Controlling Weldment and Metallurgical Properties Through Process Control in Rotary Friction Welding

Brandon Scott Boyer Taysom
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Controlling Weldment Properties Through Process Control

in Rotary Friction Welding

Brandon Scott Boyer Taysom

A dissertation submitted to the faculty of
Brigham Young University
in partial fulfillment of the requirements for the degree of

Doctor of Philosophy

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ABSTRACT

Controlling Weldment Properties Through Process Control in Rotary Friction Welding

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Doctor of Philosophy

Weld quality in the context of process control and internal conditions is studied. Several different alloys are welded including plain carbon steel, high-temperature steels, and several traditional and advanced superalloys.

Across all studied weld systems, the following conditions led to stronger welds: higher forces and feedrates, lower temperatures, and moderate or limited upsets. In the best cases, post-weld strengths were nearly equal to basemetal strength.

Tradition holds that large and symmetric upsets are necessary for good welds, but this study contradicts that notion. The fundamental requirements for bonding are two sufficiently clean surfaces in intimate contact. Only minimal upset is necessary to achieve that. In welding alloy 718, only 1 mm of feed (or 0.4 mm of sample upset) was necessary to achieve >95% of basemetal strength. In an advanced superalloy with low ductility, very low upsets were required in order to achieve high joint strength. For this low-ductility alloy, using a containing geometry increased both the internal pressure and ductility of this alloy, leading to a much larger window of sound welding conditions and stronger welds overall. In several dissimilar alloy systems, the relationship between force/feedrate and upset asymmetry varied between each alloy, but a more symmetric upset did not correlate to stronger welds.

Advanced process control in FW was also performed with closed-loop temperature control and open-loop predictive cooling rate control. Using this technique, martensitic microstructures associated with a fast natural cooling rate were avoided, and a pearlitic microstructure was obtained. The yield and tensile strength of the weld was not adversely affected, and both were within range of published values for the basemetal.

Keywords: rotary friction welding, friction welding, superalloy, process control
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CHAPTER 1. INTRODUCTION

1.1 Overview of Friction Welding

Rotary Friction Welding (RFW), often simply called Friction Welding (FW), is a solid-state joining process for axisymmetric metallic components. In FW, both parts are rigidly held, and while one is rotated the pieces are brought into contact with each other under a high axial force. The rotation under force creates high frictional heating, significantly softening the material. Once soft enough, the interface deforms, and the heat and pressure create an extremely strong joint. This process is shown graphically in Figure 1.1.

![Friction Welding Diagram](image)

Figure 1.1: Schematic showing the primary stages in FW [1].

There are several subsets of this technology. In Direct Drive Friction Welding (DDFW) or Continuous Drive Friction Welding (CDFW), a large motor directly drives the spindle axis and controls rotation during the process. In Inertial Friction Welding (IFW), a medium-sized motor is used to spin up a large flywheel which is attached to one of the pieces, and the kinetic energy of the flywheel is used for welding. IFW is particularly preferred when joining large pieces where a large enough spindle motor is not be available or feasible. Sister solid state joining technologies include Friction Stir Welding (FSW) and Linear Friction Welding (LFW). While many of the details of
these welding processes differ, all are grounded in the joining of metals at elevated temperature by causing surface friction and deformation to generate heat.

1.2 Current State of the Art in FW

FW has inherent advantages and disadvantages compared to other welding technologies. Many researchers have shown that FW experiences sub-melting temperatures, and so avoids solidification defects and minimizes the heat affected zone (HAZ) when compared to fusion welding [3, 4]. This leads to superior weld properties. It is also easier to join dissimilar alloys with FW than with arc welding, including different alloys systems [5, 6] and metallurgically incompatible materials [7]. However, FW requires large machinery and can only weld axisymmetric parts, which limits the types of parts that can be joined with this technology.

The amount and symmetry of upset required for a solid weld has been studied by many different researchers. Friction welding has traditionally utilized large upsets during the welding process. Large upsets help to clean surface contaminants [8] which can degrade weld quality [6, 9]. Weld upset generates flash curl extruding from the interface. The color and shape of the flash correlates to internal welding conditions such as temperature, and thus the visual appearance of the flash is often used as a visual indicator of weld quality. Due to these factors, and the fact that most metals have the necessary extended ductility at welding temperatures, large and symmetric upsets have become the goal in much of FW.

Welding parameters have a strong influence on final weld properties [10, 11], just like other solid state joining processes like FSW and LFW [12]. Surface cleanliness has a large impact on weld quality, particularly if the surfaces are not prepared well. [6, 9]. Force is needed both to generate heat and to prevent voids from opening at the weld interface [10, 11]. Excessive temperature, however, leads to a large HAZ and low strength [13], demonstrating some of the trade-offs that exist in welding parameters

1.3 Gaps in the State of the Art in FW

Most FW has been performed in alloys with extended high-temperature ductility. This allows these alloys to accommodate the large strains that occur during FW. Problematically, next-
generation superalloys have been designed to be creep resistant at very high temperatures, and thus have low ductility at high temperatures. During welding, the high strains lead to catastrophic failure and blowout near the weld interface instead of controlled flash generation. This behavior has made it incredibly hard to create solid and reliable welds in advanced superalloys.

Much of the research surrounding welding has focused on purposefully creating large upsets of both workpieces. The goal is to create a clean surface which will be conducive to new metallic bonding. While an common assumption, it is not clear that symmetric deformation and flash generation are necessary for a good bond, or for that matter that increased welding time and energy leads to symmetric flash. Furthermore, welds that are longer than necessary impart more thermal energy to the workpieces, thus creating more intense and larger flash regions. Despite this, creating welds that are either faster and/or lower temperature in order to achieve better welds has not been a focus of research up to this point. Temperature history is the primary driver for microstructure evolution of metals, and thus controlling the temperature history can result in more desirable microstructure.

Many solid state processes are often treated as a “black box”, in which weld parameters are used to drive final weld properties. In some cases this approach is sufficient. However, this approach breaks down in complicated cases where welding is more difficult. In this case, understanding how input parameters affect intermediary and metallurgical parameters is essential in order to achieve optimal weld properties.

1.4 Dissertation Structure

The body of this dissertation is composed of a series of independent but interrelated papers. At the time of writing, each of these papers is either accepted by a journal, under review, or being prepared to be submitted to a journal. Each paper is included in its entirety, including introduction and FW overview sections, with small changes to formatting and style.

The first paper in Chapter 2 covers the joining of dissimilar materials. This paper sets forth many general welding principles, such as that welding with higher forces and lower temperatures usually leads to stronger welds. These principles are further studied and reinforced in later chapters. Chapter 3 goes into further depth of a single dissimilar steel welding system. An analysis of the microstructure is presented, and this is shown to directly affect the strength and failure lo-
cations of welds. Chapter 4 discusses the fundamental requirements for bonding in FW. Building on what is and isn’t required for creating bonds in FW, it is shown that previously “un-weldable” low-ductility advanced superalloys can in fact be welded. Emphasis is given to the fact that large upsets, which are common in traditional FW, are not only unnecessary but can in fact weaken a joint. Chapter 5 goes into detail about process control in FW and microstructure. Closed loop control of FW is developed and demonstrated, and with this the properties and microstructure of the weldment are carefully controlled.
CHAPTER 2. ROTARY FRICTION WELDING OF DISSIMILAR NICKEL-BASED SUPERALLOYS

2.1 Introduction

Rotary Friction Welding (RFW or FW) is a solid-state joining process used to join axisymmetric parts [14]. In FW, one workpiece is rotated with respect to another, and the two pieces are then pushed together under high axial loads. The pressure and friction create heat, which leads to surface deformation and eventually to a strong metallic joint between the two components. With proper welding parameters, FW can produce higher-strength, defect-free joints with lower peak temperatures and smaller HAZ regions than can fusion-based welding [3, 4].

Welding parameters have a strong influence on final weld properties. Surface cleanliness has a large impact on weld quality, particularly if the surfaces are not prepared well. [6, 9]. Multiple researchers have shown that welding input parameters have a large effect on weld quality [10, 11]. Force is needed both to generate heat and to prevent voids at the weld interface [10, 11]. Excessive temperatures, however, lead to a large HAZ and low strength [13], and thus trade-offs exist in welding parameters.

While the welding of workpieces with different characteristics is more difficult than that of two identical workpieces, solid state joining processes have inherent advantages in dissimilar workpiece welding. In fusion welding, both workpieces must be melted, which is problematic for alloys with vastly different melting temperatures. In FW, workpieces are never melted, and thus many different alloys can be joined together [5]. Even alloys with drastically different melting temperatures such as aluminum and stainless steels have been successfully welded with FW [7].

While most of the body of FW research agrees on several points (such as force being required for a solid weld), research has focused on different things being required for a good weld. Bell found that while the appearance of flash did not correspond to weld quality, a low amount of flash/upset was nevertheless indicative of a bad weld [9]. While symmetry of deformation is some-
times though of as desirable in FE, several studies have not identified uniformity of deformation as requirement for good bonding [7,10,15]. Muralimohan’s work found that long welding times lead to better welds [11], whereas Sahin’s work found that (after a minimum time) extra welding time decreased weld strength [10]. This work specifically seeks to investigate the important welding factors for dissimilar nickel-based superalloys, particularly which parameters influence welding strength and which do not.

2.2 Materials and Methods

Alloy 600, alloy 625, alloy 718, Waspaloy, and Udimet 720 were used in this study. Two specimen geometries were used. Solid cylinders were machined with a 1.000 in (25.4 mm) diameter, and tubular samples were machined to an outer diameter of 1.000 in (25.4 mm) and an inner diameter of 0.800 in (20.3 mm) about 1 in (25 mm) deep. K-type thermocouples were spot-welded onto the exterior of the stationary samples at distances of 1, 2, 3, 4, and 5 mm away from the interface. Fiducial marks were placed 10 mm away from the interface to enable measuring the post-weld upset of each specimen.

All welds were performed on a TTI RM2 High-Stiffness FSW machine. The machine was retrofitted with a Bond Technologies B&R based controller. A torque meter was attached to the welding bed to hold the lower non-rotating sample and to measure torque. Other recorded measurements include spindle speed, triaxial forces, thermocouple temperatures, axial position, and velocity.

Prior to welding, all specimens were brought into contact with 40 kN of force, a zero-position reference was established at 0 kN, and the samples were then moved 1 mm apart. The spindle was started, and the machine moved with a constant feedrate until the total feed (5 mm) was reached. Starting at about 0.6 seconds before the total feed was reached, the spindle decelerated from 1000 rpm to 0 rpm over about 0.4 seconds.

After welding, the distance from the fiducial marks to the final weld interface was measured in order to calculate the sample upset ratio. As-welded samples were pulled in tension until failure in a 450 kN capacity MTS tensile tester. The ultimate tensile strength (UTS) and fracture location were recorded.
The logarithm of feedrate was used in statistical analysis instead of feedrate itself. Normalized strength was calculated as the ratio of as-welded UTS vs. basemetal UTS of the weaker alloy at room temperature. Normalized force was calculated by dividing the peak welding force by the original cross section area, and then dividing this by the basemetal UTS of the weaker alloy at room temperature. This was done in order to more easily make comparisons across alloy systems.

2.3 Experiments and Results

Welds were performed at feedrates of 0.5, 1, 2, and 4 mm/s, a total feed of 5 mm, and a spindle speed of 1000 rpm. This was done for all of the following alloy/geometry combinations: tube 718 - solid 718, tube 600 - tube 718, tube 718 - tube 625, tube 718 - Waspaloy, tube 600 - tube Udimet 720. The normalized welding force, peak temperature, and normalized ultimate strength are plotted against the varied parameter (welding feedrate, on a logarithmic scale) for each alloy/geometry system, and are shown in Figures 2.1, 2.2, and 2.3 respectively.

