavior of AA5083 and the novel high-strength aluminum alloy, AA2139. The FSW behavior of the alloys mentioned above is investigated in the present work, primarily because these alloys are often friction stir welded.

The discovery of the FSW process in 1991 [3.6], has resulted in the FSW process being the most commonly used joining technique for high-strength aluminum components as well as other difficult-to-weld metals. Several industrial sectors such as the shipbuilding and marine, aerospace, railway, land transportation, etc. are currently using the FSW process for material fabrication.

Metallic as well as polymeric materials are frequently welded using a solid-state joining process called friction stir welding. The FSW process is employed in applications in which the original material microstructure/properties must remain unchanged as much as possible after the welding process [3.6–3.8]. The process consists of a rotating FSW tool moving along the butting surfaces of two rigidly butt-clamped plates, as shown in Figure 3-1(a). The FSW tool consists of two main components, a threaded cylindrical pin on one end, and a shoulder at the other. Also, during the joining process, the work-piece (i.e. the two rigidly butt-clamped plates) is generally placed on a rigid backing plate for support. In addition, the tool shoulder is forced onto the work-piece so that a firm contact is made with the top surface of the work-piece. As the tool is rotating about its axis, it is also traversed along the butting surfaces, resulting in the generation of heat at the shoulder/work-piece and, to a lesser extent, at the pin/work-piece interface, as a result of the frictional energy dissipated. The resulting increase in temperature causes softening of the work-piece material adjacent to the contacting surfaces. As the tool moves along the butting surfaces, thermally-softened material in front of the tool is significantly deformed, extruded around the tool to the region behind the tool and compacted/forged at the wake of the tool to form a joint/weld.
FSW process has a number of advantages when compared to the conventional fusion-welding technologies, such as: (a) superior mechanical properties in the as-welded condition; (b) the lack of toxic fumes as well as spatter of molten material improves safety of the process; (c) no consumables such as the filler metal or gas shield are required; (d) ease of process automation; (e) the absence of a weld pool provides the process with an ability to operate in all positions, (horizontal, vertical, overhead, orbital, etc.); (f) extensive post-weld machining is not necessary, since the work-piece thickness under/over-matching is reduced; and (g) low environmental impact. However, some disadvantages of the FSW process have also been identified such as: (a) presence of an exit hole upon FSW tool withdrawal; (b) excessive tool press-down and plates-clamping forces are essential; (c) variable-thickness and non-linear welds are difficult to fabricate using the FSW process; and (d) lower welding rates are required when compared to the traditional fusion-welding techniques, although this shortcoming is somewhat lessened since fewer welding passes are required.

The FSW process involves the presence of an advancing side of the weld (i.e. a side in the work piece whose circumferential velocity of the rotating tool is in the same direction as the tool traverse direction) and the retreating side (i.e. the side in the work piece on which the two velocities are in opposite directions). The presence of the advancing and retreating sides in the work piece results in an asymmetry in the heat transfer, material flow and weld microstructure-properties [3.9].

The FSW process involves extremely complex relationships and contest between the associated thermo-mechanical processes involving frictional energy dissipation, plastic deformation, heat dissipation, material flow, dynamic recrystallization, etc. [3.10-3.17]. Examinations of the weld region generally reveal the presence of the following four zones, Figure 3-1(b): (a) a base-metal/un-effected zone far away from the weld where no material
microstructure/property changes take place; (b) the heat-affected zone (HAZ) in which the material microstructure/properties are influenced by the thermal effects associated with FSW process. The HAZ zone is mostly found in the case of fusion-welds, the material microstructural changes in the case of FSW is quite different due to the presence of lower temperatures and a diffuse heat source; (c) the thermo-mechanically affected zone (TMAZ) which is located closer to the butting surfaces compared to the HAZ. The material microstructure/properties in the TMAZ are affected by the thermal as well as the mechanical aspects of the FSW process. Generally, the original grains within this weld-zone undergo severe plastic deformation; and (d) the weld-nugget which is the innermost zone of the FSW joint. The weld-nugget contains the so-called “onion-ring” features which are a result of the way material is transported from the regions ahead of the tool to the wake regions behind the tool. The work piece material within the weld-nugget contains a very-fine dynamically-recrystallized (i.e. equiaxed grain microstructure) since this zone is subjected to the most extreme conditions of plastic deformation and high temperature exposure.

An important feature in the FSW process is that heat transfer takes place through thermal conduction as well as transport of the work-piece material from the region in front of the tool to the region behind the translating tool. The work-piece material properties, FSW tool geometry and the FSW process parameters significantly influence the heat and the mass transfer process. The material transport in the FSW process is accompanied by severe plastic deformation and dynamic recrystallization of the transported material. The material strain rates involved the FSW process [3.18, 3.19] is high as $30\text{s}^{-1}$.

The weld quality and process efficiency are influenced the following key FSW process parameters: (a) FSW-tool rotation and traverse velocities; (b) tool plunge depth; (c) tool tilt-angle; and (d) tool-design and material. It is critical to achieve a delicate balance between the FSW-tool rotation and traverse speeds, since, low work piece temperatures results in insufficient
material softening causing material flaws due to low ductility of the material. On the other hand, high workpiece temperatures results in significant material microstructure/property changes as well as incipient-melting flaws. The FSW-tool plunge depth (defined as the depth of the lowest point of the shoulder) is an important process parameter which ensures that the required level of shoulder/work-piece contact pressure is attained so that the tool completely penetrates the weld. Insufficient tool-plunge depths typically results in low-quality welds due to inadequate material forging at the rear of the tool, while excessive tool-plunge depths usually leads to weld under-matching i.e. the weld thickness is low compared to the base material thickness. It has been found experimentally, that rearward tilting of the tool by about 2-4 degrees improves the effect of the forging process.

The past two decades have seen considerable experimental research efforts towards gaining a better understanding of the FSW joining mechanisms and the evolution of the welded-materials microstructure/properties [3.20-3.23] as well as to explain the effect of various FSW process parameters on the weld quality/integrity [3.24-3.27]. Although the experimental efforts were able to correlate the welded-materials properties/microstructure with the FSW process parameters they provide very little real time understanding of the physics of heat/mass transfer and microstructure-evolution processes. As shown in the prior work [3.28], this insight can be gained by carrying out a detailed physically-based computational investigation of the FSW process. An overview of the prior computational FSW research efforts is not provided here, since; a detailed review of the prior research efforts dealing with numerical investigations of the FSW process reported in the public domain literature was conducted within the prior work [3.28].

The main objective is to combine the basic physical metallurgy of the two wrought aluminum alloys considered with a fully-coupled thermo-mechanical finite-element analysis of the FSW process developed in the prior work [3.28] in order to predict/assess their FSW
behaviors. The two alloys considered in the present work are AA5083-H131 (a solid-solution strengthened and strain-hardened/stabilized Al-Mg alloy) and AA2139 (a precipitation hardened quaternary Al-Cu-Mg-Ag alloy). The two alloys are currently being utilized for the design of military-vehicle hull structural and armor systems. An assessment of the various microstructure-evolution processes taking place during FSW (e.g. extensive plastic deformation, dynamic recrystallization, and precipitates coarsening, over-aging, dissolution and re-precipitation) and the relations between them will be considered in order to predict the material microstructure/properties in the various FSW zones of the two alloys.

The organization of this chapter is as follows: The main physical-metallurgy aspects of the two alloys (AA5083 and AA2139) are reviewed in section 3.3. The fully-coupled thermo-mechanical analysis used in the computational investigation of the FSW process is presented in section 3.4. Development and parameterization of two hardness models one for AA5083 and the other for AA2139 proposed within the present work and a comparison between the corresponding computed results and their experimental counterparts are discussed in section 3.5. The main conclusions resulting from the present study are summarized in section 3.6.
Figure 3-1. (a) A typical Friction Stir Welding (FSW) process; and (b) The key weld zones associated with the typical FSW joint.
3.3. Basic Physical Metallurgy of AA2139 and AA5083

3.3.1. Microstructure and Properties of AA5083-H131 Alloy

The seven major classes of wrought aluminum alloys (AA) are divided according to the principle alloying elements present. The Al-Mg AA5xxx alloy considered in the present work are often employed for the design of various structural and armor systems, since, they possess very good rollability, availability in the form of rolled plates, excellent corrosion resistance and relatively high strength and good weldability.

During friction-stir welding, the microstructure evolution process of age-hardened Al-alloys such as AA2139 are considered to be very complex due to the unstable nature of its precipitates (i.e. precipitates can coarsen, transform into more stable precipitates, or undergo partial or complete dissolution during alloy exposure to high temperature and can reappear upon cooling in different morphologies and number densities, and even precipitates with different crystal structures may appear). The non-age-hardenable AA5083 aluminum alloy used in the present work is an Mg/Mn solid-solution strengthened alloy. In addition to its solid-solution strengthening, in its H131 temper state it is strain-hardened and stabilized (to obtain a needed level of ageing/over-aging resistance). The above mentioned stabilizing heat-treatment results in the precipitation of Al₆Mn precipitates within this alloy. These precipitates are relatively resistant to both dissolution and coarsening so that precipitate-portion of the material microstructure can be taken as mainly unchanged during FSW.

As mentioned earlier, the aerospace and automotive industries employ AA5083 extensively for production of highly complex structural components of different shapes. The production of these complex shaped structural components is accomplished using the superplastic forming (a high-temperature, low-deformation-rate, low-forming-pressure, open/close-die forming process) process. The AA5083 alloy is recrystallized, after extreme cold-working
treatment. The grain nucleation during the recrystallization process results in an ultra-fine grain microstructure due to the presence of very fine Al\textsubscript{6}Mn precipitates. Under very low deformation rate and high-temperature conditions the presence of the fine grained microstructure enables plastic deformation through grain-boundary sliding, thereby, providing the super-plastic behavior to the material. The extreme levels of plastic deformation along with the recrystallization process occurring dynamically within the weld nugget, results in the formation of a very fine grained material microstructure in this region.

3.3.2. Age Hardening Behavior of AA2139 Alloy

The chemical composition of AA2139, which is an age-hardenable quaternary Al-Cu-Mg-Ag alloy (4-10 Cu/Mg ratio) consists of an Al-based solid solution, \( \alpha \), an Al-Cu-Mg based precipitate, S, and a Cu\textsubscript{2}Al based precipitate, \( \theta \), in the equilibrium phase region. During the artificial aging process, it has been found that the additions of Ag significantly promote the formation of metastable \( \Omega \) precipitates over other competing precipitates such as \( S' \) and \( \theta' \) [e.g. 3.30]. The highest levels of material strength are imparted by the \( \Omega \) precipitates which tend to form on the \{111\}_\alpha planes (the slip planes in the Al-based alloys) in these alloys [e.g. 3.31-3.35].

It has been found that during aging of Al-Cu-Mg-Ag alloys the formation of metastable and stable precipitates is followed [e.g. 3.30, 3.36] in the following sequence:

\[
\text{GP-zones} \rightarrow \theta'' \rightarrow \theta' + \Omega \rightarrow \theta' + S' \rightarrow S + \theta
\]

where GP-zones stands for the Guinier-Preston zones, i.e. the clusters of Cu atoms on \{100\}_\alpha planes which form in the earliest stages of aging of the supersaturated \( \alpha \) solid solution. It is also well established that the relative stability of the \( \Omega \) phase when compared to the \( S' \) phase [e.g. 3.30, 3.36] is generally enhanced with higher Cu/Mg ratios in these alloys. This finding is highly critical since, the best overall combination of mechanical properties in AA2139 is associated with the presence of \( \Omega \)-phase precipitates.
Over the last ten years, the following defining features of the \( \Omega \)-phase precipitates have been established through several microstructural investigations of this precipitate in AA2139 and other related alloys: (a) Within the \( \alpha \)-solid-solution, the \( \Omega \)-phase precipitates form coherent \{111\}_\alpha planes acting as the habit planes; (b) The crystal structure of this phase has been determined as being an Al\(_2\)Cu-based orthorhombic structure [3.32,3.34,3.37]; (c) Within the interior regions of the grains and mainly in the dislocation-free regions, the \( \Omega \) phase tends to precipitate mainly in a homogeneous manner; (d) The \( \Omega \) phase generally exist until a maximum temperature of about 250 °C [3.38]; and (e) \( \Omega \)-phase precipitates are most often present in hexagonal plate-like form with a typical thickness and in-plane dimensions of 2-3nm and 100-200nm, respectively [3.30-3.34, 3.37, 3.39, 3.40].

In the peak age-hardened temper condition the AA2139 alloy consists of \( \theta' \)-phase precipitates in addition to the \( \Omega \)-phase precipitates. Over the last ten years, a detailed examination of this type of precipitate has established the following [e.g. 3.45]: (a) These precipitates mainly form on (100)\( \alpha \) habit planes. [3.33, 3.41]; (b) The size of the \( \theta' \)-phase precipitates are comparable to that of the \( \Omega \)-phase precipitates [3.42, 3.43] and are mostly of octagonal-platelet or ellipsoidal shapes; (c) The \( \theta' \) phase possesses a body-centered tetragonal crystal structure; and (d) The \( S' \) precipitates tend to prefer formation on the dislocations and low-angle grain boundaries [3.32, 3.44], since they are semi-coherent with the \( \alpha \)-matrix.

A major concern during the FSW process is the replacement of the \( \Omega \)-phase precipitates with \( S' \)-phase precipitates (after prolonged aging), while the \( \Omega \)-phase and \( \theta' \)-phase precipitates can normally co-exist in the AA2139 type of alloys. Also, the \( S' \)-phase precipitates form and gradually evolve into \( S \)-phase precipitates and tend to take away Mg-Ag co-clustering surrounding the \( \Omega \)-phase precipitates leading to gradual dissolution of the \( \Omega \)-phase precipitates [3.36]. Several investigations of the \( S' \)-phase precipitates in AA2139 and related alloys revealed
the following defining features of this microstructural constituent: (a) The formation of the S’-phase precipitates on dislocations are generally of heterogeneous nature, while, within the grain interior [3.34] their formation is of a homogeneous nature; (b) They generally appear as laths and are sometimes associated with (120)\(\alpha\) habit planes [3.32]; and (c) the average S’-phase precipitate size is generally comparable to that of the \(\Omega\)-phase and \(\theta’\)-phase precipitates, [3.34].

In addition to the metastable and stable precipitates mentioned above whose formation is driven by the thermodynamic driving forces to reduce the extent of super saturation from the as-quenched \(\alpha\)-phase solid solution, fine-scale Mn- or Zr-rich dispersoids are also present in AA2139 type alloys. Due to the relatively low solubility of Mn and Zr in Al, the dispersoids generally form (and hence survive) at substantially higher temperatures compared to the above mentioned precipitates. The so-called T-phase dispersoids are generally found in AA2139. The main defining features of this phase are: (a) Its stoichiometric formula is Al\(_{20}\)Mg\(_2\)Mn\(_3\) [3.34]; (b) The T phase possesses an orthorhombic crystal structure [3.34]; (c) It is generally present in the form of a rod with a typical rod length between 50 and 500nm [3.30]; (d) It has been generally found that the T-phase dispersoids in fine form generally lead to higher static strength levels in the AA2139-type alloys, while, coarser T-phase dispersoids tend to improve strain-localization resistance and, thus, improve dynamic strength of the material [3.30]; and (e) While both Zr and Mn tend to promote formation of the T-phase dispersoids, Mn generally yields coarser dispersoids and is, hence, a preferred alloying element from the standpoint of achieving improved dynamic strength in AA2139.

3.4. FSW Process Computational Modeling

As described earlier, the computational-based finite-element procedure developed in the prior work [3.28] was utilized for the modeling of the FSW process. In this section, a brief
overview of the prior computational procedure developed is provided, since, a detailed account of the procedure was provided in Chapter 2.

3.4.1. FSW Computational Model

The computational model employed for the FSW process analysis consists of a (40.0mm-radius, 3.0mm thickness) circular work-piece in the form of a plate (with a concentric through-the-thickness 3.0mm-radius circular hole) and a two-part tool (consisting of a 3.0mm-radius, 3.0mm-length solid right circular cylinder, at the bottom, and a 9.0mm-radius, 3.0mm-thickness circular-plate section, on the top), Figures 3-2(a)-(b). The finite-element domain for the FSW
computational model consists of ~20,000 first-order eight-node reduced-integration hexahedral thermo-mechanically coupled solid elements (the meshed model is not shown for brevity).

### 3.4.2. Computational Algorithm

A fully-coupled thermo-mechanical finite-element analysis was utilized to investigate the FSW process. Within this analysis, the nodal degrees of freedom include both the nodal velocities and nodal temperatures. Furthermore, solid-mechanics and heat-transfer aspects of the analysis are fully-coupled. That is, the work of plastic deformation and that associated with frictional sliding are considered as heat sources within the thermal analysis, while the effect of local temperature on the mechanical aspect of the analysis is accounted for through the use of temperature-dependant work-piece material properties.

At the beginning of the computational analysis, the following (initial) conditions are employed: the tool is assigned a fixed rotational speed in a range of 200-400rpm and a zero translational velocity, while the work-piece is assumed to be stationary.

