The Effects of Tool Texture on Tool Wear in Friction Stir Welding of X-70 Steel

THESIS

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Friction Stir Welding (FSW) was invented by TWI in 1991 and primarily used to join aluminum (Al) alloys. The cost benefit of FSW in Al has led to an increased focus on the use of FSW to join hard metals such as steel, titanium (Ti), and nickel (Ni) based alloys and the materials used to join hard metals. The welding of hard metals creates an extremely high stress region coupled with high temperatures (up to 1300 °C in some regions) which promotes tool degradation. Tool degradation and cost of the material are currently the largest challenges to the implantation of FSW of hard metals in industry. Previous studies have focused upon investigating alloys suitable to withstand the harsh environment associated with joining hard metal and tool degradation mechanisms. The studies have found that W-based alloys are suitable for joining steel alloys up to 19-mm thick. The studies have also displayed that tools degrade by deformation, abrasive wear, and adhesive wear. Other studies displayed that tool material composition has a large effect of the tool’s susceptibility to tool degradation and degradation mechanisms.

Previous work has failed to account for the effects of tool microstructure on tool degradation. This study focused upon the effects of tool microstructure on tool material performance. W-Re was selected for this study. Two microstructures were investigated in this study and were an extruded microstructure and a recrystallized microstructure. The investigation was conducted by first performing a short parameter development, followed by a tool life investigation on X-70 steel at identical welding parameters and identical
tool designs for each tool material. Subsequent weld passes were ran until a defect was formed, and then the tool was redressed (remachined) to its original profile. The process was repeated until the tool could no longer be redressed. Tool degradation was characterized by pre and post weld microstructure characterization and geometry measurements taken after each weld and redress. The study demonstrated that an extruded microstructure was 27.5% more resistant to tool degradation than the recrystallized microstructure. Both tools lost the majority of the material from the material taper, and primarily by abrasive wear. The recrystallized microstructure also experienced intergranular failure, which may have accelerated the rate of the abrasive wear.
Dedication

To my hero, my Dad
Acknowledgments

I would like thank and acknowledge the following individuals and institutions for their contributions to this work. I would to thank EWI for the uses of their facilities, equipment, met lab, mechanical test lab, and knowledge staff. I would like to thank Todd Leonhardt and Rhenium Alloys Inc. for the donation of the tool material. I would also like to thank Min Hyun Cho and POSCO for the donation of the steel for this project. Without their donations, a project of this size would not have been possible and for this, I am eternally indebted and grateful. Several individuals also deserve recognition. First, I would to thank Brian Thompson for his guidance and knowledge throughout the entire process; you are a great friend and mentor. I also would to thank Tim Moore for his tireless efforts throughout this study. Additionally, I would to thank James Cruz for helping me achieve a balance throughout this work. Lastly, I want to thank my advisor, Dr. Suresh Babu. Thank you for tireless pushing me, and bestowing me with your patience and drive. To all of you, I cannot express how much I appreciate your contributions and guidance, thank you.
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<tr>
<td>Al</td>
<td>Aluminum</td>
</tr>
<tr>
<td>AS</td>
<td>Advancing Side</td>
</tr>
<tr>
<td>BCC</td>
<td>Body-Center Cubic</td>
</tr>
<tr>
<td>C</td>
<td>Celsius</td>
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<tr>
<td>CIP</td>
<td>Cold Isostatic Press</td>
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<tr>
<td>Cr</td>
<td>Chromium</td>
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<tr>
<td>CTOD</td>
<td>Crack Tip Opening Displacement</td>
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<tr>
<td>Cu</td>
<td>Copper</td>
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<tr>
<td>CVN</td>
<td>Charpy V-Notch</td>
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<tr>
<td>EDS</td>
<td>Energy-dispersive x-ray spectrometry</td>
</tr>
<tr>
<td>Fe</td>
<td>Iron</td>
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<tr>
<td>FEA</td>
<td>Finite-Element Analysis</td>
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<tr>
<td>FS(A)</td>
<td>Ferrite with Aligned Second Phase</td>
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<tr>
<td>FSW</td>
<td>Friction Stir Welding</td>
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<tr>
<td>GMAW</td>
<td>Gas Metal Arc Welding</td>
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<tr>
<td>HAZ</td>
<td>Heat Affected Zone</td>
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<tr>
<td>HfC</td>
<td>Hafnium Carbide</td>
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<tr>
<td>HIP</td>
<td>Hot Isostatic Press</td>
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<tr>
<td>HSLA</td>
<td>High Strength Low Alloy</td>
</tr>
<tr>
<td>HSS</td>
<td>High Strength Steel</td>
</tr>
<tr>
<td>HV</td>
<td>Vickers Hardness</td>
</tr>
<tr>
<td>ipm</td>
<td>inch per minute</td>
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<tr>
<td>ksi</td>
<td>kips per square inch</td>
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<tr>
<td>La</td>
<td>Lanthanum</td>
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<tr>
<td>Lbf</td>
<td>Pound Force</td>
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<tr>
<td>M</td>
<td>Martensite</td>
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<td>Mo</td>
<td>Molybdenum</td>
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<tr>
<td>Ni</td>
<td>Nickel</td>
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<tr>
<td>PcBN</td>
<td>Polycrystalline cubic Boron Nitride</td>
</tr>
<tr>
<td>PF</td>
<td>Primary Ferrite</td>
</tr>
<tr>
<td>RAI</td>
<td>Rhenuim Alloys Incorporated</td>
</tr>
<tr>
<td>Re</td>
<td>Rhenuim</td>
</tr>
<tr>
<td>ROW</td>
<td>Right of Way</td>
</tr>
<tr>
<td>RPM</td>
<td>Revolutions per minute</td>
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<tr>
<td>RS</td>
<td>Retreating Side</td>
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<tr>
<td>RX</td>
<td>Recrystallized tool material</td>
</tr>
<tr>
<td>SEM</td>
<td>Scanning electron microscopy</td>
</tr>
<tr>
<td>σ</td>
<td>Sigma Phase</td>
</tr>
<tr>
<td>Si</td>
<td>Silicon</td>
</tr>
<tr>
<td>UTS</td>
<td>Ultimate Tensile Strength</td>
</tr>
<tr>
<td>VPT</td>
<td>Variable-penetration tool</td>
</tr>
<tr>
<td>W</td>
<td>Tungsten</td>
</tr>
<tr>
<td>XT</td>
<td>Extruded tool material</td>
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1.0 Introduction