For all of the alloy and geometry systems studied, the highest strength welds occurred at high feedrates, which also correlated with high welding forces and low peak temperatures. These variables all have a strong covariance regardless of the alloy/geometry used. Figure 2.1 is also notable in that welding force trends were extremely similar once the forces were normalized against room temperature alloy strength, despite a nearly two times strength difference between the strongest and weakest alloy.

![Figure 2.1: Normalized peak axial forces versus feedrate. Force increases with feedrate for all alloy/geometry systems.](image)
Figure 2.2: Maximum observed thermocouple temperature versus feedrate. For each alloy system, as feedrate increased, temperature decreased.

Figure 2.3: Normalized weld tensile strength versus feedrate. Strength increased with feedrate, with strengths approaching basemetal strength for some alloy systems. The highest feedrate T 718 - T 625 and T 600 - T Ud720 welds experienced extensive necking in the softer basemetal and failed in the softer basemetal instead of in the weld joint.

Weld upset asymmetry also changed with operating conditions. Cross sections of welds between alloys 625 and 718 at approximately 1 and 4 mm/s are shown in Figure 2.4. In these welds, upset asymmetry was mostly eliminated by welding at higher feedrates. The relationship between weld upset asymmetry and feedrate is shown quantitatively for all alloys’ welds in Figure 2.5. For some of the alloy sets, upset asymmetry increased with feedrate, whereas for others it decreased.
Figure 2.4: Cross section of welds between alloys 625 (top, black) and 718 (bottom, gray). As the weld moved from about 1 mm/s (a) to about 4 mm/s (b) the level of asymmetry present dramatically decreased.

Figure 2.5: Ratio of specimen upsets. Values closer to 1 indicate a more symmetric weld.

2.4 Analysis and Discussion

Statistical analysis was performed on the weld data. Multiple covariate parameters were considered separately to avoid the artificially increased uncertainty that occurs from having multiple covariate parameters in a model. The three highly covariate input/intermediary parameters (log(feedrate), normalized force, and temperature) were each used in different mixed linear models along with the upset asymmetry and alloy/geometry combination to predict normalized weld strength. Plots of weld strength against each of these variables are shown in Figures 2.3, 2.6, and 2.7.

The p-value estimates for terms in three separate models containing the weld upset asymmetry and the three covariate parameters are given in Table 2.1. Numerically, this suggests that the three covariate parameters affect weld strength, whereas weld upset asymmetry may not have a statistically significant influence on weld strength. A plot of normalized weld strength vs. upset
Figure 2.6: Normalized weld strength plotted against normalized peak welding force. Strength increases with increasing welding force.

Figure 2.7: Normalized weld strength plotted against peak measured welding temperature. Higher strength welds occurred at lower temperatures.

Potential trends exist within each alloy/geometry system. However, no trend exists between the weld strength and specimen upset asymmetry that bridge across different alloy/geometry systems.

Table 2.1: p-value Estimates for Parameters in Three Different Models

<table>
<thead>
<tr>
<th></th>
<th>log(Feedrate) &amp; Asymmetry</th>
<th>N. Force &amp; Asymmetry</th>
<th>N. Strength &amp; Asymmetry</th>
</tr>
</thead>
<tbody>
<tr>
<td>$p_{\text{log(Feedrate)}}$</td>
<td>5.7e-6</td>
<td>1.0e-4</td>
<td>5.0e-5</td>
</tr>
<tr>
<td>$p_{\text{Asymmetry}}$</td>
<td>0.14</td>
<td>0.21</td>
<td>0.045</td>
</tr>
</tbody>
</table>
All possible combinations of models including terms of log(feedrate), normalized force, peak temperature, upset asymmetry, and alloy/geometry system were considered to predict normalized welding force. Different statistical fitness metrics were calculated for each model including Cp, AiCC, BIC, and RMSE. The model with the terms of log(feedrate) and alloy/geometry system had the highest adjusted $R^2$ at 0.77. This model also had the lowest Cp of 0.33, lowest AICc of -55.9, lowest BIC of -54.6, and lowest RMSE of 0.049. Adding additional terms made the adjusted $R^2$ (and all other weighted fit parameters) worse because these parameters are either covariate with log(feedrate) or don’t predict the strength well.

![Figure 2.8: Normalized weld strength vs. upset asymmetry. Although most individual systems might have trends, no trends across the different alloy/geometry systems exist.](image)

Fundamentally, a difference in strength causes upset asymmetry of the weld pieces. This can be due to a difference in alloy, or in the case of geometry different thermal gradients. Either way, when a strength difference exists, the weaker alloy will deform faster than the stronger alloy. This increase of deformation increases internal strainrates, leading to strainrate hardening which then leads to a balance of forces during welding.

Symmetry of deformation is often believed to be evidence of equal cleaning of both weld surfaces, thus increasing weld strength (ref). However, this study found no correlation between weld upset symmetry and weld strength that was consistent between multiple alloy/geometry systems.
2.5 Conclusions

Welds were created in several dissimilar superalloy systems, with differences in both weld alloy and weld geometry. Sound welds were created in all alloy and geometry systems. Input parameters and intermediary measured variables that correlated with the strong welds were high feedrates, high forces, and low temperatures. These factors led to good welds across all of the studied weld systems, regardless of the source of the weld asymmetry. These factors are linked together by the fundamental physics of the process, so studying them independently is difficult; i.e. a slower feedrate leads to more heat accumulation and higher temperatures, which in turn leads to a lower force-carrying capacity.

The upset asymmetry did not have a correlation to weld strength across the studied alloy systems. In some systems, a more asymmetric flash was associated with higher weld strength, whereas the inverse was true in different alloy or geometry systems. Reducing upset asymmetry should not be used as a means to improve weld strength. Other factors such as welding forces, feedrates, and temperature are far more important for producing high quality welds.
CHAPTER 3. INVESTIGATION OF STRENGTH IN DISSIMILAR ALLOY ROTARY FRICTION WELDS

3.1 Introduction

Rotary Friction Welding (FW) is a solid-state joining process for axisymmetric components. In FW, both parts are brought into contact with a high axial force while one part is rotating. The rotation under force creates high frictional heating which significantly softens the material. Once the material is soft enough the interfaces deform, and the heat and pressure create an extremely strong joint. FW is a well established technology with many uses and applications as [14].

FW has inherent advantages and disadvantages compared to other welding technologies. Many researchers have shown that FW experiences sub-melting temperatures, and so avoids solidification defects and minimizes the heat affected zone (HAZ) [3, 4]. This leads to superior weld properties. It is also easier to join dissimilar alloys with FW than with arc welding, including different alloys systems [5], and metallurgically incompatible materials [7]. However, FW requires large machinery and can only weld axisymmetric parts, which limits the types of parts that can be joined with this technology.

Many external, input, and intermediary factors affect the quality and performance of friction welds. Surface condition and contaminants can have a stronger effect on weld quality welding parameters in some cases [6,9]. Nevertheless, weld input parameters greatly influence the resultant weld [10], and pressure is needed not only to generate heat, but to stop voids from generating at the interface [11]. Welding energy directly affects temperature, and temperature is necessary for material ductility and diffusion-based bonding [10,11]. However, increased temperature makes the HAZ more severe [13]. Consequently, parameters must be balanced so that the resultant thermo-mechanical cycle results in a high strength and quality joint.

The steel alloys CostE and NFA are of interest in FW because they have potential application for high temperature structural components in power generation systems. CostE is a
high-temperature martensitic steel, and is mainly used by the land-based power industry. NFA is a nanostructured ferritic alloy, and is strengthened by grain size refinement and small oxides [16]. NFA is often considered unweldable by conventional techniques due to melting of the weld pool which results in loss of small oxides and severe grain coarsening [17]. However, multiple individuals have had success using solid-state welding for NFA and other oxide-dispersion strengthened (ODS) alloys. These include friction stir welding [18], rotary friction welding [19], and pressurized resistance welding [20].

This paper focuses on dissimilar alloy welds between Cost-E and NFA steel alloys. Welds were performed under different conditions, and weld quality and performance was analyzed using microhardness, microscopy, and tensile testing. Important parameters and their effects are highlighted. From these results, insight is gained about fundamental characteristics of dissimilar metal joints and the strength limits of these joints.

3.2 Experimental Setup

3.2.1 Materials

The welding materials (CostE and NFA) were chosen because of their applications in land-based power generation systems. The as-received NFA material had an average grain size of about 0.5 \( \mu \text{m} \), and the microstructure of both alloys is shown in Figures 3.1 and 3.2. The composition of both alloys is shown in Table 3.1, and basic mechanical properties are shown in Table 3.2.

![Figure 3.1: As-received NFA basemetal with fine ferritic grain structure.](image)
Figure 3.2: As-received CostE basemetal, showing large prior austenite grains and a martensitic lath structure.

Table 3.1: Nominal Composition of Alloys (wt%)

<table>
<thead>
<tr>
<th></th>
<th>Fe</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>Nb</th>
<th>N</th>
<th>W</th>
<th>Ti</th>
<th>Y</th>
<th>O</th>
</tr>
</thead>
<tbody>
<tr>
<td>CostE</td>
<td>Ball.</td>
<td>0.12</td>
<td>0.5</td>
<td>0.075</td>
<td>0.7</td>
<td>10.3</td>
<td>1.1</td>
<td>0.2</td>
<td>0.055</td>
<td>0.055</td>
<td>1</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>NFA</td>
<td>Ball.</td>
<td>0.01</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>11.25</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.02</td>
<td>3</td>
<td>0.8</td>
<td>0.4</td>
<td>0.2</td>
</tr>
</tbody>
</table>

Table 3.2: Mechanical Properties of NFA and CostE

<table>
<thead>
<tr>
<th></th>
<th>Ultimate Strength</th>
<th>Elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td>CostE</td>
<td>1090 MPa</td>
<td>6%</td>
</tr>
<tr>
<td>NFA</td>
<td>1275 MPa</td>
<td>9%</td>
</tr>
</tbody>
</table>

Material was machined into solid cylinders with a 0.75 in (19 mm) diameter. Due to very limited sample quantity, after most welds the remaining specimen material (unaffected by the weld) was re-machined for use in subsequent welds once the welded portion was separated for analysis.

3.2.2 Welding Setup and Instrumentation

Welds were performed with a TTI High Stiffness RM2 FSW machine with 10 ton spindle and direct drive motor, with a Bond Technologies B&R based high-speed controller and data acquisition system. A custom torque meter was attached to the machine to measure process torque on the sample [21]. Time series information that was collected included: x, y, and z forces; z position and velocity; spindle and motor torque, speed, and power; specimen torque; and specimen thermocouple temperature.
In all experiments, the CostE specimen was mounted to the torque meter, and the NFA specimen was rotated in the spindle. Welds were performed under a constant rotational speed and axial feedrate. After the spindle stopped, feed was continued for a short while in order to maintain pressure at the welding interface.

### 3.2.3 Design of Experiment Setup

Three series of welds were performed: initial screening, a formal design of experiments (DOE), and repeats for tensile test. First, screening welds were performed to determine parameters that could both produce sound welds and remain within the machine capabilities. Second, a three-level independent three-factor DOE was performed using feedrate, feed, and spindle speed as the control parameters. The DOE input parameters are shown below in Table 3.3.

