Inertia Friction Welded Ni-Base Superalloys: Process Examination, Modeling and Microstructure

DISSENTATION

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

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2016

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Abstract

Inertia friction welding of Ni-base superalloys is of increasing importance for future designs of gas turbine engines where the overall pressure ratio and operating temperatures are increased in an effort to improve efficiency. Ni-base superalloys are chosen for the hottest sections in turbine engines due to their inherent ability to retain good strength levels to very high fractions of their melting point. This inherent strength at high temperatures is due to the strengthening contributions of the ordered $\gamma'$ precipitates. As model systems, the Ni-base superalloys LSHR and Mar-M247 were chosen for further welding and investigation. The LSHR is a powder-metallurgy alloy that is a surrogate for a bore material in a multi-alloy disk system. The Mar-M247 is a coarse-grain cast alloy that is a surrogate for a rim alloy.

As the strength retention at elevated temperatures is advantageous for engine applications, it is disadvantageous for the production of sound welds. The retention of strength at elevated temperatures effectively reduces the size of the processing window for these alloys, further complicating the welding process. These alloys that are high in $\gamma'$ formers (Ti and Al specifically) are also prone to strain-age cracking and other post-weld cracking issues. These factors make it necessary to utilize a process model to improve weld outcomes and reduce the amount of trial-and-error experimentation.
Finite element modeling techniques and experimental weld process examination have led to insights into the complex interplay between the weld process parameters and their impact on post-weld microstructure, bond quality & character as well as mechanical properties. A number of modeling approaches have been explored in the literature, finite element, analytical, numerical etc, however all have limitations to their usefulness. None of the modeling approaches have demonstrated true predictive capability, or general applicability. Also, none of these approaches have eliminated the need for experimental weld process data as input.

In effort to improve upon the predictive capability of the finite element process model, a number of weld parameters were examined. A minimum energy input bond criteria was formulated along with empirical relationships between the process parameters and weld behavior. Key input data for finite element process modeling was shown to include flow stress, coefficient of friction, and weld process efficiency. Strategies to estimate appropriate values for these key parameters were demonstrated and validated within the process model. The methodology to identify appropriate welding parameters, as well as their relative importance, was determined based on weld-trial data and resultant bond quality. The Ni-base superalloys LSHR and MarM247 were welded under varying conditions to provide experimental validation data. The finite element model results indicated the importance of the weld variables outlined above for accurate prediction of weld quality and upset.

The effects of inertia-friction-welding process parameters on the microstructure, weld-plane quality, and tensile behavior of welds between dissimilar nickel-base
superalloys were established. For this purpose, the fine-grain, powder-metallurgy alloy LSHR was joined to coarse-grain, cast Mar-M247 using a fixed level of initial kinetic energy, but different combinations of the flywheel moment of inertia and initial rotation speed. It was found that welds using the largest moment of inertia produced a sound bond with the best microstructure and room-temperature tensile strength equal to or greater than that of the parent materials. The post-weld tensile behavior was interpreted in the context of observed microstructure gradients and weld-line defects.

The investigation showed that the weld kinetic energy was not a sufficient criterion to set weld process parameters, and the effects of the efficiency of the weld equipment can have a profound impact on both the interpretation of model results and weld outcomes. It was also established that the weld upset behavior changes as a function of the energy multiplied by axial load. This parameter was also related to a minimum upset criterion.
Dedication

This document is dedicated to my wife Andrea, and children Abigail & Madelyn.
Acknowledgments

I would like to acknowledge the support of my advisor and committee members, Professors Shivpuri, Zhang and Fraser. Your input, encouragement and patience during this journey are very much appreciated. Without you this would not have been possible.

I must also thank Lee Semiatin, Oleg Senkov, Suresh Babu and Shesh Srivatsa for the numerous challenging and fruitful discussions along the way.

I would also like to thank Dan Evans and Paul Ret for setting me on this course and their encouragement throughout the process. The staff of the Metals Branch has been invaluable.

I also want to acknowledge Jay Tiley and Don Weaver for their friendship and support over the years.

I gratefully acknowledge the support from Scientific Forming Technologies Corp and use of their DEFORM finite element software.

This research was supported by the Materials and Manufacturing Directorate of the Air Force Research Laboratory.
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Fields of Study

Major Field: Industrial and Systems Engineering
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Chapter 1: Introduction and Problem Statement

Gas turbine engines utilized in aerospace applications and land-based energy generation operate at extremely high temperatures and pressures for long durations in order to perform with adequate efficiency and thrust [1-4]. These conditions levy significant challenges to the metals utilized within the hottest sections of the engine. Temperatures at the compressor exit or turbine inlet can exceed 760°C routinely [4]. The conditions demand that the metals utilized in these applications provide inherent corrosion resistance, good creep strength, oxidation resistance, fatigue resistance, and yield strength all while operating at high temperatures for extended periods of time[3, 4]. This greatly reduces the field of candidate materials that are capable of handling such extreme environments.

Traditional design methods for turbine engine applications utilize monolithic materials such as titanium alloys and nickel-base superalloys for the rotating components. Various product forms of these alloys have been utilized which include, cast & wrought, and powder metallurgy methods [1]. Cast product forms have been favored in blade applications due to the inherent creep strength at very high temperatures (few grain boundaries); these include equiaxed cast, directionally solidified, and single crystal. Wrought product forms have been favored for disk applications due to their inherent
toughness and strength at intermediate temperatures (due to fine grain structure and chemical homogeneity); these include traditional cast/wrought and powder metallurgy forms. Nickel-base superalloys are a class of metallic alloys that provide good strength at high fractions of their homologous temperature, creep strength and inherent oxidation resistance[1, 2].

Nickel-base superalloys are a class of metallic materials that are based on face-centered-cubic (FCC) Ni, and contain γ’, carbide and oxide precipitates in a gamma matrix. The FCC structure provides ample slip systems which generally translate to good toughness and strength. The γ’ particles are an ordered L12 structure based on Ni3Al, and provide small misfit with the FCC gamma matrix. The degree of misfit between the γ’ and gamma grains is important in that it controls the stability and driving force for coarsening of the γ’ precipitates, which in turn accounts for the retention of strength at high temperatures. Ni-base superalloys generally contain numerous constituent alloying elements. These fall into a couple categories, those that partition to γ’ (Al, Ti, Nb, Ta, Hf), gamma (Ni, Co, Fe, Cr, Mo, W), or produce oxide and carbide particles (Cr, Mo, Nb, Ta, W, Ti). Ni-base superalloys are generally precipitation strengthened, although benefit from solid solution effects, and grain refinement strengthening as well. The unique nature of the γ’ precipitates produce an increase in yield strength with increased temperature which allows these alloys to be used at such a high fraction of their homologous temperature (0.85Tm) and make them most suitable for high temperature applications such as in gas turbine engines. The normally slight misfit between the γ’ and gamma matrix reduces the coherency strains associated with the precipitate which in turn,
promotes stability of the precipitate, i.e. sluggish coarsening. Increasing the misfit however, promotes stronger interactions between dislocations and the $\gamma'$ precipitates which improves strength levels. Therefore, tailoring the misfit for the application is an important aspect for design with Ni-base superalloys.

Ni-base superalloys which contain many constituent alloy elements can present problems for traditional cast and wrought product forms. Significant segregation, coarse grain sizes and deleterious phase formation are just a few of the challenges for cast and wrought processes. In order to provide chemically homogeneous product forms, powder metallurgy processes are common. In these processes, the alloy is gas-atomized into spherical powder, which is then typically consolidated via hot isostatic pressing and further processed via extrusion and forging into a near-net shape form. The fine grain sizes that are typical from powder metallurgy processing routes provide further benefit in that the material can be superplastically formed into complex shapes relatively easily as compared to the inherent strength at elevated temperatures typical of these alloys [1].

Utilizing a monolithic design and therefore the assumption of uniform mechanical properties across the component in gas turbine engine disks allows for reduced complexity in the manufacturing and lifting approaches, however places challenging and divergent requirements on the alloy chosen. For turbine disks, the inner bore operates at intermediate temperatures and high stress, while the outer rim requires creep resistance at high temperatures and intermediate stresses [4]. These competing requirements force designers to make trade-offs in performance capability when selecting appropriate alloys.
In other words, for monolithic construction, non-optimal alloys are chosen that balance the mechanical properties required for both locations in the disk.

An alternative approach to produce a disk with mechanical properties tailored for specific locations would be to join two metals with dissimilar mechanical behavior. For nickel-base superalloys, solid state joining processes would be required due to the limited weldability of high-temperature capable metals [5, 6]. Many solidification related defects are prevalent when melting nickel-base superalloys that typically have many constituent elements that make up the alloy [3]. Therefore it is generally preferred to utilize joining methods that do not require melting of the materials to form a bond. Within the solid state joining processes, friction welding techniques would be most suitable for this type of application, although diffusion-based techniques such as diffusion bonding, or transient liquid phase bonding could be suitable [7, 8]. For axisymmetric components such as an engine disk, inertia friction welding (IFW) is the obvious choice, as direct-drive friction welding may present issues with motor limitations with respect to the size and power requirements for full-scale production [5]. Another key comparison with IFW and direct-drive friction welding is that the energy input with direct-drive is constant until the braking mechanism is applied. In IFW, the energy input is regulated by the energy consumed at the weld interface and within the machine bearings. The energy input rate is continuously decreasing until the process is complete and therefore exhibits a less abrupt stop than continuous-drive. Diffusion based techniques also can prove disadvantageous due to numerous reasons for these types of applications. Most notably, the formation of deleterious phases at the weld interface, due to extended times at high temperatures.
involved, as well as the retention of oxides at the interface due to no bulk deformation or material flow are both detrimental to developing high strength bonds between these types of materials [9, 10].

IFW is a solid state process that involves rotating a flywheel of known mass to a pre-determined initial rotational velocity. Once this initial rotation speed is reached, the initial kinetic energy of the weld is set, and the weld samples are brought together under a fixed axial load. The weld faces of the samples increase in temperature due to frictional heating and once this thermal field expands across the weld face and reaches a critical value where the shear yield strength of the metals is reached by the applied shear stress, bulk deformation begins.

In the specific process known as inertia-friction welding (IFW), the energy is supplied by a rotating flywheel, and the primary process parameters are the flywheel moment of inertia ($I$), the initial flywheel rotation speed ($\omega_0$), and the applied axial (forging) force ($P$). $I$ and $\omega_0$ define the initial kinetic energy of the flywheel $E_{ko}$ (also referred to as the welding energy):

$$E_{ko} = \frac{I\omega_0^2}{2}$$  \hspace{1cm} (1)

During the IFW process, the kinetic energy of the flywheel is transformed into heat via friction at the weld interface. The energy required to produce a sound weld is generally considered a sufficient criterion for a given material combination and weld geometry [11]. However, Equation (1) indicates that numerous combinations of $I$ and $\omega_0$ can produce the same value of $E_{ko}$. 
As long as the energy input into the weld process is balanced by the energy sinks within the weld machine, and energy consumed by the plasticized material moving into the flash, then the weld process continues to reduce the length of the weld samples until rotation stops. This process is generally considered ‘self-limiting’ as much as only the energy required by the process to produce the temperature required at the interface sufficient to allow material flow into the flash is utilized at any given time step. This process is typically relatively short, on the order of a few seconds from maximum rotation until rotation stops. Due to this short duration, the thermal profile is steep as a function of distance from the weld, and therefore all bulk deformation takes place in a relatively thin layer adjacent to the weld face. There is a characteristic intermediate torque peak at the beginning of the weld process (due to the initial coupling of the weld faces, and deformation of the surface asperities), followed by a period of near steady-state torque values (where the material flow at the weld face is balanced by frictional and deformation heat input), and finally a large torque peak occurs upon the rotation speed of the flywheel reaching zero rad/s. The IFW process is fairly simple, in that there are only a few independent variables. These include, sample geometry, alloy, moment of inertia (flywheel size), initial rotational velocity, and axial load. The typical time-dependent output parameters that are measured include, upset length, rotation speed, energy decay, and weld temperature [12-14].

IFW was utilized to weld the forged, PM superalloy LSHR to coarse-grain, cast Mar-M247 [15]. The welds exhibited two different types of behavior during post-weld
tension testing [16]: (1) deformation and fracture in the parent material or (2) fracture at the weld interface due to weld-related defects [15, 16]. The particle clustering associated with the defects was suggested to have resulted from insufficient radial plastic flow of Mar-M247 during IFW. This limited plastic flow inhibited “self-cleaning” at the weld interface and insufficient mechanical mixing of the mating surface layers (which tends to disperse undesirable inclusions into the bulk), both of which are characteristic of friction-welding processes. The radial cracks at/near the weld interface were likely associated with insufficient ductility of the weld material adjacent to the weld line which was forced to twist further immediately prior to cessation of flywheel rotation. In this regard, it was suggested that the propensity for the formation of radial cracks could be reduced if the flywheel rotation were stopped before the critical shear stress/strain for fracture at the bond line was reached.

A limited amount of research has been published in the open literature regarding IFW of Ni-base superalloys. Most of this work has involved microstructure analysis (with particular attention to the gamma grain size and $\gamma'$ precipitate behavior) as well as mechanical response of the weld zone both before and after post weld heat treatment [17-24]. Of those references listed most did not provide specific details regarding the welding parameters utilized. This fact makes drawing comparisons with the conclusions drawn difficult. It is also interesting to note that only a few Ni-base superalloys have been reported, these include IN718, 720Li, RR1000, LSHR and Mar-M247. Both similar and dissimilar welds were performed and limited results were published. In general, it was found that Ni-base superalloys that have a high fraction of $\gamma'$ precipitates as a main
strengthener tended to exhibit near-weld line regions of fine, re-precipitated $\gamma'$ upon cooling from the weld. Alloys such as IN718 tended to have a denuded region adjacent to the weld line [17-19]. This difference in precipitation behavior was reported to be caused by the relative amounts of Al and Ti in the alloys. For the high fraction $\gamma'$ alloys (720Li, RR1000, LSHR) the large amount of Al and Ti in solution near the weld line tends to cause a fine, evenly dispersed precipitation of on-cooling $\gamma'$, while in alloys like IN718, the reduced amount of $\gamma'$ formers (Al, Ti) in solution provides a reduced driving force for on-cooling precipitation of $\gamma'$, and therefore a denuded zone near the weld. Atom probe tomography was utilized in LSHR near-weld line and showed the chemistry of the fine $\gamma'$ did not change as a function of distance from the weld line, which indicated that there was insufficient time at elevated temperatures to allow significant diffusion [22]. It was also noted that in these efforts the as-welded condition hardness profiles mirrored the $\gamma'$ area fraction profiles. This trend was most notably captured by the use of x-ray synchrotron diffraction techniques to capture the fraction of $\gamma'$ as a function of distance from the weld line [25]. This technique eliminates the ambiguity in measuring area fractions via SEM-based techniques that is due to the relative size of the tertiary particles and size registration due to imaging techniques. This however, does not differentiate the $\gamma'$ into its families, i.e. primary, secondary, and tertiary. It was also demonstrated that a post weld heat treatment (stress relief and age) tended to recover the near-weld hardness values back to those approaching parent material values [18, 19].

Due to the complex behavior of the near-weld $\gamma'$, high strength at elevated temperatures, and complex chemistries, developing the welding parameters for these
high-temperature capable alloys can prove challenging. Typical trial-and-error approaches although time consuming and expensive, are common [12-14]. Arriving at weld parameters that provide sufficient deformation within both materials at the weld line is crucial to producing defect-free welds. This can be challenging due to the inherent nature of nickel-base superalloys to retain strength at very high fractions of their homologous temperature. Another challenge is resident in the IFW equipment itself. It is common to have variable weld outcomes from the same IFW machine operated under the same conditions as a function of time. This can be due to a number of factors to include the condition of the drive train & bearings, hydraulic fluid state, recent maintenance etc.

To overcome these process challenges, a robust & validated process model that can accurately predict the weld zone temperatures, flash formation, and total reduction in length would provide a major step forward. Various approaches to model the IFW process have been reported in the literature [26-40] but no industrial standard model or approach has been settled. Various finite element modeling approaches have been reported but none have been shown to provide a truly predictive capability along with general applicability across alloys and geometries [26-35, 38, 40]. Also, these models all rely on some form of experimental weld process data as input and show reasonable agreement with experimental values. Another shortcoming is that the process models do not accurately account for energy efficiency within the IFW machine itself; in fact three recent literature reviews cite numerous articles describing the IFW process with no mention of welding process efficiency [7, 8, 41]. The articles that do mention process efficiency assign an arbitrarily high value with little or no justification [27, 28, 42].
Process modeling holds promise to reduce reliance on Edisonian trial-and-error approaches but significant challenges remain to provide a truly predictive capability.

First off, no standard method to determine the effective friction coefficient at the weld interface has been defined. There are various methods that have been developed to address this issue of effective friction coefficient. Energy input-based calculations can provide an estimate of the effective friction coefficient [28, 35], shear or coulomb relationships [26, 29, 32, 34, 38], direct measurement of pin-on-disk sliding friction, as well as methods prescribing a heat flux at the weld interface in order to remove the requirement for friction have all been used [27, 30, 31, 33, 36, 37, 39, 40]. The friction coefficient is challenging as many values have been reported in the literature and few agree. This is due to the nature of the coefficient, as the temperature, surface finish, applied load, and possibly numerous other factors impact the estimated values[12]. Also, specific to the IFW process it appears that the effective coefficient changes from a coulomb-based behavior early on in the process (prior to any significant deformation) to a shear-based behavior when the weld line material flows radially into the flash. The apparent reliance on experimental weld data to estimate the effective friction coefficient is one of the main limiting factors for predictive finite element modeling of the IFW process.