Friction Stir welding (FSW) was invented by TWI in 1991 and used primarily join aluminum alloys.\(^{(1)}\) Since the invention of FSW, extensive studies have been done investigating the benefits of joining aluminum using FSW. A few major benefits of using FSW as opposed to arc welding include the reduction in the HAZ size, as a result of lower peak temperatures during welding, the ability to join alloys incapable of being fusion welded, autogenous welding, increased corrosion resistance in the weld zone, and lastly increased strength ductility and toughness.\(^{(2-5)}\) Deflaco conducted a cost analysis of using comparing GMAW and FSW to join 2XXX and 7XXX series Al alloys in a butt joint configuration. The results of the study showed that FSW possesses a large economical advantage, in addition to providing superior mechanical properties, such as a stronger HAZ, no need for filler metal, and a greater joint efficiency. The study suggests that FSW is approximately 40% more cost effective per unit weld length and is capable of producing over twice the weld length per hour when compared to arc welding.\(^{(5)}\) The success and benefits seen in joining aluminum led to research in other areas such as dissimilar material welds, and other soft metals such as magnesium, and copper.\(^{(6,7)}\) FSW shows a large advantage for dissimilar welding applications due to the lower welding temperatures leading to significantly less intermetallic formation.\(^{(7)}\)
Recently, there has been an increased interest in the FSW of hard metals. In recent studies, steel, stainless steel, titanium, and nickel have all been joined using FSW.\(^{(8-10)}\) Welds in thickness up to 19mm in steel, 25mm in Ti, and 6.5mm in Ni-based alloys have been completed using FSW. In steel alone, the stir zone often exhibits overmatching tensile strengths due to the reduction in grain size in the weld nugget with very little distortion.\(^{(10)}\) The main limitation in adapting FSW for use in hard metals has been finding a tool material suitable to withstand welding temperatures (above 1200ºC), high flow stresses, and large loads implicit in these applications.\(^{(11)}\) Such conditions can cause excessive wear, deformation, or fracture of the tool itself.\(^{(12)}\) Despite these challenges, there are considerable benefits associated with friction stir welding hard metals. Specifically, there are a number of cost benefits associated with using FSW to join aluminum, and there is interest in applying these benefits to steel.\(^{(5)}\) However, unlike aluminum, there exists a plethora of arc welding knowledge and capability to join steel. Arc welding offers a low initial cost, and is generally accepted by industry. To be competitive, FSW must show a large economical promise in addition to its demonstrated advantage in mechanical properties.