<table>
<thead>
<tr>
<th>Feedrate mm/s</th>
<th>Feed mm</th>
<th>Spindle Speed rpm</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>5</td>
<td>1000</td>
</tr>
<tr>
<td>0.5</td>
<td>5</td>
<td>1000</td>
</tr>
<tr>
<td>2</td>
<td>5</td>
<td>1000</td>
</tr>
<tr>
<td>1</td>
<td>2.5</td>
<td>1000</td>
</tr>
<tr>
<td>1</td>
<td>10</td>
<td>1000</td>
</tr>
<tr>
<td>1</td>
<td>5</td>
<td>750</td>
</tr>
<tr>
<td>1</td>
<td>5</td>
<td>1250</td>
</tr>
</tbody>
</table>

Due to very limited material, tensile tests could be performed at only a few weld conditions. Microhardness and microscopy results from the original DOE were used to determine the most influential input parameter (feedrate) on microstructure. For the third part of the welding experiments, repeats were done at each of the three feedrates (0.5, 1, and 2 mm/s), and tensile samples were extracted axially from the welds.
3.2.4 Analysis Methods

High-resolution microhardness measurements were performed on each of the seven DOE welds. Metallographic samples were cut longitudinally from the weld, and hardness was measured from the center of the weld to the outer surface. A 50g load was used to allow an axial spacing of 50 μm and radial spacing of 100 μm in order to characterize HAZ regions as small as 200 μm wide. In order to reduce measurement time, symmetry was leveraged so that only half of each sample was measured.

A 2D total variation filter [22] was used to smooth the data while maintaining important features such as the HAZ. Thickness and peak hardness of HAZ regions were calculated from the data. As some of the hardness changes from base material to HAZ are gradual, a cutoff of <325 HV and >450 HV were used to calculate the boundaries of the NFA and CostE HAZ, respectively. These values correspond to the minimum CostE tempered hardness and average NFA basemetal hardness, respectively.

Micro-tensile dogbones with a nominal cross section of 1 mm x 2 mm were cut from the repeat welds, netting 12-13 dogbones per weld. The layout of the samples is shown in Figure 3.3. Tensile tests were performed in an Instron with a 5 kN load cell, after which load-displacement data was transformed into stress-strain data. Some tensile samples on the outside of the welds had flash-induced cracks, and are excluded from the results. Micro-tensile samples were also cut from NFA and CostE basemetal, and average properties are shown in Table 3.2.

Scanning Electron Microscopy utilizing channeling contrast was used to determine grain size in NFA near and away from the weld interface. Optical microscopy was used to determine the location of failure in tensile samples relative to the weld interface.

3.3 Results

3.3.1 Microhardness and HAZ Profiles

The microhardness map from the 1 mm/s weld is shown below in Figure 5.10 with select annotations for important weld features. HAZ thicknesses at all radii positions can be determined from the microhardness maps, as is shown for two radii in Figure 3.5. Notably, the CostE quenched HAZ is far wider than the NFA softened HAZ. This is because any portion of the CostE that was
significantly above the A1 temperature will transform into martensite during cooling, whereas the kinetics of NFA grain growth require relatively more time at temperature for significant softening to occur. The NFA HAZ thickness as a function of radius was calculated for all of the welds, and is plotted for the three repeat welds in Figure 3.6. CostE HAZ thickness as a function of radius is not plotted as it did not vary as much nor was correlated with failure mode.

Notably, the 1 and 2 mm/s welds have a narrow NFA HAZ in the center which grows towards the outside, whereas the 0.5 mm/s weld has a NFA HAZ that is thicker and roughly uniform throughout the weld. Non-uniformity of the HAZ is believed to be caused by variable surface velocity which is proportional to radius, thus generating more heat and a thicker HAZ at the outer diameter in the 1 and 2 mm/s weld. More welding time in the 0.5 mm/s weld allowed heat to diffuse to the weld center, thus raising temperatures there and causing significant growth of the HAZ near the weld center.
Figure 3.4: Microhardness maps of the 1 mm/s weld with annotations for important features.

Figure 3.5: Vertical hardness profiles of the 1 mm/s weld taken at radii of 2 and 7 mm. The calculated thickness of the HAZ at each radial distance is indicated with curly braces.

3.3.2 Weld Parameters and HAZ Thickness

The width of NFA and CostE HAZ were compared to input, measured, and calculated process variables. Total weld energy explained the majority of variance of the HAZ thickness with $R^2 = 0.79$ and $R^2 = 0.91$ for NFA and CostE respectively, and is shown in Figure 3.7. Adding feed as a describing parameter was statistically significant and improved the fit to $R^2 = 0.98$ and $R^2 =$
Figure 3.6: Calculated thickness of the NFA HAZ as a function of weld radius for the three repeat welds. Symbols show the center of tensile samples on each weld. Open symbols indicate tensile failure occurring in NFA side of weld, filled symbols indicate tensile failure in CostE side of weld, and "x" symbols indicate brittle failure near weld interface.

0.98 respectively, with model parameters shown in Table 3.4. Time and maximum temperature modeled the HAZ thickness about as well as energy and feed, with $R^2 = 0.97$ and $R^2 = 0.99$ for the NFA and CostE HAZ. These models demonstrate the underlying physics: energy drives temperature, temperature causes the HAZ, and extruding hot material from the interface results in a smaller remaining HAZ.

Figure 3.7: Plot of average HAZ thickness vs weld energy. As shown in Table 3.4, adding feed significantly improves fit.
Table 3.4: Parameter and p-value Estimates for Models Describing the NFA and CostE HAZ Thickness

<table>
<thead>
<tr>
<th></th>
<th>NFA - $R^2 = 0.98$</th>
<th></th>
<th>CostE - $R^2 = 0.98$</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Intercept $mm$</td>
<td>Energy $mm/kJ$</td>
<td>Feed $unitless$</td>
<td>Intercept $mm$</td>
</tr>
<tr>
<td>value</td>
<td>0.069</td>
<td>0.030</td>
<td>-0.077</td>
<td>0.401</td>
</tr>
<tr>
<td>p-val</td>
<td>0.22</td>
<td>0.0003</td>
<td>0.0047</td>
<td>0.083</td>
</tr>
</tbody>
</table>

3.3.3 Tensile Results

Ultimate tensile strengths (UTS) of the three repeat welds are shown in Figure 3.8. The 0.5 and 2 mm/s welds had approximately uniform strengths, with the 2 mm/s weld strengths approaching that of CostE basemetal. The 1 mm/s weld exhibited a much greater spread in strength values than the 0.5 and 2 mm/s welds, with standard deviations of 88, 16.2, and 11.4 MPa respectively. Notably, weld strength heavily correlated with weld radius for the 1 mm/s weld but not for the other two welds.

Several specimens failed within the elastic range. Some of these specimens were taken from the outer diameter of the weld, and the premature failure was caused by cracks from the flash separating as material extruded from the weld. These specimens are excluded from all analysis, as failure was due to by sample geometry and not internal welding conditions. Additionally, four samples from the center of the 2 mm/s weld also failed before yielding, with one having near zero strength. These samples are discussed further in Section 3.4.4.

Figure 3.9 shows typical fracture surfaces occurring in the NFA HAZ and CostE portions of the weld. NFA fractures occur very close to the weld interface. These specimens exhibit no deformation on the CostE side of the joint, with necking between the CostE and failure location occurring over a very short distance. The NFA side of the failure exhibits some necking. In contrast, failures in CostE experienced extensive necking before failure. The necking appears to have a definitive start and stop as the tempered CostE transitions to quenched CostE, whereas the transition is much more smooth in the un-tempered basemetal.
3.4 Discussion

3.4.1 Tensile Strength

Weld strength is affected by process and intermediary parameters. Figure 3.8 shows that the higher feedrate welds had higher strengths. Notably, the 1 mm/s weld strength exhibited con-
siderable spread compared to the other two welds. This is because tensile specimens were taken at different radii as shown in Figure 3.3, and weld properties are not constant across the weld radius as Figures 5.10-3.6 show. When the strength is plotted against NFA HAZ thickness as shown in Figure 3.10, the spread in the 1 mm/s weld is not random, but strongly correlates with the HAZ thickness. Above 350 \(\mu m\), strength increased as HAZ thickness decreased. However, for HAZ thicknesses less than about 350 \(\mu m\), the strength was constant at 1050±25 MPa, which is just slightly below the CostE basemetal strength. The change from increasing strength to constant strength correlates to a change of failure mode.

![Figure 3.10: UTS vs NFA HAZ thickness. Strength increases with decreasing NFA HAZ thickness, then saturates slightly below CostE basemetal strength. Open symbols indicate tensile failure occurring in NFA side of weld, filled symbols indicate tensile failure in CostE side of weld, and \(\times\) symbols indicate brittle failure near weld interface.](image)

The failure mode and location correlates strongly with NFA HAZ thickness. Figure 3.11 shows specimen failure location plotted against the NFA HAZ thickness. Tensile specimens with a NFA HAZ thicknesses greater than about 350 \(\mu m\) failed near or in the NFA HAZ. When failure occurred in the NFA HAZ, the HAZ thickness had a direct effect on its strength. Fully joined specimens with NFA HAZ thicknesses less than about 350 \(\mu m\) failed in the tempered CostE portion of the weld. CostE failures had uniform strength, failing slightly below basemetal strength. Some specimens failed prematurely (marked with a “\(\times\)” symbol in Figure 3.11), and these are discussed below in Section 3.4.4.
Figure 3.11: Location of failure of tensile samples. Open symbols indicate tensile failure occurring in NFA side of weld, filled symbols indicate tensile failure in CostE side of weld, and × symbols indicate brittle failure near weld interface. Dotted lines were added to show the approximate functional relationship between positions of NFA and CostE HAZ locations in the 2.0 mm/s weld. Samples with thinner NFA HAZ regions failed in the tempered CostE region, whereas a thicker NFA HAZ led to failure in or adjacent to the NFA HAZ.

At first glance, there is an apparent discrepancy between HAZ strength and tensile strength. The softest part of the 0.5 mm/s weld’s NFA HAZ is about 175 HV (600 MPa), and the 2 mm/s weld’s NFA HAZ minimum hardness ranges from about 225 to 300 HV (750 to 940 MPa). CostE basemetal hardness is about 350 HV (1100 MPa), and the tempered CostE HAZ is about 310 HV (1000 MPa). Given this, it might be expected that failure would always occur in the NFA HAZ below 1000 MPa. However, the tensile data shows failures occurring above 1000 MPa and in the CostE section of the weld. The explanation for these discrepancies is “contact strengthening”, or strengthening by plane strain, and has been observed by others in traditional welding [23].

3.4.2 Plane and Antiplane Strain Strengthening

When material is loaded uniaxially with an applied stress of $\sigma_a$, the materials experience a strain along the loading direction of $\varepsilon_1 = E\sigma_a$, and Poisson contraction along the other two dimensions of $\varepsilon_2 = \varepsilon_3 = -\nu \varepsilon_1$. Plane strain is a condition in solid mechanics where the material is constrained in one dimension such that strain cannot develop in that direction, and so the actual developed strains are $\varepsilon_1 = E\sigma_a$, $\varepsilon_2 = 0$, and $\varepsilon_3 = -\nu \varepsilon_1$. Similarly, antiplane strain occurs when
both principal strains normal to the loading direction are zero, with the only non-zero strain in the direction of loading such that $\varepsilon_1 = E\sigma_a$ and $\varepsilon_2 = \varepsilon_3 = 0$. Due to this, smaller shear stresses develop under plane and antiplane strain conditions, and failure may occur at a higher than expected loads. This is shown well by the von Mises equivalent stress.

The von Mises equivalent stress is $\sigma_v = \sqrt{\left( (\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2 \right)/2}$. For an applied load of $\sigma_a$, principal stresses are: $\sigma_1 = \sigma_a$, $\sigma_2 = \sigma_3 = 0$ in uniaxial tensile; $\sigma_1 = \sigma_a$, $\sigma_2 = \nu\sigma_a$ and $\sigma_3 = 0$ in plane strain; and $\sigma_1 = \sigma_a$ and $\sigma_2 = \sigma_3 = \nu\sigma_a$ in antiplane strain. Accordingly, the von Mises stresses for the same applied load are $\sigma_a$, $\sigma_a\sqrt{1 - \nu + \nu^2}$, and $\sigma_a(1 - \nu)$ in uniaxial tension, plane strain, and antiplane strain respectively. For a Poisson’s ratio typical of steel of $\nu = 0.3$, the resultant von Mises stress is $0.89\sigma_a$ in plane strain, and $0.7\sigma_a$ in antiplane strain, requiring an applied stress of $1.13\sigma_y$ and $1.42\sigma_y$ for yielding respectively. This is the principle behind the strength of brazing. Braze filler metals are usually much weaker than the parent material, but because they are orders of magnitude thinner than wide, they are able to withstand much higher stresses than a bulk piece of the braze material could.