Another challenge to process modeling is the input flow stress data. Finite element modeling requires input flow stress data as a function of temperature, strain and strain rate. For IFW, this data set must cover the range from room temperature up to those approaching the melting point of the alloys involved. The strain rates estimated during
IFW have been reported up to 1000 s\(^{-1}\) [41]. This poses a particular challenge in that developing the flow stress data across such a wide range of temperatures and strain rates is both time consuming and expensive. This also requires specialized test equipment to reach the highest temperatures and strain rates. Traditional isothermal compression testing can provide flow stress data up to approximately 10 s\(^{-1}\) strain rates (and high temperatures), while torsion testing can cover a higher range of strain rates (~ 50 s\(^{-1}\)). Higher strain rates yet require impact test methods such as split Hopkinson bar testing. The wide range of flow stress data required poses a particular challenge as finite element codes such as DEFORM will interpolate flow stress data needed during the simulation from the input data set. Therefore, the sparser the input data set, the more likely errors from interpolation will enter into the analysis. Programs such as JMatPro can predict input flow stress data sets formatted for DEFORM which evenly cover the temperature and strain rate ranges mentioned previously. However, these predicted data sets require experimental validation to make sure that the yield strength, ultimate strength and strain hardening/flow softening behavior are correct.

A third challenge to effective process modeling is accounting for the energy partitioning during the weld process. During IFW, the initial kinetic energy of the flywheel (flywheel mass rotating at the specified initial velocity) steadily reduces to zero as the weld couple is brought together under the applied axial load. The energy supplied to the system is reduced by a number of factors which include, frictional heating at the weld interface, thermal gain in the hydraulic fluid, and frictional heating of the axial and thrust bearings. For a given IFW machine and set of weld parameters, a characteristic
fraction of the supplied energy actually participates in producing the weld. The ability to accurately account for the energy losses not part of producing the weld is critical to effective modeling and process scale-up and has not been published previously. The only references that mention process efficiency apply a constant value regardless of process parameters (or materials) and is in the range of 85 – 95% [27, 28].

Another challenge with respect to process modeling of the IFW process is the lack of predictive capability for the actual bonding process. In FEM, there is no direct prediction of the bonding state at the weld interface. One can make inferences as to when bonding may occur by carefully examining upset rate data, experimental weld data, and estimated effective coefficient of friction data, but no direct method of modeling the actual bonding process exists. In addition, there is no direct method to predict the presence of weld-line defects. Examining the amount of material removed from the weld zone into the flash may be the only metric that provides a reasonable estimation of the likelihood of weld defects and therefore state of bonding.

Based on the challenges detailed above, it is evident that the need exists to further develop the modeling capabilities for the IFW process, as well as to perform a careful analysis & characterization of the weld equipment, process and resultant weld zone microstructure and mechanical behavior of Ni-base superalloys. Enabling the ability to efficiently determine and/or optimize welding conditions for a dissimilar set of nickel base superalloys while providing insights into the effects of the individual welding parameters on the weld outcome is critical to advancing the state of the art in solid state joining of these difficult to weld materials.
Chapter 2: Methodology & Procedures

The program materials consisted of two nickel-base superalloys, isothermally-forged LSHR (low-solvus, high refractory) and cast Mar-M247, whose compositions are given in Table 1. The LSHR had an initial average gamma grain size of 3.5 µm. The microstructure of Mar-M247 consisted of coarse grains between 1000 and 6000 µm. The $\gamma'$ solvus temperature of LSHR is 1157°C; and that for Mar-M247 is 1225°C. The yield strength of LSHR is approximately 50% higher than that of Mar-M247 at temperatures up to 750°C; at yet higher temperatures Mar-M247 exhibits higher yield strength [15, 16].

<table>
<thead>
<tr>
<th></th>
<th>Al</th>
<th>B</th>
<th>C</th>
<th>Co</th>
<th>Cr</th>
<th>Hf</th>
<th>Mo</th>
<th>Nb</th>
<th>Ti</th>
<th>Ta</th>
<th>W</th>
<th>Zr</th>
<th>Ni</th>
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<tr>
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<td>0.03</td>
<td>0.03</td>
<td>21.4</td>
<td>12.3</td>
<td>0.06</td>
<td>2.66</td>
<td>1.45</td>
<td>3.48</td>
<td>1.58</td>
<td>4.48</td>
<td>0.05</td>
<td>49.3</td>
</tr>
<tr>
<td>Mar-</td>
<td>5.10</td>
<td>-</td>
<td>0.07</td>
<td>10.4</td>
<td>8.51</td>
<td>1.49</td>
<td>0.73</td>
<td>0.00</td>
<td>0.94</td>
<td>2.64</td>
<td>10.0</td>
<td>0.01</td>
<td>60.0</td>
</tr>
<tr>
<td>M247</td>
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<td></td>
<td></td>
<td></td>
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</table>

Table 1: Average Composition (wt.%) of LSHR and Mar-M247 Program Materials.

LSHR samples measuring 12.7-mm diameter x 51-mm length were extracted from the center section of a 330-mm-diameter x 51-mm-thick pancake forging using electric-discharge machining (EDM). The longitudinal direction of the extracted samples was parallel to the short transverse direction of the forging; the weld face of each sample coincided with the surface of the forging. Samples of Mar-M247 with the same
dimensions were EDM’ed from a cast, 14.5-mm-thick, rectangular plate. The longitudinal
direction of each Mar-M247 sample was parallel to the solidification direction. The EDM
recast layer was removed by mechanical grinding.

Inertia friction welding trials were conducted to assess the effect of process
parameters (Independent parameters labeled as “Ind” in Table 2) on the weld output
measures (dependent parameters labeled as “Dep” in Table 2) as well as the quality of
LSHR/Mar-M247 welds. The weld quality was determined from the mechanical test data
of the samples in the as-welded form as well as a relative comparison of the weld line
defects present.

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>( I ) (kg.m(^2))</th>
<th>( \omega_0 ) (rad/s)</th>
<th>( E_o ) (kJ)</th>
<th>( P ) (kN)</th>
<th>( \Delta l ) (mm)</th>
<th>( T_{ss} ) (N.m)</th>
<th>( \mu_{ss} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>Parameter state</td>
<td>Ind</td>
<td>Ind</td>
<td>Ind</td>
<td>Ind</td>
<td>Dep</td>
<td>Dep</td>
<td>Dep</td>
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<tr>
<td>LM01</td>
<td>0.166</td>
<td>518</td>
<td>22.3</td>
<td>60.0±1.5</td>
<td>2.86</td>
<td>20.3±1.7</td>
<td>0.074±0.006</td>
</tr>
<tr>
<td>LM02</td>
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<td>340</td>
<td>22.0</td>
<td>60.0±1.5</td>
<td>3.80</td>
<td>17.0±1.7</td>
<td>0.064±0.006</td>
</tr>
<tr>
<td>LM03</td>
<td>0.802</td>
<td>235</td>
<td>22.1</td>
<td>60.0±1.5</td>
<td>5.78</td>
<td>18.5±1.0</td>
<td>0.069±0.004</td>
</tr>
</tbody>
</table>

Table 2: Inertia Friction Welding Conditions (\( I \): flywheel moment of inertia, \( \omega_0 \): initial
rotation speed, \( P \): axial compression load, \( \Delta l \): total upset, \( T_{ss} \): steady-state total torque,
and \( \mu_{ss} \): apparent friction coefficient.)

The welded samples were sectioned longitudinally, and the microstructure,
defects, phases, chemical composition, and hardness were determined as a function of
axial location relative to the weld interface. In particular, energy-dispersive spectroscopy
(EDS), backscatter-electron (BSE) imaging, and electron backscatter diffraction (EBSD)
were performed in a scanning electron microscope (SEM) for microstructure and phase-
composition analyses. Electron probe micro analysis (EPMA) was used to determine
concentration profiles as a function of distance from the weld interface. To accomplish
this task, longitudinal EPMA scans at various radial locations were conducted
perpendicular to the weld interface to a distance of 300 µm into each alloy using a
spacing of 1 µm. Beyond 300 µm, the spacing was increased to 100 µm. Hardness
measurements were conducted in a microhardness testing machine using a Vickers
(diamond-pyramid) indenter and a 500-g load held for 20 s.

To establish post-weld mechanical properties, three subscale tension specimens
were excised via EDM from samples produced with each set of welding conditions. The
samples were extracted at three radial locations relative to the axial center line, 0.7 mm,
2.5 mm, and 4.3 mm (Figure 1). In all cases, the tension axis was parallel to the
longitudinal axis of the welded samples, and the weld interface was located in the middle
of the gage section. Following EDM, the lateral surfaces of each tension sample were
ground to a 600-grit finish. Fiducial marks having a spacing of 1 mm and covering 26
mm in total length were laser inscribed on the gage section. Tension testing was
conducted at room temperature using a constant ram speed of 0.02 mm/s, which
corresponded to an initial strain rate of 10^{-3} \text{s}^{-1}. A differential-image-correlation (DIC)
technique was used to capture the development of strain non-uniformity within each of
the inscribed sections during the tension tests. Details of this procedure were reported
previously [16].
A 2-D axisymmetric finite element model of the IFW process using the DEFORM™ software was utilized to explore the effect of the weld process parameters (including process efficiency) on weld upset and peak interface temperature. Trends in predicted upset and peak interface temperature from the computer simulations were compared to experimental results to aid in the interpretation and analysis of the welding process.

**Experimental IFW Trials**

In order to provide validation data for finite element modeling, as well as as-welded materials for metallographic investigation and mechanical testing, welding trials were performed at EWI utilizing their MTI model 120 inertia friction welding machine. The machine could apply up to 8000 RPM, and moment of inertia up to 19 wk². All
similar welds were performed with LSHR. Dissimilar welds were performed with LSHR and Mar-M247. Both superalloys were machined to the same dimensions (12.7mm diameter by 50.8 mm long). A significant number of weld trials were performed with both LSHR to LSHR and LSHR to Mar-M247 combinations. Weld parameter combinations were selected to sample a large range of energy, rotation speed, and axial load combinations. A number of welds were performed to examine the repeatability of the process. The LSHR to LSHR welds were performed primarily to provide input and validation data for the finite element process model, while the dissimilar welds were performed in effort to optimize the weld process for that combination as well as explore the ability to develop sufficient upset in both sides of the weld couple simultaneously.

The experimental welds provided a number of important sets of data used for in-depth analysis of the process. First, the welded samples provided a measure of total upset (loss of length), flash curl formation, as well as microstructural and mechanical property gradients as a function of distance from both the axial centerline and weld interface. Further weld information was obtained from the welding equipment itself. This time-dependent data includes axial force, displacement (upset), and rotation speed. The stationary specimens were also fitted with Type K thermocouples at distances from \( \sim 1.5 \text{mm} \) to 6mm from the weld interface in order to capture the time dependent thermal field during the weld process. Later trials involved the measurement of strains via strain gages on the weld sample in order to calculate the torque imposed on the stationary sample during welding.
The weld samples were sectioned along the axial centerline to allow for metallographic examination of the weld interface, heat affected zone, and flash formation. Hardness testing along the axial direction was performed on many of the samples in order to determine the effects of the weld process on local microstructure and mechanical properties. Microprobe analysis, EBSD and SEM imaging of the near weld areas was performed as well. For a subset of the welded samples, milli-scale dog bone shaped, flat tensile bars were extracted from one half of the excised weld specimen and tension tested at room temperature.

**IFW Process Model**

The finite element code DEFORM (Scientific Forming Technologies Corp) was utilized to produce a two dimensional process model for IFW. The process model was initially set up in a descriptive manner utilizing weld process data as input and comparing weld upset for validation. In this manner, model parameters and input data such as flow stress as a function of temperature and strain rate, thermal conductivity, process efficiency and friction coefficient can be tuned to provide output results comparable to the experimental weld trials. Once the model has been calibrated the formulation was altered such that only the initial kinetic energy of the process, friction coefficient and axial load were input. This formulation allowed the total reduction in sample length (upset) as well as the rotation of they flywheel to be compared to experimental data for
validation purposes. In both formulations, the upset, temperature profile, and flash formation can be compared to and utilized to interpret the experimental weld trials.

Within DEFORM, a 2.5-D axisymmetric element was used which accounted for torsional effects while maintaining the computational efficiency of a 2-D model. A fixed mesh size was utilized for the weld samples and grips which provided sufficient spatial resolution while maintaining computational efficiency. Due to the axi-symmetric nature of the weld samples, mirror symmetry across the axial centerline was utilized to reduce the number of elements required in the model (Figure 2). The grips were idealized as meshed, but non-deformable bodies at the top and bottom of Figure 2 in order to allow heat conduction away from the weld samples. The rotation was applied to the top grip and the axial load was applied to the bottom grip in Figure 2. Sticking boundary conditions were applied to the interfaces between the weld samples and the grips, as well as at the point on the weld interface at the axial centerline. A friction coefficient was applied between the weld samples, as well as between the flash curl and the outer radius of the sample. The velocity of the nodes along the axial centerline was set to zero, and heat exchange with environment was applied to the outer radius of the samples. The time step was set at 0.005 s, and the results were saved at every 5 steps.
A mesh size of 300 microns was assigned to the non-rotating side and a mesh size of 150 microns was assigned to the rotating side (Figure 3). A sensitivity analysis was performed on the mesh size that evaluated both upset length and simulation time. There were approximately 4000 elements utilized in the rotating side, and 1500 elements on the non-rotating side.
Figure 3: 3-D schematic showing the relative mesh sizes of the weld samples and grip sections.

These mesh sizes agree with the results of sensitivity analyses presented in [35, 38]. A larger overall mesh size was used for the master object in order to reduce over-penetration of the mesh from the adjacent object at the weld face. The grips were
idealized as tapered collars that contained sticking boundary conditions at the interface with the weld samples along with a coarse mesh. The axial load was input as a constant value, while the imposed rotation was input as either the initial kinetic energy or a time-dependent rotation speed (depending on the model formulation). The flow stress data for the LSHR material was a combination of predicted data from JMatPro software and experimental torsion and compression data. This flow stress data covered the temperature range from room temperature to 1250°C, and strain rates from 0.0005 to 100 sec⁻¹. All other thermo-physical properties (as a function of temperature where appropriate) were predicted from JMatPro, and validated with LSHR data produced by Gabb et.al. [43].

The initial model set up utilized the experimental weld process data as input. The time dependent rotation of the flywheel was input directly along with the axial load (constant with time), friction coefficient, and process efficiency. The main validation parameter with this model formulation was the total upset of the weld samples. It was particularly evident that the model was sensitive to the friction coefficient, flow stress data, and process efficiency. This model formulation was useful to narrow down the input parameters that are difficult to measure experimentally (friction coefficient, process efficiency), but provided no predictive capability for exploration of the weld process parameter space.

The second model set up utilized the initial kinetic energy of the flywheel, maximum rotation speed, axial load, coefficient of friction and process efficiency as input data. This model formulation calculated the time-dependent rotation speed for the weld process along with the sample upset, thermal field and flash formation. These parameters
could then be used in comparison with to aid interpretation of the experimental weld trials. This formulation had the ability to aid in exploration of the weld process space and was predictive in nature.

*Mechanical Testing*

Weld samples were tested in tension as well as hardness in order to ascertain bond quality and effects of welding on the resultant program materials. All tests were performed on coupons excised from as-welded materials. The as-welded samples were sectioned axially along the centerline of the 12.7 mm diameter round bars. One half of the weld sample was further prepared for metallographic analysis and hardness testing while sub-scale tension samples were excised from the other half.

Tension tests were carried out on a MTS servo-hydraulic test frame at a constant crosshead speed of 0.02 mm/s. The tension samples were dog-bone shaped with a gage length of 20 mm, and cross section of 1.8 mm by 3.4 mm (LSHR to LSHR welds), or cross section of 1.2 mm by 3.5 mm (LSHR to Mar-M247 welds). The weld interface was centered within the gage section, perpendicular to the tension axis. For the dissimilar welds, parallel gridlines laser marked on the surface to aid in determining the location of strain localization during testing. For both sets of weld samples, load versus displacement data was captured, and stress-strain data was generated using Vic-Gauge software (Correlated Solutions Inc.). Both a clip-on extensometer and virtual gauges were used to capture both macroscopic and local tension behavior. Tension test data provided a quality
metric for comparison of the resultant welds produced with a range of welding parameters. The tension specimen fracture surfaces were analyzed via optical and SEM techniques to determine the failure location and contributing features.

The hardness of the welds in the as-welded condition was determined utilizing a Vickers microhardness testing machine (500 g load for 20 s). The weld samples were tested in an array perpendicular to the weld face along the axial centerline. The hardness measurements as a function of distance across the weld line provided a clear delineation of the width of the region where the weld process had altered the structure and properties of the parent material.