One area of industry which could greatly benefit from the use of FSW is the oil and gas industry.\(^{(13)}\) Pipeline construction is key to this industry and is a critical area of growth for these companies as global demand for natural gas is continually rising and will be an estimated 55% higher than in 2005 by the year 2030.\(^{(14)}\) Pipeline construction involves many facets and stages, but can be summarized as constructing a right of way (ROW),
trenching, laying the pipe, welding, protective coating, filling the trench, and lastly restoring the land. Pipeline construction is mostly performed by specialized teams of people and is done in waves or more commonly referred to as “spreads”. Each team has a specific task to perform and moves from point to point along the pipeline performing their task. Currently mechanized GMAW welding is used to perform the welding on the pipeline. Welding stations containing mechanized welding equipment are transported using boom cranes. Each station is a protective shelter and is set over a prepared weld joint. Once set several welders and helper enter the welding station and perform several welds on the joint. Upon completion of the weld the station is removed and moved down the pipeline. Manpower and equipment are usually a large portion of cost in pipeline construction. Offshore applications are similar in nature but the pipe is fed through a barrage and then typically constructed using a J-lay process, although many other processes are used. A typical barge rental is $445,000 USD per day, thus maintaining production is crucial during offshore applications.\(^{(15)}\)

Based upon studies conducted by Kumar et al, FSW offers approximately a 7% savings for onshore applications and 25% savings for offshore applications.\(^{(15)}\) Deflaco also estimates a typical energy savings of around 65-80% over traditional arc welding and total savings of approximately 40% in general applications.\(^{(13)}\) The majority of the savings seen for onshore applications are a reduction in equipment needs and manpower. It should be noted that based upon this study, high tools cost and short tool life would result in an economic loss for use of FSW over GMAW. Thus the drive for cost effective
tool material is key to making FSW economically feasible in onshore pipeline manufacturing. Offshore applications see a majority of the savings in a large reduction in cycle time stemming from FSW’s ability to complete welds in a single pass. The study showed the savings of using the process greatly increased with thicker welding sections, faster travel speeds and longer tool life.\(^{(15)}\) The typical wall thickness used in offshore applications is around 20mm, and currently only W-based alloys have shown the ability to weld thickness greater than 6-mm. Thus there is also a large driver to produce thick section welds in steel in a single pass and maximize tool life in these welds.

Through tool material studies, two main tool materials have emerged for FSW of steels. These include ceramic based tools and refractory based tools.\(^{(13)}\) PcBN is polycrystalline boron nitride and is second only to diamond in hardness.\(^{(16)}\) Aluminum nitride is typically the binder used for PcBN. PcBN has been demonstrated to show little wear when welding steel but often fails catastrophically during welding.\(^{(11)}\) Also PcBN is generally limited to 6.4-mm of weld depth according to open literature. This however, may be changing through the use of new composite materials, such as Q-60, Q-70, and Q-80, which use W-Re as binder for the PcBN.\(^{(17)}\) These new tool materials have shown the ability of welding a wide range of industrial steels for long weld lengths (40-60-m) before fracture. According to open literature, the thickness is still currently limited up to 6-mm, however development is under way to increase this depth.\(^{(17)}\) W-Re is a much tougher material and is currently capable of welding steels up to 19-mm thick.\(^{(18)}\) W-Re has also
shown the ability to weld a variety of industrial steels, including X-80 with minimal wear.\textsuperscript{(19)}

Joining steel is in thickness of less than 12.5-mm is readily accomplished in literature and has been accomplished using both PcBN and W-based alloys.\textsuperscript{(2, 10, 20-23)} Low carbon steels such as X-65 through X-100 have easily been joined in the butt joint configuration in multiple positions using FSW.\textsuperscript{(22, 24)} Even lap T-joint configurations have successfully been joined using FSW.\textsuperscript{(26)} Some steels have actually shown superior joinability, tensile strength, and toughness, when using FSW as opposed to traditional arc welding and these alloys are nitrogen enriched steel and interstitial free steels. The benefits are largely due to the welding occurring in the single phase ferrite region even at peak temperatures.\textsuperscript{(26, 27)} Typically the weld metal yield strength and ultimate tensile strength is greater than the base material but the weld does suffer a slight loss in ductility.\textsuperscript{(10,28, 29)} Hardness is often associated with strength in steel and thus it is not surprise that the stir zone is often harder than the base metal.\textsuperscript{(11, 28, 30, 31)}