It is for this reason that NFA HAZ thickness strongly affected tensile strength. Very thin NFA HAZ regions were loaded such that the antiplane strain assumption is reasonably accurate. This strengthening effect was high enough that it exceeded the strength of the tempered and basemetal CostE, and thus failure occurred in the CostE. As the NFA HAZ became thicker relative to the tensile geometry, the material was less constrained by the neighboring quench-hardened CostE HAZ or NFA basemetal, and failure occurred in the NFA HAZ. This is the reason that necking around the NFA failures was far less than necking near CostE failures.

The experiments and antiplane strain strengthening effect reveal fundamental limits of weld strength. The upper strength limit of a weld is the strength of the weaker basemetal of the two halves. The lower limit of weld strength is the softest portion of either basemetal or HAZ.

Weld strength can approach the basemetal strength if the softened HAZ is thin enough that it is loaded in an antiplane strain condition, but as HAZ thickness increases, weld strength shifts towards the lower limit. For these experiments, the theoretical strength upper and lower limits were about 1100 (CostE basemetal) and 600 MPa (NFA HAZ). Observed strengths varied from about 1050 MPa to 750 MPa depending on the HAZ thickness, as Figure 3.10 shows, and are in line with the upper and lower limits.
3.4.3 Heat Transfer and Specimen Geometry

Heat generation and transfer explains many of the differences between and within the welds. The center of a weld has no surface velocity and generates minimal or no heat. In a longer weld there is ample time for heat to diffuse from the outside to the center of the weld. Therefore, a relatively uniform HAZ occurred in the 0.5 mm/s (10 s) weld. Diffusion is limited in a shorter weld, and thus the 2.0 mm/s (2.5 s) weld had a nonuniform HAZ and intermittent bonding in the center. This is evidenced in Figure 3.6 by the NFA HAZ size.

HAZ size and heat transfer theory leads to an important observation about welding geometries. In tubular weldments, heat is generated fairly uniformly across the welding surface, and this leads to relatively a homogeneous HAZ and bonding. In solid weldments, heat generation is non-uniform and thus either 1) bonding and HAZ are heterogeneous, or 2) welding time must be increased to form a uniform weld, but at the expense of HAZ and tensile properties.

3.4.4 Energy, Bonding, and Weld Strength

This study highlights the relationship between energy and weld strength. Sufficient weld energy is necessary to create a bond, but too much energy degrades the weld strength. As a consequence, neither low nor high energies result in optimal weld properties.

High welding energy results in a repeatable but sub-optimal weld. Energy drives temperature, which is necessary for diffusion and new bonding. As a consequence, high-energy welds typically have consistent bonding. However, weld energy also directly influences HAZ thickness (Figure 3.7), and HAZ thickness can have an inverse effect on the weld strength (Figure 3.10) with failure occurring in the HAZ (Figure 3.11). Therefore, even though welds with high energies have full bonds, the strength is reduced due to their HAZ.

In contrast, very low welding energy can have negative effects on weld strength and consistency. The weld with the lowest energy and thinnest HAZ also had several abnormal failures, indicated with an "x" in Figures 3.10 and 3.11. An investigation of these failures determined that 1) all specimens experienced pre-yield brittle failure, 2) all of these welds failed at or immediately adjacent to the weld interface, and not in the NFA HAZ or CostE basemetal, and 3) all of these
failures occurred near the center of 2 mm/s weld. Even though the basemetal was not weakened by the HAZ, having a small HAZ does not matter if a consistent bond cannot be formed.

Moderate weld energies form the highest quality welds. Moderate energies ensure that a bond is fully formed. However, by using only as much energy as is necessary, HAZ growth and HAZ strength loss is minimized. This is shown schematically in Figure 3.12. An ideal friction weld uses welding parameters (i.e., energy) and/or control schemes to provide a consistent bond with minimal HAZ strength drop.

![Figure 3.12: Schematic showing how weld quality (such as strength) often relates to an input parameter (such as energy). Shortly after a bond is formed, quality is at a maximum, but decreases as further work occurs.](image)

For the studied alloys and geometries, peak weld strength and reliability occurred when the NFA HAZ thickness was 300-400 μm. This condition occurred in the weld performed at 1000 rpm, feedrate of 2 mm/s, and a feed of 5 mm at a radius of about 7 mm.

### 3.5 Conclusion

Successful welds of the alloys NFA and CostE were created over multiple speeds, feedrates, and upsets. Weld energy and upset determined the HAZ thickness in both NFA and CostE. A thicker NFA HAZ reduced the tensile strength of the welded joint. Below about 350 μm, the NFA HAZ was effectively loaded in an antiplane strain state, which resulted in tensile properties nearly
equal to the unwelded CostE alloy. Low local weld energy and HAZ thicknesses were associated with an unreliable joint.

An ideal weld has sufficient energy to guarantee a bond every time, but not significantly more because the HAZ increases and tensile strengths decrease thereafter. Future work can explore this concept more, especially for cases where the space to create optimal conditions is very narrow, and thus no bond and weak bond conditions are very close to each other. Future work could also explore the effects of residual stress and post-weld heat treatments on properties in addition to the as-welded state that this research investigates.

3.6 Acknowledgements

Material and funding for this research was provided by General Electric Global Research.
CHAPTER 4. ACHIEVING FULLY CONSOLIDATED HIGH-STRENGTH FRICTION WELDS IN TRADITIONAL VS ADVANCED SUPERALLOYS

4.1 Introduction

Rotary friction welding (RFW or FW), is a solid-state joining process for axisymmetric components. FW results in high strength joints with a smaller HAZ than traditional fusion welding [3, 4]. In FW, both parts are brought into contact with a high axial force while one part is rotating. The rotation under force creates high frictional heating which softens the material. Once the material is soft enough the interfaces deform, and the heat and pressure create an extremely strong joint. FW is a well established and mature technology [14].

Friction welding has traditionally utilized large upsets during the welding process. Large upsets are thought to help clean surface contaminants which can degrade weld quality [6, 9]. Weld upset generates flash curl extruding from the interface. The color and shape of the flash correlates to internal welding conditions such as temperature, and thus the visual appearance of the flash is often used as a visual indicator of weld quality. Due to these factors, and the fact that most metals have the necessary extended ductility at welding temperatures, large upsets have become the norm when producing repeatable high-quality friction welds in traditional superalloys such as 718.

Forming repeatable fully consolidated welds in many advanced Ni-based superalloys using conventional welding techniques has been extremely difficult. These alloys have been specifically designed not to creep or undergo plastic flow even at high temperatures. The metallurgical design choices mean that many of these alloys have limited ductility even at high temperatures. Cracking, void formation, and bulk failure can occur when these alloys are hot deformed. When these alloys are welded using traditionally techniques, the extreme deformation at the weld interface can lead to intermittent interface failure and/or blow-out.

This paper shows how welding process can be altered to change material deformation, thus resulting in more consolidated friction welds in advanced superalloy alloys. First, the necessary
and sufficient requirements for welding and bonding are reviewed. Comparisons to current welding techniques are made, and contrasts are drawn especially in light of advanced limited-ductility superalloys. Second, changes in welding methodology are shown, such as limiting feed and changing welding force and rotational speed. Third, ductility-enhancing changes to the weld geometry are shown which increase the reliability and strength of welds in advanced superalloys. These principles produce great welds in high ductility alloys, and are essential for repeatedly producing good welds in low-ductility advanced superalloys. The principles and techniques set forth in this paper are separate but synergistic, thus allowing the application of some or all of these techniques depending on the situation.

4.2 Bonding and Joining of Metals

At a fundamental level, creating new metallurgical bonds is simple: two compatible metal atoms will spontaneously create a metallic bond if they are close enough to each other. In practice, this can be difficult. All metal surfaces are imperfect, and factors such as oxides, greases, contaminants, and surface asperities prevent two pieces from welding on contact. Limiting these factors makes creating quality welds easier [6,9], but it is impossible to completely eliminate these obstacles.

Solid state joining processes create strong metallurgical joints by overcoming two primary types of obstacles. First, asperities and surface roughness must be eliminated in order to have complete contact between two workpieces. This is typically done through deformation and pressure. Second, oxides and contaminants must be removed or broken up. This is usually accomplished through diffusion and deformation.

Solid state joining processes overcome these obstacles to bonding in different ways. Diffusion welding induces no bulk deformation, instead using pressure and time at elevated temperatures to bond carefully prepared surfaces [24]. In contrast, explosion bonding needs virtually no time or externally elevated temperature, and instead utilizes extreme pressures to cause a moving explosion wavefront and deformation [25,26]. In between these two extremes lie technologies such as rotary friction weld, friction stir welding, roll bonding, and forging. All solid state joining processes use some combination of time, temperature, pressure, and bulk/surface deformation in order to achieve bonding [26].
4.3 Experimental Methods and Materials

4.3.1 Materials

Two main alloys were used for these experiments. Precipitation hardened alloy 718 is a commonly used superalloy used for rotating disk applications that require high strength at temperatures not exceeding 600 °C. Alloy 718 is known for its relatively good weldability and wide thermomechanical processing window and slow strengthening phase precipitation kinetics. Alloy A is a Ni-based alloy designed for rotating components that operate at higher temperatures. Alloy A has excellent high temperature properties but as a result has a narrow thermomechanical processing window and quick precipitation kinetics that pose a challenge to conventional joining methods.

All samples were machined into a cylinder with a 25.4 mm (1.00 in) outer diameter. Tubular samples were made with an inner diameter of 20.3 mm (0.80 in) about 25 mm (1 in) deep. Samples for quasi-hydrostatic conditions were created by making female pieces with flat bottom holes of 12.75 mm (0.502 in) inner diameter and 6.35 mm (0.25 in) deep, and male pieces with solid cylindrical tips 12.57 mm (0.495 in) wide and at least 10 mm (0.4 in) long. The diameter tolerance was chosen to prevent sidewall rubbing in the presence of thermal expansion. A schematic of the weld samples is shown in Figure 4.1.

4.3.2 Welding Equipment and Procedures

Welds were performed with a TTI High Stiffness RM2 FSW machine with 10 ton spindle and direct drive motor, with a Bond Technologies B&R-based high-speed controller and data acquisition system. Time series information that was collected included: x, y, and z forces; z position and velocity; spindle and motor torque, speed, and power; and specimen torque of the stationary sample [21].

The axial stiffness of the machine is approximately 80 kN/mm. In this paper, the term “feed” is used to denote total crosshead movement of the machine, and “upset” is used to denote shortening of the welding samples. Typical welding loads can result in feeds about 1 mm larger than the sample upset. This is a non-trivial difference in low-upset welding (< 1 mm upset), and axial force data was used to compensate the feed data to calculate samples’ upset.
Sample interfaces were cleaned with acetone before welds to remove any residual oils and grease remaining from machining and handling. Samples were welded under constant spindle speed conditions. Welds were performed at either a constant axial feedrate or force. Feed/force were maintained during and after spindle stop to ensure pressure was always present in the weld.

Tensile tests were performed on as-welded samples in an MTS tensile tester with 105 klbf (467 kN) capacity. For the male-female quasi-hydrostatic welds, the thick-wall of the female sample was machined away prior to tensile testing. This is because even slight welding of the side walls can significantly increase bonding area, leading to an artificially high tensile strength. Machine displacement rate was 0.005 in/s (0.13 mm/s), and load-displacement data were acquired at 50 Hz. All stresses are calculated with respect to original cross section of the weld specimens.