_Metallographic Examination_

The sectioned weld samples were prepared for further analysis via standard metallographic preparation techniques, grinding through 600 grit media, followed by polishing via diamond slurry, and finished with 0.5 micron colloidal silica solution. The metallographically prepared specimens were then imaged via optical techniques in order to show the morphology of the flash formation and compare the effects of various welding parameters. The optical techniques were followed-up with SEM based imaging techniques. Both secondary (SEI) and backscattered (BSE) imaging conditions were performed on the weld samples in order to highlight various microstructure features such as gamma grains, γ’ precipitates, carbides, oxides, and weld line defects. Electron backscattered diffraction (EBSD) was performed to elucidate changes in the gamma grain
structure as a function of axial distance from the weld line. EBSD was useful to show the grain size, as well as the degree of recrystallization and deformation of the grain structure that took place near the weld line. Electron Probe Micro Analysis (EPMA) was utilized to determine the changes in chemistry in the weld samples as a function of distance from the bond line. Various line-scans were performed at the radial centerline and near the outer radius of the weld samples. The line-scans highlighted the heavily deformed & recrystallized region at the bond line where both diffusion and mechanical mixing of the weld materials took place.
Chapter 3: Results and Discussion

Experimental IFW Trials

In order to explore the IFW process as well as provide input and validation data for the finite element process model, various welding trials were performed across a wide range of welding conditions (Figure 4). These welds were performed with LSHR samples and spanned axial loads from 5k lbs to 22.5k lbs. The initial rotation speeds ranged from 1500 rpm to 5500 rpm. Most welds were performed with moments of inertia of 0.381 kgm$^2$, however a few welds were performed at lower moments of inertia for specific comparisons. In an effort to explore process optimization of welding two dissimilar superalloys, various moments of inertia were used, which ranged from 0.166 kgm$^2$ to 0.802 kgm$^2$. 
Figure 4: IFW process parameters for the LSHR to LSHR welds. The welds indicated in red exhibited single-step upset behavior, while those in blue exhibited multi-step upset.

Time-dependent experimental data was obtained from each weld trial. This data was in the form of rotation speed, axial force, and upset length (Figure 5). A select number of welding trials were performed which utilized thermocouples attached to specific locations along the outer radius of the weld samples. This provided time-dependent temperature data at multiple locations. The closest thermocouple was placed approximately 1.5 mm from the initial weld surface. This temperature data provided correlation challenges as the distance between the thermocouple and the weld interface changed during the weld due to the reduction in height of the weld sample. The initial and final locations of the thermocouple can be determined by simple measurement, but the exact location during welding is more challenging to determine. Therefore specific
correlations between the measured temperature and axial location as well as comparisons with simulation results are difficult.

Figure 5: Time-dependent IFW output data highlighting the applied load and rotation speed.

The first observation from the LSHR similar welds was the effects of weld input parameters (axial load and initial rotational velocity) on upset length. The upset length is the total reduction in height of the weld samples during the weld process. In order to visualize the effects of the welding process parameters, the total upset length data was plotted as a function of both initial rpm and axial load (Figure 6). To produce this figure,
the weld data was interpolated from an irregular grid to a regular grid, which was followed by piecewise, local linear interpolation between the grid points. Based on this analysis, the upset length increased directly as a function of initial rotation speed. There also appeared to be a minor effect of axial load on the upset length.

Figure 6: LSHR to LSHR weld upset length (experimental) as a function of axial load and initial rotation speed.

Based upon the review of this experimental data, the upset length was plotted as a function of rotational speed (Figure 7). A second order polynomial was used to fit the data. In this figure, within each discreet rotational velocity ‘bin’ the upset increased with...
increased axial load. This figure highlights the complex interrelation of the weld process parameters and the challenges in developing simple relationships between weld input parameters and measurable output metrics.

Figure 7: LSHR to LSHR experimental weld upset as a function of rotation speed showing a second order polynomial fit.

The LSHR similar welds exhibited two characteristic upset behaviors as a function of time. For the conditions shown in red in Figure 4, the upset as a function of time was smooth and continuous (denoted as single-step) while the upset was discontinuous for the conditions shown in blue (denoted as multi-step). The continuous upset behavior was characterized by a smooth, continuous increase in upset length as a function of time. Once the rotation stopped there was a significant decrease in the rate of
upset length as a function of time. This secondary upset is due to contraction of the weld samples during cooling (Figure 8). The discontinuous upset behavior was characterized by periods of smooth, continuous upset (similar in form to the single-step behavior) followed by periods of little or no upset. This behavior repeated until the rotation stopped. A secondary reduction in length due to contraction of the samples upon cooling followed the end of rotation. As a general rule for these welding conditions, the axial load appeared to have little effect on the occurrence of the single- versus multi-step upset behavior, and the initial kinetic energy or rotation speed was the primary factor. It was postulated that the multi-step upset phenomenon was repeated cycles of the same behavior that transpired for the single-step upset. What was interesting was that there were no experimental examples of welds with large upset lengths that exhibited the single-step upset behavior. Likewise, there were no examples of low upset welds with multi-step behavior. Based on this observation and that shown in Figure 4, there exists an inflection point in upset behavior between single-step and multi-step.
From the similar weld experiments, there appears to be a transition at approximately 3k rpm between the two upset behaviors. Based on the curvature of the fitted data, there may be a slight secondary effect from the axial load on the upset length. One weld performed at 4k rpm and 5k lbs load exhibited smooth, continuous single-step upset, while additional welds performed at 4k rpm and higher loads exhibited multi-step upset. The single-step upset could be due to the increased duration of the weld as well as increased peak temperatures at the weld line that are due to the low load value. Increasing the weld duration through lower applied loads tends to provide more complete thermal diffusion radially along the weld line and axially away from the weld line as compared to
welds produced with high loads. This is due to the increased duration of the welds performed at low loads. This results in a larger volume of material near the weld line that reaches a sufficient temperature to allow for radial plastic flow. This larger, heated volume of material effectively stabilizes the upset and promotes single-step behavior. Also, decreasing the axial load allows the material to reach higher temperatures at the weld line, albeit more slowly as compared to welds with increased axial loads.

It is clear from the experimental observations that this single-step versus multi-step upset behavior is due to the balance of energy/heat input versus output at the weld zone. Once the material near the weld line reaches a temperature high enough to allow radial plastic flow under the stresses developed due to the applied load, the weld couple begins to upset and the characteristic flash evolves. If the thermal input due to both friction at the weld interface, and deformation heating near the weld line, is balanced with both the thermal convection into the flash, and the thermal conduction away from the weld line into the parent material, then the upset will tend to be single-step in nature as evidenced by the weld parameters marked with red in Figure 4. If the upset rate is too rapid, heat input rate is too rapid, or the thermal field is too shallow, then the weld zone will be out of balance. Once the system is out of balance, then the plasticized material will flow into the flash. This is accompanied with a rapid increase in upset length, followed by a period of little to no upset. Then the material now at the weld line must heat up via friction and deformation heating until it reaches a sufficient temperature to allow radial flow again into the flash. This repeated deformation and heating cycle is evident in the weld conditions marked with blue in Figure 4 (multi-step). When
comparing the multi-step upset welds with the single-step welds, the major differentiating factor is that the multi-step welds were performed at relatively high initial kinetic energy levels, and resulted in significant upset lengths. For the single-step upset welds, these were performed at low initial kinetic energy and provided more modest upset length values.

This change in behavior also highlights the need to develop weld parameters that provide sufficient upset to clear the weld line of any oxides, films, particles or anything else that would be a barrier to developing a solid state bond without producing excessive amounts of flash [44]. For efficient bonding in the solid state, any surface contamination that could be a barrier to bringing the two weld samples within the inter-atomic spacing of the constituent unit cell must be removed in order to produce a quality weld. Any remnant particles, films, oxides, etc at the weld interface promote weld defect formation and inhibit bonding. In IFW, the radial flow of material into the flash accomplishes two major factors that promote joinability. The first factor is that the radial flow breaks up oxides and other surface films that are resident on the weld samples. Second, the radial flow brings nascent material into intimate contact at the weld line where subsequent bonding occurs. Without both of these two factors taking place, joinability is compromised.

From an industrial standpoint, performing welds with very large upset lengths may be non-optimal. Welds performed with the minimum energy required to produce enough upset to only clear away the initial material at the weld line, as well as any surface contaminants will reduce the costs associated with the weld process. This reduced
cost is associated with the cost of energy to perform the weld, as well as the cost of the material that is expelled into the flash. For most applications, the flash must be removed via a machining process, and the weld materials are typically high value added at the point of welding. Therefore, reducing the amount of material that ends up as scrap in the flash can reduce the cost of production. This highlights the need to determine weld parameters that provide excellent bond quality with a minimum amount of upset and energy utilized.

One method to determine the minimum upset required to clear all of the material away from the weld line and bring nascent material together to form a bond is to use a process model. Once a process model of the particular weld conditions has been run, it is possible via post-processing, to track the movement of the individual nodes within the finite element model at the weld interface. With nodal tracking, one can estimate the amount of upset required for the initial material at the weld face to flow out into the flash [27]. By tracking the movement of a node that begins very close to the axial centerline of the weld sample, one can determine the amount of upset length to target for a particular material and geometry combination. A well-calibrated process model can then provide the ability to perform a sensitivity analysis on the weld parameters in order to reach these optimum conditions.

If a process model is not available, or practical, another method to estimate the minimum upset required to clear the bond line is based on visual examination of the weld and geometric considerations. First, one must assume that all deformation during welding is limited to a volume of material at either side of the weld line that is the same thickness
as the flash. Based on optical examination of various welds, this assumption appears valid, although is an approximation as the actual cross-section area will at times be concave, or convex depending on the specific welding conditions (Figure 9) [10].

![Figure 9: Schematic showing the effects of energy (top row), initial velocity (middle row) and pressure (bottom row). The values increase from left to right. [10]](image)

Next, if all of the deformed material within that initial volume flows out into the flash, then the minimum upset required to clear the initial bond line is equal to two times the flash thickness. This is a reasonable assumption as no bulk deformation appears to
take place away from the weld (due to the steep temperature gradients found in IFW). For the similar weld parameters utilized in this work, the total range of flash thickness measured from ~500 microns to ~3500 microns (~1800 microns average). Doubling the average flash thickness results in an estimated minimum upset length in the range of 3 to 4mm. This metric is only pertinent to welds utilizing LSHR material in 12.7mm diameter solid cylinders. Changes in weld sample size, and material will alter the deformation volume and flash thickness. Examples of the flash size and single- versus multi-step morphology are shown in Figure 10 below. For dissimilar welds, this methodology is also applicable, however one must estimate the amount of deformation required on both sides of the weld independently and combine that value to get the minimum upset as the materials may upset at different rates to different total lengths.

![Figure 10: Optical micrographs of (a) multi-step, and (b) single-step flash morphology.](image)

In order to better understand the contributions of the weld process parameters on weld behavior, another method to visualize the relationship between upset length, kinetic energy and applied load is shown in Figure 11. There appears to be a general linear
relationship between the upset length of the welds and initial kinetic energy as shown with the linear trend line shown. As the initial kinetic energy increases, so does the weld upset length.

![Graph showing linear relationship between upset length and initial kinetic energy.](image)

**Figure 11:** Experimental LSHR to LSHR weld upset length as a function of initial kinetic energy.

This data was fit with a simple linear relationship as shown in Figure 11 with an $R^2$ value of 0.9528, indicating reasonable agreement with the linear fit. There does appear to be some scatter in upset length, which is greater in magnitude at the higher initial kinetic energy values. This scatter at each energy level is primarily due to changes in applied axial load. As the axial load increases so does the upset length for a particular initial kinetic energy level. In other words, the upset length for an initial kinetic energy can be modified with changes in the applied load.
To visualize the complex interplay of initial kinetic energy and axial load on the weld upset length, the weld data was evaluated as a function of load multiplied by energy (P*E). This plot is similar to that found in [26] and shown in Figure 12.

![Figure 12: Weld upset length as a function of pressure multiplied by initial kinetic energy from [26].](image)

In this figure, the axial shortening (upset length) follows a power law relationship with the load multiplied by the initial kinetic energy. It is important to note that this was data produced via a finite element process model for a different nickel-base superalloy over a relative small range of weld parameters. When the LSHR weld data was evaluated in this
manner, a similar relationship was evident for higher values of $P^*E$, but a different behavior was shown for low $P^*E$ values (Figure 13).

![Figure 13: Experimental LSHR to LSHR upset length as a function of pressure multiplied by initial kinetic energy.](image)

This data is shown in a log-log graph to highlight the data trends more clearly. For higher $P^*E$ values, a similar power law relationship to that in Figure 12 is evident, however at low $P^*E$ values, a linear relationship is shown. The apparent change from one population to the other occurs at approximately 3 to 4 mm upset length and approximately $5E6$ MPa*J. This change in behavior was not predicted by the model in [26] even though that particular range of $P^*E$ was evaluated. This may be due to the model not accounting for changes in IFW process efficiency as a function of the weld parameters. It is also
important to note that the welds that fall into the lower curve are all on the low initial kinetic energy side which exhibited single-step upset. The welds falling on the upper curve all fall into the multi-step upset regime. Based on the previously mentioned minimum upset estimation method, this change in behavior falls right at the range of the estimated minimum upset for LSHR welds of this geometry.

The effects of axial load on weld behavior were examined within the context of the finite element process model. To this end, two simulations were performed; both will equivalent weld process parameters with the exception of axial load. The axial load was varied from 48.6 kN to 75 kN. Increased axial load had a marked effect on the weld process. First, the rate of deceleration of the flywheel increased with increased axial load (Figure 14). Next the upset length increased with increased axial load (Figure 14). This increase in deceleration rate and upset length indicates that the energy input rate increased, and the weld duration decreased as a result of the axial load.
The increased load also increased the rate of upset, and reduced the incubation time prior to the onset of upset. In other words, the increased load promoted a more rapid conversion of kinetic energy to heat at the weld interface, and allowed the weld materials to begin to upset more quickly (at a lower temperature). This effect on temperature was also shown by the model (Figure 15). The increased axial load resulted in an increased heating rate (due to the faster conversion of kinetic energy to heat), but a reduced peak temperature (1135°C as compared to 1171°C). The increased axial load also reduced the duration of the weld where the weld line was above 1000°C. This reduced duration was from 5.3 s at 75 kN load to 7.9 s at 48.6 kN load. For these difficult to weld superalloys, time at high temperature is a key to promoting joinability. This time at elevated
temperatures (approaching the $\gamma'$ solvus ~1160°C) shows that as the $\gamma'$ particles begin to go into solution, the strength of the material decreases, allowing rapid deformation and flash formation.

Figure 15: Finite element results showing the effect of increased axial load on weld line temperature.

Tension tests were conducted for a select number of LSHR welds to provide a quantifiable metric for bond quality. These tests were performed at room temperature, and two test specimens were extracted from each weld. Three welds from the lower curve and three welds from the upper curve in Figure 13 were chosen for tension testing, and covered the full range of P*E values available. The first observation from the room temperature tension tests was the limited variability of the yield strength, ultimate tensile
strength and elongation to failure for all samples that failed in the parent material away from the weld line. In other words, there was no improvement in tensile properties once a defect-free weld line was produced (i.e. beyond a minimum threshold value of P*E). For the low P*E welds, the tension test results were mixed. In some cases, the two tension samples from a single weld would provide mixed results. In one case the tension sample would exhibit good strength and elongation values, while the adjacent tension sample would fail at the bond line with limited strength or ductility. Figure 16 shows the tensile data for the weld samples that failed away from the bond line. There is a conspicuous lack of variability in the tensile values for these samples. This indicates that the weld parameters have little or no effect on the room temperature tensile strength of the welds once a defect-free bond line was produced. For the few welds that failed at the bond line, the values were slightly lower on average, but still exhibited reasonable values.
Figure 16: Tension test data for the defect-free LSHR weld samples.

See Table 3 for results from the tension tests. For the tension samples that failed at reduced strength values as compared to the parent material, the failure initiated at the bond line in large areas contaminated with oxides.

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Table 3: Tension test data for all LSHR weld samples.
This failure location at the weld line is directly related to the limited upset length, and the fact that not all surface contaminants and oxides were removed during welding. All other weld samples failed away from the bond line, and at strength values that were comparable to or greater than the parent materials. The slight increase in strength values and ductility is most likely related to the grain refinement and increased number of $\gamma'$ precipitates that are present in the heat affected zone during the weld [17, 19, 22-24].

Hardness testing across the weld zone for these specimens showed a general decrease in hardness with distance from the weld line. The peak hardness values were typically 25% higher than the parent material values and corresponded well with the reduced gamma grain size in the heat affected zone (Figure 17).

Figure 17: Typical hardness and grain size as a function of distance in LSHR to LSHR weld samples.
Recrystallization due to the elevated temperature and strain near the weld line produced reduced-size and relatively strain-free gamma grains on cooling from the weld process. The elevated temperatures and strain also allowed the $\gamma'$ precipitates to go into solution, and the rapid cooling after welding promoted a fine, even dispersion of $\gamma'$ particles. These precipitates do not have sufficient time to coarsen prior to the weld reaching room temperature.

*IFW Microstructure*

The near-weld microstructure of the LSHR to LSHR welds was largely unremarkable. As shown in Figure 17 the gamma grain size decreased from ASTM grain size number 14 to $\sim$18 ($\sim$2.8 micron diameter and below) with decreased distance from the weld line. This behavior was typical in the LSHR to LSHR welds. Upon closer examination of the bond line in the SEM (Figure 18), there was a slight change in the character of the $\gamma'$ near the interface. The $\gamma'$ particles were imaged in the SEM in an etched condition (utilizing Kalling’s $\gamma'$ etchant). The bond line area is shown in Figure 18a (backscattered SEM) and Figure 18b (secondary electron SEM). The parent material is shown in Figure 18c (backscattered SEM) and Figure 18d (secondary electron SEM) for comparison.
Nearer to the bond line, the smaller secondary and tertiary $\gamma'$ particles were less pronounced and the primary $\gamma'$ dominated the population. In the parent material it is clear that all three families of $\gamma'$ were present from the forge and heat treatment processes of the powder metallurgy product form. During the welding process, the steep temperature
gradients (above the $\gamma'$ solvus \textasciitilde 1160°C) and short weld durations allow the smaller $\gamma'$ to go into solution, but in this case, there was not sufficient time at supersolvus temperatures to solutionize the larger primary $\gamma'$ particles.