The toughness of friction stir welded steel has conflicting data in the current literature. Some welds in thicknesses less than 6-mm seem to produce desirable toughness properties with near base metal toughness.\textsuperscript{(27, 32)} Feng et. al. place the CVN in TL position with the center of the CVN in the center of the advancing side and excellent CVN properties were recorded in the weld metal.\textsuperscript{(32)} Lakshiminarayanan et al. did not list the location of the notch, but also recorded desirable fracture toughness values.\textsuperscript{(27)} In
more recent work by Lakshminarayanan et al., AISI 1018 was welded using FSW and the notch was placed at the weld centerline. The CVN values were at best 34% lower than the base metal toughness.\textsuperscript{(29)} At larger thickness up to 12mm, similar toughness data exists. Work done by Matsushita et al. demonstrated that the toughness in the weld varied based upon notch location with the advancing side generally showing more toughness than the retreating side.\textsuperscript{(23)} Fairchild et al. performed CTOD work in 12-mm thick steel also. They tested a variety of steels at -20 °C at the weld centerline and the results varied from acceptable toughness values to very low and unacceptable toughness values. The work also stated that the centerline typically had the lowest toughness values and hence why the CVN notches were placed on the centerline for their study.\textsuperscript{(33)} Similar work done by Kumar et al. also demonstrated that FSW increases the ductile to brittle transition temperature in the weld metal. The work compared PcBN and W-Re and it appears that the PcBN tools cause a larger change in the ductile to brittle transition, although both produce ductile to brittle transition temperatures which are not desirable by industrial standards.\textsuperscript{(10)} Double-sided welds on 12-mm steel were performed by Santos et al. and these welds produced acceptable toughness in both the stir zone and HAZ. Friction stir welds experience the highest temperature during welding within the weld nugget near the surface of the weld.\textsuperscript{(33)} In thin section welding, conduction can easily transmit this heat throughout the thickness of the material to produce a small temperature gradient throughout the weld. As section thickness increases, so does the heat imbalance and thus higher welding temperatures may be seen near the surface of the weld to achieve the necessary heat throughout the remainder of the weld. The increased heat input in weld
could then explain the decreased toughness values as current views in literature that suggest the reduction in toughness is likely related to large prior austenite grain sizes, and the subsequent transformations which occur during cooling. Thus the higher heats would result in larger prior austenite grain boundaries and reduce the weld metal toughness. The reduction in toughness is currently not understood by open literature, however it is believed to be related to either impurities present at prior austenite grain boundaries, or the resulting morphologies seen due to larger prior austenite grain boundaries.\(^{(10, 33)}\)

Nelson et al. used orientation imaging microscopy and optical microscopy on HSLA-65 steel to identify the microstructure in friction stir welds consists of primarily lath upper bainite and prior austenite grain sizes in the stir zone were as large as 50 μm.\(^{(34)}\)

Friction stir welding of steel with thicknesses greater than 12-mm is not readily understood by literature. Only limited examples can be found in literature of single pass joints greater than 12-mm and all welds were made using W-Re.\(^{(10, 18, 33)}\) PcBN cannot currently be used for thick section (>6-mm) steel welding according to open literature due to the size limitations of the HIPing process. The technical challenges of welding thick section steel are examined in depth by Thompson and Seaman and will be summarized below.\(^{(19)}\) First, thick section steel produces larger amounts of heat and the issue of thermal management and how to adequately remove heat from the joint becomes key. Several methods of removing heat were examined and they included running a cooling medium such as forced air or a cooling liquid over the surface, internal cooling of the FSW tool, and cooling the welding anvil itself.\(^{(19)}\) The most common method of
thermal management is cooling the tool itself, and is commonly done by drilling an internal passage within the tool and running chilled water through the passage.\textsuperscript{35} If too much heat is removed from the weld, a defect will form within the weld. Additionally, due to the higher heat input, tool design becomes crucial. The tool must be able to withstand high torsional, compressive and bending stress, while still provide additional down force to assist in consolidation at temperatures exceeding 1300 °C.\textsuperscript{19, 33} Wear is even more prevalent at these stages so the tool must be resistant to multiple types of wear. Design can help minimize these factors but the material itself must be able to withstand the process at a fundamental level. To summarize, finding a durable, wear resistant, and cost effective tool material is the largest challenge currently present in the welding of steel.\textsuperscript{36} Currently, W-Re is the only material that has successfully welded steel in a single pass at thicknesses greater than 12-mm and met the requirements outlined above.\textsuperscript{18}

W-based alloys offer the most promise for joining thick section steel. Currently the high cost of tool material and the limited life of the tool are the largest challenges to implementing FSW of thick section steel in industry.\textsuperscript{19} The W-based alloys do not typically fail during welding, but rather degrade until the tool can no create a defect free weld.\textsuperscript{12} Several studies have been conducted in the area of tool wear and the focus has varied from gross geometrical changes to identifying the mechanisms by which tool wear occurs.
Early work in the study of tool wear focused mostly on geometrical changes which occurred to the tool during welding.\(^{(11, 28, 37, 38)}\) These studies investigated the effects of plastic deformation and lost material due to wear on a FSW tool. Wei Gan et al. investigated the welding of L80 steel using FSW. Their study focused upon modeling a W-based tool material using finite element analysis. The finite element model estimated a 7 percent loss in volume for a commercial pure tungsten tool used to weld the L80 steel. In order to resist the deformation, the work suggested that a tool material would need at least 400 MPa of yield strength at 1000\(^\circ\)C. The study also concluded that the loss in volume was caused by both deformation and wear.\(^{(39)}\)