4.3.3 Calculations and Plotting

All force and speed data for alloy A is analyzed and plotted in normalized form. Normalized strength is calculated as the failure load of the weld divided by the original cross section, normalized by basemetal ultimate tensile strength (UTS). Similarly, normalized welding force is calculated by dividing the axial force by the original cross section, and dividing that by the basemetal UTS. For alloy A, normalized spindle speed and feed are always calculated by dividing
the speed or feed of a weld by the maximum speed of the welds. Normalized average surface speed is calculated in several steps. First, the average radius of the specimen is calculated, which is $2/3r$ for a circle and approximately $(r_1 + r_2)/2$ for a thin-walled tube. Second, the spindle speed is divided by this area to find the average surface speed. Lastly, for alloy A this is normalized by the maximum of the average surface speeds for that weld series.

For all plots, the ○ symbol is used for tensile strength of a tube weld, whereas a ● denotes a quasi-hydrostatic containment weld geometry. A * through a plotting symbol indicates that blowout of the weld interface was clearly visible, whereas a × through the plotting symbol indicates lack of bonding is visible at the fracture surface.

### 4.4 Minimizing Weld Upset

Once a bond has occurred, further weld action is unnecessary. Metallurgical bonding occurs when surfaces are free from asperities and contaminants, and the atoms are in intimate contact. Any extra weld upset, energy, temperature, time, etc is unnecessary. Furthermore, longer welds can lead to sub-optimal final weld properties (cite NFA paper).

#### 4.4.1 Traditional Superalloys

The initiation and completion of weld bonding was studied by creating many similar welds at different total feeds. Alloy 718 was welded at a constant spindle speed of 800 rpm and a feedrate of 2 mm/s, and with total feeds of 0.9, 1.1, 1.3 1.5, 2, 3, 5, and 10 mm. The force-displacement curve of the 10 mm weld is shown in Figure 4.2, where the welding force is normalized against the basemetal tensile strength. Normalized strength is superimposed on the plot corresponding to when each weld stopped.

During the first 1 mm of feed, axial elastic loading occurs while the weld interface heats up. Welds that finished in the elastic loading portion of the weld had subpar strength due to incomplete bonding.

The welds with the highest ultimate strengths (>95% of published UTS) occurred immediately after the weld started to yield, which is indicated by the force plateauing. Flash production started at this point, which indicates that the interface was beginning to flow, and thus intimate
Figure 4.2: Weld force and tensile strength vs. feed of welds in alloy 718. The welding force versus position profile of the 10 mm weld is shown for context of the typical force-feed history. Normalized strength of the different welds are superimposed on the plot with diamond markers, with × markers denoting a lack of full bonding visible at the fracture surface.

contact could be achieved. Also at this point sufficient diffusion and mechanical scouring of the surface and occurred to break up existing oxides and contaminants. With the surface free of contaminants and the surface in intimate contact, the fundamental conditions for bonding are met, and a strong weld occurs. As further axial feed occurred, more flash extruded and the welds approached a thermal and force equilibrium. The UTS decreased slightly, but quickly approached a constant value of 87% of basemetal UTS.

A dissimilar alloy weld was performed using alloys 718 and 625, at a spindle speed of 1000 rpm, a feed rate of 1 mm/s, and a total feed of 5 mm. The weld was cross-sectioned, polished, and etched with Kalling’s Waterless Reagent, which etches alloy 718 but not alloy 625. The micrograph is shown in Figure 4.3. Notably, alloy 718 can be seen attached to the very tip of alloy 625. The bi-metal joint was strong enough that separation occurred on one side of the joint rather than in the joint itself. This indicates that a quality bond was present in the beginning of the weld just as bulk deformation started to occur.

These experiments show that a complete joint can be formed once a small amount of upset occurs. Further welding is unnecessary and may gradually degrade weld strength. An incomplete bond due to insufficient welding results in a dramatic loss of strength. Consequently, it is prudent to err on the side of too much welding rather than too little welding for many traditional superalloys,
as is later shown schematically in 4.5. However, welding with more upset than is needed is not always viable in advanced superalloys.

4.4.2 Advanced Superalloys

The same basic experimental procedures were performed for alloy A, but at a different spindle speed, feedrate, and total feeds in order to compensate for the enhanced high-temperature properties of alloy A versus alloy 718. Welds were again pulled in tension until failure. A plot showing the normalized tensile strength superimposed over the highest feed weld’s force-feed curve is shown in Figure 4.4, with the feed for all welds normalized against the total feed of the highest-feed weld.

Similarly to alloy 718, a period of elastic loading occurred before bulk deformation. Unlike alloy 718, alloy A reached a higher normalized welding force before yielding, at which point the force suddenly dropped due to localized catastrophic failure and material ejection. This loading and ejection cycle continued, with the highest feed weld experiencing a less dramatic and almost continuous shedding of material by the time the weld ended.
Figure 4.4: Weld force and tensile strength vs normalized feed in alloy A. Weld UTS of the different welds are superimposed on the plot with diamond markers, which is superimposed on the welding force versus position profile of the highest feed weld.

The normalized strength of all alloy A welds is lower than alloy 718 welds. Instead of dropping from >95% basemetal UTS to 87%, alloy A drops from about 50% of basemetal UTS down to an abysmal 2%. This large decrease in strength is due to the limited ductility of alloy A. During welding, sample upset and flash generation cause high strains which the material is unable to accommodate. As noted above, once yielding occurs, catastrophic failure and ejection of material at the weld interface follows. This loss of material results in the sudden drops of force shown in Figure 4.4.

4.4.3 Differences between Traditional and Advanced Superalloys

The relationship between the total amount of welding upset and weld quality is shown schematically in 4.5. In the beginning of a weld, no bond is present. As additional force is applied and the material begins to heat, intermittent bonding occurs. When bulk upset starts to occur, the weld is usually complete. In traditional high-ductility superalloys, additional welding can increase weld consolidation reliability by sacrificing some strength. In low-ductility alloy A, additional welding can cause intermittent interface failure and blowout, which drastically reduces the weld quality. Thus in alloy A, additional welding not only doesn’t improve weld quality, but rather it leads directly to blowout and weld defects.
Figure 4.5: Schematic of weld quality versus feed. Welds transition rapidly from no bonding to a fully bonded interface. In high ductility alloys, strength drops off slowly as further welding occurs. In low ductility alloys, strength drops off quickly after the initial bond is formed.

4.5 Changing Speed and Force under Limited Upset Conditions

Weld parameters can have a significant effect on the quality of welds. Two of the most common control parameters are axial force and spindle speed. Spindle speed determines total weld power in inertial friction welding, and affects the heat generation rate in direct-drive friction welds. This affects the deformation temperature and flow of material.

Increased axial force causes deformation at lower temperatures. The flow stress of metals is temperature dependent, with higher strengths occurring at lower temperatures. In order to achieve force equilibrium in a weld, low temperatures and high upset rates occur in response to high axial forces. In addition to lower welding temperatures, high axial force can mitigate void formation and lack of bonding at the weld interface.

A two-factor three-level DOE with a triple repeat of the center point was performed in tubular samples of alloy A. Total weld upset was kept constant between welds, and was similar to the best welds in Figure 4.4. The normalized DOE input parameters and results are shown in Table 4.1. The data points are later shown superimposed over a fit surface in Figure 4.6.

Statistical analysis was performed on the results in Table 4.1. Fitting the data to a full second-order model resulted in four statistically significant non-intercept terms, as shown in Table 4.2, with force and speed represented as “F” and “S”. All possible second-order models
Table 4.1: Normalized Parameters and Results for DOE of Tubular Alloy A

<table>
<thead>
<tr>
<th>Speed</th>
<th>Force</th>
<th>Strength</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.07</td>
<td>0.14</td>
<td>0.03</td>
</tr>
<tr>
<td>0.37</td>
<td>0.14</td>
<td>0.16</td>
</tr>
<tr>
<td>0.07</td>
<td>0.36</td>
<td>0.50</td>
</tr>
<tr>
<td>0.37</td>
<td>0.36</td>
<td>0.51</td>
</tr>
<tr>
<td>0.22</td>
<td>0.14</td>
<td>0.14</td>
</tr>
<tr>
<td>0.07</td>
<td>0.25</td>
<td>0.45</td>
</tr>
<tr>
<td>0.22</td>
<td>0.36</td>
<td>0.65</td>
</tr>
<tr>
<td>0.37</td>
<td>0.25</td>
<td>0.57</td>
</tr>
<tr>
<td>0.22</td>
<td>0.25</td>
<td>0.53</td>
</tr>
<tr>
<td>0.22</td>
<td>0.25</td>
<td>0.59</td>
</tr>
<tr>
<td>0.22</td>
<td>0.25</td>
<td>0.57</td>
</tr>
</tbody>
</table>

were screened to determine which terms were not only statistically significant, but also notably improved fit. The best model of a given total order is given in Table 4.3, along with the adjusted $R^2$ value and the AICc fit parameters. Given the fit parameters, a model with 2 to 4 terms would be appropriate to model the process.

Table 4.2: Model Terms and Corresponding p-values for DOE

<table>
<thead>
<tr>
<th>Term</th>
<th>p-value</th>
</tr>
</thead>
<tbody>
<tr>
<td>F</td>
<td>0.00004</td>
</tr>
<tr>
<td>$F^2$</td>
<td>0.00057</td>
</tr>
<tr>
<td>S</td>
<td>0.041</td>
</tr>
<tr>
<td>$S^2$</td>
<td>0.028</td>
</tr>
<tr>
<td>F*S</td>
<td>0.23</td>
</tr>
</tbody>
</table>

The DOE input and output variables from Table 4.1 are plotted in Figure 4.6, with the best third-order model of Table 4.3 plotted as a mesh surface. Spindle speed had a positive yet minor affect on weld strength. In contrast, welding force had a large and non-linear affect on weld strength. At the lowest forces, weld strength was very low averaging about 11% of base metal strength. Increasing force led to large increases in strength, with smaller strength increases occurring as force increased further. The fracture surface of the medium and high force welds indicated a full or nearly full weld, whereas at the lowest force interface blowout and lack of bonding was prevalent.
Table 4.3: Model Terms and Statistical Model Quality Criteria

<table>
<thead>
<tr>
<th># of Terms</th>
<th>Terms</th>
<th>Adj R²</th>
<th>AICc</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>F</td>
<td>0.612</td>
<td>-5.92</td>
</tr>
<tr>
<td>2</td>
<td>F, F²</td>
<td>0.899</td>
<td>-16.82</td>
</tr>
<tr>
<td>3</td>
<td>F, F², S</td>
<td>0.921</td>
<td>-13.65</td>
</tr>
<tr>
<td>4</td>
<td>F, F², S, S²</td>
<td>0.961</td>
<td>-12.05</td>
</tr>
<tr>
<td>5</td>
<td>F, F², S, S², F*S</td>
<td>0.966</td>
<td>2.73</td>
</tr>
</tbody>
</table>

Figure 4.6: Plot of weld strength versus spindle speed and welding force, showing the large non-linear effects of force on weld strength and consolidation. Individual data points from Table 4.1 are plotted with a ◦. Welds which had lack of bonding or material blowout are marked with an × and * symbol respectively. The predicted strength from the 3rd order model (terms of F, F², and S) is plotted as a mesh surface.

4.6 Quasi-Hydrostatic Pressure

4.6.1 Effects of Hydrostatic Pressure on Material Failure

The von Mises yield criterion is useful in predict yielding under complex stress states. One limitation is that it predicts no changes in failure under hydrostatic tension or compression. This is not quite accurate. Under sufficient hydrostatic tension, void cavitation occurs even without a deviatoric stress. The opposite effect occurs with hydrostatic compressive pressure.

The work of Bridgmann [27,28] shows that ductility can be greatly enhanced by hydrostatic pressure, with different materials showing different sensitives to hydrostatic pressure. Greater
hydrostatic pressure resulted in fewer voids in the center of necked tensile samples and mildly improved fracture stress. Most importantly, hydrostatic stress leads to an increase of fracture strain: over four times the failure strain versus non-hydrostatic tension in many of the cases Bridgman presents.