The gamma grain size was characterized by imaging with EBSD in the SEM. Using a unique grain color filter, the gamma grains were clearly delineated based on differences in orientations as measured with the EBSD technique. The parent material gamma grain size is highlighted in Figure 19a and the gamma grain size near the weld line is shown in Figure 19b.

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{figure19.png}
\caption{Unique grain color maps produced via EBSD in the SEM for (a) parent LSHR and (b) weld line LSHR.}
\end{figure}

From these images it is clear that significant recrystallization occurs during the weld process and upon cooling from the end of the weld process. These refined grains, provide much of the strengthening near the weld line as discussed in [22].
As a comparison, the near weld microstructure in a dissimilar weld between LSHR and Mar-M247 is shown in Figure 20.

The LSHR is on the left side and Mar-M247 is on the right in Figure 20. This sample was etched with the same $\gamma'$ etchant as used on the LSHR welds, and showed significantly different behavior. The as-cast microstructure of the Mar-M247 exhibited a large fraction of $\gamma'$ present near the weld line, and near uniform size distribution of the same as a function of distance away from the weld line. The LSHR microstructure however is quite different than what was observed in the similar LSHR welds. Here, there appeared to be a zone adjacent to the weld line of approximately 80 to 100 microns in width where no large $\gamma'$ particles were present. Chemical analysis of this area utilizing electron probe micro analysis (EPMA) showed that this area is predominately equivalent to the LSHR
base chemistry. It has been noted that the driving force for precipitation of γ′ upon cooling from the weld process is quite high, and therefore difficult if not impossible to suppress [17, 45, 46]. In order to determine if there is any γ′ present in this zone, TEM analysis techniques were employed in order to determine their presence. Selected area diffraction techniques were used to determine if there are any superlattice reflections present from the ordered γ′ particles (Figure 21).

Figure 21: TEM selected area diffraction pattern showing the superlattice reflections from the ordered γ′ particles [22].
The finer points between the large reflections are the superlattice reflections from the $\gamma'$ particles. These are present due to the ordered structure of the $\gamma'$, and indicate that discreet $\gamma'$ particles are indeed present in the 80 to 100 micron region adjacent to the weld line where the large $\gamma'$ are absent. However, the particles must be too small to image with conventional SEM-based techniques. Utilizing dark field imaging in the TEM, the fine $\gamma'$ were imaged showing their size to be sub 20 nm in diameter (Figure 22).

Figure 22: TEM dark field image of the $\gamma'$ particles in the ~80 micron region highlighted in Figure 21 [22].
These particles re-precipitated as a fine, even dispersion of particles across the near-weld line during cooling from the weld process, but the severe cooling rates and therefore short duration at elevated temperatures retard the coarsening of the particles.

Another microstructure observation relates to the whether or not any significant melting occurs during the IFW process [8, 39]. IFW is generally considered a bonding process that takes place in the solid state, therefore no gross melting and subsequent solidification occurs. It has been suggested however, that the very low effective friction coefficients that have been reported could be due to a thin film of molten metal at the weld interface that would act as a lubricant [39]. This however, has been suggested as unlikely as it is considered that the molten material would either be expelled out with the flash, or that a solidified structure would be apparent at the bond line post welding. One reference [39] suggested that a hydrostatic stress state at the weld interface could constrain any molten film within the bond plane thus suggesting it is indeed possible to have limited melting. Imaging of the weld line flash for a high energy input dissimilar LSHR to Mar-M247 weld revealed interesting microstructure features. First there was significant cracking and voids present along the grain boundaries in the LSHR flash. Second, there appear to be large (~10 micron diameter) spherical particles attached to the outer edges of the flash (Figure 23).
Figure 23: SEM backscattered image of the LSHR flash from a LSHR to Mar-M247 weld produced with high initial kinetic energy showing voids along the grain boundaries and spherical particles that appear to be solidified.

This figure may indicate that limited melting, and possibly incipient melting at grain boundaries can indeed occur. As well as the molten film can be expelled with the flash. Upon closer inspection of these spherical particles it is clear that there was entrapped gas porosity present (Figure 24).
This entrapped gas porosity may indicate that the molten droplets solidified quickly during expulsion from the weld line with the flash while trapping ambient air. Based upon these observations, one might enquire as to why these features aren’t present on all welds performed between LSHR and Mar-M247 or any other welds reported. First it is most likely very difficult to achieve localized melting with similar welds, as both sides of the weld couple will deform at the same temperature. The local flow stress of the material on both sides of the weld is surpassed by the shear stress applied due to the weld process and therefore both sides upset. In general, Ni-base superalloys are unique in that they retain a high fraction of their strength to very large fractions of their melting
temperatures. This high strength at high temperatures allows the weld material to approach the onset of insipient melting. Second, with dissimilar welds, the material that is stronger at elevated temperatures will cause the weaker material to deform first while retaining extremely high temperatures. This could allow some limited melting at the interface. Third, limited short-range diffusion across the weld line could provide a local chemistry that promotes a low melting point eutectic or other feature that could preferentially liquate. In this particular case where the spherical solidified particles were found, two factors were present which may have promoted limited liquation. The first factor was relatively high energy input with two dissimilar Ni-base superalloys. The second factor was a slight misalignment of the weld samples during IFW. It was observed during the IFW process that significant amounts of sparks were thrown as compared to all other weld performed. The material that was thrown from the weld as sparks was gathered and imaged via SEM. A large fraction of the spallation looked very similar to brittle chips from a machining process (Figure 25a), but also present were spherical particles that exhibited a solidified dendritic structure (Figure 25b,c,d).
Figure 25: SEM secondary electron images of material ejected from the weld line during a high energy LSHR to Mar-M247 weld (a) brittle chips (b, c, d) spherical particles with solidified dendritic structure.

**IFW Process Model**

An IFW process model utilizing DEFORM Finite Element software (Scientific Forming Technologies Corp.) was developed in order to provide insights into the thermo-
mechanical behavior of the materials during welding. LSHR to LSHR weld data was utilized as both input and later as validation data for the process model. The model provided time dependent output data in the form of upset, temperature, strain, strain rate, and rotation speed. The development of the process model allowed in-depth comparison of weld data with experimental observations, as well as allowed exploration of factors that are either difficult to control or cannot be modified experimentally, such as weld process efficiency, thermal conductivity, gripping geometry etc.

The initial process model was formulated in a descriptive capacity. In other words, time-dependent input data was utilized to validate the time-dependent upset and peak temperature data. Various material parameters such as the friction coefficient, thermal conductivity, flow stress, and model parameters such as number of elements, element size, interference depth, remesh criteria etc. were modified during this validation process. This version of the model significantly over-predicted both the upset length and maximum temperatures in the weld zone. A sensitivity analysis of the descriptive process model showed that in order to get the predicted upset length to match the experimental conditions, an efficiency factor of 70% was necessary. To get the predicted peak temperature into a reasonable range (below the solidus for LSHR) an efficiency factor of 40% was necessary (Figure 26). These results highlighted the need to further investigate the input data as well as the model parameters in order to produce a model that provided a useful result as compared to the experimental weld data.
In order to tune the model and reduce the magnitude of errors in predicted temperatures and upset lengths, an iterative process whereby the friction coefficient, flow stress, weld efficiency and thermal conductivity were modified. Once the model was fine-tuned, i.e. produced weld zone temperatures below the solidus temperature for LSHR, and upsets that were reasonable in comparison with the experiments, the various parameters were set, and then used for further model exploration and development.

The first model parameter that required further analysis was the friction coefficient between the weld pieces. This time-dependent friction coefficient was estimated from the experimental weld trial data. This is an energy balance method.

Figure 26: Results from a DEFORM FEM sensitivity analysis of the effects of efficiency on peak weld line temperature and upset length.
heat generated at the weld interface is equal to the work done by the friction force. This can be represented by the friction force multiplied by the sliding velocity [35].

\[ Q = \int_{r_0}^{r_1} \mu p r \omega 2\pi r \, dr \]  

(2)

Here, \( Q \) is the heat generated at the weld interface, \( \mu \) is the friction coefficient, \( p \) is the axial pressure, \( r \) is the radius of the sample, and \( \omega \) is the rotational velocity. For modeling purposes, we are only interested in an ‘effective’ friction coefficient for the entire weld surface, not a location-specific value. Therefore, assuming that both friction coefficient and applied load are both independent of radius, the following derivation provides an approximation of the friction coefficient. Integrating equation (2) yields:

\[ Q = \left( \frac{4}{3} \right) \pi^2 \mu p n (r_1^3 - r_0^3) \]  

(3)

Equation (4) relates the kinetic energy of the flywheel to the moment of inertia \( I \), and the rotation speed \( \omega \) (or number of rotations \( n \) through \( \omega = 2\pi n \)).

\[ E = \frac{1}{2} I \omega^2 = 2\pi^2 n^2 I \]  

(4)

Taking the derivative of equation (4) with time yields equation (5).

\[ \frac{dE}{dt} = 4\pi^2 I n \frac{dn}{dt} \]  

(5)

Equating the rate of energy input (\( dE/dt \)) with the heat generated at the weld interface (\( Q \)) yields equation (6):

\[ 4\pi^2 I n \frac{dn}{dt} = \frac{4}{3} \pi^2 \mu p n (r_1^3 - r_0^3) \]  

(6)

Substituting for \( I \) and rearranging provides the final equation for the apparent (time-dependent) friction coefficient, equation (7).
\[ \mu = \frac{3E_0 \frac{dn}{dt}}{2\pi^2 n^2 P (r_1^3 - r_0^3)} \]  

(7)

Utilizing equation (7) to evaluate the apparent friction coefficient for the experimental LSHR weld trials produced the time-dependent data similar to that shown in Figure 27.

Figure 27: Effective friction coefficient calculated from LSHR to LSHR weld data (equation 7).

An interesting observation regarding the effective friction coefficient was that the behavior appeared to segregate based upon initial rotation speed. As the initial rotation
speed was increased, the steady-state behavior increased in duration, and the peak values decreased. It was also noteworthy that within each rotation speed bin, the effective friction values increased with increased axial force. This effective friction coefficient data was useful to calibrate the process model while operating in the descriptive format (based on the experimental behavior).

To gain a wider perspective on the changes in effective friction coefficient as a function of the weld process parameters, the steady-state behavior was calculated as indicated above for each of the experimental LSHR to LSHR weld trials. This effective friction coefficient data spanned the range of 0.04 to 0.20 (Figure 28).

Figure 28: Effective steady-state friction coefficient calculated from the experimental LSHR to LSHR weld data.
The analysis of this data involved interpolation from an irregular grid to a regular grid, followed by piecewise, local linear interpolation between the grid points. It is evident that both increased axial force and initial rotation speed decreased the effective steady-state friction coefficient. However, the initial rotation speed appears to have a greater effect on the steady-state coefficient. These steady-state values can then be used for the predictive model formulation as the time-dependent nature of the data has been removed.

Estimating the friction coefficient allowed the model to more closely describe the weld behavior, but more improvements remained. The next input parameter that was examined was the LSHR flow stress. The flow stress data is a set of stress – strain curves as a function of temperature and strain rate. This data is typically produced via isothermal compression or torsion testing. For modeling the IFW process, which is considered to take place entirely in the solid state, the temperature range of interest is therefore room temperature up to the solidus temperature of LSHR (approximately 1280°C as measured via Differential Thermal Analysis). There is much disagreement in the literature about the actual strain rates present during IFW, some reporting estimates as high as 1000 s⁻¹[7]. Based on experimental observations and comparison with finite element results suggest the range is limited to values approaching 20 s⁻¹ on the high end.

Producing a complete flow stress dataset covering room temperature to 1280°C and strain rates up to 20 s⁻¹ is both time and material consuming, as well as costly. In order to produce a complete dataset, the software package JMatPro version 5 was utilized to predict the flow stress data for LSHR. JMatPro utilizes chemistry, phase fraction, heat treatment condition, room temperature tension properties, grain size, precipitate size and
fraction to predict the flow stress data as a function of temperature and strain rate. This method of producing the data was advantageous as it was very rapid and produced a complete data set across the ranges stated above. This dataset was specifically formatted for use in DEFORM finite element software. The main drawback to this method was that the magnitude of the flow stress values were off by a significant margin in some cases (up to three times), however the dataset appeared to capture the correct form of the flow stress curves (i.e. strain hardening or flow softening as appropriate). To address the magnitude issue with the JMatPro predicted flow stress data, a number of isothermal compression tests, and torsion tests were performed on the LSHR material and the flow stress data produced experimentally was utilized to correct the magnitude of the predicted data set. This corrected data was then utilized in the finite element process model.

The next model-based activity was to explore the multi-step upset behavior that occurred experimentally with the high energy input welds. The working theory was that this multi-step upset was due to an imbalance of the thermal input and output from the weld line. Effectively, this could be explained by a repeating increase and then decrease of the effective coefficient of friction at the weld interface (based on temperature effects). To explore this phenomenon further, the effective friction coefficient was modulated with a sine wave to various magnitudes. A series of weld simulations were then performed starting with the baseline effective friction coefficient, followed by modulated friction coefficients (Figure 29). The effective friction coefficient, which was estimated from the steady-state portion of the time-dependent rotation data, was constant. The sine wave
modulation overlay of the friction coefficient followed the same general behavior with the addition of local maxima and minima following a typical sine wave.

![Effective friction coefficient modulation for input into DEFORM FEM process model.](image)

The predicted time-dependent upset behavior was collected and compared amongst the five simulations (Figure 30). It is clear from the simulated time-dependent upset data, which was produced with repeated increases and decreases of the friction coefficient, resulted in periods of rapid upset followed by periods of little to no upset. This observation of the simulation data supports the explanation of the multi-step upset in as
much as the weld line material at high temperatures will exhibit a lower effective friction coefficient, while material at a lower temperature, which will not flow into the flash, will exhibit a higher effective friction coefficient. Thus after periods of rapid upset, the thermal field will need to be reset, and there will be a period where the weld line material will heat up due to friction at the interface. Once the thermal field has reached a critical temperature, bulk plastic flow will again occur as material moves out into the flash. This cyclic behavior repeats until the energy of the flywheel is expended and rotation stops.

Figure 30: Upset length predicted by FEM utilizing the modulated friction coefficient input.

The temperature profile at the $R = 0$ mm location (axial centerline) of the weld interface also exhibited a similar modulated behavior (Figure 31). During the weld process the
temperature increased rapidly, which corresponded minimal or no upset. This is followed by a period of temperature decrease as material at a lower temperature moved into the weld line due to radial flow into the flash of the high temperature material. As the new material at the weld line increases in temperature, due to frictional heating and deformation heating, radial flow begins again. This cyclic behavior repeated until the rotation of the weld machine stopped and all input kinetic energy was expended.

Figure 31: FEM predicted time-dependent temperature data for the axial centerline of the weld face using modulated friction coefficient input data.

As per the discussion previously regarding the factors influencing the multi-step upset that occurred with large energy input welds, the weld line temperature does indeed modulate along with the deformation behavior. The thermal profile from the model
supports the theory that the multi-step upset is based on periods of rapid heating and little
deformation followed by cooling and deformation; thus cycling until rotation stopped.

IFW Machine Characteristics & Efficiency Estimation

The next major challenge to producing a process model and fully exploring the weld process parameter space is to determine the fraction of energy consumed during welding by parasitic energy sinks. This efficiency of the weld equipment is an important factor that has received nearly no attention in the IFW literature to date. There has been no attempt to determine the efficiency of the IFW equipment published, the only mention was to assign a constant value which was usually around 85 to 95% [27, 28, 42]. Understanding the efficiency of the IFW equipment and how it changes based on weld process parameters is critical to effective modeling of the process, scale-up of the weld process as well as transferring the weld process to multiple weld machines.

In essence, the total kinetic energy of the weld machine is determined by the size of the flywheel, and its rotational velocity. This energy is believed to dissipate through frictional heating at the weld interface during IFW. This is in part true, but not a complete picture of the energy segmentation during IFW. To support the rotating mass and drive train, as well as react the axial forces involved in IFW, both journal and thrust bearings are contained in the IFW equipment. These are typically hydrodynamic bearings, meaning that hydraulic fluid is utilized to both cool and reduce friction between the bearing faces. These types of bearings are typically employed in applications where
speeds, and loads are too large for traditional ball bearing designs, or when vibration and noise would be detrimental [47].

Once the rate of parasitic losses to the IFW equipment has been determined, then the characteristic fraction of the initial kinetic energy that is required to produce a bond in the solid state can be specified. This information is important in a number of ways. First this information can be utilized to monitor the weld equipment itself to make sure that the condition of the drive train, bearings, hydraulic system etc. are nominal and promote reproducible weld results. The second important aspect of this information is that it allows one to effectively scale the weld parameters from lab-scale specimens to large scale specimens. If the scale-up process involves two different IFW machines, the efficiencies of both can be determined and then the actual fraction of energy utilized to make the solid state bond can be scaled to the larger machine. This has the possibility to significantly reduce the amount of trial-and-error required to determine process parameters, as well as reduce the amount of in-process variability in weld outcomes currently experienced.

In an effort to characterize the impact of the bearings on the IFW machine behavior, a number of experiments were devised. The journal bearing supports the rotational components of the IFW equipment and the thrust bearing supports the axial force involved. The behavior of the journal bearings can be examined via rotation-only spin down tests of the equipment (under no axial load). This is advantageous as these do not require any modification of the IFW machine or specialized ancillary equipment. Unfortunately, examining the behavior of the thrust bearing would require the design and
fabrication of a fixture with known frictional response that would be allowed to rotate with the flywheel while under an axial load. A series of spin-down tests were run and the effects of both initial rotation speed and moment of inertia were examined. To examine the effects of initial rotation speed four tests were run, each at \( I = 0.380 \text{ kgm}^2 \) and four different rotation speeds (480 rad/s, 300 rad/s, 245 rad/s, 165 rad/s). It was observed that the flywheel decelerated at the same rate regardless of the initial rotational velocity, and appropriately, the energy decreased at the same rate as well (Figure 32a,b).