Weinberger et al. investigated the main drivers of the welding process upon tool wear and the failure mechanism of W-based tool material when welding steel. The study focused upon characterizing tool wear through observations, modeling of the W-based material, the relationship between tool wear and welding parameters, fracture, and identifying failure mechanisms. The study had several interesting conclusions. The first is that in this study, the majority of the wear of the W-based tools occurred at the pin of the tool. Secondly, that the spindle speed had a much larger effect on the amount of wear than the travel speed, although both did affect tool wear. Increased spindle speed increased the wear in the tool, while increasing the travel speed increased weld forces, but decreased the amount of wear on the tool. Based upon optical microscopy, the study also concluded that abrasive and adhesive wear were the main causes of wear, with abrasive wear being the larger contributor of the two.\(^{(38)}\)
Abrasive wear is characterized by gouging, grinding, or scratching abrasion.\(^{(40)}\) During FSW, abrasive wear can occur in two ways. The first is that the substrate material directly abrades the tool material. The second manner, involves the tool material, the substrate, and particles removed from the tool itself. The particles themselves abrade the tool material and remain in the flowing material. As a result of abrasive wear, particles of the tool material will be deposited into the substrate and this can be seen in literature\(^{(29)}\). Typically abrasive wear leaves a very distinct surface upon the material. These distinct regions can be seen in Figure 1.\(^{(41)}\)

Adhesive wear is commonly known as galling or scuffing.\(^{(40)}\) This wear is largely driven by diffusion of substrate material with the tool material, when the substrate is stuck to the tool. This stuck material can form intermetallics on the surface of the tool.\(^{(42)}\) This intermetallic is brittle and easily broken off, and thus reducing the volume of the tool. Stuck material can also form a stress concentration, which in turns increases the likelihood of material being removed from the tool. When welding steel, adhesive wear is less prevalent than abrasive wear and deformation. In work done by Thompson et al., diffusion models between tungsten and steel were created to identify tool wear due to abrasive wear in steel. The model showed minimal diffusion and interface movement between the W and steel.\(^{(42)}\) Micrographs of actual steel welding using W-Re showed no interdiffused layer between the steel and W-Re.\(^{(12)}\) The work by Weinberger et al. also
concluded that abrasive wear was the greatest cause of wear when welding steel, and adhesive wear was only a lesser part.

Lastly, plastic deformation is a major component of tool degradation during the FSW of steel. As discussed by Wei Gan et al., geometrical changes due to stress can occur in tungsten alloy when welding steels. The deformation is commonly referred to as “mushrooming” as the height of tool was depressed and the width increased. The study concluded that plastic deformation was the major cause of the tool degradation.\(^{38}\) Commonly twinning and slipping of the W-based structure is observed as a result of plastic deformation in a FSW tool.\(^{38, 39, 12}\)

Thompson and Babu investigated the effects of FSW on pre and post tool welding microstructures. The tool was welded with parameters which promoted tool degradation. Three unique tool materials were investigated in this study, W-La, W-Re, and W-Re-HfC and each was used to join 12-mm thick steel. Each material experienced tool degradation in a different form, with the W-La primarily wearing due to deformation, the W-Re was primarily due to twinning, and the W-Re-HfC wore primarily due to intergranular failure. In this study, the W-Re-HfC was the most wear resistant material; however, its grain size was significantly smaller the W-Re grains, especially near the edge.\(^{43}\) Although the HfC carbide particles did indeed strengthen the material, it is unclear if the composition or the microstructure played the larger role in resisting wear.
Although limited data exists as to the wear performance of W as a tool material, a large amount of information exists about tungsten alloys and tungsten heavy alloys. A tungsten heavy alloy is typically a W base alloyed with a mixture of Cobalt (Co), Nickel (Ni), Iron (Fe), chromium (Cr), and copper (Cu). The primarily purpose of these additions is to act as a matrix for the W grains. Studies have shown the W grains primarily deform in there alloys, until a threshold is reached (typically around 30%), and the matrix phase begins to deform. The deformation of the matrix causes W-W grain boundary decohesion and greatly accelerates deformation. Microstructures with second phase precipitates at the grain boundary typically exhibit more strength, and failure occurs by cleave of the tungsten grain. When second phase particles are not present at the grain boundary, grain boundary decohesion is the typical failure mode.\(^{(44)}\) Other tungsten heavy alloy research has concluded that a decrease in grain size increases the tensile strength of the alloy, however the studies did not quantitatively measure grain size.\(^{(45,46)}\) Research has shown that microstructure before extrusion affects properties after extrusion. A pre-extrusion fine grain alloy after extrusion has a slightly high ultimate tensile strength but only exhibits about two thirds of the ductility as seen in a pre-extrusion coarse grain alloy after extrusion.\(^{(47)}\)