4.6.2 Quasi-Hydrostatic Conditions via Containment Geometry

A quasi-hydrostatic condition at the weld interface can assist in welding of advanced alloys that have low ductility. In FW, localized yielding at the weld surface is not only acceptable but also necessary to create intimate contact. However, bulk upset and yielding can often lead to catastrophic failure in low ductility alloys. Consequently hydrostatic pressure can reduce interface failure in advanced alloys by extending plasticity and minimizing voids at the weld interface. This then allows sufficient temperature, pressure, and local deformation to form a joint without material blowout.

A quasi-hydrostatic condition can be achieved via a containment geometry of the weldment instead of a pressurized fluid. This has the advantage of being much simpler to physically implement, and also keeps fluid away from the weld interface, which could interfere with bonding. As described in Section 5.4, a solid cylindrical male piece is machined to a slightly smaller diameter than a thick-walled female piece. As the interface heats up, material extrusion is mostly contained by the thick wall tube. This creates back pressure at the weld interface and a net quasi-hydrostatic pressure. The weld can be completed at higher temperatures and/or pressures, both of which aid bonding but can lead to catastrophic failure in constrained friction welding of advanced superalloys.

4.6.3 Containment Welds

Welds were performed with male and thick-walled female specimens, as described in Section 5.4. Due to a limited amount of material and samples, welds were performed at only the highest and lowest normalized speed and normalized force levels of the tubular DOE, resulting in a two-factor two-level DOE.
As shown in Figure 4.1, the containment walls of the female specimens were machined away before tensile testing to eliminate any side-wall bonding which would otherwise artificially increase the apparent weld strength. Tensile tests were performed on the welds, and the normalized strengths and input parameters are shown in Table 4.4. These values are plotted in comparison to the strength of the tubular welds in Figure 4.7.

Table 4.4: Normalized Parameters and Results for DOE of Containment-Geometry Alloy A Welds

<table>
<thead>
<tr>
<th>Speed</th>
<th>Force</th>
<th>Strength</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.07</td>
<td>0.14</td>
<td>0.72</td>
</tr>
<tr>
<td>0.37</td>
<td>0.14</td>
<td>0.80</td>
</tr>
<tr>
<td>0.07</td>
<td>0.36</td>
<td>0.57</td>
</tr>
<tr>
<td>0.37</td>
<td>0.36</td>
<td>0.76</td>
</tr>
</tbody>
</table>

Figure 4.7: Comparison of the strength of tubular welds (diamond markers and lower surface) to the strength of quasi-hydrostatic welds (circle markers and upper surface).

All quasi-hydrostatic containment welds were stronger than comparable tubular geometry welds. A one-sided t-test confirms that the quasi-hydrostatic welds are stronger than the tubular welds with a p-value of 0.0054. Even under conditions that resulted in low-strength defect-filled tubular welds, quasi-hydrostatic welds contained no voids and were multiple times stronger. The
strength of containment welds increased with axial force and decreased with spindle speed. The interaction term seems to be potentially significant, but this cannot be verified with a small sample size.

### 4.7 Discussion

One common motivation behind creating upset in friction welds is to remove contaminants from the interface. As Figure 4.2 shows, large upsets are not required for a good weld. Certainly a heavily oxidized or very dirty surface may benefit from extra cleaning, but if a welding surface is properly cleaned and prepared, then large upsets are unnecessary.

In traditional superalloys, further welding leads to a small decrease in strength, but ensures a full bond at the interface. For many industries reliability is more important than small performance improvements, and thus welding with more upset than is necessary has become the norm. In contrast, additional welding of advanced superalloys leads to deformations exceeding the hot plasticity limit, leading to catastrophic localized failure and material ejection. Increased welding of advanced alloys thus lowers the bond reliability instead of increasing it. The best welds are formed at minimal upsets, right after an initial bond has formed.

For superalloy A, increasing welding force led to stronger welds. It is believed that this is due to higher forces creating lower peak temperatures and higher interface pressures. Spindle speed had a smaller effect than did axial force, which is likely unique to continuous-drive friction welding. In inertial friction welding, spindle speed directly affects total weld energy and upset, which has a large effect on weld quality and strength.

### 4.8 Conclusions

Tradition holds that more upset creates better friction welds. In traditional alloys, while excessive upset is not necessary for bonding, it can increase reliability at the cost of minor strength decreases. In advanced superalloys, increasing upset decreases both reliability and strength. For both alloys, the best welds occurred right as bulk upset started. This study refutes the idea that performing a longer, hotter weld with lots of upset results in a better friction weld - in advanced superalloys these result in a worse weld.
Weld parameters other than upset also have an impact on weld strength and quality. Specifically, increasing welding force – while keeping upset to a minimum – increased weld strength. Under constant upset conditions, spindle speed did not have a dramatic effect on strength.

A containment geometry which adds a quasi-hydrostatic force can aid the welding of low ductility superalloys. With the containment geometry, once a small amount of bulk deformation has occurred, that displaced material fills any voids and cannot easily escape from the interface. Internal pressures increase beyond what could normally occur in the deformation of a hot metal, extending ductility and forcing the entire interface into intimate contact. Containment welds were able to consistently achieve high strengths, and were fully bonded without material blowout.

Advanced superalloys are undoubtedly more difficult to weld than traditional superalloys like alloy 718. They can, however, be welded using FW. Increasing welding force and decreasing weld upset both led to stronger welds with fewer volumetric defects. Novel weld geometries can be used to induce a quasi-hydrostatic material condition, which can eliminate blowout and volumetric defects. Increasing force, decreasing upset, and using a hydrostatic-inducing geometry are all synergistic with one other. In traditional superalloy, welds these parameters and conditions are useful. In advanced superalloys, they are essential for creating strong and reliable defect-free welds.
5.1 Introduction

5.1.1 Rotary Friction Welding

Rotary friction welding (RFW or FW) is a thermo-mechanical solid-state joining process for axisymmetric metallic components [14]. In FW, work pieces are brought together while rotating under a high axial force, creating frictional heating. The combination of heat and pressure leads to metallurgical bonding after the weld has finished.

Friction welding is an advantageous welding processes due largely to its thermal cycle. Peak temperatures in FW, often in the hot-working range [29], are lower than fusion welding. This eliminates solidification and shrinkage defects which may occur in fusion welding, and can lessen the severity and extent of the heat affected zone (HAZ) [3, 4]. Process times in FW are also typically lower, resulting in lower energy expenditure and greater process throughput [10, 11].

While a lower thermal cycle in FW is primarily advantageous, there are potential drawbacks. Due to low energies and rapid temperature changes the thermal gradients can be steep, resulting in high cooling rates [30]. While solid solution strengthened alloys are largely unaffected by a quick cooling rate, quench and precipitation strengthened alloys may have suboptimal microstructures. In another solid-state joining process, friction stir welding, cooling rate has been shown to have a far greater impact on microstructure than any other input variable [31]. High cooling rates can lead to brittle microstructures like martensite in most steels.

5.1.2 Thermal Control

A specific microstructure can be created by modifying the cooling rate. The cooling rate is determined by the temperature profile at weld completion, and thus cannot be directly controlled.
Preheating a weld can build up heat in the weldment. The cooling rate can thus be indirectly controlled as an open-loop system by controlling the thermal profile. High interface temperatures are, however, a common consequence of weld preheating. This can result in grain growth, overly large HAZ regions, and extra shrinkage during cooling. These effects can be mitigated by limiting or controlling the interface temperature during a welding preheat.

Closed-loop temperature control is common in friction stir welding (FSW), a related thermo-mechanical joining process. Factors such as temperature measurement location and controller tuning affect temperature control [32]. The large body of temperature control research in FSW shows that temperature can be reliably controlled with a variety of algorithms and methods (PID, MPC, SISO, MIMO, etc) [33–36], using different inputs as the manipulated variable(s) (spindle speed, spindle power via torque, axial force) [37–39], with different controller tuning methods (step test, relay feedback, simulation, first-principles modeling) [32, 35, 40, 41], and with different controller parameters/gains [40, 42].

5.1.3 Overview

The present work introduces cooling rate and temperature control in FW with the goal of controlling process temperatures and thermal profiles in order to achieve specific cooling rates, and thus desired microstructure and properties. Heat transfer analysis is used to determine approximate upper and lower bounds on cooling rates. Controller operation, tuning, and setpoints are presented and discussed. The impacts of controlled and uncontrolled welding is shown in both simulation and welding. Microstructural analysis corroborates simulation data, showing that controlling temperature and cooling rates can result in drastically different microstructures.

5.2 Limits of Achievable Cooling Rates

Material kinetics determine the microstructure(s) that is formed during cooling. A cooling rate in excess of 1000 °C/s is needed to form martensite in a low-carbon steel, while a cooling rate slower than 0.006 °C/s is needed to form a fully pearlitic structure in 4340 steel. Only a limited range of cooling rates can be realistically achieved in FW. This range determines if a specific transformation can be achieved through FW temperature control.
5.2.1 Minimum Cooling Rate

In FW, the dominant cooling mechanism for the weld zone is conduction into the part being welded. The slowest realistic conductive cooling rate is achieved when a temperature gradient is fully linear, with the maximum temperature at the weld interface, and room temperature at the fixed end. In this case the thermal diffusivity and specimen gauge length determine the cooling rate.

Multiple simulations were performed while varying thermal diffusivity and preheated gauge length. The material was preheated to a linear gradient, with the cold end fixed to an infinite heat sink. Temperature was non-dimensionalized to be equal to the difference between the hot and cold end. Convective and radiative losses are ignored, and thermal diffusivity is constant. The results from these simulations are shown in Figure 5.1.

![Figure 5.1: Slowest achievable cooling rate for different ratios of thermal diffusivity to preheat gauge length.](image)

CCT diagrams often show superimposed constant cooling rate contours, but truly constant cooling rates rarely occur outside of slow furnace-controlled situations. Instead, the average cooling rate down to 50% of the initial temperature was calculated by first measuring the half-cooling time ($t_{50}$), or the time to cool to 50% of the initial temperature as shown in Figure 5.1. A constitutive relationship between thermal diffusivity ($\alpha$), preheated length ($l$), and the half-cooling time was determined to be
\[ t_{50} = 0.196 \frac{l^2}{\alpha} \] (5.1)

With an initial temperature differential between the hot and cold ends of \( \Delta T \), the average cooling rate \( (CR_{\text{min}}) \) over the range of \( \Delta T/2 \) is

\[ CR_{\text{min}} = 2.55 \Delta T \frac{\alpha}{l^2} \] (5.2)

Accordingly, a mild steel with a thermal diffusivity of about 1.2e-5 m\(^2\)/s that is preheated to 750 °C above room temperature over a length of 50 mm would take approximately 40 s to reach about 400 °C and have an expected minimum cooling rate of about 9 °C/s.

These relationships are merely estimates to check feasibility. Convection cooling and the temperature-dependence of thermal priorities are ignored, and \( CR_{\text{min}} \) is therefore merely an estimate. If \( CR_{\text{min}} \ll CR_{\text{needed}} \) temperature control is viable; if \( CR_{\text{min}} \approx CR_{\text{needed}} \) then the method may work or may only be marginally effective; if \( CR_{\text{min}} \gg CR_{\text{needed}} \) then this method is not viable. Extremely slow cooling rates and workpieces with a high surface area to volume ratio may also be nonviable due to convection and radiation becoming the dominant cooling mechanism.

### 5.2.2 Maximum Cooling Rate

The fastest possible cooling rate will occur when the interface temperature is reached as quickly and with as little thermal energy as possible. An infinite cooling rate would develop if an impulse of thermal energy were added to the welding interface of a semi-infinite solid. Such a situation is only possible in mathematical treatments of heat transfer, not when heat is generated due to surface friction as with FW.