The analysis is carried out by prescribing from the onset a constant rotational velocity and a constant downward pressure to the tool. Instead of assigning a travel velocity to the tool along the (postulated) butting surfaces of the work-piece, the work-piece material is forced to move through the work-piece computational domain at the same velocity but in the opposite direction. The work-piece displayed in Figure 3-2(b), represents only a small circular region around the tool in the otherwise infinitely long work-piece. During the FSW process simulation, the effect of the rigid work-piece backing plate is accounted for by preventing the work-piece material from flowing through its bottom face, the free surfaces of the work-piece and the tool are provided with standard convective boundary conditions while enhanced convection boundary conditions are applied over the bottom face of the work-piece (to mimic the effect of enhanced heat extraction through the work-piece backing plate).
Within the present work, the work-piece/tool interactions is accounted for through the use of a penalty algorithm which is employed to compute the contact pressure (between the contacting bodies) which is governed by the local surface penetrations. On the other hand, the shear stresses between the contacting bodies are transferred via a “slip/stick” algorithm, according to which the two contacting bodies do not slide (otherwise interface sliding takes place) over one another if the developed shear stresses are lower than the frictional shear stress. A modified Coulomb law is employed to compute the frictional shear stress within which there is an upper limit to the frictional shear stress which is equal to the work-piece material shear strength. The frictional shear stress is then defined as a smaller of the product between the static/kinetic friction coefficient and the contact pressure, on one hand, and the work-piece material shear strength, on the other.

As mentioned earlier, both plastic deformation of the material and frictional sliding are treated as heat sources. It was assumed that 95% of the work of plastic deformation was assumed to be dissipated in the form of heat, while, a small fraction of the plastic-deformation work is stored in the form of crystal defects. The rate of heat generation at the tool/work-piece interface due to frictional sliding is assumed to scale with the product of local shear stress and the rate of sliding, and that 100% of this energy is dissipated in the form of heat. The thermal properties of the two materials are then utilized to determine the partitioning of the heat between the tool and the work-piece.

As discussed earlier, large amounts of plastic deformations are experienced by the weld-nugget and the TMAZ, along with large-scale movement/extrusion of the work-piece material from the regions in front of the tool to the region behind the tool are encountered during the FSW process. Under such material processing conditions, the traditional pure-Lagrangian based formulation, in which the computational domain/mesh is attached to the material and
moves/deforms with the material, can encounter extreme numerical difficulties. To overcome this numerical problem, a finite-element analysis procedure based on the Arbitrary Lagrangian-Eulerian (ALE) formulation was employed for the FSW process modeling.

The computational analysis of the FSW problem is solved using an explicit solution method implemented in ABAQUS/Explicit [3.47], a general purpose finite element solver. A mass scaling algorithm is used in order to reduce the computational cost while ensuring a stable solution. This algorithm adaptively adjusts material density in the critical finite elements without significantly affecting accuracy of the computational results.

3.4.3. Material Models

The tool was modeled as a rigid material, since it undergoes relatively lower deformation during the FSW process. The density and thermal properties of the tool are next set to that of AISI-H13, hot-worked tool steel, frequently used as the FSW-tool material.

The work-piece is modeled as an isotropic linear-elastic material and the materials plastic response is assumed to be strain-rate sensitive, strain-hardenable and (reversibly) thermally-softenable material through the use of a Johnson-Cook material model [3.48]. Standard density and thermal properties for AA5083 and AA2139 alloys are used to define the thermal-portion of the material model.

The temperature in the material is assumed to affect the material strength through its effect on the thermal activation of dislocation motion. This is effect is accounted for within the original Johnson-Cook material model. However, during the FSW process the weld-nugget material exposure to high temperature results in the dynamic recrystallization and this phenomenon is not accounted for within the original Johnson-Cook model. Hence, a modified version of this material model was proposed in the prior work [3.28]. Essentially, strain hardening is still assumed to be related to the effective plastic strain, \( \varepsilon_{pl} \), via a parabolic relation,
$B\bar{\varepsilon}_{pl}^n$, where $B$ and $n$ are material parameters. However, $\bar{\varepsilon}_{pl}$ is taken to be composed of two terms: one (positive) associated with the operation of plastic deformation and the other (negative) resulting from the operation of dynamic recrystallization.

3.4.4. Results of the Computational Analysis

A few FSW process simulation results are presented and briefly discussed within this section.

3.4.4.1. Equivalent Plastic Strain Field

The spatial distribution of the equivalent plastic strain within the work-piece during FSW is displayed in Figures 3-3(a)-(d). An examination of the results shown in these figures and those obtained in the present work (but not shown for brevity) reveals that: (a) The equivalent plastic strains in a range between 0 and 50 are observed, depending on the FSW process conditions such as tool contact pressure, tool rotational and translational speeds; (b) The distribution of the equivalent plastic strain showed a high level of asymmetry relative to the initial location of the butting surfaces. The main reasons for the observed asymmetry was due to the differences in the material transport (at the advancing and the retreating sides of the weld) from the region ahead of the tool to the region behind the tool; (c) The equivalent plastic strains gradually decreased from the region below the tool shoulder as a function of the distance in the radial and through-the-thickness directions with the highest equivalent plastic strains found in the work-piece material right below the tool shoulder; and (d) The differences in the equivalent plastic strains between the top and bottom surfaces of the work piece are reduced as the tool translational speed is decreased and the tool/work-piece contact pressure is increased. These results suggest that under the FSW process conditions used, the extent of material stirring/mixing (which plays a critical role in weld quality/joint-strength) is increased.

3.4.4.2. Nodal Velocity Field
The spatial variation of the work-piece nodal velocities on the boundary regions of the work-piece at times of 0.0s and 0.5s are displayed in Figures 3-4(a)-(b). For clarity, the tool is not shown. As seen in the figures, the initial conditions assigned in the form of an unidirectional velocity field in the welding direction, transforms into a very complex velocity field in the region right below the tool shoulder (within which the material is stirred around the pin) and the remainder of the field (within which the material flows around the stir region). As shown in Figures 3-4(a)-(b), the initially unfilled region underneath the tool shoulder becomes filled as the FSW process proceeds (an increase in the work-piece hole upper-rim altitude is seen). Thereafter, the region underneath the tool shoulder remains completely filled throughout the FSW process. As the FSW tool traverses along the welding direction, the work-piece material under the tool is continuously refreshed.
Figure 3-3. The spatial distribution and temporal evolution of the equivalent plastic strain during FSW: (a) zero-time step; (b) at the end of tool plunging; (c) after 7s; and (d) after 14s. Range of equivalent-plastic strain: 0.0 (blue) to 50.0 (red).
Figure 3-3. Continued
Figure 3-4. The spatial distribution of the nodal-velocity during the FSW process: (a) the initial state; (b) the fully developed state.
3.4.4.3. Material/Tracer Particle Trajectories

The spatial distribution and temporal evolution of the nodal velocities are shown in Figures 3-4(a)-(b). Since, finite-element analysis employed in the present work is based on the ALE method in which the motion of the computational mesh is not completely tied to the motion of the material. Hence, the nodal velocity results displayed in Figures 3-4(a)-(b) are associated with the velocity of the material points passing through the nodes at that instant. It should be noted that different material points are associated with the same nodes. Material-particle trajectories are employed within the present work to observe the material extrusion around the tool pin and its subsequent forging at the tool wake. This is accomplished in the ABAQUS/Explicit program through the use of tracer particles which are attached to the material points and not to the nodes points in the mesh.

The results obtained though the use of the tracer particles on the retreating-side and advancing-side are displayed in Figures 3-5(a)-(b), respectively. The initial location of the tracer particles shown in these figures is halfway between the top and bottom surfaces of the work-piece. The tracer-particle trajectories are shown in color for clarity. The following key aspects of the FSW process are observed from the results displayed in Figures 3-5(a)-(b): (a) The retreating side of the work-piece material, as shown by the yellow and green tracer-particle trajectories, Figure 3-5(a)), for the most part, does not enter the stir zone under the tool-shoulder and flows around it; (b) The work-piece material on the advancing side (as represented by the white and cyan tracer-particle trajectories, Figure 3-5(b)), is extruded to the retreating side and is stirred along with the retreating side material to form the welded joint; and (c) The advancing-side material further away from the initial butting surfaces remains on the advancing side and either enters the stir region on the advancing side or flows around it.
Figure 3-5. Trajectories of the material tracer particles originating from the: (a) retreating-side and (b) advancing-side
3.5. Evolution of the Material Microstructure/Hardness

3.5.1. Qualitative Analysis of FSW Joint Material Hardening Mechanisms

3.5.1.1 AA5083

Based on the discussion presented in section 3.3.1, as far as the microstructure/property relations in AA5083 is concerned, the following main strengthening mechanisms are expected to be present in this alloy: (a) solid-solution strengthening; (b) strain-hardening; and (c) grain-size refinement. Within this section, the relative importance of the above mentioned strengthening mechanisms within the four different weld-zones (e.g. the weld nugget, the TMAZ, the HAZ and the base material) are discussed.

**Solid Solution Strengthening**

Within all the four different weld-zones this hardening mechanism is present and its contribution to the material hardness is expected to be fairly uniform across the entire weld region.

**Strain Hardening**

The contribution of the strain-hardening mechanism towards the overall material hardness of the AA5083 base-metal in its H131 temper condition, is significantly larger than the contributions of the other two material hardening mechanisms mentioned above. Some annealing occurs within the HAZ. The annealing occurring within the HAZ is mainly due to recovery or polygonization. Hence, the strain-hardening contribution to the material hardness in the HAZ is comparable to that in the base-metal region. During the FSW process, the material within the TMAZ undergoes extreme plastic deformation, due to which the contribution of strain-hardening to the overall material hardness in the TMAZ is expected to increase. The contribution of strain-hardening towards the overall material hardness within the weld nugget is very low, since, the
dynamic recrystallization process controls the material microstructure/property evolution in this weld-zone.

**Grain Size Refinement**

To a first order approximation, the average grain size within the base-metal, the HAZ and the TMAZ weld-zones are expected to be not significantly different, hence, the contribution of the grain-size refinement towards the overall material strength in these weld-zones is expected to be comparable. However, within the weld-nugget, the dynamic-recrystallization process produces a very fine grain microstructure. Therefore, the contribution of the grain-refinement mechanism to the overall material hardness is expected to be largest in this weld-zone.

**3.5.1.2. AA2139**

Based on the discussion presented in section 3.3.2, as far as the microstructure/property relations in AA2139 is concerned, the following main strengthening mechanisms are expected to be present in this alloy: (a) precipitation-hardening; (b) strain-hardening; and (c) grain-size refinement. The importance of the strain-hardening and the grain-size refinement mechanisms within the four weld-zones was discussed earlier in the context of AA5083. The key points made earlier are equally valid in the case of AA2139. The following key observations are made with respect to the role of the precipitation hardening mechanism in AA2139. The precipitation hardened AA2139 alloy in its T8 (quenched + cold-worked + artificially-aged) temper condition, provides a significant contribution to the overall material hardness in the base-metal zone which is greater than the contributions of the other two hardening mechanisms. During the FSW process, the remaining three weld-zones (the weld nugget, the TMAZ, the HAZ) experience high temperatures resulting in the material over-aging along with a significant loss in the material
strength. As one moves from the HAZ zone to the weld-nugget, the loss in material strength significantly increases.

3.5.2. Parameterization of Simple Models for the Hardening Mechanisms within the FSW Joint

3.5.2.1. AA5083

In the present work, a simple hardness model for the AA5083 alloy is proposed and the overall material hardness, $H$, is computed

$$
H = H_C(c) + \frac{H_d}{d^{1/2}} + H_\varepsilon \varepsilon^n
$$

(1)

where, the terms on the right-hand-side of Eq. (1) represent the contributions of the solid-solution strengthening, grain-size refinement and strain-hardening to overall material hardness, respectively, $C$ is the content of alloying elements in the alloy, $d$ the average grain-size and $\varepsilon$ the equivalent plastic strain, while, the hardness model parameters are $H_C$, $H_d$, $H_\varepsilon$ and $n$.

As discussed earlier, within the four weld-zones, the contribution of solid-solution strengthening is uniform, hence, the solid-solution hardness parameter $H_C$ is considered constant. The hardness data for the fully-annealed coarse grained AA5083 [3.49] is used to determine the value of $H_C$ (410MPa) for which the grain-size refinement and strain-hardening contributions are very low.

Within Eq. (1), the grain-refinement hardening term is written according to the Hall-Petch relation [3.50]. Using the results regarding dependence of material hardness on the grain-size in fully-annealed AA5083 [3.51], $H_d$ is evaluated as 340MPa.
As shown in Eq. (1), a parabolic strain-hardening law [3.48] is employed to model the dependence of the material hardness on the equivalent plastic strain. A regression analysis of the strain-hardening data reported in Ref. [3.52], yielded $H_d = 620$ MPa and $n=0.23$.

### 3.5.2.1. AA2139

The hardness model for AA2139 consists of a contribution from the precipitation-hardening mechanism, as shown below:

$$H = [H_{O} + \Delta H_{PA}(1-\eta)] + \frac{H_d}{d^{1/2}} + H_{\varepsilon} \bar{\varepsilon}^n$$

(2)

where, the contribution of precipitation hardening mechanism to the overall material hardness is represented by the first term on the right hand side of Eq. (2) and $H_{O}$ and $\Delta H_{PA}$ are the hardness levels in the over-aged condition and hardness increment at the peak-aged condition respectively and $\eta$ is the extent of over-aging. $H_{O}$ and $\Delta H_{PA}$ are assessed as 420 MPa and 790 MPa, respectively, using the available hardness variation data for different aging heat treatments [3.53].

The parameters for the hardness model which describe the effect of grain-size refinement and strain-hardening on the overall material hardness are set equal to their AA5083 counterparts reported earlier, due to lack of available data in the open literature pertaining AA2139. Also, the effect of solid-solution strengthening towards the overall material hardness for AA2139 is neglected since it is expected to be very small in comparison to the contributions associated with the other three strengthening mechanisms.
3.5.3. Evolution Equations for the Material State-variables

Two hardness models, one for AA5083 and the other for AA2139 were parameterized within the previous section. The following three state variables describe the material state with respect to the microstructure for the alloys considered: (a) extent of material over-aging, \( \eta \), (applicable only in case of AA2139); (b) the average grain-size, \( d \); and (c) the equivalent plastic strain, \( \varepsilon \). In order to compute the overall material hardness at different locations within the weld using Eqs. (1)-(2), the final values of the three state-variables mentioned above must be computed. This is accomplished by integrating the appropriate evolution equations (provided below) for the three state variables starting from their initial values (in the base-metal before welding) at each material-point, over the entire thermo-mechanical history.

3.5.3.1. Extent of Material Over-aging

The temporal evolution of the degree of material over-aging under isothermal conditions was computed using a simple inverse exponential law according to which, \( \eta = e^{-t/\tau_0} \), where \( t \) is time and \( \tau_0 \) a temperature-dependent relaxation time. The following evolution equation for the degree of over-aging is proposed by employing the appropriate chain-rule differentiation and simplification:

\[
\frac{d\eta}{dt} = e^{\frac{-C_1\eta}{1-\eta}} \frac{C_2}{1-T_H^m} \quad 0 \leq \eta \leq 1.0
\]  

where, \( 0 \leq T_H \leq 1.0 \) is a room/melting temperature based homologous temperature, \( T_H = (T - T_{Room})/(T_{Melt} - T_{Room}) \) and \( C_1 \), \( C_2 \) and \( m \) are material parameters. Using available aging kinetics data at different temperatures [3.2] \( C_1 \), \( C_2 \) and \( m \) are assessed as 0.8, 0.00035 and 9.6, respectively.
Within the present model, the material over-aging is assumed to be dependent on the exposure of the material to high temperatures while the potential effect of plastic deformation on the over-aging kinetics is treated as a second-order effect and, hence, ignored.

3.5.3.2. Evolution of the Grain-Size

Both the plastic deformation as well as the dynamic-recrystallization process is assumed to control the grain-size evolution. Plastic deformation does not per-se alter the grain size but creates dislocations which rearrange themselves into low-angle grain boundaries to form sub-grains. The mis-orientations between the sub-grains increase due to generation and incorporation of new dislocations into the sub-grain boundaries. At some point, the degree of mis-orientation becomes large enough to convert a sub-grain into a grain (with large-angle grain boundaries) which then begins to consume the surrounding sub-grains until it encounters another “recrystallized” grain. Both, the dislocation incorporation rate into the sub-grain boundaries and the rate of growth of “recrystallized” grains are thermally-activated processes which depend on temperature via an Arrhenius type relation. Following a similar procedure to that employed in the case of over-aging the following grain-size evolution law is proposed here:

\[
\frac{dd}{dt} = (1 - e^{-C_3d}) \frac{C_4 \epsilon_p}{1 - T_H^q}
\]

(4)

where, \(d\) is the average grain-size and \(C_3, C_4, p\) and \(q\) are material parameters. Using available recrystallization kinetics data at different temperatures [3.54] \(C_3, C_4, p\) and \(q\) are assessed as 0.00051/\(\mu m\), 0.24\(\mu m/s\), 0.71 and 0.97, respectively.
3.5.3.3. Evolution of the Equivalent Plastic Strain

In the present work, the calculation of the equivalent plastic-strain is identical to the one developed in the recent work [3.28]. In remainder of this section, an overview of this procedure will be provided, since a detail account of the method can be found in Ref. [3.28].