Briant investigated the effect of microstructure of pure tungsten on failure mode. The investigation examined tungsten rods in the swaged and drawn condition, and focused upon the effect of heat treating these microstructures on fracture stress and mode of failure. In general, heat treating was detrimental to strength and the higher the heat
treatment hold temperature, the less stress which was required to fracture the sample. The failure mode was always cleavage but the crack initiated at patches of traverse grain boundaries, (boundaries perpendicular to the drawing direction) which were nearly in plane. Heat treating above 1000°C effectively reduces the grain length and increases the odds of traverse grain boundaries being nearly in a plane in any given location (as shown in the second image in Figure 2). At more extremely heat treatments (temperatures above 1600°C), secondary grain growth occurred. The stress to failure was lower in the samples in which secondary grain growth occurred. The alignment of the grain boundaries after secondary grain growth is much more random than the perpendicular grain boundary intersections seen in the drawn samples. The random orientation of the grain boundaries allow for a lower energy crack propagation path than the drawn microstructure. The difference in crack propagation path can easily be seen in Figure 2. The study could not identify why grain boundaries were always the crack initiation point. Potassium was the only additional element besides tungsten which was present at the grain boundaries and currently no correlation can be made between potassium and grain boundary failure in tungsten. Overall the study showed that traverse grain boundaries lower the strength of the material and secondary grain growth further reduces the strength of the material.\(^{(S44)}\)

Lee et al. also investigated failure behavior of pure tungsten in both the wrought and recrystallized form. The study found that in general the flow stress-strain behavior is dependent on the strain rate. Higher strain rates allow tungsten to accommodate more stress and less strain. As to the effect of microstructure, the study found that at a given
strain rate, the recrystallized material exhibits a higher ductility than the wrought material, but the recrystallized material exhibits a lower flow stress due to the decrease in dislocation density during recrystallization. Lastly, the study concluded that cracking in the wrought material was dominated by transgranular cleavage, while the recrystallized mainly failure by intergranular cleavage. This may once again be due to the tortuous path the crack must follow along the wrought grain structure, as opposed to the relatively straight propagation path present in a recrystallized microstructure. It is clear that grain boundaries and microstructure play a large role in the strength and failure characteristics of any tungsten alloy.

W-Re is currently the preferred refractory tool material for welding steel. A typical composition used is W-25wt%Re and thus this composition was selected for this study. The addition of Re has several benefits as opposed to pure W. The additions increase recrystallization temperature to 1900°C, improves ductility, lowers to ductile to brittle temperature to -50°C, and increases the ultimate tensile strength of the material. A swage bar of W-25Re tested at 1371°C has a UTS of 546 MPa, a yield strength of 382 MPa, and has a 6.64% strain at maximum stress.

Re is typical added in concentrations up to 27% but room temperature solubility of Re in W is only 20%. At a typical FSW welding temperature of 1200-1300°C would increase this solubility to 27 wt% Re. The Re-W phase diagram can be seen in Figure 3. Thus at room temperature sigma phase (σ) can be present, which has a chemical composition
of W$_2$Re$_3$. Once the solubility limit is reached, a large increase occurs in the material due to the precipitation of $\sigma$ phase along the grain boundaries. Sigma phase is extremely brittle and can have hardness varying from 1300-2000HV which is at least two times the hardness of the alpha phase. The presence of sigma phase also increases the microstructure’s susceptibility to cleavage failure.$^{(54, 55)}$ Re additions also make the microstructure more prone to oxidations. Rhenium additions of 20% have shown large increase of the rate of oxidation at temperatures above 700°C. The study also shows that microstructure can also affect the rate of oxidation as deformed or worked microstructures appear to more susceptible than a recrystallized microstructure.$^{(56)}$

The focus of this study will be to investigated the effect of tool microstructure on tool wear and performance. An extruded and recrystallized microstructure will be investigated. The extruded microstructure is typically used due to the better mechanical properties described above. The recrystallized microstructure is being investigated as a fine grain microstructure may be easier to produce. If the fine grain microstructure is acceptable, the potential to reduce tool manufacturing costs could be reduced by near net shaped HIPing. Near net shape HIPing would eliminate the extrusion step and produce a geometry close to the final tool geometry. This could eliminate 90% of the machining needs and greatly reduce the amount of material lost due to machining, further reducing the cost of the tools. A common practice is to machine a worn W- based tool back to its original profile and continue welding, which is commonly referred to as redressing. Surprisingly, data on the performance of this practice is not readily available in literature.
If this process works successfully, it would greatly increase the useful life of a single piece of tool stock. Thus the performance of redressing will also be investigated in this study in an effort to increase tool life and cost effectiveness.
2.0 Objectives

The objective of this study was to determine the effect of W-Re tool material’s microstructure on tool wear, when used to friction stir weld 19-mm thick X80 steel. Tool wear was investigated by using digital profilometry scans of each tool before and after each weld, pre- and post-weld tool material microstructures, non-destructive evaluations, and fractography of the worn tool surfaces. Additional goals of this study were to establish robust welding parameter of 19-mm thick X-80 steel and observe wear mechanisms and wear patterns over extended weld lengths.
3.0 Approach

Two compositionally identical W-Re based tool materials were selected for this study. The material originated from the same batch with a nominal composition of W-25%Re and each underwent a distinctly different thermo mechanical process to produce two unique microstructures with identical compositions. The microstructures examined in this study were an extruded (XT) microstructure and a recrystallized (RX) microstructure. The extruded microstructure was chosen as it is currently the most commonly used microstructure of W-Re when welding steel. The recrystallized microstructure was chosen because is also used when welding steel, but more importantly it offers alternative processing methods which could significantly reduce the cost of the tool material and machining of the tool. All tool material was provided courtesy of Rhenium Alloys Incorporated (RAI).