Processes such as FW are self regulating, thus limiting the maximum heating rate. If the process could be maintained at the maximum heat rate until and only until the transformation temperature was reached, a maximum quench rate at the interface would occur. A normalized heat-flux rate \( (\bar{Q}) \) is defined as

\[ \bar{Q} = q''/(k \Delta T) \] (5.3)
where \( q'' \) is the heat flux, \( k \) is the thermal conductivity, and \( \Delta T \) is the temperature difference. Simulations were performed at different values of \( \alpha \) and \( \bar{Q} \). A precise relationship between \( \alpha \), \( \Theta \), and \( t_{50} \) was determined to be

\[
t_{50} = 0.452/(\bar{Q}^2 \alpha)
\]

from which the maximum average cooling rate to the half-temperature is then

\[
CR_{max} = 2.21 \Delta T \bar{Q}^2 \alpha
\]

Determining an appropriate maximum heat flux rate is difficult but critical to approximating the maximum cooling rate. Factors such as temperature-dependent flow stress, friction constitutive law, surface velocity and axial force all affect the maximum heat flux that can be generated at the sample interface [4]. Table 5.1 shows peak values that the authors have observed in FW for different alloys.

Table 5.1: Observed Peak Heat Generation Rates and Estimated Max Cooling Rates in Various Alloys

<table>
<thead>
<tr>
<th>Alloys</th>
<th>( q'' ) (( W/m^2 ))</th>
<th>( \bar{Q} )</th>
<th>( CR_{max} ) (( ^\circ C/s ))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Alloy 718</td>
<td>8e7</td>
<td>3e3</td>
<td>2e5</td>
</tr>
<tr>
<td>NFA</td>
<td>6e7</td>
<td>2e3</td>
<td>5e4</td>
</tr>
<tr>
<td>Plain Carbon Steel</td>
<td>4.5e7</td>
<td>1.2e3</td>
<td>3e4</td>
</tr>
</tbody>
</table>

It must be stressed that the max cooling rate calculated in Equation 5.5 is merely an estimate. In particular, this assumes a very high heating rate this is constant but brief, which may be difficult to achieve in some welding conditions. Accordingly, the max cooling rate \( CR_{max} \) should only be used as a feasibility check to to ensure that a sufficient fast cooling rate can likely be achieved. For almost all pearlitic transformations \( CR_{min} \) will be the limiting factor, whereas \( CR_{max} \) is the limiting factor if martensite is desired.
5.3 Thermal Control in Friction Welding

Controlling the thermal profile and subsequent cooling in FW to obtain a desired microstructure is comprised of two separate but interrelated aspects. A closed-loop preheat is used to raise the thermal profile, thus effectively controlling cooling rate in an open-loop manner. At the same time, temperature control is used in a closed-loop manner to control the weld interface temperature and prevent excessively high temperatures.

5.3.1 Weld Preheat

A "preheat" section in FW is traditionally a low-force and high-temperature segment of a weld. Bulk deformation occurs, and the weld interface can easily exceed 1200-1300 °C in steels and superalloys. Extremely high temperatures generally create negative effects in the resultant microstructure.

The preheat objective in this case, however, is to raise the temperature at a distance away from the weld interface, in order to prevent rapid quenching. While doing so, the weld interface should be kept at moderate temperatures. Temperature control during preheat thus relies on using low power and prevents bulk upset and deformation.

5.3.2 Controlling Weld Power

Weld temperature is controlled by interface power, which is primarily dependent upon surface velocity and axial force. A weld was performed with step changes in spindle speed and axial force, with the output and inputs shown in Figure 5.2. Both force and power had a direct effect on the weld power, which can be approximated by the equation \( P(W) = S(rpm) \times F(N)/2000 \) for the alloy and geometry described in Section 5.4.1. Minor changes in axial force are easier to achieve than changes in spindle speed (due to inertial mass of the spindle), and so axial force was chosen as the control variable for power and temperature. Given a spindle speed and desired power, the required force can be solved using the above equation. The constant in this equation are dependent on the alloy and geometry being welded.
Figure 5.2: Weld power output in response to step test changing spindle speed and axial force.

5.3.3 PID Temperature Controller

PID is a commonly used classical control scheme. FW PID temperature control requires both an interface setpoint and a trigger value. This is in contrast to PID temperature control in other thermo-mechanical process such as FSW which only references the setpoint value. The PID algorithm controls the interface temperature to a setpoint, and the trigger temperature determines when to stop the preheat segment of the weld. The control objectives are shown schematically in Figure 5.3.

The interface temperature is measured near the weld interface. This provides quick response and helps to avoid excessive surface temperatures. The PID controller adjusts the welding power in response to the measured interface temperature in a closed-loop manner.

The body temperature is measured farther away from the interface, and together with the setpoint value is used to estimate the thermal profile in the weldment. Once the body temperature reaches the trigger value, the weld is stopped. The trigger value is determined in advance by FEA simulations such that once this value is reached, a sufficient thermal profile has developed that will result in the desired cooling rate upon weld completion. Consequently this is open-loop predictive control of the cooling rate: once the weld is finished it is impossible to affect the cooling rate by adding or removing thermal energy.
A PID controller was coupled in simulation to a thermal FEA model of the material. The PID controller commanded a power, which was used as a boundary condition in the thermal model. For actual welds, the controller’s commanded power was implemented via a machine axial force setpoint. Both in simulation and actual welds, an artificial power limit of 1 kW was given, during which time the PID integrator action is held off. Without a power limit, a PID controller will attempt to reach the setpoint as quickly as possible by requesting high power and forces. Premature high power and force can accidentally result in a weld, instead of the desired preheat segment with zero bulk deformation.

In simulation, PID controller parameters were manually varied. The system showed a wide range of acceptable PID parameters. PID gains of $K_p = 10 \ W/\degree C$, $K_i = 1 \ W/(s/\degree C)$, and $K_d = 0$ resulted in good performance without excessive manipulated variable movement in simulation.

Coupled PID - FEA simulations were performed using the above PID gains, a setpoint temperature of 900 °C 1mm away from the interface, and a trigger value of 700 °C 5 mm away from the interface. The same FEA routine was also used to simulate a standard uncontrolled weld, but with a constant power of 10 kW and a trigger value of 150 °C. The cooling rates from these simulations are shown in Figure 5.4. A short uncontrolled weld is predicted to largely transform...
into martensite, whereas a temperature controlled weld should almost completely avoid martensitic transformations.

Figure 5.4: Simulation cooling curves for a quick uncontrolled and temperature controlled friction weld in mild steel, superimposed on a TTT diagram for 0.50% C steel [2].

5.4 Experimental Methods and Materials

5.4.1 Materials

AISI 1045 steel was selected as the welding alloy. Bar stock measuring 25.4 mm (1 inch) was machined into tube samples having a 19 mm (0.75 inch) inner diameter 50 mm (2 in) deep. K-type thermocouple beads were created and welded onto the sample at locations 1, 2, 3, 4, and 5 mm away from the interface.

5.4.2 Welding Procedure and Control

All welds were performed on a TTI High Stiffness RM2 FSW machine with 10 ton spindle and direct drive motor, and upgraded with a Bond Technologies B&R high-speed controller and data acquisition system. Force, position, spindle speed, specimen torque, and external thermocouple temperatures were recorded during welding and for 60 s after each weld. Duplicate
weld sets were created to allow for both tensile tests and microstructure analysis as described in Section 5.4.3.

All welds were run to a low upset condition (1-2 mm total upset) as more upset is unnecessary to achieve good bonding. Additional welding generates excess heat and degrade post-weld properties, and is more likely to detach thermocouples from the surface. Uncontrolled welds were run at a spindle speed of 1000 rpm, whereas temperature control welds were performed at 200 rpm in order to achieve the lower welding powers which are required for temperature control. The nominal feedrate during welding (excluding preheat) was set at 4 mm/s, with a force maximum set to 40 kN for both welds.

For uncontrolled welds, rotation continued through a feed of 1.5 mm at which point the spindle was given a stop command. The machine adjusted feed and position to attempt to maintain a constant 40 kN of force during spindle stop and cooling. For temperature controlled welds, the axial force was varied to match appropriate power setpoints commanded by the PID controller, with the PID controller referencing the 1 mm thermocouple for the setpoint and controlling to 900 °C. Once the 5 mm thermocouple reached 700 °C, the axial force was set to 40 kN (with a maximum of 4 mm/s) for 2 mm of feed, after which the spindle stop command was given.

5.4.3 Testing and Microstructure Analysis

Tensile testing was performed on one specimen of each welding set to confirm full weld consolidation and bonding. Specimens were tested in the as-welded condition. Testing was done on an MTS machine with 450 kN (100 klbf) capacity, with a displacement rate of 0.0002 mm/s (0.005 in/s).

The other specimen of each weld set was used for microstructural analysis. Welds were sectioned along the long axis and polished to a 0.05 μm finish. Specimens were tested in a LECO microhardness indentor, with a loading of 200 g, 100 μm spacing in the axial direction, and 5 columns spaced μm 500 apart in the radial direction. Hardness values were averaged at each axial distance after confirming a lack of radial dependence on the measurements. After indenting, imaging was performed in an FEI Aprio SEM using a backscatter detector to produce channeling contrast, with a 30 kV accelerating voltage, 1.6 nA beam current, and a nominal working distance of 5 mm.
5.5 Results

5.5.1 Tensile

The engineering stress-strain curve of both welds is shown in Figure 5.5. Both welds had a fully consolidated joint, and the stress-strain curves of both welds were nearly identical through 1.5% plastic elongation. The uncontrolled weld experienced brittle failure at and around the weld interface, whereas the temperature controlled weld experienced severe necking and failed about 25 mm away from the weld interface in base metal. Both welds exceeded the published yield and tensile strength of approximately 450 and 600 MPa respectively. The presence or absence of temperature control did not negatively affect the usage for yield-based design criteria.

![Figure 5.5: Engineering stress-strain curves of both welds. Stress is calculated with respect to the original cross section area. Strain is calculated with respect to the tubular gauge section, which is approximately 100 mm (4 in).](image)

5.5.2 Temperature

The temperature history of the temperature controlled weld is shown in Figure 5.6. The observed temperature variations in the 1 mm thermocouple are due to intermittent stick-slip power generation at the interface and changing force commands given by the temperature controller.
Figure 5.6: Thermocouple temperatures for the temperature controlled weld. The setpoint and trigger temperatures of 900 °C and 700 °C are shown in horizontal dashed lines. The end of the preheat and weld sections are shown with two vertical dash-dotted lines.

Cooling profiles for both welds are superimposed on TTT diagrams in Figure 5.7. Measured weld power was used by the FEA model to simulate the temperature history of the welds. Both the interface and farthest position that reached the austenitic temperature are plotted for the FEA simulations. If weld thermocouples reached the austenitic temperature, the closest and farthest thermocouples that reached the austenitic temperature are also shown in the plots.

The FEA model predicts that material within 0.75 / 3 mm from the interface in the uncontrolled / controlled welds reached the austenitic temperature of 766 °C. In line with this, the uncontrolled weld did not have any thermocouples that reached 766 °C, whereas several thermocouples of the temperature controlled weld reached the austenitic temperature. The FEA simulations predict a fully martensitic structure near the interface of the uncontrolled weld, and a fully pearlitic structure through the entirety of the temperature controlled weld.

5.5.3 Microscopy

Scanning electron micrographs were taken of both samples, both at the interface and in non-HAZ base metal, and are shown in 5.8. The base metal of both welds shows pearlite and ferrite grains as expected for a sub-eutectic carbon steel. The interface of the controlled welds shows deformed ferrite and fine pearlite/discontinuous cementite grains. In contrast, the uncontrolled...
Figure 5.7: TTT diagram for 1045 steel with cooling curves for the (a) uncontrolled quicker weld and (b) temperature controlled weld.