![Figure 32: Results of spin-down IFW tests, highlighting the deceleration rate as a function of various starting rotational velocities (a) and initial kinetic energies (b) all maintaining one flywheel size.](image)

To examine the effects of changing the flywheel moment of inertia, similar spin-down tests were run at moment of inertia values ranging from 0.051 kgm\(^2\) to 0.802 kgm\(^2\) (Figure 33).
Figure 33: Results of spin-down IFW tests, highlighting the deceleration rate as a function of various flywheel sizes (a) and initial kinetic energy values (b).

In contrast to Figure 32a, the deceleration data contained in Figure 33a shows a clear dependence on moment of inertia. The rate of deceleration increases with decreased moment of inertia. Another observation is that the total rotation time increases with increased moment of inertia. Also the kinetic energy decay scales inversely to the moment of inertia (Figure 33b). From this data it is clear that the moment of inertia has a profound impact on the rotation and energy decay rates of the IFW equipment examined. This indicates that the losses of energy within the journal bearings are dependent on the moment of inertia. However one must take caution in drawing conclusions from this data, as no axial loading was present, and therefore any synergistic effects on the energy losses within the welding equipment will not be accounted for and further investigation is necessary.

Another method to visualize this spin-down data is to plot the steady-state deceleration rate as a function of rotation speed (Figure 34). For a given flywheel
moment of inertia, the deceleration rate increases with an increase in rotation speed. This effect is strongest at the lowest moment of inertia values, and becomes nearly zero slope for the highest moment of inertia tested.

Figure 34: Deceleration rate as a function of rotational velocity for five flywheel sizes.

One can calculate the torque due to the rotating flywheel by utilizing the relationship in equation 8.
When the torque due to the rotating flywheel is examined as a function of rotation speed for these spin-down tests, the data compresses to a general linear relationship (Figure 35). Here, the torque increases with increased rotation speed, and decreased moment of inertia.

Figure 35: Torque as a function of rotational velocity for five flywheel sizes.
As a first order approximation, this relationship can be described with a simple linear form.

\[ y = mx + b \]  

(9)

This results in a linear form of the equation showing that the torque due to the journal bearing is linearly proportional to the rotation speed.

\[ T = -I \frac{d\omega}{dt} = m\omega + b \]  

(10)

Armed with this relationship and spin-down data for the IFW machine, one can estimate the process efficiency. It is important to note, that this will not account for any energy losses due to the thrust bearings or any synergistic effects between the journal and thrust bearings. Therefore, this methodology is a reasonable first-order approximation, but should not be considered as definitive. The total energy of the weld is calculated from the rotational velocity data and moment of inertia from a weld trial according to equation (1).

\[ \Delta E(t) = I(\omega_0^2 - \omega_t^2)/2 \]  

(11)

The energy lost to the journal bearings can also be calculated utilizing equation (11), however the rotational velocity data from a spin-down test must be utilized instead. Subtracting the energy lost to the journal bearings from the total energy of the weld produces the energy utilized to produce the weld itself (Figure 36).
Figure 36: Estimation of the energy partitioning between the weld sample and machine bearings during IFW.

For this particular weld, a significant fraction of the total weld energy is consumed via the weld itself, and only a minor fraction is lost to the journal bearings. And again, this estimation method does not account for any energy losses to the thrust bearings.

The next step in this efficiency estimation method is to calculate the fraction of energy utilized to produce the weld versus the total energy. For this weld, the efficiency as a function of time is nearly constant, with a slight positive slope (Figure 37). The
efficiency value is between 0.77 and 0.8 for this case. Keep in mind that this is a conservative estimate, as it does not account for losses to the thrust bearings or any synergistic effects between the bearings, and therefore the actual efficiency for this case should be lower. This value however, is significantly lower than those reported in literature (0.85 – 0.95) [27, 28, 42]. This however, provides the first real insight into the energy partitioning during IFW, and suggests that the efficiency may change during a weld, as well as with different weld parameters and most likely with different weld equipment.

Figure 37: IFW process efficiency as estimated utilizing the spin-down test data along with experimental weld data.
The next step in this analysis is to calculate the torque associated with the weld, weld sample, and journal bearings. This can be calculated for each case using equation (10) with weld trial data and spin-down data (in the same manner as calculating energy partitioning above). The resultant data for this weld trial highlights nearly constant torque values, with slight negative slopes for each (Figure 38). It is important to mention that these torque curves do not show the same characteristic peaks at the beginning and ending of the weld process as it typical for IFW. The reason for this is that these values were calculated from the ‘steady-state’ portion of the deceleration curve, which omits the behavior at the immediate beginning and ending of rotation (where the slope changes dramatically) and the torque spikes due to the initial contact at the beginning of the process and the final stopping of the system.
Figure 38: Estimated torque as a function of time for the weld interface, IFW machine, and total process.

With the torque data in hand, one can then calculate the effective friction coefficient for this weld trial utilizing equation (12).

\[ \mu = \frac{1.5T_s}{r_0 p} \]  

(12)

The resultant effective friction coefficient (as a function of time) during this weld trial is rather noisy (due to the modulation of the axial force registered during the weld process), but otherwise presents a relatively constant value between 0.05 and 0.06 (Figure 39).
Using the methodology outlined above, one can effectively utilize weld trial data along with spin-down data to estimate the energy efficiency of the weld machine as well as the effective friction coefficient. This data can be utilized in a finite element process model and eliminate some of the uncertainty in the input parameters and process variables. However, since this data was calculated from welds, the model will be limited in its true predictive capability. Also, due to the spin-down tests not engaging the thrust bearings,
the resultant estimates of efficiency are too high. Further refinement of the analysis is required in order to more accurately determine the efficiency of the weld process.

In an effort to more accurately determine the efficiency of the weld process, an additional experimental technique was devised. In order to account for the energy losses within the IFW machine (to both the journal and thrust bearings), a technique to estimate the energy partitioning must be developed based on data derived from experiments where both sets of bearings are activated. Two possible concepts were initially explored. The first concept was a device that could be clamped into the rotating side of the IFW machine, and a coupling device clamped into the non-rotating side of the IFW machine. The two pieces could be considered a hydrodynamic bearing with an incompressible fluid between. Knife-edge seals would need to be employed to retain the hydraulic fluid in the reservoir, and allow the bearing to rotate freely (Figure 40). For this concept to work, the friction response of the bearing would have to be calibrated independently and this data used to correct data from subsequent spin-down tests.
This apparatus would allow an analogous experiment to the ‘spin-down’ tests to be run while allowing an axial load to engage the thrust bearings. Eliminating the friction (or independently determining the friction of the apparatus) associated with the application of the axial force while rotating would enable measurement of the rotation speed of the system as a function of time without any friction-induced energy losses due to production of the weld and heating of the weld samples. Using this devise would enable a similar analysis as that used above with the ‘spin-down’ test data to estimate the IFW machine efficiency and effective coefficient of friction.

The second concept to fully engage both the thrust and axial bearings while providing data to determine the energy efficiency is to use strain gages on the stationary side of the weld-couple. Strain gages were attached to the outer radius of the weld sample in the axial (0 degrees), torsional (45 degrees), and orthogonal (90 degrees)
configuration. The strain gages provide real-time strain data as a function of time during the weld process. From the calibration data for these strain gages, the shear modulus of the weld samples can be determined, as the applied load is known, and the shear strain response is measured from the gage. Based on the equation (13) the shear modulus can be calculated.

$$\tau = G \gamma$$  \hspace{1cm} (13)

During IFW, the total torque during welding can be calculated based on equation (10) and then the shear stress on the stationary sample can be calculated based on the measured strain from the gages and the pre-calculated shear modulus (equation 13). If one assumes that the shear stress is not a function of the radius of the weld sample, the torque on the weld sample can be calculated from the following (equations 14, 15)

$$dT = \tau \cdot 2\pi r dr \cdot r$$  \hspace{1cm} (14)

After integration and solving for torque, the following relationship provides the means to calculate the torque on the sample based on shear stress.

$$T = \frac{2}{3} \pi r^3 \tau_0$$  \hspace{1cm} (15)

The resultant torque data for the total weld process and the sample, based on equations 10 and 15 show the typical behavior for the IFW process, initial and final torque peaks with a region of relative ‘steady-state’ between. The time dependent torque for one weld condition is shown in Figure 41. This figure also depicts the time-dependent rotation speed as well.
Now that the torque on the weld sample and the total torque for the weld process are determined, one can calculate the total energy of the system as well as the energy utilized to produce the deformation and bonding of the sample. The efficiency of the weld machine for these conditions is given as the ratio of the energy utilized in heating & deforming the weld sample to the total energy consumed during the weld process (equation 16).

$$E_{ff} = \frac{E_s}{E_{tot}}$$  \hspace{1cm} (16)

The energy of the system is determined from equation (1) while the energy consumed in the weld sample in heating the weld face and bonding is determined from equation (17).
\[ E_s(t) = \int_0^t T_s \omega dt \]  

(17)

The time-dependent energy partitioning and process efficiency are shown for this dissimilar LSHR to MarM-247 weld in Figure 42.

![Figure 42: Time-dependent energy partitioning during IFW and associated process efficiency.](image)

Utilizing this methodology, one can effectively calculate the weld process efficiency for various weld parameters. It is important to note that the weld process efficiency utilizing this method is approximately 0.5 while the previous estimate using the spin-down method was in the range of 0.78 to 0.8. Once again, efficiency values in the range of 0.5 are much below what has been previously presented in the literature. It is therefore very important
to accurately account for the energy partitioning during IFW, especially if multiple weld machines will be utilized or scale-up in weld geometry will be involved in order to accurately estimate weld parameters and reduce the need for trial-and-error approaches.

Dissimilar IFW

The principal results from this portion of the investigation consisted of quantitative measurements of the various process parameters, the evolution of macrostructure and microstructure, and post-welding mechanical properties. Portions of this chapter are under consideration for publication [48].

Inertia Friction Welding Process Measurements

The rotation velocity, instantaneous kinetic energy, torque, duration, upset, and temperature transients varied noticeably for the different IFW trials. Due to friction between the mating surfaces of the weld samples as well as the rotating parts of the welding machine itself, the rotation velocity, $\omega$, and kinetic energy of the flywheel, $E_k$, decreased continuously with time until the rotation stopped (Figure 43).
The rate of change in rotational velocity, $a = -\frac{d\omega}{dt}$, (deceleration) was inversely proportional to $I$ (Figure 43a). In addition, an increase in $I$ from 0.166 kg·m$^2$ to 0.802 kg·m$^2$ resulted in an approximate doubling of the duration of welding (i.e., ~4 to ~9 seconds) (Figure 43).

For each value of $I$, the deceleration decreased as the processing time increased and reached a minimum value. However, at the end of the process, when extensive sticking of the friction surfaces occurred, the deceleration rapidly increased until the IFW process stopped (Figure 44a).
The corresponding total torque $T$ (= sum of the torque $T_S$ due to friction between the mating weld-faces and the torque $T_M$ due to friction in the bearings of the IFW machine), which is associated with the deceleration of the flywheel, and the apparent friction coefficient $\mu$, were determined using the following relations [15]:

$$T = Ia,$$  \hspace{1cm} (18)
\[ \mu = 1.5T/(Pr_o) \]  

Here, \( r_o \) is the outer radius of the contacting surfaces of the cylindrical samples. In the present experiments, both \( P \) and \( r_o \) can be considered constant. Therefore, \( \mu \) is linearly proportional to \( T \) and exhibits a similar dependence on processing time \( t \). Specifically during each trial, \( T \) decreased slightly, reached a minimum value at approximately two-thirds of the total duration of the IFW process, and then increased continuously, approaching a maximum value at the end (Figure 44c). In contrast to the deceleration behavior (Figure 44a, b); the total torque exhibited a rather weak dependence on \( I \) (Figure 44c, d). The dependence of the total torque on \( \omega \) during the initial steady-state stage (during which \( T \) decreased at a constant rate with \( \omega \) ) was nearly identical for different values of \( I \) and can be described by the following relationship:

\[ T_{ss} = T_o(1 + \tau \omega) \]  

Here, \( T_{ss} \) is the total torque during the steady-state stage, and \( T_o \) and \( \tau \) are \( \omega \)-independent parameters, which were determined from a linear fit (fine dotted line in Figure 44d) to be \( T_o = 13 \pm 1 \) Nm, and \( \tau = 1.5 \times 10^{-3} \) s. The average values of \( T_{ss} \) and friction coefficient \( \mu_{ss} \) varied from 17.0 to 20.5 Nm and from 0.064 to 0.074 respectively (Table 2).

The length of upset during the experimental welding trials increased with an increase in the flywheel moment of inertia (Table 2). The upset length nearly doubled with a ~5 times increase in moment of inertia.

Temperature transients measured on the Mar-M247 side at axial distances \( L \) between ~2.5 mm and 7 mm from the weld interface revealed a striking dependence on the flywheel moment of inertia (Figure 45). Although the temperature increased at nearly
the same rate for all three welds (at $L = 4.7$ mm), the peak temperature increased with moment of inertia (Figure 45a). Furthermore, the peak temperature decreased nearly linearly with an increase in the axial distance from the weld interface (Figure 45b).

Figure 45: (a) Temperature, $T$, versus time, $t$, profiles registered on the Mar-M247 side at the distance $L = 4.7$ mm from the weld interface during welding with three different flywheel moments of inertia. (b) Peak temperatures recorded on the Mar-M247 side at different distances from the weld interface during welding with three different flywheel moments of inertia.

Temperature transients for the weld interface at the radial centerline ($L = r = 0$ mm) predicted from DEFORM simulations showed an increasing heating rate to the peak temperature with a decrease in moment of inertia followed by a region of nearly-steady state behavior for the three welding conditions (Figure 46a). As the moment of inertia increased, the duration of the nearly-steady state temperature increased, while the predicted maximum temperatures decreased slightly.
Figure 46: FEM predicted (a) temperature profiles at the weld interface centerline locations and (b) upset profiles for three weld conditions with the energy held constant and the efficiency adjusted according to the moment of inertia.

The predicted upset was also shown to increase with an increase in moment of inertia (Figure 46b); the upset nearly doubled in magnitude for welds performed from $I = 0.166$, to $0.802 \text{ kg.m}^2$. The actual upset from the experimental IFW trials exhibited a
similar trend; ranging from 2.86 to 5.78 mm for $I = 0.166$ and 0.802 kg.m$^2$, respectively (Table 2).

Metal Flow, Macrostructure, and Microstructure Observations

The rate and duration of the energy input strongly affected metal-flow behavior and the evolution of macrostructure and microstructure in the weld zone. For example, the shape and size of the zone of highly-localized deformation at the weld line changed noticeably as a function of energy-input rate (Figure 47).
Figure 47: Optical images of slightly etched IFW Mar-M247/LSHR weld cross-sections highlighting the changes in morphology of the narrow bands of localized deformation at the weld interface (a) LM01 (b) LM02 (c) LM03.

Due to the differences in flow behavior at elevated temperatures between the two alloys, most of the deformation during the weld process took place on the LSHR side. Sample LM01, which was welded with the smallest moment of inertia, shortest weld duration, and highest initial rotation speed, exhibited the smallest upset on both LSHR and Mar-M sides. The thickness of the radially-deformed region (extending beyond the original diameter) was ~ 1.1 mm and 2.2 mm on the LSHR and Mar-M247 sides, respectively (Figure 47a). Sample LM02 exhibited an intermediate degree of upset during welding.
This upset was mainly due to extensive plastic deformation and material flow within ~1.6 mm region near the weld interface on the LSHR side. The Mar-M247 side of LM02 exhibited limited plasticity, similar to that observed in sample LM01, with a slight barreling at the weld line. The thickness of the deformed region of LM02 was ~2.2 mm (Figure 47b). Sample LM03, which was welded using the highest $I$ and lowest $\omega_o$, revealed the largest upset of the three welds with flash formation on the LSHR side and significant barreling of the Mar-M247 (Figure 47c). The thickness of the highly-deformed region on the LSHR side was ~1.4 mm and on Mar-M247 side was ~3.2 mm.

The metal-flow observations (Figure 47) also revealed that the weld lines in samples LM01 and LM02 were predominantly linear, while sample LM03 exhibited non-linear morphology. This is an important finding in that straight/flat weld interfaces in these materials often indicate a lack of plastic deformation, are usually decorated with fine carbide and oxide particles, and typically yield poor bond quality [15, 16]. It is also important to note that all three of the welds had the same applied axial load and initial flywheel kinetic energy, thus indicating that bond quality depended primarily on the flywheel moment of inertia and its impact on the weld duration. This observation is intuitive in that the weld times are longer and therefore the weld interface is held at an elevated temperature longer with a larger moment of inertia. By this means, frictional heat is able to conduct further axially (and radially) into the material, thereby increasing the volume of the plastically-deformed region.

More-detailed examination of the weld line revealed a range of location-dependent characteristics which varied with welding conditions. For example, in sample
LM01, well-welded regions with fine recrystallized $\gamma$ grains and no (or very few) welding defects, (Figure 48a), occupied less than 30% of the weld interface area and were mainly observed in the mid-radius regions. Typical welding defects in LM01 were submicron carbide/oxide particles or films which decorated the weld interface area (Figure 48b,c) and cracks along the weld interface (Figure 48d). These defects were present at all radial locations. A tabular description of bond-line defects encountered in the three welds is contained in Table 4.