When the dynamic-recrystallization process does not exist, the equivalent plastic strain evolution is computed by satisfying the Hooke’s law, yield criterion and flow rule relations simultaneously, for each material-point at each time increment [3.28]. Therefore, in the absence of the dynamic-recrystallization process, only the effect of strain-hardening (due to an increase in the dislocation density) and the associated increase in resistance to any dislocation-motion imposed by the neighboring dislocations on the material hardness/strength is accounted for. When the dynamic-recrystallization process accompanies plastic deformation of the material, the evolution equation for the equivalent plastic strain is modified to account for the annealing effects. That is, the dynamic-recrystallization effects are accounted for by incorporating an additional (negative) equivalent plastic strain rate term. This additional term is based on the following physics-based arguments:

(a) The correction to the equivalent plastic strain evolution equation should contain a Boltzmann probability term in the form \( \exp(-\frac{Q}{RT}) \), where \( Q \) is an activation energy while \( R \) is the universal gas-constant, since dynamic-recrystallization is a thermally activated process. That is, an Arrhenius-type function should be used to denote the dynamic-recrystallization correction term in the equivalent plastic strain evolution equation.

(b) Since the rate of recrystallization across various alloy systems appear to scale with the previously defined homologous temperature, \( T_h \), this term was replaced with \( \frac{Q}{RT} \) term in the Boltzmann probability relation with \( \frac{q}{T_h} \), where \( q \) is a dimensionless activation energy; and
(c) It should be noted that as the amount of cold work is increased the rate at which material tends to recrystallize also increases. Hence, \( q \) was set to be a decreasing function of the equivalent plastic strain, \( \varepsilon \).

From the above arguments, the contribution of the dynamic-recrystallization process towards the evolution of the equivalent plastic strain, is expressed as:

\[
\dot{\varepsilon}_{\text{pl,dyn,rec}} = \dot{\varepsilon}_{\text{pl,dyn,rec}} e^{-q(\varepsilon_{\text{pl}})/T_s}
\]  

(5)

where, \( \dot{\varepsilon}_{\text{pl,dyn,rec}} \) is a dynamic-recrystallization pre-exponential term. An analysis of the available experimental data pertaining to the kinetics of recrystallization of AA5083 [3.28] showed that \( q \) scales inversely with \( \varepsilon_{\text{pl}} \) raised to a power of 2.9. Based on above finding and using the regression-analysis results for the experimental recrystallization-kinetics data reported in Ref. [3.54], it is found that Eq. (5) can be rewritten as:

\[
\dot{\varepsilon}_{\text{pl,dyn,rec}} = 21.5 e^{-1/(\varepsilon_{\text{pl}}^{2.9}T_s)}
\]  

(6)

3.5.4. Comparison of the Computational and Experimental Results

The FSW process computational results (provides the material thermo-mechanical history input, i.e. the time-based variation of the temperature and equivalent plastic strain of the material points within the weld) and the material hardness models (as well as that for the evolution of the material average grain size) are employed to determine the variations in the material-hardness and grain-size across the four weld zones. The validation of the models developed within the present work is accomplished by comparing them with their experimental counterparts.

3.5.4.1. AA5083
Figures 3-6(a)-(b) shows the variation in the material-hardness measured across a transverse section of a friction stir weld over the top surface of AA5083-H131 welded plates. The results shown in these figures are for a constant tool rotation speed, shoulder diameter and threaded pin diameter which are 350rpm, 18mm and 5mm, respectively, but for two different FSW tool traverse speeds: (a) Figure 3-6(a) 100mm/min; and (b) Figure 3-6(b), 150mm/min.

The experimental results obtained in Ref. [3.55] are displayed in Figures 3-6(a)-(b) and are utilized for comparison with the computational results. The original hardness results in Ref. [3.55] were reported using the Vicker’s hardness units. They were converted using the known indentation loads and indentor geometry data to the SI stress units before including in Figures 3-6(a)-(b).

An examination of the results displayed in Figures 3-6(a)-(b) shows that:

(a) The calculated hardness profiles clearly show the four different weld zones. Also, the hardness model employed yields a physically realistic variation in material hardness across the FSW joints.

(b) The concurrence between the computational results and their experimental counterparts reported in Ref. [3.55] on a quantitative level can be characterized as being good to fair. The primary reasons for the observed discrepancy may be: (i) the functional relations used to describe the contribution of various mechanisms to material hardness can be further improved; (ii) the experimental data used for model parameterization were relatively scarce and came from different sources; and (iii) potential inaccuracies associated with hardness measurements in Ref. [3.55].

A comparison of the grain-size results obtained using the computational procedure and the experimental counterparts (obtained in Ref. [3.56]) is displayed in Figure 3-7. The level of agreement obtained between the computational results and the experimental counterparts is quite
encouraging, considering the fact that not all the FSW process parameters were specified in Ref. [3.56].
Figure 3-6. Computational and experimental hardness (transverse) profiles comparison over the top surface of the AA5083 work piece. Please refer to the text for information regarding the FSW process parameters used for (a) and (b). The advancing side of the weld joint is on the right-hand side of the plot.
3.5.4.2. AA2139

Figures 3-8(a)-(c) shows the variation in the material-hardness measured across a transverse section of a friction stir weld over the top surface of AA2139-T8 welded plates. The results displayed in these figures correspond to the hardness measurements over the top surface of the work piece, intermediate surface of the work piece and over the bottom surface, respectively. In all three cases the same FSW process parameters (welding speed: 100mm/min; tool rotational speed: 350rpm; shoulder diameter: 18mm; pin diameter: 5mm) were employed.

Similar to the case of the AA5083 alloy, the results displayed in Figures 3-8(a)-(c) shows that the material-hardness model employed provides physically-realistic hardness profiles over a transverse cross-section of the weld (at different locations through the thickness of the work-piece) and that the computational/experimental agreement is good to fair.

A comparison of the grain-size results obtained using the computational procedure and the experimental counterparts (obtained in Ref. [3.58]) is displayed in Figure 3-9. The results
displayed in the figures pertain to the top surface of the work-piece. As seen in Figure 3-9, the level of agreement between the computational and experimental results is comparable to that obtained in the case of AA5083.

Figure 3-8. Computational and experimental hardness profiles (transverse) through the AA213 work-piece weld over the: (a) topmost surface of the work-piece; (b) mid-section; and (c) the bottom surface of the work piece. Please refer to the text for information regarding the FSW process parameters. The advancing side of the weld joint is on the right-hand side of the plot.
Figure 3-8. Contd…
Figure 3-9. Computational and experimental grain-size profiles (transverse) over the top surface of the AA2139 work piece. The advancing side of the weld joint is on the right-hand side of the plot.

3.6. Summary and Conclusions

The main summary remarks and conclusions based on the work presented and discussed in the present work are:

1. The main aspects of the physical-metallurgy of AA5083 (a solid-solution strengthened and strain-hardened and stabilized Al-Mg-Mn alloy) and AA2139 (a precipitation-hardened quaternary Al-Cu-Mg-Ag alloy) is brief provided.

2. For the two aluminum alloys considered in the present work, simple hardness evolution models were developed and parameterized for the various weld-zones (e.g. the weld-nugget, the thermo-mechanically affected zone and the heat-affected zone).

3. The overall material-hardness and the grain-size evolution equations were integrated over the entire thermo-mechanical history of the work-piece material points. The hardness and the
grain-size profiles (one for each alloy) were obtained along a direction transverse to the weld-line. The information about the thermo-mechanical history of the material was obtained by performing a fully-coupled thermo-mechanical finite-element analysis of the Friction Stir Welding (FSW) process.

4. A comparison between the computationally obtained material-hardness and grain-size profiles with their experimental counterparts revealed that the approach can account qualitatively quite well for the measured experimental response, while, the quantitative comparison between the computational and experimental results is only fair.
3.7. References


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CHAPTER 4

A CONCURRENT DESIGN, MANUFACTURING AND TESTING

PRODUCT-DEVELOPMENT APPROACH FOR FRICTION-STIR WELDED

VEHICLE-UNDERBODY STRUCTURES

4.1. Abstract

High strength aluminum and titanium alloys with superior blast/ballistic resistance against armor piercing (AP) threats and with high vehicle light-weighing potential are being increasingly used as military-vehicle armor. Due to the complex structure of these vehicles, they are commonly constructed through joining (mainly welding) of the individual components. Unfortunately, these alloys are not very amenable to conventional fusion based welding technologies (e.g. Gas Metal Arc Welding (GMAW)) and in-order to obtain high-quality welds, solid-state joining technologies such as Friction Stir Welding (FSW) have to be employed. However, since FSW is a relatively new and fairly complex joining technology, its introduction into advanced military vehicle underbody structures is not straight forward and entails a comprehensive multi-prong approach which addresses concurrently and interactively all the aspects associated with the components/vehicle-underbody design, fabrication and testing. One such approach is developed and applied in the present work. The approach consists of a number of well-defined steps taking place concurrently and relies on two-way interactions between various steps. The approach is critically assessed using a SWOT (Strengths, Weaknesses, Opportunities and Threats) analysis.

4.2. Introduction

Friction stir welding (FSW) is a solid-state metal-joining process [4.1]. The basic concept behind FSW is described using the example of flat butt weld, Figure 4-1. As shown in
Figure 4-1, a non-consumable rotating tool moves along the contacting surfaces of two rigidly butt-clamped plates. As seen in this figure, the tool consists of a threaded conical pin with four flutes. During welding, the workpiece (i.e. the two clamped plates) is placed on a rigid backing support, the shoulder is forced to make a firm contact with the top surface of the workpiece while the tool is rotated and advanced along the butting surfaces. Due to frictional sliding, heat is generated at the shoulder/work-piece and at the pin/work-piece contact surfaces. This, in turn, causes an increase in the workpiece/tool temperature and gives rise to pronounced softening of the workpiece material adjacent to these contacting surfaces. As the tool advances along the butting surfaces, thermally-softened workpiece material in front of the tool is back-extruded around the tool, stirred/heavily deformed (this process also generates heat) and ultimately compacted/forged into the tool-wake region to form a joint/weld.

Figure 4-1. A schematic of the Friction Stir Welding (FSW) process used to fabricate a flat-butt joint. Four typical microstructural zones associated with the FSW process are also labeled.
<table>
<thead>
<tr>
<th>Advantages</th>
<th>Shortcomings</th>
</tr>
</thead>
<tbody>
<tr>
<td>Good as-weld mechanical properties and joint quality even in alloys</td>
<td>An exit hole is left after the tool is withdrawn from the work-piece</td>
</tr>
<tr>
<td>unweldable by conventional techniques</td>
<td>Relatively large tool press-down and plates-clamping forces required</td>
</tr>
<tr>
<td>Improved safety due to the absence of toxic fumes or the spatter of</td>
<td>Lower flexibility of the process with respect to variable-thickness and non-linear welds</td>
</tr>
<tr>
<td>molten material</td>
<td></td>
</tr>
<tr>
<td>No consumables such as the filler metal or gas shield are required</td>
<td></td>
</tr>
<tr>
<td>Ease of process automation</td>
<td>Lower welding rates than conventional fusion-welding techniques. This shortcoming is somewhat lessened since fewer welding passes are required</td>
</tr>
<tr>
<td>Ability to operate in horizontal, vertical, overhead, and orbital positions</td>
<td></td>
</tr>
<tr>
<td>as there is no weld pool</td>
<td>It is a relatively costly process</td>
</tr>
<tr>
<td>Minimal thickness under/over-matching which reduces the need for</td>
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<tr>
<td>expensive post-weld machining</td>
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<tr>
<td>Low environmental impact</td>
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<tr>
<td>Ability to produce aluminum-alloy welds in a 0.02-3.0in range in a single</td>
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</tr>
<tr>
<td>pass</td>
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<tr>
<td>Dissimilar aluminum-alloy grades can be readily FSWed (e.g. AA6061 to AA5083, wrought and cast aluminum alloys as well as aluminum matrix composites)</td>
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<tr>
<td>Substantially lower attendant temperatures, residual stresses and</td>
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<tr>
<td>distortions in comparison to those encountered in traditional arc</td>
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<tr>
<td>welding processes</td>
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<tr>
<td>Superior impact resistance property of the FSW joint due to a fine</td>
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<tr>
<td>equiaxed grain structure in the innermost zone</td>
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</tr>
<tr>
<td>Complete absence of filler-induced defects (no fillers used) and</td>
<td></td>
</tr>
<tr>
<td>hydrogen-embrittlement cracking (no hydrocarbon fuel used)</td>
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</tr>
<tr>
<td>Conventional milling machines can be converted into FSW machines</td>
<td></td>
</tr>
<tr>
<td>Fastened joints can be replaced of with FSW joints leading to significant</td>
<td></td>
</tr>
<tr>
<td>savings in weight reduction and cost</td>
<td></td>
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<tr>
<td>Difficult to join 2xxx and 7xxx aluminum alloys can be joined by FSW</td>
<td></td>
</tr>
<tr>
<td>without any solidification-induced defects</td>
<td></td>
</tr>
<tr>
<td>Particularly suited for butt and lap joining of difficult-to-join</td>
<td></td>
</tr>
<tr>
<td>aluminum alloys</td>
<td></td>
</tr>
</tbody>
</table>
Relative to the traditional fusion-welding technologies such as gas metal arc welding (GMAW), FSW offers a number of advantages. Unfortunately, there are also several potential challenges associated with the use of FSW. Since a detailed discussion pertaining to the main advantages and shortcomings of FSW was presented in our prior work [4.2-4.7], only a summary of these is provided in Table 4-1.

FSW has established itself as a preferred joining technique for aluminum components and its applications for joining other difficult-to-weld metals (e.g. titanium-based alloys) is gradually expanding. Currently, FSW is being widely used in many industrial sectors such as shipbuilding and marine, aerospace, railway, land transportation, etc. This joining technology is, in principle, suitable for the fabrication of the welds of different topologies such as: 90° corner, flat-butt, lap, T, spot, fillet and hem joints, as well as to weld hollow objects, such as tanks and tubes/pipes, stock with different thicknesses, tapered sections and parts with three-dimensional contours. A collage of the most frequently encountered FSW joints is provided in Figure 4-2.

In order to respond to the new enemy threats and warfare tactics, military systems, in particular those supporting the U.S. ground forces, are being continuously transformed to become faster, more agile, and more mobile so that they can be quickly transported to operations conducted throughout the world. Consequently, an increased emphasis is being placed on the development of improved lightweight body-armor and lightweight vehicle-armor systems as well as on the development of new high-performance armor materials/structures. Therefore, a number of research and development programs are under way to engineer light-weight, highly mobile, transportable and lethal battlefield vehicles with a target weight under 20 tons. To attain these goals, significant advances are needed in the areas of light-weight structural- and armor-materials development (including aluminum and titanium-based structural/armor-grade materials). Due to the complex structure of the military battle-field and tactical vehicle underbodies, the use of
aluminum- and titanium-alloy components generally requires component joining by welding. Unfortunately, the high-performance aluminum and titanium alloy grades used in vehicle-armor applications are normally not very amenable to conventional fusion-based welding technologies with the weld-zone and/or heat-affected zone mechanical (and often corrosion) properties being quite deficient in comparison to those found in the base-metal.

In principle, many problems associated with fusion welding of the advanced high-strength aluminum and titanium alloys used in military-vehicle applications can be overcome through the use of FSW. However, since FSW is a relatively new and fairly complex joining technology, its introduction into advanced military vehicle structures is not straightforward and entails a comprehensive multi-prong approach. Development and application of one such approach is the subject of the present work. As will be presented in the next section, the present approach requires concurrent and interactive considerations of the key aspects associated with the components/vehicle design/manufacturing and testing. Since blast-survivability and ballistic resistance (destructive) testing of full-size military vehicle-underbodies is quite costly and time consuming, it is commonly replaced with the corresponding fabrication/testing of sub-scale (look-alike) test structures. Consequently, within the present work attention will be given to the fabrication and testing of such sub-scale structures and not to the full-scale vehicle-underbodies.