The two tool materials each had the same nominal compositions but each has a unique thermomechanical history. The extruded tool material was produced in the following way under powder metallurgy and a variety of thermo mechanical processing techniques. First the material began in powder form and the W was dry blended with Re to produce a mixture of 75 wt% W and 25 wt% Re. The material then underwent a cold isostatic press to consolidate the powder. The compressed powder then underwent sintering, a process without pressure, at 2475°C (4352°F) for 12 hours. The diameter of the bar at this point is in the process is approximately 76-mm (3.0-in) in diameter. Following sintering, the bar
undergoes a hot extrusion process. The bar is heated to 1900°C (3452°F) and is then extruded from its 76-mm diameter to a 43-mm diameter and undergoes approximately 5400-kN (600 tons) of force during extrusion. The extruded bar was completed after the extrusion process. A 102-mm long section was removed from this bar and this portion underwent an additional annealing heat treatment at 2000°C (3635°F) for a time of 1 hour. In other words, the recrystallized material was blended, CIPed, sintered, extruded, and then annealed for 1 hour to produce its unique microstructure, as opposed to the extruded material, which did not undergo an annealing treatment following the extrusion.

Two bars nominally 43-mm (1.7-in) in diameter and 102-mm (4.0-in) long were delivered by RAI with identical nominal compositions and unique microstructures. For microstructural analysis, a 6.35-mm (0.25-in) section was taken from both ends of each bar. These portions were sectioned down the central axis to give a cross section of the pre-weld tool microstructure. One cross section from each piece will be used for characterization, resulting in 4 samples; 2 of the XT material and two of the RX material. These 4 samples were then mounted and used for pre welding characterization of the two materials. The remaining two pieces of 88.9-mm (3.5-in) of tool material were each machined into identical designs. The designs used were a EWI patented variable penetration tool (VPT) design (EWI Patent Nos. 7,234,626, 7,404,510, and 7,416,102). A typical VPT design can be seen in Figure 4. Note the labels on the tool, as these names will be referenced accordingly throughout the remainder of the report.
The substrate material selected for this study was X-70 steel as qualified by API 5L and was donated by POSCO. This material is classified based upon its yield strength 70 ksi (550.58 MPa) and meets compositional standards and mechanical properties as outlined by API 5L. The plate composition as determined by OES Laser is displayed in Table 1. The plate donated by POSCO originally had the dimensions of 4m x 1m x 26mm (157.48-in x 39.37-in x 1.02-in). The plates were not of usable size and were sectioned into usable sizes using water jet cutting. The tool life plates were nominally 1.52-m x 0.50-m x 26 mm (60-in x 19.7-in x 1-in) and are displayed in Figure 5. All mechanical test plates had the nominal dimensions of 0.33-m x 0.18-m x 19-mm (13-in x 7.25-in x 0.75-in) and are displayed in Figure 6. The thickness was reduced by milling both the top and the bottom equally. The edges of the plates were also milled to remove mill scale and any oxide layers. The thickness was reduced for mechanical test plates to ensure full thickness welds for testing. All welding was conducted using EWI’s FSW #2 machine, which a converted milling machine. Thus the machine utilizes a stationary spindle and moves the table to create relative movement between the tool and substrate. The machine is able to withstand up to 178-kN (40,000 lbf) of vertical force, up to 35.5-kN (8,000 lbf) in the travel direction, and torque up to 2,040-N·m (1,500 ft-lbs). The working envelope for this machine is 1.5-m x 0.92-m (60-in x 36-in).

A parameter development was then conducted. These welds were conducted using the XT tool material, as this the common material used for welding steel. The trials were done in an iterative fashion with focus both upon process parameters and tool design.
The parameters were developed on the 25-mm thick material, as the majority of the welds done on this study will be the partial penetration tool life welds. The tool design and parameters resulting from this study are by no means optimized but represent a “best” parameter set and tool design based upon this small portion of the study. Wear data was not taken during this portion of the study due to the large range of parameters and the focus on producing an acceptable repeatable weld. A typical in process weld is displayed in Figure 7.