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>Oxide/Carbide Precipitates</th>
<th>Linear Bond line</th>
<th>Porosity/ Cracks</th>
</tr>
</thead>
<tbody>
<tr>
<td>LM01</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>r = 0.7 mm</td>
<td>High</td>
<td>Medium</td>
<td>Low</td>
</tr>
<tr>
<td>r = 2.5 mm</td>
<td>High</td>
<td>Medium</td>
<td>High</td>
</tr>
<tr>
<td>r = 4.3 mm</td>
<td>Low</td>
<td>Medium</td>
<td>Low</td>
</tr>
<tr>
<td>LM02</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>r = 0.7 mm</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>r = 2.5 mm</td>
<td>Low</td>
<td>Low</td>
<td>-</td>
</tr>
<tr>
<td>r = 4.3 mm</td>
<td>High</td>
<td>High</td>
<td>Low</td>
</tr>
<tr>
<td>LM03</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>r = 0.7 mm</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>r = 2.5 mm</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>r = 4.3 mm</td>
<td>-</td>
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<td>-</td>
</tr>
</tbody>
</table>

Table 4: Incidence of Defect Occurrence at the Weld Line.

The weld quality of sample LM02 was better. More than 70% of the weld interface contained no welding defects (Figure 49a,b). Typical welding defects at the bond line in LM02 were carbide films, carbide and oxide particles (Figure 49c), and radial cracks (Figure 49d). These defects were observed mainly in the outer-diameter (OD) regions. The carbides that decorated the linear regions of the weld line appeared as
continuous films or stringers, which tended to have a smooth surface adjacent to the LSHR side, and a lobed surface that protruded into the Mar-M247 (Figure 49c).

Figure 48: SEM backscattered electron images of weld sample LM01 showing different regions of the weld interface. (a) A defect-free, dynamically recrystallized layer at the weld interface in a mid-radius region; (b) a weld line region agglomerated with carbide submicron-sized particles; (c) a chain of oxide particles; (d) a crack at the weld interface.
This observation may be indicative of bond line temperatures during welding that were high enough to partially dissolve the carbides in the parent materials and then re-precipitate carbide films upon cooling.

Sample LM03, which exhibited the largest upset of the three welds with flash formation on the LSHR side and significant barreling of the Mar-M247 (Figure 50), exhibited a weld interface that was largely free of defects and had been extensively hot worked during IFW (Figure 50c,d).
Figure 49: SEM backscattered electron images of weld sample LM02 highlighting the changes in weld line morphology; (a) & (b) defect-free, dynamically recrystallized layer at the weld interface; (c) long, semi-continuous carbide particles decorating the weld interface; (d) large remnant carbide particles and cracks along the weld interface.

Metal flow at the OD (Figure 50d) highlighted the development in LSHR of secondary flash as well as a thin ribbon that remained bonded to the Mar-M247 flash. Both the high degree of weld interface non-linearity and the absence of oxide and carbide particles at the interface indicated that the bond quality of sample LM03 was likely better than LM01 and LM02.
Figure 50: SEM backscattered electron images of weld sample LM03 highlighting the changes in weld line morphology; (a), (b), (c) defect-free dynamically recrystallized layer at the weld interface; (d) secondary flash formation and remnant, bonded LSHR to the MarM barreled section.

EBSD inverse-pole-figure (IPF) maps (for the tangential direction of each sample) determined at and near the weld line showed that the degree of recrystallization of the initially coarse-grain Mar-M247 was proportional to the moment of inertia (Figure 51). Specifically, the IPF map for the Mar-M247 material welded using the flywheel with the
smallest moment of inertia (sample LM01) showed only slight color variation within the large grains near the weld line. This slight color variation and lack of associated fine grain regions indicated minimal deformation and limited recrystallization (Figure 51a). With an increase in \( I \), the degree of deformation of the large grains (indicated by color variations within the remnant grains) and the amount of recrystallization increased (Figure 51b). For the sample welded with the flywheel with the largest moment of inertia (LM03), the large grains were fragmented into smaller grains which contained necklace-like recrystallization along the boundaries (Figure 51c).

Figure 51: EBSD inverse pole figure maps highlighting the effects of moment of inertia on the recrystallization behavior of Mar-M247 (left) and LSHR (right) at the weld line. (a) LM01 \((I = 0.166 \text{ kg.m}^2)\), (b) LM02 \((I = 0.380 \text{ kg.m}^2)\), (c) LM03 \((I = 0.802 \text{ kg.m}^2)\).
The LSHR material also exhibited different behaviors depending on the flywheel moment of inertia. In particular, the degree of recrystallization increased significantly with increasing $I$. This behavior can be attributed to the fact that with an increase in moment of inertia, higher hot-working temperatures and longer weld durations were attained. This resulted in accommodation of more of the imposed deformation by LSHR.

**Chemical Mixing and Interdiffusion during IFW**

A comparison of the baseline composition of the program alloys (Table 1) indicated that Mar-M247 has higher levels of Ni (60.3 vs 49.4%) and W (10.1 vs 4.3%) and a lower amount of Co (10.4 vs 21.4%) relative to LSHR.

Friction-induced heating and extensive plastic deformation during IFW would thus be expected to bring about interdiffusion and mechanical mixing of these alloying elements and the formation of a transition zone with an intermediate composition. Experimental results in terms of EPMA composition profiles across the weld line at two different radial positions [center ($r = 0$) and OD ($r = 4.2$ mm)] quantified these phenomena (Figure 52). In these figures, the weld interface was taken to be the position at which the BSE contrast changed sharply from lighter/higher Z (Mar-M247) to darker/lower Z (LSHR) (e.g. Figure 50).
Figure 52: Concentration profiles of Ni, Co and W near the weld interface of (a,b) LM01, (c,d) LM02, and (e,f) LM03 samples at different radial distances: r = 0 mm (center) and 4.2 mm (OD).

The thickness of the transition zone from the Mar-M247 to the LSHR composition depended on processing conditions, radial position, and the thickness of the
heavily-deformed Mar-M247 layer near the interface. For example, at the center \((r = 0)\) of samples LM01, LM02, and LM03 (the latter containing the heavily-deformed layer on the Mar-M side of the interface), the thickness of this transition zone was similar, i.e., \(16 \pm 2 \, \mu m, 15 \pm 2 \, \mu m, \) and \(13 \pm 2 \, \mu m\), respectively (Figure 52a,c,e). In this region, the composition of Mar-M247 started to change inside the fine-grain layer, \(~5-7 \, \mu m\) from the weld interface for LM01 and LM02 and \(~10 \, \mu m\) for LM03. The composition of LSHR showed a more abrupt change within \(~9-10 \, \mu m\) from the weld interface for LM01 and LM02 and only \(~3 \, \mu m\) for LM03.

At \(r = 4.2 \, mm\) (OD), no fine-grain layer was present on the Mar-M247 side of either LM01 or LM02, and the weld interface was contaminated with oxide and carbide particles. Here, the transition from the Mar-M247 to LSHR compositions occurred entirely on the LSHR side, and the thickness of the transition zone increased with an increase in the moment of inertia (i.e., an increase in welding time), viz., \(15 \pm 2 \, \mu m\) for LM01 (Figure 52b) and \(29 \pm 2 \, \mu m\) for LM02 (Figure 52d). In contrast with LM01 and LM02, sample LM03 exhibited a continuous fine-grain layer from the center to the OD on the Mar-M247 side of the weld interface, and this layer was wider at the OD than in the center. The transition from the Mar-M247 to the LSHR composition at the OD of sample LM03 occurred mainly inside the fine-grain region. The thickness of the transition zone inside the fine-grain Mar-M247 side was estimated to be \(~50 \, \mu m\), and only \(~4 \, \mu m\) on the LSHR side (Figure 52f). These observations suggested that LSHR was softer during trials LM01 and LM02, deformed more severely than Mar-M247, and thus
experienced more extensive mechanical mixing into Mar-M247. Enhanced (pipe) diffusion associated with the finer grain size and associated increased grain boundary length of LSHR may also have contributed to the formation of the transition zone on the LSHR side. By similar reasoning, the deep transition zone in LM03 on the Mar-M247 side and a thinner zone on the LSHR side, relative to the LM01 and LM02 samples, can be associated with heavier deformation in Mar-M247, and a higher degree of mechanical mixing of the two alloys during IFW.

The EPMA data also revealed extensive scatter about the average concentration values in both Mar-M247 and LSHR far from the weld interface (Figure 52). This scatter arose mainly from the presence of large primary (or secondary) $\gamma'$ precipitates having higher Ni and lower Co than the $\gamma$ matrix. Carbide particles enriched in Ta, Hf, Ti, and/or W and depleted in Ni also contributed to the scatter in the data; the position of these minor-phase particles can be correlated to locations which exhibited a considerable drop in Ni. In LSHR near the IFW interface, the composition scatter associated with the $\gamma'$ phase was noticeably lower. The formation of this homogeneous region likely resulted from dynamic dissolution of large primary (and secondary) $\gamma'$ particles during IFW and subsequent re-precipitation of finer $\gamma'$ during rapid cooling upon completion of the weld. [15, 49, 50]. The thickness of the homogeneous region tended to increase with radial distance from the center and a decrease in the flywheel moment of inertia. For example, near the center of LM01, LM02, and LM03, the homogenized region was $\sim$50 µm, 33
μm, and 22 μm, while at the OD it increased to 135 μm, 114 μm, and 105 μm, respectively (Figure 52).

Mechanical Properties

Mirroring the microstructure results, the microhardness measurements also varied axially and radially (Figure 53(r = 0 mm) and Figure 54(r = 3.3 mm)). In each of the figures, LSHR is on the right side of the weld interface annotated by a vertical dotted line. The data showed that the axial hardness profiles of the welded samples depended strongly on the flywheel moment of inertia and the radial position. Sample LM01, which used the flywheel with the lowest moment of inertia, showed a sharp hardness maximum at the weld interface and two local hardness minima, one on each side of the interface, approximately 0.75 mm from the weld line on the LSHR side and 0.1-0.2 mm from the weld line on the Mar-M247 side (Figure 53a).
Figure 53: Hardness as a function of distance along the weld centerline in samples (a) LM01, (b) LM02, and (c) LM03 (Mar-M247 is on the left side and LSHR is on the right side of each graph).

The minimum on the LSHR side was deeper at the center of the weld \( (r = 0 \text{ mm}) \) than near the OD (Figure 53a, Figure 54a). The minimum on the Mar-M247 side was deeper and more pronounced at \( r = 3.3 \text{ mm} \) (Figure 54a). The hardness minima disappeared on Mar-M247 side and became very shallow on LSHR side, while the height and breadth of the hardness maximum near the weld interface increased with increased moment of
inertia, i.e., samples LM02 and LM03 (Figure 53 and Figure 54). At distances greater than 2 mm from the weld interface, the hardness on the LSHR side was Hv=361±5, comparable to that of the parent LSHR, in all of the welded samples. On the Mar-M247 side, the width of the heat affected zone (HAZ), in which the hardness after welding differed from that of the parent metal (397±10 Hv), was ~2 - 3 mm at r = 0 mm, and ~3 - 4 mm at r = 3.3 mm.

![Graphs showing hardness as a function of distance across the weld OD in samples LM01, LM02, and LM03.](image)

Figure 54: Hardness as a function of distance across the weld OD in samples (a) LM01, (b) LM02, and (c) LM03 (Mar-M247 is on the left side and LSHR is on the right side of each graph).
The tensile properties of the welded samples at three different radial locations (Table 5, Figure 55) provided quantitative insight into the mechanical integrity of the IFW bonds. This information was complemented by measurements of the local axial strain as a function of position along the tension axis (Figure 56); the position of the weld line was indicated by a vertical dashed line with LSHR on the right. In each of the tension tests, LSHR did not yield, and all strain was localized within the Mar-M247 portion of the sample (Figure 56).

Figure 55: Tensile stress-strain curves of (a) LM01, (b) LM02 and (c) LM03 welds at different radial locations: \( r = 0.7 \) mm, \( r = 2.5 \) mm and \( r = 4.3 \) mm.
Tension samples extracted from the center and middle of sample LM01 (i.e., $r = 0.7$ and 2.5 mm, respectively) failed along the weld interface at relatively low stress levels (559 MPa and 517 MPa respectively) and did not show any macroscopic plastic strain (Figure 55a). The tension sample extracted from the $r = 4.3$ mm location of weld LM01 showed noticeable plastic strain. The 0.2% yield stress (YS) of this sample was 724 MPa, and the sample exhibited continuous strain hardening at the rate $d\sigma/d\varepsilon = 2800 \pm 200$ MPa, total elongation (El) of 3.8%, and an ultimate tensile strength (UTS) of 840 MPa (Figure 55a). Extensive strain localization preceded ductile fracture on the Mar-M247 side approximately 7 mm away from the weld interface (Figure 56a).
Among the tension samples extracted from LM02 at three different radial positions, those at $r = 0.7$ mm and 2.5 mm showed identical ductile behavior, with $YS = 727 \pm 6$ MPa, $d\sigma/d\varepsilon = 2700$ MPa, $UTS = 827 \pm 15$ MPa, and $El = 3.7 \pm 0.3\%$ (Figure 55b, Table 5). These samples showed noticeable strain localization on the Mar-M247 side and fractured away from the weld interface and the heat affected zone (Figure 56b). However, the sample extracted at $r = 4.3$ mm failed along the weld interface at a very low stress level (301 MPa) and did not show any macroscopic strain.
All tension samples extracted from LM03 showed similar ductile behavior, with
YS = 733±5 MPa, dσ/dε = 2700 MPa, UTS = 838±15 MPa, and El = 2.9±0.5% (Figure 55, Table 5). In these samples, the strain localized and failure occurred on the Mar-M247 size outside the HAZ (Figure 56c). Failure outside the HAZ was an indication of complete bonding across the weld interface area.

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>E (GPa)</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>Elongation %</th>
<th>Fracture Location</th>
</tr>
</thead>
<tbody>
<tr>
<td>LM01</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>r = 0.7 mm</td>
<td>211</td>
<td>-</td>
<td>559</td>
<td>0.3</td>
<td>Weld Line</td>
</tr>
<tr>
<td>r = 2.5 mm</td>
<td>220</td>
<td>-</td>
<td>517</td>
<td>0.22</td>
<td>Weld Line</td>
</tr>
<tr>
<td>r = 4.3 mm</td>
<td>216</td>
<td>724</td>
<td>840</td>
<td>3.8</td>
<td>Outside HAZ</td>
</tr>
<tr>
<td>LM02</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>r = 0.7 mm</td>
<td>210</td>
<td>722</td>
<td>830</td>
<td>3.9</td>
<td>Outside HAZ</td>
</tr>
<tr>
<td>r = 2.5 mm</td>
<td>206</td>
<td>733</td>
<td>824</td>
<td>3.5</td>
<td>Outside HAZ</td>
</tr>
<tr>
<td>r = 4.3 mm</td>
<td>201</td>
<td>-</td>
<td>301</td>
<td>0.2</td>
<td>Weld Line</td>
</tr>
<tr>
<td>LM03</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>r = 0.7 mm</td>
<td>205</td>
<td>733</td>
<td>834</td>
<td>2.9</td>
<td>Outside HAZ</td>
</tr>
<tr>
<td>r = 2.5 mm</td>
<td>181</td>
<td>735</td>
<td>853</td>
<td>3.4</td>
<td>Outside HAZ</td>
</tr>
<tr>
<td>r = 4.3 mm</td>
<td>203</td>
<td>732</td>
<td>828</td>
<td>2.4</td>
<td>Outside HAZ</td>
</tr>
</tbody>
</table>

Table 5: Tensile Properties of IFW Samples as a Function of Radial Location.

Two markedly-different fracture-surface morphologies were noted in SEM secondary-electron (SE) images of the failed tension specimens. For samples which failed in the Mar-M247 side outside the HAZ, the fracture morphology was typical of a moderately-ductile material (Figure 57a). The fracture surface had a blocky, faceted, and layered appearance. Cleaved surfaces of large fractured carbide particles were also observed suggesting that cracks were initiated inside these particles. The presence of
dimples on the faceted surfaces (Figure 57b,c) suggested that the ductile failure mechanism of the γ matrix was by cavitation.

By contrast, the fracture surfaces of LM01 ($r = 2.7$ mm and 4.3 mm) and LM02 ($r = 4.3$ mm), which failed at the weld interface, exhibited a large area of un-bonded material (Figure 58a). The un-bonded regions appeared to exhibit wear/rubbing features in a circular pattern suggestive of the rotational motion imposed during IFW. Higher magnification examination revealed the presence of oxide and carbide particles, as well as fine porosity, in these un-bonded regions (Figure 58b,c). The fracture-surface regions adjacent to these defects contained a refined grain structure of Mar-M247 and numerous ductile dimples (Figure 58d,e).
Figure 57: SEM secondary electron images of the fracture surface of a tensile sample fractured outside HAZ on Mar-M247 side (LM03-2); (a) Entire fracture surface at low magnification; (b-c) higher magnification images illustrating (b) a cellular, faceted appearance of fracture and (c) dimples on the faceted surfaces. A fractured carbide particle is shown by an arrow in figure (b).
Figure 58: SEM secondary electron images of the fracture surface of a tensile sample fractured at the weld interface (LM21-3); (a) Entire fracture surface with a circular welding defect at low magnification; (b, c) submicron-sized carbide and oxide particles on the surface of the welding defect; (d) transition from poorly bonded (bottom) to bonded (top) region; (e) ductile fracture of the bonded region.
The experimental results provided broad insight into the mechanics of the IFW process and the effect of IFW process variables on metal flow, microstructure evolution, and post-formed mechanical properties. These aspects are discussed and interpreted in the following sections.