The controlled weld interface contains martensite with small pockets of retained austenite/ferrite, but is completely void of pearlite structures.

Figure 5.9 shows that the base metal abruptly starts to transition from ferrite and pearlite at 0.7 mm from the interface, with the transition finishing at approximately 0.2 - 0.3 mm away from the interface. Lath structures are faintly visible in many of the dark areas of Figure 5.9 in the 0.7 - 0.4 mm range under higher magnification. It is thus theorized that the transition zone between 0.7 and 0.3 mm is due to incomplete transformation of ferrite and pearlite to austenite during due to...
insufficient time and temperature above the austenitic transformation temperature. Any austenite formed during welding transformed into martensite during cooling.

### 5.5.4 Microhardness

The microhardness profiles of both welds are shown in 5.10. The faster uncontrolled weld has an area of extreme hardness extending about 0.7 mm on each side of the center of the weld. The temperature controlled weld shows slight hardening at the center, but has mostly uniform hardness. Both welds show evidence of slight tempering around the weld zone, about 20 HV softer than the base metal.

Estimated martensitic percentage was determined by interpolating the hardness values from the intent columns visible in Figure 5.9 against hardness vs martensite percentage data for 0.45%
Figure 5.9: SEM micrograph of the uncontrolled weld near the interface. The image spans from 0 to 0.9 mm away from the weld interface, with the weld interface at the top of the image. The microhardness grid is spaced 100μm vertically and 500 μm horizontally. The middle three columns of indents are visible in the micrograph.

Figure 5.10: Microhardness profiles of the uncontrolled quicker weld and temperature controlled weld. The middle three columns of indents are averaged together in the above figure.

carbon steel [43]. The calculated martensite percentage is plotted against distance from interface in Figure 5.11.

5.6 Discussion

Pre-weld thermal simulations in Figure 5.4 suggest that a shift from martensitic to pearlitic microstructures can be achieved by controlling temperature. Both thermocouple data and post-
Figure 5.11: Calculated martensite percentage versus distance from the interface of the uncontrolled weld. Data is calculated from the three middle microhardness traces visible in 5.9. A drop to almost 0% martensite occurs about 0.7 mm away from the interface.

Weld FEA simulations support this conclusion (Figure 5.7). SEM images show that grains near the weld interface are broken up and refined, which is to be expected for a thermo-mechanical deformation process (Figure 5.8). Even at the interface, grains of both ferrite and lath pearlite with discontinuous cementite are present. Microhardness measurements confirm that the hardness throughout the entire weldment is consistent with a pearlitic structure (Figure 5.10).

In contrast, the thermal data from Figure 5.7 predict that austenitized material in the uncontrolled weld will transform into martensite upon cooling. FEA simulations predict that only material within 0.75 of the interface reached austenitic temperatures. This is in line with the microhardness data of Figure 5.10, which shows a dramatic increase in hardness at 0.7 mm. Between 0.3 and 0.6 mm of the interface, the martensitic percentage decreases from over 95% to about 80%, before dropping off entirely at 0.7 mm (Figure 5.11).

The macrograph of Figure 5.9 shows a change in microstructure sharply occurring at about 0.7 mm, and then gradually fading until saturating at about 0.3 mm. This is in-line with the estimates of martensitic percentage in Figure 5.11. Cooling rates from Figure 5.7 are sufficient to form martensite within 0.75 mm from the interface, and some martensitic structures can be seen as low as 0.6 - 0.7 mm away from the interface. Therefore it is believed that the region near 0.7 mm did not have sufficient time and temperature during heating to fully transform into austenite in the
first place. Very small pockets of retained ferrite near the weld interface in Figure 5.8 support this hypothesis. Yet any austenite that did form during welding completely transformed into martensite upon cooling.

Without temperature control, cooling rates were sufficiently high that martensite formed in any material that had sufficient time and temperature to transform into austenite during welding. In contrast, temperature control was able to control the cooling process such that brittle martensitic microstructures were avoided, and post-weld microstructure and properties without any heat treatment were pearlitic and similar to the base metal.

5.7 Conclusions and Future Work

Cooling rate and temperature control is developed, and is shown to be a way to obtain desired microstructures in FW. No negative affects occurred due to temperature and cooling rate control in 1045 steel – no weld defects were present, yield strength was the same for both welds, and no part geometry modifications were needed. Multiple evidence sources including thermal FEA, SEM, and microhardness all show that cooling rate and temperature control was able to completely stop the formation of martensite in this 1045 steel. The only necessary modifications to enable this control were (1) adding thermocouples to the parts, and (2) changing machine programming.

Inherent limitations exist for FW thermal control, with thermal and geometric properties limiting what cooling rates and microstructures can be formed in a given weld. The metal’s thermal diffusivity and part gauge length primarily determine the minimum cooling rate. At extremely slow cooling rates, convective and radiative cooling may instead dominate cooling. Consequently this method will not work for all microstructures of all alloys, such as forming pearlite in an M grade steel.

A large amount of potential future work exists in the field of temperature and cooling rate control in rotary friction welding. Temperature control may be able to help mitigate ultra-fine grain structures at weld interfaces where creep is a potential concern, such as in superalloys jet engine applications. A more in-depth mathematical treatment could be given to the limits of cooling rates, including for instance convective cooling, temperature dependent material properties, different heat-generation assumptions, etc. An induction coil could be used to to achieve ultra-slow cooling rates that cannot be achieved solely by building up interface temperature. More research could be
done in generating power at the sample interface, such as a more in-depth look at using spindle speed vs. axial force as the manipulated variable, and also smoother power generation which would produce steadier interface temperatures.
6.1 Discussion

Friction welding is a thermo-mechanical process, and thus the input and measurable parameters are interconnected. Metal alloys lose strength with increasing temperature, so welding forces decrease with increasing temperatures. Energy is the fundamental driver of temperature. A higher power results in higher temperatures if it is not offset by larger losses. In friction welding, welding with a higher force does result in higher power generation, but also results in a much higher feedrate. This means that the specific power (power per unit material displaced) is much lower, and lower temperatures result as a consequence. Similarly in the sister technology FSW, heat input (power per distance traveled) is a more reliable indicator of weld temperature than is power or total weld energy [31]. For the same total upset, a higher welding force leads to a higher feedrate, higher power, lower specific power, lower total energy, and lower temperature.

Symmetry of both weld pieces has been studied in the context of dissimilar welds. A higher welding power or temperature is sometime sought in order to lower the strength of the stronger work piece, and thus produce a more symmetric and stronger weld. However, higher temperature also lowers the strength of the weaker work piece. Therefore, it is not the strength of the stronger alloy, but the ratio of strengths of the two work pieces that determines upset and flash asymmetry in a dissimilar alloy/geometry weld. In some alloy systems, asymmetry increases with increasing temperature, and in other cases the exact opposite occurs.

The goal of all friction welding is to join two work pieces together - this is the driving force behind the selection of welding parameters. On a fundamental level, all that is required to bond two metallic atoms is proximity. To do so, enough force must be applied to overcome any repulsive forces, with temperature aiding in the bond formation. In real-world cases however, surface roughness prevents all but a few high points from coming into contact with each other.
Furthermore, contaminants and oxides are present on metal surfaces, and these prevent the metal atoms from coming close enough to form a bond.

Overcoming the obstacles of roughness and contamination is critical to friction welding. Increased welding force helps overcome surface roughness, but is insufficient to completely put the surfaces into atomic-level intimate contact. Creating small amounts of plastic deformation at the surface allows the surfaces to perfectly match when put under pressure. To remove contaminants and tenacious oxides, large upsets are often employed to expel the original surfaces from the welding interface. However, the bifurcation point in the joint is never expelled [44], and so attempting to extrude literally 100% of the interface is impossible. This is, however, not actually necessary. The sliding action and deformation of the interface breaks up oxides and contaminants, and diffusion acts to disperse these very small amounts of containment atoms into the metal body instead of extruding them from the interface. Therefore, a small amount of deformation and some heat are all that is actually required in order to form metallic bonds at the interface.

Welds can often be created by a trial and error approach treating FW as a “black box” process. Although this is an inefficient approach, it is sufficient for many alloys. Selecting welding parameters while considering the process fundamentals can lead to better welds much more easily. A high welding force helps bring the two welding surfaces together into intimate contact. Some surface deformation is required, but not extensive amounts. An unnecessarily high temperature lowers the force carrying capacity of the metal and worsens the HAZ in the finished weld. In contrast, a low temperature results in cold metal without sufficient ductility to accommodate the strains induced in FW. A low temperature also slows down diffusion which significantly aids forming a new bond. For this reason, a balanced temperature results in an ideal weld due to competing temperature-dependent mechanisms.

Application of process control and material science fundamentals enables better welds to be created in adverse situations. For example, many advanced nickel-base superalloys have extremely low ductility, and as a consequence experience catastrophic failure near the interface when welded with large upsets. The solution is therefore to produce welds with high forces and low upsets. Alternatively, if the ductility of the alloy can be improved (i.e. though a specially designed geometry), then the risk of catastrophic failure can be reduced. Similarly, some steels are quench
sensitive and form martensite during rapid cooling. Consequently by finely controlling the process, the cooling rate can be slowed down by building up temperature in advance of the actual weld.

6.2 Conclusions

In all cases studied in this dissertation, welding with a higher force and lower temperature improved weld quality. HAZ regions were made smaller. Defects at the surface were minimized or eliminated, and consolidation improved. The measured tensile strength of the welds also improved, which is primarily determined by both the weld’s annealed state and any defects that are or are not present. Not only did weld parameters not correlate to weld upset asymmetry across alloy systems, but weld asymmetry did not correlate to weld strength.

Large upsets were not required for good friction welds. Surface deformation is required to form a full bond, but this often is satisfied by a total upset of 0.2 - 0.4 mm instead of the approximately 5 mm in a “normal” weld. In the traditional superalloy alloy 718, minimal upsets corresponded to the highest strength welds. In the advanced superalloy A, minimal upset welds were the only way that reasonable strengths were obtained with normal weldment geometries. In alloy A, the highest strengths were observed when a specialized containment geometry was used, which increased the hydrostatic component of pressure at the weld joint. Not only were the highest strengths observed, but weld input conditions that previously resulted in strengths of 3 - 16% in normal geometry weld pieces tensile strengths of >70% with the hydrostatic-inducing containment geometry.

Using process control to improve welds also extends to closed-loop control. Using a closed-loop temperature controller, thermal energy was built up in weld pieces of 1045 steel while mitigating peak temperatures. Once the weld was completed and cooling occurred, favorable microstructures (pearlite instead of martensite) were generated, and despite the increased total energy, weld strength was equal to or better than the shorter uncontrolled weld.

6.3 Future Work

In all of the welds in this dissertation, higher force led to better welds. There exist limits to this phenomena however, and future work could explore this. For example, studies on the minimum
amount of heat needed to form a solid bond at high forces would be complimentary to this research. In a more general sense, this thesis suggests that many commonly held assumptions of welding and parameters could be revisited in light of advancements made in this thesis - very high force, low upset, closed loop control, etc. With these advances, knowing the limits of these approaches would be very helpful. In addition, lessons learned from this dissertation could be applied to completely different alloy systems such as aluminum to determine if/how they may be similar or different for a completely different alloy system.

This work has shown that low ductility alloys can be welded with proper technique, particularly using: low upsets, high forces, and a containment geometry. This could be further tested for additional alloys. Some interesting possibilities would be ceramic compounds that are sintered together and have less than 1% ductility. Additionally, the exact geometry used in hydrostatic inducing containment welds can be modified and likely improved further.

This dissertation introduces for the first time closed-loop temperature control in FW. This may have other applications, such as limiting cooling to mitigate ultra-fine grain regions. The prediction algorithms could likely be improved, and the mathematics behind both the limits and control could be advanced. This could enable the creation of consistently better welds for a variety of materials that currently are not considered weldable in FW.
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