To critically assess the potential of the proposed approach, the so-called “SWOT” (strength, weaknesses, opportunities and threats) analysis is employed. For a well-defined goal/objective, this analysis (frequently used in projects and business ventures) allows for the identification of the internal and external factors that are favorable and unfavorable with respect to the attainment of the goal. The key objective in the present work is to develop a computational approach which will enable the low-cost, short lead-time development of blast-resistant vehicle underbodies.
Figure 4-2. Typical joint/weld geometries/designs fabricated using the FSW process: (a) flat-butt joint; (b) unequal thickness flat-butt joint; (c) 90° corner butt joint; (d) 90° corner rabbeted joint; (e) angle joint; (f) transition weld joint; (g) T- joint; and (h) lap joint.
Figure 4-3. A flow chart of the proposed concurrent design, manufacturing and testing approach.
4.3. Concurrent Vehicle-Underbody Design, Fabrication and Testing

Design, manufacturing and blast-survivability performance testing of military vehicle-underbody sub-scale test structures is a highly complex and time consuming process. It is generally recognized that the lead-time and the cost of this process can be greatly reduced by addressing the issues related to designing, manufacturing and testing concurrently and interactively. In this section, a new fully-integrated approach for the concurrent design, FSW-based manufacturing and testing of high-survivability military vehicle-underbodies is introduced. As will be seen, while this approach contains a number of discrete steps, these steps are most often carried out concurrently and multiple iterations/interactions between different steps are encountered. To help understanding of the proposed approach, a flowchart is provided in Figure 4-3. As seen in this figure, the main steps encountered in the present approach include:

(a) Step 1: Preliminary/modified Design: Within this step, legacy knowledge related to the performance of the vehicles during combat operations or field testing are combined with the results of preliminary studies pertaining to blast-survivability of different FSW joint configurations and the design for manufacturing principles, to arrive at a preliminary (and, subsequently modified) design. All three (conceptual, embodiment and detailed) design stages are included and the topological (e.g. flat butt, 90° corner butt, etc.) and geometrical (e.g. linearity, depth, etc.) details related to different FSW joints are identified and passed to the next step;

(b) Step 2: FSW Process Modeling: Within this step, input FSW weld topologies and geometries from step 1 are combined with FSW process parameters (e.g. tool geometry, tool material, tool rotational and travel speeds, etc.), legacy knowledge and the results of preliminary tests pertaining to the correlation between FSW process parameters and the weld microstructure/properties. These are next used within a FSW process model [e.g. 4.2-4.7] to
determine spatial distribution of the workpiece material microstructure (as well as properties and residual stresses) within different weld zones (i.e. weld nugget, thermo-mechanically affected zone and heat-affected zone);

(c) Step 3: Weld-zone Delineation and Homogenization: The results obtained in step 2 are used within a weld-scanning and homogenization procedure to delineate the boundaries between the different weld zones and to compute the average values of the microstructural parameters (e.g. grain-size, degree of recrystallization, equivalent plastic strain, etc.) within each zone;

(d) Step 4: Re-parameterization of the Weld Material Model(s): Within this step, average values of the microstructural parameters for each of the weld zones, as obtained in step 3, are used to appropriately adjust the corresponding material model parameters relative to their base-metal counterparts in order to include the effect of FSW-induced changes in the material microstructure and properties within each zone. This is a very critical step and typically its success depends on the availability and the quality of the open-literature, legacy and proprietary results relating the microstructure and properties of the materials in question;

(e) Step 5: Definition of the Weld-zone Geometries and Materials: The results obtained in step 3 which pertain to the geometry of different weld-zones are combined with the material model re-parameterization results obtained in step 4, and used to define the components and joints geometries and materials as needed in a transient non-linear dynamics analysis of blast loaded sub-scale test structures;

(f) Step 6: Sub-scale Test Structure Survivability: The designs obtained in step 1 and step 5 are pre-processed (e.g. meshed, fixtured, etc.) and subjected to blast loading within a transient non-linear dynamics analysis and the results obtained used to quantify vehicle-underbody sub-scale test structure survivability;
(g) Inner-loop: FSW Process/structure Testing Iterations: While keeping the preliminary design obtained in step 1 unchanged, FSW process parameters are systematically varied within an optimization scheme in order to maximize vehicle-underbody sub-scale test structure blast survivability; and

(h) Outer-loop: Preliminary-design Modifications: The results obtained in the previous steps are utilized collectively to identify potential modifications in the preliminary design and the process is continued starting with step 2. Modifications in the design are carried out until further design changes do not any longer appreciably affect the blast-survivability of the sub-scale test structure. At this point, the design is “frozen”.

(i) Step 7: Test-structure Fabrication and Testing: Following the final design obtained within the outer iteration loop, the sub-scale test structure is fabricated and tested for blast/ballistic impact survivability in order to provide the proof of concept.

Each of the aforementioned steps is associated with a consideration of important design, manufacturing and testing aspects as related to the vehicle-underbody sub-scale test structures. The most important of these aspects which were not considered in our prior work [4.2-4.7] are analyzed in the remainder of this manuscript.

4.4. Step 1

As stated earlier, within this step, legacy knowledge is combined with the results of preliminary studies pertaining to blast-survivability of different FSW joint configurations and the design for manufacturing principles, to arrive at a preliminary (and, subsequently modified) vehicle-underbody design. When designing the test structures, it is critical to ensure that their topology and design (e.g. plates, stiffeners, and structural details) closely resemble those of a prototypical military vehicle so that the results obtained can be used to judge blast survivability of the vehicle structures themselves. An example of the (sub-scale) vehicle-underbody structure is
displayed in Figure 4-4. The main issues related to the use of legacy knowledge and preliminary test results have been discussed in our recent work [4.5]. In the remainder of this section, a brief discussion is provided regarding the main issues related to the consideration of test-structure manufacturability within the design step.

As discussed earlier, the manufacturing of advanced military vehicle-underbody structures capable of enduring ballistic/blast forces involves the utilization of friction stir welding (FSW) (to join the vehicle components). In general, manufacturability of the FSW weldments in question needs to be considered during the design phase of the component(s) and the vehicle. This approach, commonly referred to as “Design for Manufacturing” (DFM), is an economically attractive option since it may greatly reduce refabricating/retrofitting costs and mainly involves the conceptual and the embodiment design stages (the stages which are associated with the lowest product-development cost). In the remainder of this section, examples are provided of the most frequently encountered aspects of DFM within the context of FSW of high-survivability military vehicle-underbody structures.

4.4.1. Weld Region Accessibility to the FSW Tool

A typical FSW tool assembly consists of a circular-cylindrical flat shoulder and a pin. This tool assembly is mounted on a tool holder (also referred to as the shank) which is connected to the machine spindle. The machine spindle itself is connected to the load cell and the load cell housing making the entire tool/tool-holder assembly quite bulky. The bulky nature of the FSW tool/tool holder assembly may lead to inaccessibility of the weld region to the FSW tool. An example of the case in which the initial design may not be adequate with respect to the weld-region accessibility to the FSW tool is depicted in Figure 4-5(a). A modified design in which the problem of weld-region accessibility is corrected is provided in Figure 4-5(b). An alternative modified design in which the length of the horizontal member is increased is provided in Figure
4-5(c). It should be noted that the modified design(s), may, in general compromise the functional performance of the weldment or reduce its mass efficiency. Consequently, both the weld-region accessibility to the tool and component functional performance/mass efficiency have to be considered concurrently.

![Figure 4-4. An example of the (sub-scale) vehicle-underbody structure.](image)

4.4.2. Weld Joint Design/Configuration

The design of military vehicle structures involves the selection of appropriate weld-joint designs (e.g. butt, lap, T-joint etc.) in different sections of the vehicle. While, all these joint designs can be manufactured using FSW, flat and 900 corner-butt joints have been demonstrated to be most easily fabricated, and to yield superior static and ballistic/blast strength performance (where the latter strength performance is typically assessed using the so-called ballistic shock test procedure [4.8]). Consequently, designs involving the use of butt-joints are generally preferred. For example, a T-joint, displayed in Figure 4-6(a), of high-quality is quite challenging to produce using FSW. As shown in Figure 4-6(b), a T-joint may be replaced by a pair of more easily manufacturable 900 corner-butt joints.
Figure 4-5. (a) Original weldment design which may be difficult to fabricate due to lack of weld-region accessibility by the FSW tool; (b) and (c) two potential modified designs.
Figure 4-6. (a) Original weldment design containing a single T-joint which may require multi-pass FSW procedure; and (b) a potential modified design containing two 90° corner butt joints.
Figure 4-7. (a) Original weldment design requiring complex fixturing and non-orthogonal clamping forces; and (b) a modified design requiring simple fixturing, orthogonal clamping and geometrically simpler components.
4.4.3. Component Fixturing for FSW

The FSW process requires the use of stiff and strong fixtures in order to: (a) ensure large contact pressures along the butting surfaces; and (b) prevent welding component deflection and displacement during the welding process. In general, strict fixturing requirements must be met in order to produce good quality FSW joints. Meeting these fixturing requirements may become challenging due to the inherent shape of the welding components as well as the location of the welds. An example of the two FSW-joint weldments in which fixturing may become important is displayed in Figures 4-7(a)-(b). The design in Figure 4-7(a) is associated with geometrically more complex fixtures and with the need for the application of clamping forces in non-orthogonal directions. In addition, the shape of two out of three components is relatively complex. In the modified design in Figure 4-7(b), only geometrically-simple and orthogonal fixturing is needed and the component’s shape is simplified. An examination of Figure 4-7(b) shows that the revised design may be deficient with respect to meeting the weld-region accessibility requirements. Thus, during the components/structure design stage, all the critical FSW-based DFM aspects must be considered.

4.5. Step 2

As stated earlier, within this step, input FSW weld topologies and geometries from step 1 are combined with FSW process parameters, legacy knowledge and the results of preliminary tests and used within a FSW process model to determine spatial distribution of the workpiece material microstructure (and properties) within different weld zones. In the remainder of this section, a brief description is provided regarding the structure of a typical FSW process model. Since the FSW tool design and tool material are important FSW process-model input parameters,
and they were not considered in our prior work, they will be also briefly overviewed in this section.

4.5.1. FSW Process Modeling

FSW normally involves complex interactions and competition between various thermo-mechanical processes such as frictional-energy dissipation, plastic deformation and the associated heat dissipation, material transport/flow, material microstructure evolution (e.g. grain-growth, precipitate coarsening, recrystallization etc.) and local cooling [4.8-4.15]. A unique feature of the FSW process is that heat transfer does not only take place via thermal conduction but also via transport of the work-piece material adjacent to the tool from the region in front to the region behind the advancing tool. In general, both the heat- and mass-transfer depend on the work-piece material properties, tool geometry and the FSW process parameters. Mass transport during FSW is accompanied by extensive plastic deformation (with maximum equivalent plastic strains of the order of 10-50) of the transported material with the attendant strain rates as high as 10 s⁻¹ [4.16, 4.17].

Over the last 10-15 years, considerable effort has been expended towards developing computational methods and tools for analyzing the FSW joining process, quality of the resulting weld as well as the microstructure and properties of the workpiece material in the as-welded state. A detailed overview of the existing FSW process models was presented in our prior work [4.2, 4.3]. Hence, no similar in-depth overview will be presented here. Instead, only the aspects of a typical FSW process model which are pertinent to the present concurrent design, fabrication and testing approach will be discussed.

A typical FSW process model requires specification of a number of input parameters such as the workpiece material properties, component’s geometry, weld topology and FSW process parameters. The main FSW process parameters include: (a) tool-design/material; (b) rotational
and translational velocities of the tool; (c) tool-plunge depth; (d) tool tilt-angle; and (e) tool-dwell time (the FSW process typically involves three distinct stages: (a) tool plunging; (b) dwelling; and (c) welding).

Within a typical FSW process model, the mass, momentum and energy conservation equations are solved under the conditions specified by the aforementioned input parameters in order to determine the associated thermo-mechanical fields (e.g. temperature, equivalent plastic strain, equivalent plastic strain-rate, stress components, particle velocities, etc.). The model is frequently combined with a microstructure material model. In this case, the list of field quantities includes additional (microstructural) parameters such as grain-size, the extent of precipitate coarsening, degree of recrystallization, etc. An example of the latter type of FSW process model can be found in our recent work e.g.[4.3, 4.4], in which a fully-coupled thermo-mechanical finite-element analysis is employed to solve the governing mass, momentum and heat-transfer conservation equations combined with the microstructure evolution equations (describing the basic physical metallurgy of the aluminum alloy grades being FSWed). Within this model, various microstructure-evolution processes taking place during FSW (e.g. extensive plastic-deformation induced grain-shape distortion and dislocation-density increase, dynamic recrystallization, and precipitates coarsening, over-aging, dissolution and re-precipitation) are considered to predict the material microstructure/properties in the various FSW zones of the alloys being welded. For each of the aforementioned microstructure evolution processes, the appropriate material state variables are introduced and their evolution equations constructed and parameterized (using available open literature sources pertaining to the kinetics of the microstructure evolution processes). Next, the thermo-mechanical constitutive models for the alloys being FSWed are modified to include the effect of the local material microstructure on the material response during FSW. This approach enabled examination of the two-way interactions
between the FSW process and the weld-material microstructure evolution. In other words, both the effect of the current material microstructure on its thermo-mechanical response during the FSW process and the effects of thermo-mechanical history of a material point during the FSW process on the associated microstructure are analyzed.

4.5.2. FSW Tool Design/Material

4.5.2.1. Tool Design

Tool design is one of the most important factors that influences the FSW joint quality as well as the weld material microstructure and properties. A typical FSW tool, in its base line configuration, consists of two main sections, a solid right circular-cylindrical (RCC) shoulder and a solid RCC pin. Both the shoulder and the pin play an important role in the FSW process, affecting heat generation, material flow, weld quality as well as the power required for welding. The tool shoulder is responsible for the majority of heat generation via frictional sliding at the tool-shoulder/workpiece interface, while both the tool shoulder and the pin affect the material-flow/stirring and the weld quality. It is generally recognized that the base-line FSW tool design produces limited material flow and mixing. Consequently, in recent years several tool designs were proposed which improve the efficiency of the FSW process and the resulting weld quality over the ones obtained using the base line design. These new tools typically contain modified designs in both the shoulder and the pin sections. The two main modifications in the FSW tool shoulder are: (a) concave shoulder profile; and (b) flat shoulder with scrolls. These modifications are displayed and labeled in Figures 4-7(a)-(b) and their use is found to greatly enhance material stirring and deformation and typically results in joints of improved quality. Additionally, the concave shoulder profile reduces workpiece/weld thickness mismatch while scrolls eliminate the need for tool-tilting and, thus, promote the fabrication of non-linear (e.g. 90° turn flat butt) welds.
The main FSW tool pin modifications include: (a) non-flat bottom (lowers the wear-rate and tendency for fracture at the expense of material stirring extent); (b) taper (lowers the longitudinal loads experienced by the pin); (c) threads (promotes material mixing in the workpiece thickness direction and improves material forging in the same direction); (d) stepped spiral (performs a role similar to threads); and (d) flats and flutes (enhances the extent of material stirring, plastic deformation and thermal softening which, in turn, enables higher welding speeds). These pin-design modifications are also displayed and labeled in Figures 4-8(a)-(c).

Figure 4-8. Three FSW tool designs which collectively include most of the advanced features found in the new-generation FSW tools.
The nature and the extent of modifications of the FSW-tool base-line design is controlled by a number of factors such as: (a) the workpiece material (e.g. in the case of FSW of high-temperature materials, stepped-spirals are more frequently used than threads since the latter are prone to wear and fracture) and tool materials (e.g. threads/stepped spirals are difficult to machine in low-ductility ceramic materials and these features may result in pronounced stress concentration effects); (b) the weld joint design (e.g. in the case of lap joints, tools with two shoulders are often used. The lower shoulder is smaller in diameter and is plunged down to the joint interface while the top shoulder (larger diameter) rests on the top surface of the workpiece); (c) FSW process parameters (e.g. features which promote extensive heat generation and material softening via frictional sliding and material stirring/plastic deformation are used when larger welding speeds are desired); and (d) manufacturer’s prior experience (i.e. legacy and proprietary knowledge regarding the suitability of different tool designs for different FSW applications is still a major factor controlling the design of the FSW tool).
Friction-stir welding (FSW) is a thermo-mechanical deformation process during which the tool temperature approaches the work-piece solidus temperature (the minimum temperature at which the liquid phase is observed during heating) and the tool is subjected to large normal and shear contact stresses. In order to produce good quality welds for a particular application, not only the appropriate tool design but also the selection of the appropriate tool material is critical. The selection of FSW-tool materials is guided by the fulfillment of the functional requirements such as: (a) long service-life as governed by wear, fracture, work-piece/tool chemical-interactions and thermal-decomposition processes; (b) availability and cost; and (c) good dimensional stability under high-temperature working conditions. By employing the conventional material selection principles [4.18], the following thermo-mechano-physical properties are identified as being the most critical in the case of FSW tools: (a) strength at elevated as well as ambient temperatures; (b) thermal and chemical stability at elevated-temperatures; (c) wear resistance; (d) workpiece/tool chemical reactivity; (e) material fracture toughness; (f) coefficient of thermal expansion (in the case of multi-material tools); (g) machinability; and (h) uniformity in microstructure, density and property distributions (primarily in the case of powder metallurgy fabricated FSW tools). While, ranking of these material properties may be highly subjective, the order in which the properties are listed above is consistent with the most commonly used FSW tool-material property ranking.

The tool materials most commonly used in the FSW-tool applications are as follows: (a) tool steels, (e.g. AISI H13); (b) nickel- and cobalt-base alloys (e.g. Inconel738LC and MP 159); (c) refractory metals (e.g. tungsten, molybdenum, niobium and tantalum); (d) crystalline ceramics (e.g. carbides like titanium carbide and polycrystalline cubic-boron nitride (PCBN)); and (e) metal-matrix composites (e.g. W+1vol.%La2O3, W-Re+2vol.%HfC). A summary of the common
FSW-tool materials and the critical material properties is provided in Table 4-2, in this table, materials performance with respect to the properties in question is ranked using an excellent/good/fair/poor scale.