*Analysis of IFW with constant input energy and axial force*

The present results revealed a number of important details related to IFW process optimization in terms of metal flow, degree of upset, bond quality, and post-weld mechanical properties. It is commonly believed that IFW is controlled by two main parameters, welding energy, \( E_{ko} \), and the axial compression force, \( P \). The moment of inertia, \( I \), and the initial rotation speed, of flywheel \( \omega_0 \), are typically selected based on the required welding energy (Equation (1)) and flywheel mass available for a given IFW machine [11]. The results of the present work revealed, however, that metal flow and microstructure response during IFW of dissimilar superalloys such as LSHR and Mar-M247 depend strongly on \( I \), despite constant \( E_{ko} \) and \( P \). In particular, the processing time, deformed volume, maximum temperatures developed in HAZ, and degree of sample upset, all increased with an increase in \( I \). The quality (i.e. integrity and strength) of the welds also improved with increasing \( I \). In particular, during post-weld tension testing, welds produced with the highest moment of inertia (\( I = 0.802 \) kg·m\(^2\)) showed significant plasticity beyond the yield point and failure on the Mar-M247 side far from the interface. On the other hand, welds fabricated with the lowest moment of inertia (\( I = 0.166 \) kg·m\(^2\))...
exhibited essentially no plastic flow in tension prior to failure at the weld interface at which there were a variety of defects. In the latter samples, the process parameters resulted in a partially-bonded condition, likely due to insufficient heating and plastic deformation during welding. Specifically, the Mar-M247 material exhibited minimal upset which resulted in the retention of weld-related defects. Such plastic deformation is required to expel weld-surface contaminants into the flash; these contaminants include submicron-sized, often-agglomerated, carbide and oxide particles, which prevent bringing nascent metal into contact to form a sound metallurgical bond [7, 11]. The welds produced using the intermediate moment of inertia ($I = 0.381 \text{ kg} \cdot \text{m}^2$) showed a mixed deformation/fracture behavior during post-weld tension testing.

The noticeable effect of $I$ at fixed values of $E_{ko}$ and $P$ on welding and post-welding behavior can be rationalized by examining the rate of dissipation of the kinetic energy of the flywheel, $dE_k/dt$ (Figure 43b). The present results showed that the decrease in $E_k$ occurred more rapidly when a smaller flywheel mass was used. In general, the rate of decrease in $E_k$ is controlled by the power losses due to friction and the transformation of the kinetic energy into friction-induced heating of both (i) the workpiece samples at the weld interface and (ii) the journal and thrust bearings and the surrounding oil within IFW machine. Although the energy efficiency of IFW equipment has been mentioned as a contributing factor in the description of the IFW process, it appears that the energy losses due to friction in the IFW machine bearings have neither been analyzed nor reported in the literature [7, 28, 51].
An increase in the fraction of the kinetic energy consumed by the rotating parts of the welding machine (i.e., parasitic energy losses) results in a decrease in the efficiency of the IFW process due to the reduced level of energy available to heat the contact surfaces of the alloys to be welded. The relative contributions to the rate of energy dissipation can be expressed as:

\[
dE_S/dt = dE_k/dt - dE_M/dt
\]

Here, \(dE_S/dt\) and \(dE_M/dt\) are the rates of energy dissipation at the weld interface surface and within the welding machine, respectively. If the respective friction-induced torque values, \(T_S\) and \(T_M\), are known, these quantities can be calculated, from the general relationship:

\[
dE/dt = T \omega
\]

For cylindrical workpieces, the torque, \(T_S\), at the weld interface is determined by the product of the effective friction coefficient, \(\mu_S\), applied axial compression force \(P\), and the outer radius of the workpieces, \(r_o\) [15, 52], i.e.,

\[
T_S = 2\mu_S Pr_o/3
\]

For the machine losses, \(T_M\) is the sum of the journal bearing torque \(T_{JB}\) and thrust bearing torque \(T_{TB}\), which can be expressed as follows [53-55]:

\[
T_{JB} = (2\pi\nu R_1^3 L/h_1)\omega
\]

\[
T_{TB} = (2\pi\nu R_2^4/3h_2)\omega
\]

Here \(\nu\) is the oil viscosity, \(R_1\) is the shaft radius, \(L\) and \(h_1\) are the length and radial clearance of the journal bearing, and \(R_2\) and \(h_2\) are the effective surface radius and oil
film thickness of the thrust bearing. If it is assumed that these values are constant during
the steady-state portion of welding, \( T_M \) can be expressed as a linear function of \( \omega \), i.e.,

\[
T_M = C_1 \omega
\]  
(26)

Here, \( C_1 = 2\pi \nu (R_1^3 L/h_1 + R_2^4/3h_2) \) is a parameter that depends on the configuration of the
bearings, oil viscosity, and, perhaps, the axial compression force \( P \) (through its effect on \( h_2 \)).

Combining Equations (22) and (26), the parasitic energy losses inside the welding
machine can be calculated as:

\[
E_M = \int_{0}^{t_{max}} C_1 \omega^2 dt
\]  
(27)

In Equation (27), \( t_{max} \) denotes the duration of the IFW process (i.e., the time interval
between the instant when the sample surfaces are brought together at the rotational
velocity \( \omega = \omega_0 \) and that when the flywheel rotation stops, \( \omega = 0 \)). To perform the
integration in Equation (27), it was assumed as a first approximation that \( \omega \) decreases
linearly with time, or

\[
\omega = \omega_0 - at
\]  
(28)

Inserting this expression into Equation (27), integrating, and applying Equation (1) yields
the following relation:

\[
E_M = \frac{C_1 \omega_0^2 t_{max}}{3} \equiv \frac{2C_1 E_{ko} t_{max}}{3I}
\]  
(29)

Taking into account the fact that the welding trials were conducted with identical values
of \( E_{ko} \), the process efficiency, \( \eta \), can then be estimated using the following formula:
\[
\eta = 1 - \frac{E_M}{E_{ko}} = 1 - \frac{2Ct_{\text{max}}}{3I}
\]  
(30)

Equation (30) reveals that \(\eta\) increases with decreasing \(t_{\text{max}}\) and increasing \(I\). The present experimental data (Figure 43a) indicated that when \(I\) was increased \(\sim 4.83\) times (from \(0.166\) kg·m\(^2\) to \(0.802\) kg·m\(^2\)), \(t_{\text{max}}\) increased \(\sim 2.45\) times (from \(4\) s to \(9.8\) s). Thus the quotient \(t_{\text{max}}/I\) decreased by a factor of \(\sim 2\). Therefore, the efficiency of the IFW process increased with an increase in the flywheel moment of inertia, even though the total welding energy, \(E_{ko}\), had remained constant. This analysis thus provides a plausible explanation why increasing \(I\) at constant \(E_{ko}\) results in welds with more pronounced flash, more extensive deformation, and improved weld quality.

The analysis above also enables quantitative estimation of \(\mu_S, C_1,\) and \(\eta\). Specifically, combining Equations (23) and (26), the following relation for the total torque is obtained:

\[
T = \frac{2\mu_S r_s P}{3} + C_1 \omega
\]  
(31)

A comparison of Equations (20) and (31) reveals that the friction coefficient at the weld interface during the steady-state stage of the IFW process (\(\mu_{SS}\)) is constant and exhibits a weak dependence on the IFW parameters, at least for the present IFW conditions and alloys, i.e.,

\[
\mu_{SS} = \frac{3T_o}{2r_s P} \approx 0.050 \pm 0.005
\]  
(32)

The comparison of equations (20) and (31) also enables an estimation of the coefficient \(C_1\), i.e.,
\[ C_1 = \tau T_o \approx 0.020 \pm 0.003 \text{ kg}\cdot\text{m}^2\cdot\text{s} \]  

(33)

Assuming that \( C_1 \) is time-independent, the temporal dependence of the friction coefficient at the weld interface, \( \mu_s \), is determined from Equation (31) by subtracting the machine bearing torque, \( T_M = C_1 \omega \), from the total torque, \( T = Ia \), thereby resulting in the following expression

\[ \mu_s = \frac{3}{2} \frac{Ia - C_1 \omega}{r_s P} \]  

(34)

As expected, the value of \( \mu_s \) during IFW (with constant \( E_{ko} \) and \( P \)) was only weakly dependent on the moment of inertia of the flywheel during the first half of the welding period; i.e., \( \mu_s = \mu_{ss} \approx 0.05 \) (Figure 59).

Figure 59: The dependence of the effective friction coefficient, \( \mu S \), between the welding surfaces of LSHR and Mar-M247 on (a) the processing time and (b) number of flywheel revolutions during IFW process occurring at three different flywheel moments of inertia (given in the figure) and a constant weld energy, \( E_{ko} = 22.2 \text{ kJ} \). The friction coefficient was calculated using Equation (34) after subtracting torque due to the welding machine bearings.
Subsequently, it increased rapidly and approached a maximum value of ~ 0.12 - 0.14 at the end of welding. This increase in $\mu_S$ has been shown previously to be due to a change in the nature of interface friction between LSHR and Mar-M247 from sliding to sticking and plastic flow rather than simply a decrease in rotation speed [15]. Indeed, when $\mu_S$ is plotted as a function of the number of revolutions of the flywheel ($N$), a similar evolution of $\mu_S$ is noted for the various values of $I$ and therefore different $\omega_b$ and $a$ values (Figure 59b). This result may therefore indicate that the transition from sliding to incipient sticking, as well as the start of upset and flash formation, occurs after approximately the same number of revolutions, i.e., after the same amount of relative sliding of the faying surfaces.

Using Equation (30) with $C_1 = 0.02 \text{ kg} \cdot \text{m}^2 \cdot \text{s}$ and experimental values for $t_{\text{max}}$ (Figure 43), the efficiency of the IFW process was estimated to be $\eta = 0.68, 0.74, \text{ and } 0.84$ for $I = 0.166, 0.381 \text{ and } 0.802 \text{ kg} \cdot \text{m}^2$, respectively. These values for the IFW process efficiency are significantly lower than estimates (0.9-1.0) reported previously [56-58]. However, the present FEM results for IFW do predict the observed upset behaviors (Figure 46b) if $\eta$ is assumed to increase with $I$.

The noticeable dependence of $\eta$ on IFW process parameters identified in this work has not been considered in previous publications [7, 11, 28, 51, 56]. Rather, a constant efficiency has typically been assumed in order to interpret and model IFW and to quantify process parameters such as required energy input, degree of flash formation, and extent of sample shortening [56, 58, 59]. It appears that this assumption has thus led to incorrect conclusions comprising the following: (1) $E_{ko}$ is the principal parameter
controlling IFW, and (2) different combinations of \( I \) and \( \alpha_0 \), which provide the same value of \( E_{ko} \) have a negligible effect on the thermal and deformation behavior during IFW. For example, Wang, et al. [56] assumed that \( \eta \) is independent of IFW processing parameters and is equal to 0.9. Based on their analysis, it was concluded that use of a flywheel with a smaller moment of inertia (and thus higher initial rotation speed) is beneficial inasmuch as this will increase the temperature at the weld interface more quickly and lead to shorter welding time while providing the same amount of workpiece upset. Such conclusions contradict the present experimental observations which have revealed that welding with the smallest moment of inertia provides the least efficient conversion of \( E_{ko} \) to heat at the weld interface, thereby producing the least flash and poorest weld quality. By contrast, trial LM03 in the present work, conducted with the highest moment of inertia and lowest maximum rotation velocity, led to the only weld that did not exhibit failure at the bond line during subsequent room-temperature tension testing. This sample also exhibited the largest amount of total upset during welding.

It is also important to note that the maximum moment of inertia (0.802 kg\( \cdot \)m\(^2\)) in the present work was below the minimum moment of inertia used by Wang et al. [56], while the initial kinetic energy was almost twice as high (22.1 kJ vs. 13.75 kJ). The higher \( I \) and lower \( E_{ko} \) in [56, 58] should have resulted in slower rotational velocity and therefore smaller parasitic energy losses with higher process efficiency (Equations (29) and (30)). Another difference in the work done in [56, 58] and the present effort was that welding was performed using hollow (tubular) samples and, in the case of [18], the welded area was approximately an order of magnitude greater than that of the solid
cylinders utilized herein. The differences in reported welding time, upset, etc. for weld trials performed at constant energy in [18] as compared to the results summarized here could thus be due in part to the large differences in weld area and sample geometry. However, specific comparisons cannot be made because the specific weld conditions were not provided in [18].

For welds performed at constant $E_{ko}$ and $P$, an increase in $I$ increases the duration of the IFW process, which occurs at slower rotation speeds. Therefore, a smaller fraction of $E_{ko}$ is consumed by the parasitic (friction) work associated with the drive shaft and machine bearings, inasmuch as this work is the product of the square of the rotation speed and duration Equation (29). At least for the specific combination of superalloys used here, the simultaneous increase in efficiency and the duration of IFW is beneficial in developing plastic flow in the harder Mar-M247 side. Indeed, increasing $\eta$ indicates that a higher fraction of $E_{ko}$ is used to heat the weld surfaces and, therefore, a higher peak temperature is achieved [56, 58]. Also, increasing $t_{\text{max}}$ results in increased weld duration in the processing temperature range (Figure 46), thereby leading to a larger volume of material that is deformed plastically. The longer duration of the IFW process at higher $I$ also gives rise to an increased degree of recrystallization in Mar-M247 near the weld line (Figure 51) and to an increased width of the transition region inside which mechanical mixing of LSHR and Mar-M247 occurs (Figure 52). Additional increases in weld duration could further increase the level of deformation and recrystallization in Mar-M247 and promote improved bonding. Another attractive process alternative might include a reduction in the applied axial load in order to increase the temperature at the
weld interface and further promote plasticity in Mar-M247 [11]. Reducing the axial load, however, may not have a significant impact on weld temperature, inasmuch as LSHR has a lower \( \gamma' \) solvus temperature and reduced strength at elevated temperatures compared to Mar-M247. Therefore, the maximum temperature achievable at the weld interface would likely be limited by plastic flow of LSHR. Nevertheless, methods to reduce the flow-stress difference between the workpiece materials could prove useful and are worthy of further investigation.

*Post-weld Properties*

From a broad perspective, the results of the hardness measurements for samples LM01, LM02, and LM03 were similar to previous observations [15] and can be explained on the basis of the evolution of the \( \gamma' \) precipitate size and \( \gamma \) grain size during welding and post-weld cooling. The higher hardness in the HAZ region compared to the parent Mar-M247 and LSHR alloys was probably due to grain refinement as well as dissolution of the coarse primary and secondary \( \gamma' \) particles during the welding process and re-precipitation of finer particles from the supersaturated \( \gamma \) matrix during cool-down after welding. This change in size, distribution, and volume fraction of \( \gamma' \) particles along with a noticeable decrease in the \( \gamma \) matrix grain size could account for the observed hardness maximum at the weld interface [22]. The formation of a ~10-100 \( \mu \)m wide, apparently precipitation-free, layer in LSHR at the weld line indicates that the peak
temperature at the weld interface was above the non-equilibrium, on-heating $\gamma'$ solvus and thus all of the $\gamma'$ precipitates went into solution inside this layer (Figure 52). With an increase in the distance from the weld interface, the peak temperature gradually decreased, which resulted in a smaller fraction of $\gamma'$ which had dissolved in the matrix and then re-precipitated during cooling. This reasoning thus explains the observed continuous decrease in hardness from the peak value at the weld interface to the value(s) characteristic of the parent alloy(s) beyond the HAZ. The small local minimum in the hardness of LSHR observed near the center ($r = 0$) at an axial distance of $\sim$1.0 mm from the weld interface (Figure 53) was likely related to coarsening of the secondary and tertiary $\gamma'$ particles relative to the initial condition. This situation can happen when the welding temperature in this region is slightly below the isothermal forging temperature. The minimum hardness in LSHR became less pronounced with increases in the flywheel $I$ and radial distance $r$ (Figure 53 and Figure 54); both factors favor higher peak temperatures and longer-duration at peak temperature. Similar reasoning can be used to explain the drop in hardness at the weld line for the Mar-M247 side of sample LM01 at $r = 3.3$ mm. Here, it appears that the temperatures achieved at all regions near the interface were below the $\gamma'$ solvus, so the overall observations can be explained by the competition between coarsening of secondary $\gamma'$, partial dissolution of primary $\gamma'$ during heating, and re-precipitation of finer particles during cooling. Because the volume fraction of secondary $\gamma'$ was very small, the overall trend was an increase in hardness within the HAZ after welding relative to the parent Mar-M247.
The tensile properties provided clues to the effect of defects on weld quality. Samples LM01 and LM02 contained bond-line defects across a portion of their weld-interface area, whereas no apparent defects were detected at the weld interface of sample LM03. The lack of bonding apparent on the fracture surfaces of several tension samples extracted from the welds correlated well with as-welded defects. Furthermore, there appeared to be a correlation between the quality of the bond and the rotation speed/moment of inertia during welding. For the conditions investigated in the present work, the fraction of un-bonded region decreased as the flywheel moment of inertia increased (at constant initial kinetic energy of the flywheel $E_{k0}$). The decrease in weld-line defects can be directly related to an increase in the process efficiency. Indeed, the source of the un-bonded regions was the apparent lack of gross plasticity and deformation along the weld interface in Mar-M247. With an increase in the flywheel moment of inertia, the efficiency of the IFW machine in converting the kinetic energy of the flywheel to frictional heating at the weld interface increased from 68% for weld LM01 to 85% in weld LM03. The increased heating at the weld interface along with increased weld processing time resulted in a higher peak temperature at the weld interface and deeper heating axially from the weld line. As a result, a larger volume of Mar-M247 experienced plastic deformation and mechanical mixing with LSHR. Although the radial flow of Mar-M247 was limited, extensive mechanical mixing of the mating surface layers due to circumferential and axial plastic flow in weld LM03 was effective in dispersing submicron-size oxide and carbide particles from the interface into the bulk. This dispersal reduced the amount of un-bonded area due to these defects.
Another important observation from the tension tests was the change in the radial dependence of the weld-interface strength with a change in $I$. For example, fracture occurred at the weld interface without any evidence of plastic strain in weld LM01 at $r < 4.3$ mm. However the tension sample extracted at $r = 4.3$ mm showed noticeable strength/ductility and fractured outside the HAZ. The weld interface of LM02 exhibited some plasticity at $r = 0.7$ mm and 2.5 mm, but none at $r = 4.3$ mm. The weld interface of LM03 exhibited plasticity at all radial locations. Detailed analysis of the fracture surfaces of LM01 revealed that the defects responsible for the fracture of LM01 at the weld interface were extensive porosity and agglomerated submicron-size carbide/oxide particles at $r = 0.7$ mm and a continuous carbide layer, as well as cracking along this layer, at $r = 2.5$ mm. The weld interface of the tension sample extracted at $r = 4.3$ mm was almost defect free. In tension samples extracted from LM02, welding defects were identified at $r = 4.3$ mm only. Such differences in behavior can be explained by changes in weld duration. The short duration of the IFW process in LM01 did not allow sufficient heating of the center region because of limited thermal diffusion from the OD toward the center, whereas the OD region was heated faster due to faster linear rotation speed. In LM02, longer welding time and extensive plastic flow resulted in hotter OD material moving into the flash and a more homogeneous radial distribution of temperature. The radial plastic flow also moved contaminants from the center of the weld interface toward OD thus weakening the OD region.
Impact of Preheating on IFW Behavior