<table>
<thead>
<tr>
<th>Material</th>
<th>Property</th>
<th>High-temperature Strength</th>
<th>Wear Resistance</th>
<th>Fracture Toughness</th>
<th>Machinability</th>
<th>High-temperature Chemical Stability</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tool Steels (e.g. AISI H13, High-speed Grades)</td>
<td></td>
<td>Good</td>
<td>Good</td>
<td>Good-to-Excellent</td>
<td>Excellent</td>
<td>Good</td>
</tr>
<tr>
<td>Ni-Co-based Alloys (e.g. Inconel738LC and MP 159)</td>
<td></td>
<td>Good</td>
<td>Good</td>
<td>Good-to-Excellent</td>
<td>Good-to-Excellent</td>
<td>Good</td>
</tr>
<tr>
<td>Refractory Metals (e.g. W, W-Re, Mo, Nb, Ta)</td>
<td></td>
<td>Excellent</td>
<td>Good</td>
<td>Fair-to-Poor</td>
<td>Poor</td>
<td>Fair</td>
</tr>
<tr>
<td>Crystalline Ceramics (e.g. TiC, PCBN, WC)</td>
<td></td>
<td>Good-to-Excellent</td>
<td>Good-to-Excellent</td>
<td>Fair-to-Poor</td>
<td>Poor</td>
<td>Excellent</td>
</tr>
<tr>
<td>Metal Matrix Composites (e.g. W+1vol.%La₂O₃, W-Re+2vol.%HfC)</td>
<td></td>
<td>Excellent</td>
<td>Excellent</td>
<td>Good</td>
<td>Fair</td>
<td>Fair-to-Good</td>
</tr>
</tbody>
</table>

4.6. Step 3

As discussed earlier, the application of a typical FSW process model produces a number of thermo-mechanical fields (e.g. temperature, equivalent plastic strain, equivalent plastic strain-rate, residual stress components, particle velocities, etc.) associated with the formation of the FSW joint in question. In addition, such a model may produce a number of microstructural fields (e.g. grain-size, the extent of precipitate coarsening, degree of recrystallization, etc.) in the final joint. Here, the latter fields can be used to define the boundaries between the base-metal and the
weld as well as to define the boundary between different zones of the weld. Typically, a FSW weldment contains four distinct microstructural zones:

(a) a base-metal zone which is far enough from the weld so that material microstructure/properties are not altered by the joining process;

(b) the heat-affected zone (HAZ) in which material microstructure/properties are affected only by the thermal effects associated with FSW. While this zone is normally found in the case of fusion-welds, the nature of the microstructural changes may be different in the FSW case due to generally lower temperatures and a more diffuse heat source;

(c) the thermo-mechanically affected zone (TMAZ) which is located closer than the HAZ zone to the butting surfaces. Consequently, both the thermal and the mechanical aspects of the FSW affect the material microstructure/properties in this zone. Typically, the original grains are retained in this zone although they may have undergone severe plastic deformation; and

(d) the weld nugget is the innermost zone of an FSW joint. As a result of the way the material is transported from the regions ahead of the tool to the wake regions behind the tool, this zone typically contains the so called “onion-ring” features. The material in this region has been subjected to most severe conditions of plastic deformation and high temperature exposure and consequently contains a very-fine dynamically-recrystallized equiaxed grain microstructure.

Before one can define the boundaries between the four microstructural zones, the key thermo-mechanical and microstructural parameter(s) for the alloy in question must be identified. For example, aluminum alloys can be broadly classified as non-heat treatable (non age-hardenable) and heat treatable (age/precipitate hardenable) aluminum-alloys. In the case of non-heat treatable aluminum alloys, material strength and ductility is mainly controlled by the grain size and the extent of strain hardening (as defined by the competition between plastic deformation and dynamic recrystallization). Thus, the main parameters used to delineate different
microstructural zones are, in this case, the grain-size, the equivalent plastic strain and the degree of recrystallization. In the case of heat-treatable alloys, on the other hand, as-welded material mechanical properties are mainly controlled by age or precipitate hardening. Hence, the key microstructural parameters include the extent of precipitate over-aging/dissolution as well as the ones mentioned in the context of non-heat treatable alloys.

Another critical step in the weld-zone delineation process is the definition of the threshold values for the parameters identified above. This is important since the thermo-mechanical and microstructural fields are generally smooth and the use of such threshold values helps decision making regarding the position of the inter-zone boundaries. For example, one must define the minimal (threshold) increase in the local grain-size at a material point, relative to that in the base-metal zone, for the point to be considered a part of the HAZ. Similarly, a minimal threshold value for the degree of recrystallization must be defined for the definition of the TMAZ/weld-nugget boundary.

Once the key microstructural parameters are identified and the threshold values selected, a simple microstructure scanning algorithm can be utilized in order to delineate the four microstructural zones. This is demonstrated in Figures 4-9(a)-(e).
Figure 4-9. Typical distributions of the: (a) grain size; (b) equivalent plastic strain; and (c) degree of recrystallization over a transverse section of a flat-but FSW weld; (d) the grid used for identification of weld inter-zone boundaries; and (e) the resulting weld decomposition into three distinct zones.
Figures 4-9(a)-(c) show examples of the field plots pertaining respectively to the grain-size, equivalent plastic strain and the degree of recrystallization distributions over a transverse section of the single flat-butt joint weldment. A fine quadrilateral grid, Figure 4-9(d), is placed over the field plots and combined with the grain-size, equivalent plastic strain and the degree of recrystallization threshold values to define the boundaries between the four microstructural zones, Figure 4-9(e). In Figure 4-9(e), the base-metal/HAZ boundary is defined by a 61.2 μm grain-size contour (a 20% increase relative to the base-metal grain-size), the HAZ/TMAZ boundary by a 0.3 equivalent plastic-strain contour, while the TMAZ/weld-nugget boundary is defined by a 0.7 degree of recrystallization contour line.

Once the HAZ, TMAZ and the weld-nugget are defined, one can calculate an average value of the thermo-mechanical and microstructural parameters within each of these three zones. Following the procedure described in the next section, these average values are next used to re-parameterize the workpiece material model within each of the weld zones.

4.7. Step 4

As discussed earlier, within this step, average values of the microstructural parameters for each of the weld zones, as obtained in step 3, are used to appropriately adjust the corresponding material model parameters relative to their base-metal counterparts in order to include the effect of FSW-induced changes in the material microstructure and properties within each zone. While there is a relatively large selection of material models that can be used to describe the mechanical behavior of metallic systems, the Johnson-Cook deformation and fracture model [4.19, 4.20] is most frequently used. This model is capable of representing the material behavior displayed under large-strain, high deformation rate, high-temperature conditions, of the type encountered in the problem of computational modeling of both the FSW process and the ballistic/blast loading of
a vehicle sub-scale test structure. Deformation and failure components of this model are briefly reviewed below.

4.7.1. Deformation

Within this model, the (workpiece) material is considered as being an isotropic linear-elastic and a strain-rate sensitive, strain-hardenable and (reversibly) thermally-softenable plastic. The deformation response of the material is defined using the following three relations: (a) a yield criterion, i.e. a mathematical relation which defines the condition which must be satisfied for the onset (and continuation) of plastic deformation; (b) a flow rule, i.e. a relation which describes the rate of change of different plastic-strain components during plastic deformation; and (c) a constitutive law, i.e. a relation which describes how the material-strength changes as a function of the extent of plastic deformation, the rate of deformation and temperature. For most aluminum and titanium alloy grades used in military-vehicle FSWed structures, a von Misses yield criterion and a normality flow-rule are used. The von Misses yield criterion states that equivalent stress must be equal to the material yield strength for plastic deformation to occur. The normality flow-rule states that the plastic flow takes place in the direction of the stress-gradient of the yield surface (i.e. in a direction normal to the yield surface, when the latter is defined in the stress space). The Johnson-Cook strength constitutive law is defined as:

\[
\sigma_y = \left[A + B(\bar{\varepsilon}^{pl})^n\right] \left[1 + C_1 \log\left(\frac{\dot{\varepsilon}^{pl}}{\dot{\varepsilon}_0}\right)\right] \left[1 - T_H^m\right] 
\]

where \(\bar{\varepsilon}^{pl}\) is the equivalent plastic strain, \(\dot{\varepsilon}^{pl}\) the equivalent plastic strain rate, \(\dot{\varepsilon}_0\) a reference equivalent plastic strain rate, A the zero-plastic-strain, unit-plastic-strain-rate, room-temperature yield strength, B the strain-hardening constant, n the strain-hardening exponent, C1 the strain-rate constant, m the thermal-softening exponent and \(T_H = (T - T_{room})/(T_{melt} - T_{room})\) a room-temperature
(T_{room}) based homologous temperature while T_{melt} is the melting temperature. All temperatures are given in Kelvin.

4.7.2. Failure

Within this model, the (workpiece) material is considered as being an isotropic linear-elastic and a strain-rate sensitive, strain-hardenable and within this model, the material failure is assumed to be of a ductile character.

And the progress of failure is defined by the following cumulative damage law:

\[ D = \sum \frac{\Delta \varepsilon}{\varepsilon_f} \]

(2)

where \( \Delta \varepsilon \) is the increment in effective plastic strain with an increment in loading and \( \varepsilon_f \) is the failure strain at the current state of loading which is a function of the mean stress, the effective stress, the strain rate and the homologous temperature, given by:

\[ \varepsilon_f = D_1 \left[ 1 + \frac{D_2}{D_1} \exp(-D_3 \sigma^*) \right] \left[ 1 + D_4 \ln \lambda \tau \right] \left[ 1 + D_5 T_n \right] \]

(3)

where \( \sigma^* \) is mean stress normalized by the effective stress. The parameters \( D_1, D_2, D_3, D_4 \) and \( D_5 \) are all material specific constants. Failure is assumed to occur when \( D \) as defined in Eq. (2) is equal to 1.0.

4.7.3. Model Re-parameterization

In a typical situation, the Johnson-Cook model for the workpiece base-metal is available, i.e. the material model parameters A, B, n, etc. are known. The challenge then is to re-parameterize this model for the remaining three microstructural zones in order to account for the FSW-induced changes in the respective material microstructures. While, in principle, all the Johnson-Cook material model parameters are expected to be microstructure dependent, it is a
common practice to identify and re-parameterize only those material model parameters which are most sensitive to the changes in the material microstructure. The two material parameters generally considered to be belonging to this class are $A$ (the initial material yield strength) and $D1$ (material ductility, while the $D2/D1$ ratio is kept constant).

Revaluation of the parameter $A$ will, in general, depend on the type of the workpiece material in question. Specifically, in non-heat treatable alloys changes in the yield strength within the three weld zones is controlled by grain-size and strain-hardening effects, with the grain-size effects being dominant in the HAZ and in the weld-nugget while strain-hardening provides a major contribution in the TMAZ. Consequently, parameter $A$ is redefined in this case as

$$A = A_{WZ} \left( \frac{d_{FSW,WZ}}{d_{BM}} \right)^{-\frac{1}{2}} + B \left( \varepsilon_{FSW,WZ}^p - \varepsilon_{Re crystallized,WZ}^p \right)$$

(4)

where subscripts $WZ$ and $BM$ are used to denote weldzone and base metal, respectively, the first term on the right hand side accounts for the Hall-Petch-type [4.21] grain-size effect while the second term defines the net effect of FSW-induced strain hardening (resulting from the competition between plastic deformation and dynamic recrystallization). The term $\varepsilon_{Re crystallized,WZ}^p$ denotes the fraction of the FSW-induced plastic strain whose effect on the material strength has been eliminated by dynamic recrystallization. A functional relationship between this quantity and the degree of recrystallization can be found in our prior work [4.4].

In the case of heat-treatable workpiece materials in which age or precipitation hardening controls material strength, the $A$ parameter is redefined as:

$$A = A_{WZ} \left( \frac{l_{BM}}{l_{FSW,WZ}} \right) + B \left( \varepsilon_{FSW,WZ}^p - \varepsilon_{Re crystallized,WZ}^p \right)$$

(5)
where \( l \) denotes inter-precipitate spacing and the first term on the right hand side is defined using an Orowan-type [4.21] equation.

As far as the \( D_1 \) parameter is concerned, it is first recognized that it is a measure of material ductility. It is, in general, a more challenging task to establish a correlation between material’s ductility and its different microstructural features. It is also generally expected that these correlations will depend on the type of workpiece material and that they will be different in the case of heat-treatable and non heat-treatable alloys. In the absence of these correlations and through recognition that microstructural changes which improve strength generally degrade material ductility (and vice-versa), one can assume that the product of the material’s strength and ductility raised to a power \( (q) \) is nearly constant within a given alloy grade. Based on this assumption, parameter \( A \) in different weld zones can be calculated as:

\[
D_{1,wz} = \left( \frac{A_{BM}}{A_{WZ}} \right)^{\frac{1}{q}} D_{BM}
\]  

(6)

It should be noted that Eq. (6) may not be valid in the case when the grain-size has a dominant effect on the material strength and ductility since the aforementioned strength/ductility trade-off is usually not observed in this case.

4.8. Step 5

Within step 1, only the geometries of the components to be welded but not the geometries of the welds (and their zones) were defined. In addition, the weld-zone material properties were not available. These deficiencies are eliminated during this step through the use of weld geometries obtained in step 3 and weld material properties obtained in step 4. In addition, the computed FSW-induced residual stresses can be used to properly define the initial stress state of all components/welds. The vehicle-underbody test-structure computational model is now ready.
for use in the subsequent non-linear dynamics computational analysis of its blast/ballistic-impact resistance/survivability.

4.9. Step 6

The updated and preprocessed (meshed, fixtured, with assigned initial, boundary, loading and contact conditions) design of the vehicle-underbody test structures obtained in step 5 is used next within a transient non-linear dynamics computational analysis to assess its blast/ballistic impact survivability. Typically, survivability is characterized by the lack of penetration and/or of excessive deflection of the test structure. Details regarding the nature of the governing equations and the auxiliary equations which are solved during a typical analysis discussed here, as well as, of the mine, soil and air material models and contact/solution algorithms can be found in our prior work [e.g. 4.22-4.24]. An example of the qualitative results obtained in this portion of the work is displayed in Figure 4-10. Quantitative details regarding the nature of the results obtained and their interpretation cannot be presented or discussed here due to the sensitive character of the subject matter. It is important to emphasize that the computational analysis utilized in this step must, as closely as possible, match the test structure geometry, joining, material properties, fixturing for testing and blast/ballistic-impact test conditions that will be used in step 7 (the test-structure fabrication and testing step).
4.10. Step 7

Within this step, a sub-scale test structure is fabricated and tested under fairly realistic buried-mine blast loading conditions. The test structure is normally required to meet stringent conditions pertaining to the absence of penetration/fragmentation and a lack of excessive deflections. This is a very critical step and must be carried out appropriately in order to ensure that the results obtained can be used to judge blast survivability of the vehicle-underbody being developed. Specifically:

(a) The manner in which the test structure is secured to the test fixture and the overall fixture weight should closely resemble their counterparts present in the vehicle. This is a critical requirement since often the performance of structures (including joints) is greatly affected by the effect of surrounding constraints/interactions;
(b) If the test structure is sub-scaled then a dimensional analysis should be employed to account for the scaling effects (e.g. [4.25]);

(c) While a full-factorial blast-testing schedule over the design/test variables (mine size, shape and explosion energy, depth of burial, stand-off distance, soil type, compaction level and degree of saturation, etc.) is preferred, in many cases blast testing under most adverse combinations of these test variables (as suggested by the computational analysis results discussed in step 6) may suffice; and

(d) A comprehensive failure analysis should be conducted following each mine-blast test. Past experience has shown that one can learn a great deal about the behavior of materials and structures by investigating the manner in which they fail in the presence of various loading and constraining conditions.

4.11. SWOT Analysis

As mentioned earlier, SWOT analysis [4.26] is a strategic planning or assessment method which identifies internal (Strengths and Weaknesses) and external (Opportunities and Threats) factors that are favorable or unfavorable to achieve a given objective. The first step in the SWOT analysis is specification of the desired goal/objective. In the present work, the main goal is to develop a fully-integrated computation-based analysis, which can be used to speed up and economize the introduction of FSW into the military vehicle-underbody manufacturing practice.

The next step is to identify the major external and internal factors which may favorably or unfavorably affect the achievement of the desired goal.

4.11.1. Strengths

Strengths are defined as internal/intrinsic factors which play a favorable role in the achievement of the set objective. For example, computational analyses of the FSW process and of
the mechanical response of vehicle-underbody structures to blast/ballistic impact loads are becoming quite mature and hence, their predictions fairly reliable).

4.11.2. Weaknesses

Weakness is an internal factor which acts unfavorably towards attaining the set goal. For example, prediction of the microstructure evolution (particularly in the case of heat-treatable alloys during FSW) is still far from being mature, yet it plays an important role in obtaining reliable predictions regarding the weld-zone geometries and material properties within the zone.

4.11.3. Opportunities

These are external factors which may play a favorable role in the attainment of the set goal. For example, in the case of non-heat treatable alloys, there is a vast source of microstructural/property and hot-working microstructure evolution data, the conditions encountered during FSW.

4.11.4. Threats

These are external factors which play an unfavorable role towards the achievement of the goal in question. For example, introduction of newer alloy grades (e.g. AA 2139) whose microstructure/property and hot-working microstructure-evolution data are either not fully defined or not available in the open literature, may limit the use of the computational approach proposed in the present work.

The results of the application of the SWOT analysis to the previously identified objective are summarized in Table 4-3. It should be noted that not all the factors appearing in this table are of the same importance. In our future communications, a more refined SWOT analysis will be presented with the proper weights attached to each strength, weakness, opportunity and threat. It should be also noted that as further progress is made in the analysis of FSW process and more
information regarding the material-microstructure of the alloys become available in the open literature, weaknesses and threats will become less significant and some will get converted into strengths and opportunities, respectively.

Table 4-3. Results of the SWOT analysis for the proposed concurrent design, manufacturing and testing approach

<table>
<thead>
<tr>
<th>Strengths</th>
<th>Weaknesses</th>
</tr>
</thead>
<tbody>
<tr>
<td>• Modeling and simulation of the FSW process are quite mature</td>
<td>• More reliable and physically based models are needed to establish the effect of microstructure on the weld-material strength and ductility</td>
</tr>
<tr>
<td>• Modeling and simulation of the blast-survivability of the sub-scale vehicle-underbody test structures are also mature</td>
<td>• The computational cost associated with the FSW process modeling and simulations and the use of non-rigid detailed design FSW tools can be quite high</td>
</tr>
<tr>
<td>• Design methodology and optimization techniques are well-established</td>
<td>• Issues related to the FSW tool degradation by wear, chemical interaction with the work-piece or thermal decomposition are not currently considered</td>
</tr>
<tr>
<td>• Data/information regarding the basic metallurgy of many commercially available alloy grades are readily available</td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Opportunities</th>
<th>Threats</th>
</tr>
</thead>
<tbody>
<tr>
<td>• FSW process simulation models are continuously being improved</td>
<td>• When new alloy grades are introduced, a relatively long lead time is required before the critical body of knowledge related to basic physical metallurgy is created and made available to the public</td>
</tr>
<tr>
<td>• Vehicle-underbody test-structure survivability modeling and simulation methods and tools are continuously being improved</td>
<td>• When survivability with respect to detonation of buried mines is of concern, reliable physically based model(s) for the soil (at different levels of compaction, clay, gravel and silt moisture contents) are needed. Such models are currently not available in the open literature</td>
</tr>
<tr>
<td>• New numerical solution algorithms with improved efficiency, stability and robustness are continuously being developed</td>
<td></td>
</tr>
<tr>
<td>• Many sources of data/information which are currently of the proprietary nature may become public domain with time</td>
<td></td>
</tr>
</tbody>
</table>
4.12. Summary

Based on the work presented and discussed in the present manuscript, the following main summary remarks and conclusions can be made:

1. A new concurrent approach to designing, manufacturing and testing of military vehicle-underbody friction stir welded structures is proposed.

2. While the proposed approach involves a number of well-defined steps, these steps are highly interactive and often occur concurrently.

3. For each of the steps and their interactions, the key issues are identified and examples of the typical results presented and discussed.

4. The proposed approach was critically assessed using the so-called SWOT (Strengths/Weaknesses/Opportunities/Threats) analysis in order to identify internal and external factors which may favorably or unfavorably affect the success of the proposed approach.
4.13. References


CHAPTER 5

TWO-LEVEL WELD-MATERIAL HOMOGENIZATION FOR EFFICIENT
COMPUTATIONAL ANALYSIS OF WELDED STRUCTURE BLAST SURVIVABILITY

5.1. Abstract

The introduction of newer joining technologies like the so-called Friction Stir Welding (FSW) into automotive engineering entails the knowledge of the joint-material microstructure and properties. Since, the development of vehicles (including military vehicles capable of surviving blast and ballistic impacts) nowadays involves extensive use of the computational engineering analyses (CEA), robust high-fidelity material models are needed for the FSW joints. A two-level material-homogenization procedure is proposed and utilized in the present work in-order to help manage computational cost and computer storage requirements for such CEAs. The method utilizes experimental (microstructure, micro-hardness, tensile testing and X-ray diffraction) data to construct: (a) the material model for each weld zone and (b) the material model for the entire weld. The procedure is validated by comparing its predictions with the available experimental results and with the predictions of more-detailed but more costly computational analyses.

5.2. Introduction

During the current decade, the U.S. military has placed increased emphasis on the development of improved lightweight body-armor and lightweight vehicle-armor systems as well as on the development of new high-performance armor materials/structures (in order to properly respond to the new enemy threats and warfare tactics). As a result, the U.S. ground forces are being continuously transformed to become faster, more agile, and more mobile so that they can be quickly transported to warfare/peace-keeping operations conducted throughout the world. As part of this effort, a number of research and development programs are under way with the main goal
to engineer light-weight, highly mobile, transportable and lethal battlefield vehicles with weight under 20 tons. To attain this goal, significant advances are needed in the areas of light-weight structural- and armor-materials development (including light-weight metallic materials such as aluminum and titanium-based structural and armor-grade alloys). Due to complex geometry/topology of the military battle-field and tactical vehicles’ (metallic-armor) body structures, these structures are typically fabricated by welding separately manufactured aluminum- and titanium-alloy components. Unfortunately, the high-performance aluminum and titanium alloy grades used in vehicle-armor applications are normally not very amenable to conventional fusion-based welding technologies, mainly due to the fact that the resulting weld-zone and/or heat-affected zone mechanical (and often corrosion) properties are quite deficient in comparison to their base-metal counterparts [5.1-5.4]. In addition, the conventional welding processes are often not very economical or environment friendly. Most of the aforementioned shortcomings of the conventional welding processes when used in armor-grade aluminum/titanium alloy-joining applications are remedied by the use of the so-called “Friction Stir Welding” (FSW) solid-state process.

FSW was invented and patented by The Welding Institute (UK) in the early 1990's [5.5]. The basic principle of FSW is demonstrated in Figure 5-1 using the example of flat butt welding. The two plates (the workpiece) to be joined are rigidly clamped and placed on a backing plate. A rotating tool, consisting of a profiled pin and a shoulder, is forced down into the workpiece until the shoulder meets the surface of the workpiece. The workpiece material adjacent to the tool is thereby frictionally heated to temperatures at which it is softened/plasticized. As the tool advances along the butting surfaces, thermally-softened workpiece material in front of the tool is back-extruded around the tool, stirred/heavily deformed (this process also generates heat) and ultimately compacted/forged into the tool-wake region to form a joint/weld.
When analyzing the weld formation during FSW, a distinction is made between the so-called “advancing side” of the weld (the side where the tangential component of the tool rotational speed is in the same direction as the tool travel direction) and the “retreating side” (the side where the tangential component of the tool rotational speed is opposite to the tool travel direction). Due to the differences in mass and heat transport and material deformation history, an FSW joint is typically asymmetric relative to the plane of the butting surfaces [5.1-5.3, 5.6-5.7].

![Figure 5-1. A schematic of the Friction Stir Welding (FSW) process used to fabricate a flat-butt joint. Four typical microstructural zones associated with the FSW process are also labeled.](image)

Macrographical and micrographical examinations of a typical friction stir welded flat butt joint reveal the presence of four distinct microstructural zones [5.7]:

(a) a base-metal or un-affected zone which is far enough from the weld so that material microstructure/properties are not altered by the joining process;

(b) the heat-affected zone (HAZ) in which material microstructure/properties are affected only by the heat generated during the FSW process. While this zone is normally found in the case of fusion-welds, the nature of the microstructural changes may be different in the FSW case due to generally lower temperatures and a more diffuse heat source. Typical microstructural changes
which influence the HAZ mechanical properties include the dissolution and coarsening of precipitates (in the case of heat-treatable aluminum/titanium alloy grades) and recovery-based dislocation density reduction (in the case of cold worked heat-treatable and non-heat treatable alloy grades);

(c) the thermo-mechanically affected zone (TMAZ) which is located closer than the HAZ zone to the butting surfaces. Consequently, both the thermal and the mechanical aspects of the FSW process affect the material microstructure/properties in this zone. Typically, the original grains are retained in this zone although they may have undergone severe plastic deformation. The dislocation density is generally increased relative to its base-metal level while the precipitates are greatly affected by the coarsening and dissolution processes; and

(d) the weld nugget is the innermost zone of an FSW joint. As a result of the way the material is transported from the regions ahead of the tool to the wake regions behind the tool, this zone typically contains the so called “onion-ring” features. The material in this region has been subjected to most severe conditions of plastic deformation and high temperature exposure and consequently contains a very-fine dynamically-recrystallized equiaxed grain microstructure. The presence of this fine-grain microstructure often has a beneficial effect in promoting fine scale re-precipitation in the case of heat-treatable alloy grades. The four aforementioned zones are sketched and labeled in Figure 5-1.

Over the last ten to fifteen years, it has been clearly established that FSW provides a number of advantages when used for joining low melting point alloys (in particular aluminum alloys, the alloys which have a great industrial importance). Among these advantages are the fact that, except for the highest-strength aluminum alloy grades, relatively inexpensive tool-steel based FSW tools could be utilized and high production rates realized while producing welds with good mechanical/structural integrity and visual appearance. Additional main advantages of the
FSW process can be summarized as follows: (a) the process can be used for all welding positions (e.g. horizontal, vertical, overhead, orbital, etc.) and can, in each case, be fully automated to ensure high productivity and repeatable quality; (b) weld thicknesses in a range between ca. 0.5 mm to 65 mm can be produced in a single pass; (c) dissimilar alloy grades which are not amenable to fusion welding can be FSWed; (d) the extent of associated thermal distortion and microstructural/property changes is greatly reduced; (e) lower weld-surface preparation requirements (no oxide layer removal necessary); (f) consumables, filler materials or shielding gases are not used; (g) No harmful environmental effects/agents present such as UV radiation, spatter, weld fume, high electric current and electromagnetic fields; (h) the process is highly energy efficient; (g) limited maintenance and spare part inventory for the FSW equipment is required; (i) due to the flat nature of the weld surfaces, less post weld machining is required.

The main limitations/shortcomings of the FSW process are generally identified as: (a) large clamping and shoulder-workpiece contact forces accompany the process which requires the use of high-stiffness clamping and FSW welding equipment; (b) at the completion of the FSW process, an exit hole is left in the weldment (c) high level of geometrical conformability between the workpiece components is critical; (d) high capital equipment, operational and licensing costs; and (e) if process parameters are not properly adjusted, defective joints may result.

The main FSW process parameters which control weld quality, process efficiency and tool longevity are: (a) tool-travel/welding speed; (b) tool rotation speed; (c) tool geometry, cooling tilt angle and plunge depth (in the case of displacement control) or plunge force (in the case of force control). Additional parameters that influence the FSW process and the weldment are weld gap, workpiece thickness variation/mismatch and clamping/welding machine stiffness. However, these parameters cannot be readily controlled [5.8].
Currently, FSW is being widely used in many industrial sectors such as shipbuilding and marine, aerospace, railway, land transportation, etc. This joining technology is, in principle, suitable for the fabrication of the welds of different topologies such as: 90° corner, flat-butt, lap, T, spot, fillet and hem joints, as well as to weld hollow objects, such as tanks and tubes/pipes, stock with different thicknesses, tapered sections and parts with three-dimensional contours [5.9].

While in principle, many problems associated with fusion welding of the advanced high-strength aluminum and titanium alloys used in military-vehicle applications can be overcome through the use of FSW, the introduction of this joining process into the fabrication of advanced military vehicle structures is not straightforward and entails a comprehensive multi-step approach. One such approach, based on the concurrent and interactive considerations of the key aspects associated with the components/vehicle design/manufacturing and testing, was recently proposed by the authors [5.9]. One of the steps in this approach involves the use of computer-aided non-linear dynamics engineering analyses in order to predict (computationally) blast-survivability of the military vehicle (look-alike) test structures. As pointed out earlier, such structures are constructed by welding separately manufactured metallic components. In order for the aforementioned computational analysis of test-structure survivability to be reliable, it is critical that all the welds (and all the zones within the welds) be represented explicitly. Due to a relatively small length-scale of the FSW weld zones, this requirement typically results in finite element models containing a large number (often in the range of several millions) of elements. The resulting large number of degrees of freedom and the associated very small computational time increments place a formidable demand on to the computational memory/storage requirements and lead to often unexpectedly long wall clock simulation times. In the present work, a new two-level homogenization procedure is proposed and implemented in order to reduce the memory/storage requirements and increase the computational speed. Within the first level of
homogenization, homogenized effective properties are determined for each FSW zone. Within the second level of homogenization, homogenized properties of the entire FSW-joint local cross-section are computed. The procedure is validated against the results of the computational analyses in which weld zones are accounted for explicitly and against the available experimental results.

The organization of the paper is as follows: A brief overview of the experimental techniques employed in the present work, and the results obtained is presented in Section II. Parameterization of the base-metal and the weld-nugget materials within an FSW joint is presented in Section III. The two-level material homogenization procedure is introduced and discussed in Section IV. Validation and verification of this procedure is presented in Section V. A brief summary of the main findings obtained in the present work is presented in Section VI.

5.3. Experimental Procedures and Results

All the experimental and the computational work carried out in the present manuscript involved AA2139 (an age-hardenable quaternary Al-Cu-Mg-Ag) alloy in a T8 (quenched + cold-worked + artificially-aged) temper condition. The experimental work involved: (a) flat-butt FSW joining of 25.4mm-thick AA2139 plates; (b) quasi-static tensile testing of the base-metal and weld-nugget material properties in the weld direction; (c) quasi-static transverse (across-the-weld) tensile properties of the weldment; (d) measurements of micro-hardness distribution over the transverse cross-section of the weld; and (e) X-ray diffraction based determination of the residual stresses within the weld and the surrounding base-metal. A brief description of each of the above mentioned experimental procedures is provided below.
Figure 5-2. (a) Top view and (b) transverse section macrograph of a AA2139-T8 flat butt joint.
5.3.1. Flat-butt Friction Stir Welding

Flat-butt friction-stir welding of AA2139-T8 plates was performed at the Edison Welding Institute [5.10]. The welding was performed under the following process parameters: (a) a two piece (flat-bottom shoulder + conical pin) four-flat left handed thread FSW tool made of 350M tool-steel; (b) tool travel and rotational speeds of 50mm/min and 150 rpm, respectively; and (c) tool vertical and traverse loads of 55,600N and 26,600N, respectively. A top-view of a typical AA2139-T8 flat-butt weld is shown in Figure 5-2(a). The corresponding macrograph of the weld transverse cut section, clearly revealing the three weld zones, is depicted in Figure 5-2(b).

5.3.2. Quasi-static Longitudinal Tensile Testing

Room-temperature quasi-static (average engineering strain rate \(\sim 8e-4s^{-1}\)) tensile mechanical properties of AA2139-T8 base-metal and weld nugget are determined using sub-size round bar specimens with a 25.4mm gauge-length and 6.35mm gauge-diameter. In the base-metal case both the longitudinal (along the weld direction) and the transverse specimens were tested, while in the weld-nugget case, due to limited extent of the weld in the transverse direction, only the longitudinal samples (with their centerline located on the weld mid-thickness plane) were used. The resulting longitudinal/transverse base-metal and longitudinal weld-nugget engineering-stress vs. engineering-strain data (averaged over three specimens, in each case) are displayed in Figure 5-3. In all the cases, necking and ultimate fracture occurred within the specimen gauge length and the fracture surface had a dimpled appearance, a defining characteristic of void-nucleation, growth and coalescence based ductile failure.

5.3.3. Quasi-static Transverse Tensile Testing

Room-temperature quasi-static (average engineering strain rate \(\sim 4e-4s^{-1}\)) transverse tensile mechanical properties of AA2139-T8 weldment are determined using square bar
specimens with a 50.8mm gauge-length and 25.4mm square cross-section edge length (to enable monitoring of strain localization during the tensile test). The gauge length was divided (using fiduciary marks) into eight 6.35mm-long segments in-order to monitor the progress of strain localization. The resulting transverse tensile engineering-stress vs. engineering-strain data (averaged over three specimens, in each case) are displayed in Figure 5-4. In all the cases, necking and ultimate fracture occurred within the HAZ and the fracture surface had a dimpled appearance as in the case of the base-metal/weld-nugget materials.

Figure 5-3. Longitudinal and transverse base-metal and longitudinal flat-butt FSW weld-nugget engineering stress vs. engineering strain tensile-test curves in AA2139-T8.
5.3.4. Micro-hardness Measurements

Vicker’s type microhardness testing was undertaken using a Buehler 1600-6100 microhardness tester, at a load of 2N and an application time of 10 to 15 seconds, in accordance with ASTM E3841. Microhardness measurements were conducted over the entire weld transverse cross-sectional area. The individual measurements were located at the nodes of a square-grid with the square edge-length of 0.5mm.