Based on the previous weld experiments where welding defects were related to limited plasticity and upset in the Mar-M247 side of the dissimilar welds, simulations and an experiment were performed in order to promote deformation during IFW on the Mar-M247 side of the weld. The main issue here is that the MarM-247 is stronger at elevated temperatures as compared to LSHR [23]. Therefore during welding when the weld zone reaches temperatures approaching 1200C, the MarM-247 maintains higher yield strength than LSHR, and subsequently most of the deformation partitions to the LSHR side of the weld. This lack of plastic flow allows any films, oxides etc. resident on the surface of the Mar-M247 to remain at the weld line which promotes weld defects. Approaches to reduce the strength difference between the two materials at elevated temperatures may promote more uniform deformation and reduce the occurrence of weld-related defects.

For the experimental approach, an induction heating coil was used to heat the weld face of the Mar-M247 sample immediately prior to welding (Figure 60). Two different pre-heat temperatures were utilized, approximately 350°C and 400°C. The samples were held at temperature for approximately 20 seconds (350°C) and 60 seconds (400°C) while the heating coil and power supply were removed from the weld machine, and the flywheel was accelerating to the maximum rotational velocity.
In all cases, the heating resulted in increased deformation on the Mar-M247 side of the weld, as well as total upset length for the weld couple. The welded samples are shown below with the LSHR side on the left and Mar-M247 on the right side of the image (Figure 61).
The upset length (overall) increased by over 2mm with an increase in preheat temperature from 350°C to 400°C. The same weld conditions without a preheat resulted in a total upset of 5.8 mm. This indicates that the 350°C preheat was not sufficient to change the total upset.

Upon examination of the weld line microstructure of these samples some striking changes were discovered. First, for the sample that was preheated to 350°C, the weld zone had many large carbide particles that were fractured. This fracturing of the large carbides was most likely due to the high torque at the end of the weld process combined with the limited ductility of MarM-247 (and especially the carbide particles). Also of note are the long stringer carbide particles that decorated the weld line. These particles exhibited a smooth, linear edge (nearest the weld line), and a lobed, irregular edge away from the weld line. This may indicate that these carbides reprecipitated and grew into the MarM-247 during the cool-down from the weld process (Figure 62). There appears to be
fine recrystallized MarM-247 grains along the bond line adjacent to coarser LSHR grains (right side of Figure 62).

Figure 62: SEM backscattered micrograph of the weld line in the 350C preheat sample with Mar-M247 on the left and LSHR on the right.

Another interesting observation is that the bond line near the outer radius exhibited some cracking due to limited ductility, but also remnant secondary flash from the LSHR side that remained bonded to the Mar-M247 barreled feature. In Figure 63 LSHR is on the right side and Mar-M247 is on the left. Large carbides and as-cast dendritic structure is
evident on the Mar-M247 material. The LSHR appears uniform in composition based on the back-scattered contrast which is expected as this is a powder metallurgy alloy.

Figure 63: SEM backscattered micrograph of the flash formation on the LSHR side (right) and slight barreling of the Mar-M247 (left).

Overall, for the 350C preheat weld trial, there appears to be improved overall deformation (based on upset length), however the bond line morphology still appears to exhibit cracking, large carbides, and linear behavior similar to the non-preheat weld condition. These factors indicate that the bond quality may not be much improved. Upon
closer inspection of the carbide particles in this sample and comparison of their morphology with literature, it appears there are multiple variants present. The large carbides (remnant from the original casting) are most likely MC type carbides, while the satellite particles that likely re-precipitated during welding around the large MC type particles are likely $M_{23}C_6$. The large stringer particles along the weld line are likely $M_{23}C_6$ or $M_6C$ type (Figure 64). Based on EPMA analysis, the primary constituents of the carbide particles are tantalum, hafnium or tungsten. A specific analysis to determine the type of carbide was not performed; however the suggested types were based on comparisons of morphology with published data [46].
Figure 64: SEM backscattered micrograph of the 350°C preheat sample weld line highlighting the likely forms of the carbide particles present.

\[ MC + \gamma = M_{23}C_6 + \gamma' \]

\[ MC + \gamma = M_6C + \gamma' \]
The identification of these particles is based on comparison with micrographs published in [60]. Here, the following MC and M_{23}C_6 carbides were identified and exhibited similar morphology in Mar-M421[60] (Figure 65).

Figure 65: SEM image of replica of etched surface showing the decomposition of carbides and their forms [55].
Once again, large MC type carbides are ringed by $\text{M}_2\text{C}_6$ carbides due to decomposition and re-precipitation.

For the sample welded with the $400^\circ\text{C}$ preheat, the bond line morphology change significantly. First to note is that there is a fine recrystallized grains obscuring the bond-line (Mar-M247 on left). The carbides are there, but not in the long stringer configuration as is prevalent in the $350^\circ\text{C}$ preheat sample. These carbides appear to have begun decomposition as evidenced by the fine particulates ringing the large carbide particles (Figure 66).

Figure 66: SEM backscattered micrograph of the weld line in the $400^\circ\text{C}$ preheat sample with Mar-M247 (left) and LSHR (right).
The sample welded with the 400°C preheat exhibited increased overall upset, and also evidence of increased deformation of the Mar-M247 material. Near the outer radius, it is clearly shown that the bond line is wavy which is an indication that significant deformation took place (which breaks up oxide particles and films which are detrimental to bond quality). It is also interesting to note that there is significant LSHR secondary flash that remained bonded to the Mar-M247 flash. The Mar-M247 material is on the left side of Figure 67. It is also evident that the MarM-247 material underwent significant deformation based on the wavy pattern of the as-cast dendritic structure shown by the contrast in the back-scattered SEM image below.
Figure 67: SEM backscattered micrograph of the 400°C preheat sample weld line highlighting the deformed structure in the Mar-M247 (left) and LSHR (right).

A higher magnification view of the bond line shows the enhanced deformation that the increased preheat temperature provided (Figure 68). Here there is heavy deformation on the Mar-M257 side (left) as indicated by the wavy morphology of the dendritic structure as well as evidence of mechanical mixing of the two materials along the bond line (as evidenced by the ‘fingers’ of the two materials that are interlinked).
Figure 68: SEM backscattered micrograph of the weld line in the 400°C preheat sample highlighting the deformed structure in the Mar-M247 (left) and fine-grain LSHR (right).

It is also noteworthy that there are no large stringer carbides present near the bond line. Multiple large carbides appear to be partially decomposed from the weld process, with many smaller carbide particles that likely reprecipitated after the welding.

In order to more fully interpret the effects of preheating on the weld behavior, finite element simulations were run with various preheat temperatures. For this analysis, simulations of LSHR to LSHR welds were performed in order to examine the effects of preheat temperature only on weld behavior. Within this context, the general trends
derived from the simulation results are applicable, even though the magnitude of the temperatures and deformation values predicted are not.

The difference in the amount of upset length between the preheated side and the non-preheated side of the weld couple during welding was significant (Figure 69).

![Figure 69: FEM simulation results for the effects of preheat temperature on the difference in upset length between the preheated side and the non-preheated side.](image)

At the early stages of the weld process (1.3 s) the difference in length is quite small between the two weld pieces. Only at the largest preheat simulated did the difference in length not approach zero. At 5 seconds, the difference in length was significant at all preheats except the lowest (150°C). For the end of the rotation period during welding, all simulations resulted in a significant difference in length except once again, the lowest
preheat (150°C). It is clear from these simulations that the effect of preheat temperature on upset is quite profound for all preheat temperatures above 150°C. The reasons for the increased upset of the preheated side are most likely related to the differences in thermal profile for the weld line. Time dependent temperature data was examined from the simulations for the axial centerline (R = 0 mm) at the weld interface (Figure 70).

A number of interesting trends are evident from the temperature differential between the preheated and non-preheated weld pieces (at R = 0 mm, weld line). First, the temperature difference for all weld simulations drops dramatically in the first second of weld duration. In all cases, the temperature differential drops by more than 50% during this first second
of weld time. After 2 seconds, the temperature differential drops to approximately 25°C for all preheat levels and remains relatively constant throughout the end of the weld duration (in this case ~ 11 seconds total). This rapid thermal homogenization across the weld plane indicates that the contribution from the preheat treatment is rapidly eclipsed by the friction and deformation heating that takes place at the weld plane. It is also evident that small increases in weld plane temperature promote significant differences in upset behavior.
Chapter 4: Conclusions

Inertia friction welding of LSHR to Mar-M247 under conditions of constant welding energy, $E_{ko}$, and axial compression force, $P$, but different flywheel moments of inertia, $I$, were conducted to establish the effect of $I$ on the efficiency of the IFW process and the quality of bonds of dissimilar superalloys. Inertia friction welding of LSHR to LSHR under conditions of constant moment of inertia, $I$ were conducted to establish the effects of kinetic energy $E$, axial load $P$, and initial rotation speed $\omega$ on the weld process and resultant material. A finite element process model was developed to provide insight into the effects of various weld parameters on weld behavior. Barriers to effective modeling were explored. From this work, the following conclusions were drawn:

1. The metal flow and microstructure response during IFW of the dissimilar superalloys depend strongly on $I$, despite constant $E_{ko}$ and $P$. In particular, the processing time, deformed volume, maximum temperature developed in HAZ, and the length of sample upset, all increase with an increase in $I$. At a given welding energy, the weld quality (i.e. integrity and strength) improves with increased $I$.

2. The lack of bonding at the weld interface, prevalent in welds where small moments of inertia are utilized, is associated with limited plasticity of Mar-M247 which has a higher $\gamma'$-solvus temperature and higher hot-working flow stress in comparison to LSHR. Limited metal flow leads to trapping of remnant oxide and carbide particles at the
weld interface which is associated with poor bond quality and poor post-IFW tensile strength.

3. The simultaneous increases in the maximum temperature near the weld interface and the duration of IFW with increased $I$ promote increased plastic flow and dynamic recrystallization of a larger volume on the Mar-M247 side. The fine-grain structure formed at the weld interface effectively reduced the high-temperature yield strength of Mar-M247 and promoted more extensive deformation and flow of the material at the weld interface. As a result, the weld interface was free of the oxide and carbide particles and associated weld defects that are detrimental to bond quality.

4. The significant effect of $I$ (at fixed values of $E_{ko}$ and $P$) on welding and post-welding behavior was rationalized by its influence on the amount of energy lost to friction of the rotating components of the welding machine. It was established that an increase in the flywheel moment of inertia decreased the fraction of the weld energy lost to parasitic sinks within the IFW machine. Therefore, increased moment of inertia increased the efficiency of the conversion of the kinetic energy of the flywheel to thermal energy at the weld interface. Therefore, careful consideration of both $I$ and $\omega$ must be given when determining weld process parameters, simply assuming parameters based only on input energy is not sufficient.

5. The efficiency of the IFW process must be quantified in order to fully define the interrelation between the IFW process parameters and their effects on weldability and weld quality.
6. The apparent change in behavior at the weld line from sliding friction to sticking condition occurred at approximately the same number of revolutions of the weld sample during IFW. During sliding friction, the apparent coefficient of friction between the welding surfaces was very low, ~ 0.05, and had a minor dependence on the rotational velocity and moment of inertia. When the sticking condition occurred at the end of the welding process, the apparent coefficient of friction increased rapidly and approached the values of ~0.12-0.14.

7. Two distinct regimes in upset behavior were discovered. The multi-step upset was associated with high initial kinetic energy parameters and large upset lengths. Single-step upset was associated with low initial kinetic energy parameters. This multi-step upset behavior was replicated in the finite element process model by modulating the friction coefficient. This behavior was rationalized by changes in weld line temperature due to rapid upset rates.

8. A sound bond was determined to be one in which no gross welding defects were present at the weld interface and during room temperature tension testing, failures occurred away from the weld line.

9. A criterion was established that linked the minimum initial kinetic energy required for sound bonding to upset length. This criterion was based on minimizing the length of upset required to remove all initial material at the weld interface (to include oxides, defects, surface films etc.). A relationship between this criterion and the change in upset behavior described by the load multiplied by kinetic energy was demonstrated.
10. It was demonstrated that the upset length was a function of the axial load multiplied by the initial kinetic energy. A change in upset behavior was discovered at the low end of this regime. The changeover from the linear behavior to the power-law behavior corresponded with the minimum upset estimated for a sound bond.

11. With increased axial load, the energy input rate increased, upset length increased and heating rate increased. It was shown that this parameter is especially important for welding difficult materials such as Ni-base superalloys as time at elevated temperature is critical to promoting increased deformation at the weld line and developing a sound bond.

12. For welds produced with sound bonds (i.e. no weld line defects), room temperature tension behavior was not a strong function of the welding parameters. Welds that exhibited upset larger than required for sound bonding did not show any significant improvement in room temperature tension properties. Welds that exhibited weld line defects tended to fail at the weld line at strength values below the parent material.

13. Evidence of melting and solidification was shown for dissimilar welds. The occurrence of this localized melting at the weld interface is likely the source of the very low effective friction coefficient values calculated.

14. A method to estimate the efficiency of the transformation of initial kinetic energy to heat at the weld interface was developed. The estimated efficiency values for the weld parameters in this work approached ~50%, which is much below those which have been estimated in the open literature (85 – 95%).
15. Improved joinability was demonstrated experimentally through preheating the Mar-M247 side immediately prior to welding. This increased heat input at the beginning of the weld process increased the amount of deformation of the Mar-M247 during IFW. This increased temperature promoted increased mechanical mixing at the weld interface. The increased upset length of the harder to deform material was attributed to higher temperatures throughout the weld process as compared to the LSHR side.

16. Regardless of the initial preheat temperature; the difference in temperature between the non-preheated and preheated sides approached a steady-state value after two seconds of weld duration. Finite element model results predicted that the majority of the deformation (upset) occurred after the temperature difference between the non-preheated and preheated sides approached a steady-state temperature (25 – 35°C). These model results suggest that a relatively modest temperature increase across the weld line can improve deformation and therefore joinability in the harder to deform material.
Chapter 5: Future Work

Finite element modeling is critical to further development and deployment of inertia friction welded applications in industry. A fully-predictive process model would greatly increase the welding engineer’s ability to develop weld parameters for difficult to weld materials. The continual push to develop engines that run hotter, with fewer emissions at higher compression ratios will push the limits of current design and materials capability. In this trend, dissimilar welds between difficult to weld materials will become more common. Utilizing process modeling to enable efficient scaling of weld parameters from lab-scale to production-scale is necessary. To this end, the weld process efficiency is an area that has been altogether ignored in the literature and requires further investigation. The second area of further study that could have profound impact on the ability to produce sound bonds between dissimilar materials is incorporation of preferential preheating immediately prior to the weld process. The third area that warrants further investigation is the development of a model for the evolution of the bond process that takes place at the weld line. Understanding the development of the bond during the weld process may provide further insights into optimization of the welding parameters. The fourth area that warrants investigation is related to the weld equipment itself. If one were able to predict the bonding behavior during the weld process it could be advantageous to either alter the parameters or stop the rotation during welding. This
would require equipment modification to allow braking the rotation and some method for real-time process control & monitoring. Parameters that could be monitored include torque, acoustic response, hydraulic fluid temperature etc. Another methodology to address improved weldability could be to add a secondary heating source to the weld process. An induction heating source could be used to control heating/cooling rates, and peak temperature during the weld process. This could be particularly effective in controlling the microstructure evolution during cooling, and reduce the occurrence of post-weld cracking within the heat affected zone. Another benefit would be in promoting deformation of the harder-to-deform material in a dissimilar weld. In essence, these suggestions would be an attempt to modify the welding process and equipment to accommodate the specific behavior of the materials to be welded. These four areas outlined above are ripe for further investigation and development.
References