The Vickers micro-hardness number (in kgf/mm²) is calculated using the following relation: \( HV_{0.200} = 1.854.\frac{F}{d^2} \) where the loading force \( F (=0.2\text{kgf}) \) and \( d \text{(in mm)} \) is the diagonal.
mean value of the projected indentation. An example of the results obtained in the form of a micro-hardness contour plot, is displayed in Figure 5-5(a). Based on the results displayed in this figure and the FSW macrograph displayed in Figure 5-2(b), a schematic of the FSW flat-butt joint is provided in Figure 5-5(b) in which different microstructural/properties zones are delineated.

Figure 5-5. (a) An example of a typical Vickers micro-hardness field plot over a transverse section of a AA2139-T8 FSW flat-butt joint; and (b) the associated partitioning of the FSW joint into separate weld zones.

5.3.5. X-ray Diffraction Residual-stress Measurements

FSW-induced residual stresses in AA2139-T8 weldments are measured by carrying out standard X-ray diffraction experiments on a Scintag Polycrystalline-Texture-Stress (PTS) four-axis goniometer for stress and texture analysis with unrestricted 2θ range (from –2 to +162°) at an operating voltage of 18kV. The corresponding CuKα X-ray wave length is 0.031nm. The reflections from the {311} family of planes, representing the local poly-crystalline material state, are used in the residual-stress measurements since these planes are known to be less sensitive to inter-granular strain development [5.11, 5.12]. The measurements were carried out over the top
and the bottom portion of the flat-butt welded plates and it is assumed that the residual stresses in the portion of the weldment sandwiched by these two surfaces can be obtained using a simple linear interpolation procedure. It should be recognized however, that the X-ray diffraction technique employed mainly characterizes the in-plane stress state in a region adjacent to the test-sample surface and that the through-the-thickness residual stresses are not quantified. To quantify the in-plane residual stresses, the so-called $\sin^2 \psi$ technique was employed [5.13], where $\psi$ is the angle between the surface normal and the normal to the diffracting ([311]) crystallographic planes. The basic premise of the this technique is that due to the presence of in-plane stress/strains, the spacing of the diffracting crystallographic planes changes continuously with the inclination angle $\psi$. To quantify the effect of in-plane directions on the accompanying normal stress/strain, a reference direction is selected in the test sample surface and the azimuthal angle $\phi$ used to specify the orientation of these directions. In the case of shear-free bi-axial (in-plane) stress field in an un-textured material, the normal engineering strain associated with an azimuthal angle $\phi$ and an inclination angle $\psi$, can be defined as:

$$
\varepsilon_{\phi\psi} = \frac{d_{\phi\psi} - d}{d_0} = \frac{s_2}{2} \sigma_\phi \sin^2 \psi + s_1 (\sigma_{11} + \sigma_{22})
$$

(1)

Where, $s_1 = -\frac{\nu}{E}, s_2 = \frac{(1+\nu)}{E}$, $E$ is the Young’s modulus, $\nu$ the Poisson’s ratio, $\sigma_\phi$ the normal stress in the azimuthal $\phi$-direction and $\sigma_{11}$ and $\sigma_{22}$ the associated principal stresses.

According to Eq. (1), $\sigma_\phi$ can be computed from the slope of the $\varepsilon_{\phi\psi}$ vs. $\sin^2 \psi$ plot. When this procedure is repeated for two or more azimuthal $\phi$ directions, the in-plane residual stress state, as defined by its principal stress components $\sigma_{11}$ and $\sigma_{22}$, can be determined. In the aforementioned procedure, it was assumed that the unstressed inter-planar spacing $d_0$ is known.
As shown by Peel et al. [5.14], $d_0$ can also be determined from the foregoing X-ray diffraction analysis provided the measurements are carried out along two mutually-orthogonal azimuthal directions.

The procedure described above was used to quantify both the longitudinal and the transverse residual stresses on the top and the bottom test-sample surfaces along a line running orthogonal to the weld direction. An example of the typical results obtained in this portion of the work is displayed in Figures 5-6(a) and (b).

![Graph](a)

**Figure 5-6.** Variation of the: (a) longitudinal and (b) transverse residual stresses as a function of the distance from the weld-line. Data pertaining to the advancing side of the weld joint are on the right-hand side of the plot.
5.4. Base-metal and Weld-nugget Material-models Parameterization

In this section, the (averaged) longitudinal/transverse base-metal and weld-nugget engineering stress vs. engineering strain curves are converted into their respective true stress vs. true strain curves and parameterized.

5.4.1. Johnson-Cook Strength and Failure Models

While there is a relatively large selection of material models that can be used for parameterization of AA2139-T8 base-metal and weld-nugget materials, the Johnson-Cook deformation/strength and fracture model [5.15, 5.16] was used. This model is capable of representing the material behavior displayed under large-strain, high deformation rate, high-temperature conditions, of the type encountered in the problem of computational modeling of the
ballistic/blast loading of a vehicle test structure. Deformation/strength and failure components of this model are briefly reviewed below.

Deformation/Strength: Within this model, the subject material is considered as an isotropic linear-elastic and a strain-rate sensitive, strain-hardenable and (reversibly) thermally-softenable plastic material. The deformation response of the material is defined using the following three relations: (a) a yield criterion, i.e. a mathematical relation which defines the condition which must be satisfied for the onset (and continuation) of plastic deformation; (b) a flow rule, i.e. a relation which describes the rate of change of different plastic-strain components during plastic deformation; and (c) a constitutive law, i.e. a relation which describes how the material-strength changes as a function of the extent of plastic deformation, the rate of deformation and temperature. For most aluminum and titanium alloy grades used in military-vehicle FSWed structures, plasticity is considered to be of a purely distortional (non-volumetric) character and a von Misses yield criterion and a normality flow-rule are used. The von Misses yield criterion states that equivalent stress must be equal to the material yield strength for plastic deformation to occur/proceed. The normality flow-rule states that the plastic flow takes place in the direction of the stress-gradient of the yield surface (i.e. in a direction normal to the yield surface, when the latter is defined in the stress space). The Johnson-Cook strength constitutive law is defined as:

\[
\sigma_y = \left[ A + B (\varepsilon^{pl}_p)^n \right] + C \log \left( \frac{\dot{\varepsilon}^{pl}}{\dot{\varepsilon}_o^{pl}} \right) \left[ 1 - T_H^m \right]
\]  

(2)

where \(\varepsilon^{pl}\) is the equivalent plastic strain, \(\dot{\varepsilon}^{pl}\) the equivalent plastic strain rate, \(\dot{\varepsilon}_o^{pl}\) a reference equivalent plastic strain rate, \(A\) the zero-plastic-strain, unit-plastic-strain-rate, room-temperature yield strength, \(B\) the strain-hardening constant, \(n\) the strain-hardening exponent, \(C\) the strain-rate constant, \(m\) the thermal-softening exponent and \(T_H = (T - T_{room}) / (T_{melt} - T_{room})\) a room-temperature
(T_{\text{room}}) based homologous temperature while T_{\text{melt}} is the melting temperature. All temperatures are given in Kelvin.

**Failure:** Within this model, the material failure is assumed to be of a ductile character and the progress of failure is defined by the following cumulative damage law:

\[
D = \sum \frac{\Delta \varepsilon}{\varepsilon_f}
\]  

(3)

where \(\Delta \varepsilon\) is the increment in effective plastic strain with an increment in loading and \(\varepsilon_f\), is the failure strain at the current state of loading which is a function of the mean stress, the effective stress, the strain rate and the homologous temperature, given by:

\[
\varepsilon_f = D_1 \left[ 1 + \frac{D_2}{D_1} \exp(-D_3 \sigma^*) \right] \left[ 1 + D_4 \ln \varepsilon_{\dot{\varepsilon}} \right] \left[ 1 + D_5 T_{\text{H}} \right]
\]  

(4)

where \(\sigma^*\) is mean stress normalized by the effective stress. The parameters \(D_1, D_2, D_3, D_4\) and \(D_5\) are all material specific constants. Failure is assumed to occur when \(D\) given in Eq. (3) is equal to 1.0. It should be noted that, in contrast to, many “damage-type” materials constitutive models, a non-zero value of the damage variable \(D\) does not degrade the material’s stiffness/strength but merely signals the moment of failure (when \(D = 1.0\)).

**5.4.2. Model Parameterization**

**Base Metal:** Due to a relatively limited extent (i.e. a single strain-rate and room-temperature) of mechanical testing, not all the Johnson-Cook parameters could be determined from the experimental stress vs. strain data. To overcome this shortcoming, the following procedure was implemented:

(a) a typical value \(m = 0.859\) is assumed for the thermal softening part of the strength model;
(b) by comparing the present initial (quasi-static) yield strength with its dynamic counterpart reported in Ref.[5.18], the strain-rate coefficient has been assessed as C=0.043; and 

(c) the remaining three strength parameters (A, B, n) are determined by a standard curve-fitting procedure to yield: A=307MPa, B=524MPa and n=0.4.

As far as the failure model parameters are concerned, the last three parameters are assigned their typical values: D_3=0.349, D_4=0.147 and D_5=16.8 [5.19]. To assess the remaining two failure parameters, D_1 and D_2, it is assumed that D_2/D_1 remains constant and equal to 28.5 [5.20]. Then using Eq. (4) and the experimentally determined value of the failure strain, the two unknown failure parameters are assessed as: D_1=0.0125 and D_2=0.3554.

**Weld Nugget:** The aforementioned procedure is next applied to the weld-nugget experimental stress vs. strain data to yield: A=178MPa, B=524MPa, n=0.4, C=0.043, m=0.859, D_1=0.0268, D_2=0.7647, D_3=0.349, D_4=0.147 and D_5=16.8. It should be noted that the same values for the strength parameters B and n were obtained as in the base-metal case. This was not fortuitous but rather the result of the fact that these two parameters were set equal in the two materials [5.20] and the material model parameterization carried out for both materials simultaneously. Likewise, the failure parameters ratio D_2/D_1 was set equal in the two materials.

### 5.5. Two-level Weld-material Homogenization Procedure

In this section, a new procedure is proposed and implemented for homogenization of the material within the individual FSW zones as well as within the entire weld.

#### 5.5.1. First-level Homogenization

As discussed earlier, only the base-metal and the weld-nugget quasi-static mechanical properties are determined experimentally in the present work. On the other hand, a complete micro-hardness field plot is determined over the entire weld region. In this section, a simple
procedure (based on the use of their experimentally measured micro-hardness values) is proposed for the assessment of the mechanical tensile properties of the remaining two weld zones, i.e. HAZ and TMAZ. The procedure is based on our recent work [5.9] which suggested that the initial yield strength (as represented by the Johnson-cook parameter A) scales with the material mean micro-hardness. This hypothesis is validated in the present work, which shows that the ratio of the initial yield stress and the mean hardness for the base-metal (=307MPa/130kgf/mm²=2.36) and the weld-nugget (=178 MPa/75kgf/mm²=2.37) are quite comparable.

Based on this finding, it is assumed that this ratio can be treated as a constant and set to an average value of 2.365. Then, using the mean micro-hardness values for the HAZ (=105 kgf/mm²) and TMAZ (=120kgf/mm²), the corresponding Johnson-cook A-parameter values are determined as 248MPa and 283MPa in the two zones, respectively. The remaining Johnson-cook strength parameters, B, n, C and m are set equal to their counterparts in the base-metal/weld-nugget regions.

As far as the Johnson-cook failure-model parameters are concerned, it is assumed, following the procedure established in our recent work [5.9], that only parameter D₁ is affected by the FSW process (while the D₂/D₁ ratio is assumed constant [5.20]). Using the D₁ values for the base-metal and weld-nugget and the corresponding mean hardness values, it is found that D₁ is proportional to the mean value of the micro hardness, HV, raised to the power –p(=-1.39). Using this relation and the respective micro-hardness values, D₁ is computed as 0.0168 and 0.0139 for the HAZ and TMAZ. Likewise, D₂ is computed from the constant ratio D₂/D₁=28.5 as 0.479 and 0.397 for the HAZ and TMAZ, respectively. Thus, the application of the first-level homogenization procedure described above yielded the previously unknown Johnson-cook strength and failure-model parameters for the HAZ and TMAZ in AA2139.
To validate the procedure described above, a simple quasi-static finite-element analysis of the transverse-tensile test is conducted in which each weld zone is represented by a single homogenized material. The resulting stress-strain curve (labeled “Computational, Without Residual Stresses”) is compared with its experimental counterpart (labeled “Experiment [5.10]”), Figure 5-7(a). It is seen that only a fair agreement is obtained between the computational and experimental curves with respect to the initial-yielding portion of the curve, the overall hardening behavior and the final strain to failure. It should be noted that until now, no consideration was given to the presence of residual stresses within the different FSW weld-zones. While, this may be justified for the residual stresses aligned with the axial direction of the tensile sample, similar stress-relaxation effects cannot be assumed in the welding direction. To determine the effects of the latter residual stresses on the stress-strain behavior of the weldment in the transverse direction, the residual stress results displayed in Figure 5-6(b) are used to define the initial-stress condition in the aforementioned quasi-static finite-element analysis. The result of this analysis is also shown in Figure 5-7(a) (the curve labeled “Computational, With Residual Stresses”). It is seen that substantial improvements in the experiment/computation agreement is obtained by accounting for the presence of residual stresses. The distribution of the Johnson-Cook damage variable, D, at the onset of fracture is displayed in Figure 5-7(b). It is seen that failure occurs in the HAZ and this finding is fully consistent with the experimental observations [5.10]. Based on the foregoing findings, it was concluded that the first-level homogenization procedure, within which each weld-zone is treated as a separate (homogenized) material and within which the effect of residual stresses is accounted for is physically sound.
Figure 5-7. (a) The predictions of the transverse stress/strain tensile curves and (b) the spatial distribution of the Johnson-Cook damage parameter at the onset of failure in the case of the first-level weld-material homogenization procedure; (c) and (d) the corresponding results for the case of the second-level weld-material homogenization procedure. Please see text for details.
Figure 5-7. continued…

(c) Engineering Stress vs. Engineering Strain

- Experiment [10]
- Computational, Without Residual Stresses
- Computational, With Residual Stresses

(d) Weld Failure in the HAZ
5.5.2. The Second-level Homogenization

In this section homogenized weld-zone properties are combined into a single homogenized material representative of the entire weld.

To determine the initial strength of the resulting material, it is taken into account that the zones are fully joined and thus their mechanical response is fully kinematically coupled, i.e. the softer material will be restrained by the bordering harder-material and will yield at a higher stress-level than its yield stress. Based on this argument, it is assumed that the Johnson-Cook strength parameter $A$ for the entire weld is a simple volume-based weighted average of the HAZ, TMAZ and the weld-nugget $A$ parameters, i.e.

$$A_{weld} = f_{HAZ} \cdot A_{HAZ} + f_{TMAZ} \cdot A_{TMAZ} + f_{Nugget} \cdot A_{Nugget}$$

where, $f$ represents the respective weld-zone volume fraction.

As far as ductility of the weld is concerned, it is assumed to be dominated by its least ductile zone and hence,

$$\frac{1}{D_{1,weld}} = \frac{f_{HAZ}}{D_{1,HAZ}} + \frac{f_{TMAZ}}{D_{1,TMAZ}} + \frac{f_{Nugget}}{D_{1,Nugget}}$$

Using the procedure described above, $A_{weld}$ and $D_{1,weld}$ are determined as 215MPa and 0.0152, respectively, while, $D_{2,weld} = 28.5D_{1,weld}$. The remaining strength and failure weld parameters are set equal to their individual weld zone counterparts.

To validate the aforementioned homogenization procedure, the entire weld is modeled using a single homogenized material and the quasi-static finite-element analyses (without and with the considerations of residual stresses) of the transverse tensile test repeated. The results of the analyses are shown in Figure 5-7(c). As in the case of Figure 5-7(a), it is seen that the
inclusion of the residual stresses improves the extent of experiment/computation agreement. In addition, as expected, this extent of agreement is somewhat compromised (but still acceptable) in the case of the second-level homogenization (Figure 5-7(a) vs. Figure 5-7(c)). The results displayed in Figure 5-7(d) show that the overall distribution of the Johnson-Cook damage parameter at the onset of failure and the fracture location are correctly predicted in the case of the second-level homogenization procedure. These finding are quite encouraging and suggests that the second-level homogenization procedure also yields physically sound results. Further validation and verification of the second-level homogenization procedure will be provided in the next section.

5.6. Validation and Verification

In this section, the foregoing two-level material homogenization procedure is validated within the context of blast-survivability computational analyses of the military-vehicle test structures.

5.6.1. Transient Non-linear Dynamics Modeling of Blast Survivability

5.6.1.1. General Considerations

First, a brief description is given of the computational analysis used to simulate the interactions between the detonation-products/soil ejecta resulting from the explosion of a mine shallow-buried in soil under a military-vehicle test structure. The computational modeling of these interactions involved two distinct steps: (a) geometrical and mesh modeling of the test structure along with the accompanying mine and soil regions, and (b) the associated transient non-linear dynamics analysis of the impulse loading (momentum transfer) from the detonation-products/soil ejecta to the test structure and the kinematic and dynamic response of the structure.
All the calculations carried out in this portion of the work were done using ABAQUS/Explicit, a general-purpose transient non-linear dynamics analysis software [5.19]. In our previous work [5.9], a detailed account was provided of the basic features of ABAQUS/Explicit, emphasizing the ones which are most relevant for modeling detonation of shallow-buried and ground-laid mines and the subsequent interactions between detonation products, soil ejecta and the test structure. Therefore, only a brief overview of ABAQUS/Explicit is given in this section.

A typical transient non-linear dynamics problem such as the interactions between shallow-buried mine detonation products and soil ejecta with the test structure is analyzed within ABAQUS/Explicit by solving simultaneously the governing partial differential equations for the conservation of mass, linear momentum and energy along with the material constitutive equations and the equations defining the initial and the boundary conditions. The aforementioned equations are solved numerically using a second-order accurate explicit scheme. The ABAQUS/Explicit computational engine solves the governing equations within a Lagrange framework, i.e. the computational finite-element grid is tied to the attendant components/materials (soil, the mine and the test structure, in the present case) and moves and deforms with them.

Interactions between the various components of the model (mine detonation products, soil and the test-structure, in the present case) are typically accounted for using the “Hard Contact Pair” type of contact algorithm. Within this algorithm, contact pressures between two bodies are not transmitted unless the nodes on the “slave surface” contact the “master surface”. No penetration/over closure is allowed and there is no limit to the magnitude of the contact pressure that could be transmitted when the surfaces are in contact. Transmission of shear stresses across the contact interfaces is defined in terms of a static and a kinematic friction coefficient and an
upper-bound shear stress limit (a maximum value of shear stress which can be transmitted before the contacting surfaces begin to slide).

In a typical blast-survivability test-structure computational analysis, the following steps are taken: (a) at the beginning of the simulation, the test structure, the mine and the soil are all assumed to be at rest (with the gravitational force acting downward); (b) mine detonation is next initiated either over the entire bottom face of the mine or at the bottom center; and (c) the mechanical response of the test structure to impact by the soil ejecta and the detonation products is monitored in order to quantify the test structure blast-survivability. To ensure fidelity of this approach, i.e. in order to ensure that the results obtained are insensitive to the size of the elements used, a standard mesh-sensitivity analysis needs to be carried out (the results not shown for brevity).

5.6.1.2. Geometrical and Meshed Models

Military Vehicle Test Structure: A geometrical model of the military vehicle test structure analyzed in the present work is depicted in Figure 5-8. The CAD model shown in this figure was created in accordance with the test structure description provided in Ref. [5.25]. It is seen that the test structure represents the forward one-third portion of a typical Advanced Amphibious Assault Vehicle (AAAV) which is designed to withstand severe ballistic/blast threats. The test structure assembly (with an overall length of ca. 2.6m and width of 1.7m) consists of the following AA2139-T8 FSWed components: (a) 25.4mm-thick floor plate; (b) 25.4mm-thick lower glacis (representing the lower forward portion of the test structure); (c) 50.8mm-thick sidewalls; (d) chine actuator mounts fabricated from 25.4mm and 50.8mm thick plates; and (e) 25.4mm-thick transition piece connecting the lower glacis and the floor plate.

The CAD model was next preprocessed (meshed) using the general purpose pre-processing program HyperMesh from Altair Inc. [5.21]. The resulting meshed model of the test
structure consists of approximately 700,000 and 9,000,000 six- and eight-node prismatic and 4-node tetrahedron first-order reduced-integration continuum elements when the weld-zone is represented as a single zone and multiple-zones, respectively.

![Image](image_url)

**Figure 5-8. An example of the (sub-scale) vehicle-underbody structure.**

*Mine and Soil Regions:* The mine and soil computational domains used in the present study are shown in Figure 5-8. The size and circular-disk shape of the mine computational domain are selected to match that of a typical 7kg anti-vehicle C4 mine used in Ref. [5.25]. The mine computational domain was meshed using eight-node first-order reduced-integration continuum elements with a typical edge length of 5mm and filled with a C4 HE material.

The soil computational domain was modeled as a solid cuboid with L x W x H = 3400mm x 3400mm x 1500mm. The domain was divided into three concentric sub-domains. All three sub-domains were meshed using eight-node reduced-integration continuum elements with a typical edge length of 5mm in the inner-most sub-domain and a typical edge length of 50mm in the outer-most sub-domain. The lateral and the bottom faces of the soil domain were subsequently surrounded with eight-node infinite elements in order to model far-field soil regions.
and avoid un-physical stress-wave reflection at the soil-domain lateral and bottom surfaces. The soil domains containing non-infinite elements were filled with CU-ARL soil material (discussed later) while the infinite elements were filled with an “elastic” soil material with a Young’s modulus and a Poisson’s ratio matching those of the CU-ARL soil.

5.6.1.3. Material Models

As discussed above, the complete definition of a transient non-linear dynamics problem entails the knowledge of the material models that define the relationships between the flow variables (pressure, mass-density, energy-density, temperature, etc.). These relations typically involve: (a) an equation of state; (b) a strength equation; (c) a failure equation and (d) an erosion equation for each constituent material. These equations arise from the fact that, in general, the total stress tensor can be decomposed into a sum of a hydrostatic stress (pressure) tensor (which causes a change in the volume/density of the material) and a deviatoric stress tensor (which is responsible for the shape change of the material). An equation of state then is used to define the corresponding functional relationship between pressure, mass density and internal energy density (temperature). Likewise, a (constitutive material) strength relation is used to define the appropriate equivalent plastic strain, equivalent plastic strain rate, and temperature dependencies of the materials yield strength. This relation, in conjunction with the appropriate yield-criterion and flow-rule relations, is used to compute the deviatoric part of stress under elastic-plastic loading conditions. In addition, a material model generally includes a failure criterion, (i.e. an equation describing the hydrostatic or deviatoric stress and/or strain condition(s) which, when attained, cause the material to fracture and lose its ability to support (abruptly in the case of brittle materials or gradually in the case of ductile materials) normal and shear stresses. Such failure criterion in combination with the corresponding material-property degradation and the flow-rule relations governs the evolution of stress during failure. The erosion equation is generally
intended for eliminating numerical solution difficulties arising from highly distorted elements. Nevertheless, the erosion equation is often used to provide additional material failure mechanism especially in materials with limited ductility.

To summarize the above, the equation of state along with the strength and failure equations (as well as with the equations governing the onset of plastic deformation and failure and the plasticity and failure induced material flow) enable assessment of the evolution of the complete stress tensor during a transient non-linear dynamics analysis. Such an assessment is needed where the governing (mass, momentum and energy) conservation equations are being solved. Separate evaluations of the pressure and the deviatoric stress enable inclusion of the nonlinear shock-effects in the equation of state.

In the present work, the following materials are utilized within the computational domain: C4 HE explosive, AA 2139-T8 (base metal, various weld-zones and the weld as a whole) and soil. Since a detailed account of the constitutive models used to represent the behavior of the materials in question can be found in our recent work [5.9], only a brief qualitative description of these models will be provided in the remainder of this section.

**C4 HE Explosive:** The Jones-Wilkins-Lee (JWL) equation of state [5.22] is used for C4 in the present work since that is the preferred choice for the equation of state for high-energy explosives in most hydrodynamic calculations involving detonation. Within a typical hydrodynamic analysis, detonation is modeled as an instantaneous process which converts un-reacted explosive into gaseous detonation products and detonation of the entire high-explosive material is typically completed at the very beginning of a given simulation. Consequently, no strength and failure models are required for high-energy explosives such as C4.

**AA 2139-T8:** Since hydrostatic stress gives rise to only minor reversible density changes in metallic materials like AA 2139-T8, a linear type of equation of state was used for AA2138-T8.
As discussed earlier, to represent the constitutive response of AA 2139-T8 (base-metal and weld) under deviatoric stress, the Johnson-Cook Strength model [5.14] is used. Since AA2139-T8 base-metal and weld both exhibit a ductile mode of failure, their failure condition was defined using the Johnson-Cook failure model [5.15]. Erosion of AA2139-T8 components is assumed to take place when the Johnson-Cook damage state-variable $D$, as defined by Eq. (3), reaches a value of 1.0. When a material element is eroded, its nodes are retained along with their masses and velocities in order to conserve momentum of the system. The momentum is conserved by distributing the mass and velocities associated with the eroded elements among the corner nodes of the remaining elements. Despite the fact that some loss of accuracy is encountered in this procedure (due to removal of the strain energy from the eroded elements), the procedure is generally found to yield reasonably accurate results [5.14].

Soil: Soil is a very complicated material whose properties vary greatly with the presence/absence and relative amounts of various constituent materials (soil particles, clay, silt, gravel, etc.), and particle sizes and particle size distribution of the materials. In addition, the moisture content and the extent of pre-compaction can profoundly affect the soil properties. To account for all these effects, Clemson University and the Army Research Laboratory (ARL), Aberdeen, Proving Ground, MD jointly developed [5.26-5.28] and subsequently parameterized (using the results of a detailed investigation of dynamic response of soil at different saturation levels, as carried out by researchers at the Cavendish Laboratory, Cambridge, UK [5.29]) the CU-ARL soil model. This model (used in the present work) is capable of capturing the effect of moisture on the dynamic behavior of soil and was named the CU-ARL soil model.

For the CU-ARL soil model, a saturation-dependant porous-material/compaction equation of state is used which, as shown in our previous work [5.26] is a particular form of the Mie-Gruneisen equation of state [5.30]. Within this equation, separate pressure vs. density
relations are defined for plastic compaction (gives rise to the densification of soil) and for unloading/elastic-reloading. Within the CU-ARL soil strength model, the yield strength is assumed to be pressure dependant and to be controlled by saturation-dependant inter-particle friction. In addition to specifying the yield stress vs. pressure relationship, the strength model entails the knowledge of the density and saturation dependent shear modulus. Within the CU-ARL soil failure model, failure is assumed to occur when the negative pressure falls below a critical saturation-dependant value, i.e. a “hydro” type failure mechanism was adopted. After failure, the failed material element loses the ability to support tensile or shear loads while its ability to support compressive loads is retained. Erosion of a soil element is assumed, within the CU-ARL soil erosion model, to take place when geometrical (i.e. elastic plus plastic plus damage) instantaneous strain reaches a maximum allowable value. The investigation reported in Ref. [5.27] established that the optimal value for the geometrical instantaneous strain is \( \sim 1.0 \).

5.6.2. Results and Discussion

The foregoing computational analysis of mine-blast and of subsequent interactions between detonation-products/soil-ejecta and the target structure was conducted in such a way that it would reveal the intrinsic blast-survivability of the structure. While the geometrical models used are somewhat simplified, they still retain the essential structural details of a vehicle underbody. Typically, blast survivability of a vehicle test-structure is judged by a lack of penetration of the structure by the soil ejecta and gaseous detonation products and by the absence of excessive deflection. In addition, in the case when the test structure has survived mine-blast impact, the extent of its damage is quantified in order to estimate the potential loss of vehicle mobility and the extent of repair needed to make the structure suitable for future use.

Examples of the typical (qualitative) results pertaining to the floor-plate total displacements and the associated extents of weld failure obtained in this portion of the work are
depicted in Figures 5-9(a)-(b). Figure 5-9(a) displays the results obtained using a computationally more expensive analysis, in which the different weld-zones are represented explicitly. For comparison, Figure 5-9(b) displays the corresponding results obtained in a computational analysis in which the weld-zones were homogenized into a single weld domain. Due to the sensitive nature of the subject matter and the potential for misuse of the quantitative results, quantitative details pertaining to the results displayed in Figures 5-9(a)-(b) could not be presented here. What could be said is that under a relatively large range of mine-blast loading conditions (associated with different mine shape and size, depth of burial, stand of distance and mine placement relative to the test structure), a fairly good agreement was obtained between the results of more detailed and the more efficient computational analyses. Typically, the penetration/no-penetration condition was correctly predicted, maximum deflection differed by less than 7%, the location of the welded structure cracking was correct and the crack propagation direction was consistent. What was not always correctly predicted by the computationally more efficient analysis was the extent of crack propagation (generally over-predicted) and the overall degree of weld cracking (generally over-predicted).
Figure 5-9. A comparison of the results obtained using: (a) a computational analysis with explicit weld-zone representation and (b) a computational analysis with homogenized weld-domains. Please check text for details.
5.7. Summary

Based on the work presented and discussed in the present manuscript, the following main summary remarks and conclusions can be made:

1. A two-step weld-material homogenization procedure is introduced in order to reduce the computational cost associated with transient non-linear dynamics analyses of military-vehicle test-structure blast survivability.

2. To demonstrate the utility of this procedure, microstructure, mechanical properties and residual stresses are characterized for the case of AA2139-T8 friction-stir weldments.

3. Homogenization of different weld-zone materials (and the weld as a whole) is carried out within the context of Johnson-Cook deformation/strength and failure material models for the vehicle test-structure.

4. The procedure is validated by comparing the associated blast-survivability vehicle test structure computational results with their computational counterparts obtained in a substantially more costly analysis in which welds are represented in more details.
5.8. References


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CHAPTER 6

CONCLUSIONS AND FUTURE WORK

6.1. General Discussion and Concluding Remarks

As will be recalled, the overall objective of the present dissertation was to enable the introduction of the friction stir welding (FSW) process into the design/development of sub-scale military-vehicle underbody structures using a time-efficient, cost-effective and robust approach. It is believed that the research contained herein has achieved this in the following ways:

1. A new concurrent approach to designing, manufacturing and testing of military vehicle-underbody friction stir welded structures is developed. The proposed approach involves a number of well-defined steps. These steps are highly interactive and often occur concurrently. This research provides confidence that employing the developed approach in the design/development of sub-scale military vehicle test structures would significantly reduce the design lead time and costs while improving the test vehicles blast/ballistic survivability.

2. The concurrent approach includes development of a comprehensive fully-coupled thermo-mechanical finite-element computational model of the friction stir welding (FSW) of prototypical solid-solution strengthened (AA5083) aluminum alloy. Initially, the developed computational model accounted for the microstructure/property evolution of the material during the FSW process through a modified Johnson-Cook strength model to account for the effects of dynamic recrystallization and the associated material softening taking place in the stir zone of the welded joint.

3. Further, the computational model of the friction stir welding process was extended through the development and parameterization of simple mathematical models for the hardness evolution within various friction-stir weld zones (e.g. the weld nugget, the thermo-mechanically affected zone and the heat affected zone) for AA5083 (a solid-solution strengthened and strain-
hardened/stabilized Al-Mg-Mn alloy) and AA2139 (a precipitation hardened quaternary Al-Cu-Mg.Ag alloy) aluminum alloys. The thermo-mechanical history information obtained from the fully-coupled thermo-mechanical finite-element analysis of the friction stir welding process was used in the integration of the hardness and grain-size evolution equations over the thermo-mechanical history of various material points within the weld to yield a hardness/grain-size profile (one for each alloy) in a direction transverse to the weld line.

4. A comparison of the FSW process computational analyses results with their experimental counterparts with respect to the material hardness, grain-size and residual stress profiles, revealed a fairly good agreement.

5. One of the steps in the developed concurrent approach involves the use of computer-aided non-linear transient dynamics engineering analyses in order to predict (computationally) blast-survivability of the military vehicle (look-alike) test structures. Towards this end, a two-level weld-material homogenization procedure was introduced in order to reduce the computational cost associated with transient non-linear dynamics analyses of military-vehicle test-structure blast survivability. Homogenization of different weld-zone materials (and the weld as a whole) was carried out within the context of Johnson-Cook deformation/strength and failure material models for the vehicle test-structure. Finally, the homogenization procedure is validated by comparing the associated blast-survivability vehicle test structure computational results with their computational counterparts obtained in a substantially more costly analysis in which welds are represented in more details.

6.2. Suggestion for Future Work

In the present work, the level of agreement between the computational results and their experimental counterparts obtained from open literature can be characterized as being only fair. The primary reasons for the observed variation are: (a) some of the FSW-tool geometric
parameters and process parameters for the publically available experimental investigations were unknown. Therefore, it is desirable to obtain all the FSW-tool geometry and process parameters for the experimental investigations utilized for validating the computational results; (b) The functional relations used to describe the contribution of various mechanisms to material hardness should be further improved; and (c) The experimental data used for model parameterization were relatively scarce and came from different sources.

In addition to the above, within the present work, only the effect of the dynamic recrystallization of the weld material during the FSW process was accounted for. However, a detailed analysis of the precipitates coarsening, over-aging, dissolution and re-precipitation occurring during the FSW process is necessary to further improve the fidelity of the present computational approach. This can be accomplished by including realistic and specific material microstructure evolution equations and by effectively utilizing advanced thermodynamics analysis programs (e.g. Thermo-Calc) along with advanced programs (e.g. DICTRA) for accurate simulations of diffusion in multicomponent alloy systems. Finally, it is essential to gain a better understanding of the relationships between the FSW process-parameters/tool-geometry, work piece material flow and weld zone microstructure/properties and the underlying phenomena in order to develop more accurate and reliable FSW-process computational models.