Following parameter development, the tool life study began. The tool life trials were conducted by the following identical procedures for both materials. Both the XT tool and RX tool were machined to the identical design of T03 (Figure 8), as determined in the parameter development. The welding parameters used were a rotational speed of 90 RPM and a traverse travel speed of 1.27-mm/s (3.0 ipm) for both materials. Argon shielding of the tool material was used and the set up can be seen in Figure 9. The argon flow rate was approximately 100 CFH Air. Internal cooling of the tool was also used by running approximately 6.8 LPM (1.5 GPM) through the internal cooling hole located in the center of the tool. All welding was conducting an a 25-mm thick steel anvil.

Before the start of each trial, the tool was remachined or redressed to the original pin profile of T03 in Figure 8. This redress removes material off of the length of tool to restore the tool to the original tool profile. After redressing, the tool dimensions were characterized using digital profilometry. The digital profilometry was conducted by use
of a laser height scanner the Micro Epsilon 2800, which has a linearity of +/- 0.2%. The output of the lasers scan is a line across the tool showing a 2-D profile of the tool. These scans were taken in the same two locations on the tool before and after each weld. Images of the laser scanning system can be seen in Figures 10 and 11.

The tool life trials were conducted in the following manner. Each trial begins with a mechanical test plate at full thickness. The mechanical test plates were done in a butt joint configuration and were approximately 0.3-m (1-ft) in length (Figure 6). Following the mechanical test plate, bead on plate welds nominally 1.5-m (5ft) long were made sequentially until a defect was observed in the weld. Between each weld a laser scan was conducted. The bead on plate welds were made on a single plate and spaced approximately 5-mm (0.2-in) apart on the plate. Defects were identified in two way, through a visual inspection and by radiographic inspection. Each weld underwent radiography after welding. The radiographic set up is displayed in Figure 12. The bead on plate welds were not inspected until the plate was completely filled with welds. Therefore visual inspection was heavily relied upon during this experiment. Once a defect was observed during a 1.5-m weld, the trial was ended and the tool underwent digital profilometry. The tool was then redressed and a new trial began and repeated the process described above. Trials were repeated until the redress would cause the shoulder of the tool to be within 3-mm of the bottom of the cooling hole to ensure the tool did not fracture. This process is outline in the flowchart displayed in Figure 13.
After the completion of the tool life trials, each tool was sectioned along the central axis with water jet. After water jetting one side of each of the tool materials was ground flat to increase the ease of mounting and polishing. This ground piece was then mounted and polished for microstructural analysis, which includes optical microscopy, scanning electron microscopy (SEM), ASTM grain size measurements, (EDS), and hardness mapping.

The scanning electron microscopy was completed using EWI’s SEM to investigate the cross sections of the tool materials. The baseline parameters used for this investigate was a voltage of approximately 20kV, a probe current of 505 pA, a working distance of 8.5-mm. These parameters may have varied for individual photos and can be seen in the desired image. Each sample was examined using secondary imaging, backscatter imaging, and energy-dispersive x-ray spectrometry (EDS). The main areas of interest included the entire pre weld microstructures, pin surfaces such as the shoulder edge, the shoulder taper, the pin center, the pin tip, the cooling hole base, and the tool shank area. All of these areas are labeled on Figure 4. Upon completion of the SEM work, optical microscopy was conducted on the 4 pre welding tool microstructures and the 2 post welding sectioned tools. Once again the tools and the pre weld samples were investigated in the same regions as the SEM. These areas underwent grain size measurements using ASTM grain size measurements.
Following microscopy, hardness testing was conducted on all six tool material samples. Hardness mapping was done on the pin of each tool. The pre weld samples were subjected to rectangle angles. Each map was conducted using a Leco AMH-43 with a load of 500-g and a 1-mm spacing between indents.

Lastly, the welds themselves were investigated. Radiographs were taken of each weld in accordance with ASME Section V Article 2 and the defect quantity and length was characterized for each weld. Following the radiography, mechanical tests were taken out of the initial test plates from the beginning of each trial. The following mechanical tests were taken out of each plate, 2 bend tests, 1 traverse tensile test, 1 macro cross section, and 5 CVN tests taken along the weld center line. Each CVN notch was through the thickness, and centered at the weld centerline, which is referred to as the TL orientation. Each CVN was tested at -20°C. All tensile tests were conducted according to ASTM E8. Bend tests were conducted according to AWS B4.0:2007 and all CVNs were done according to CSA W48-06. CVNs were not taken for welds that failed radiographic inspection. Base metal properties were obtained by performing all the mechanical tests in both the longitudinal and traverse directions. The macro cross sections were investigated using optical microscopy, SEM, and EDS. The focus of this was to characterize weld quality and identify any tool material within the weld.
4.0 Results

4.1 Parameter Development

The parameter development was completed using iterative trials. The initial tool design used was T01 and this tool is characterized by its large included angle. The tool designs can be seen in Figure 8. The parameter matrix performed during parameter development and the welding results can be seen in table T2. The plunge sequence was held constant throughout all weld trials. Weld 1 was performed at 80-RPM, and a traverse travel speed of 76.2-mm/min (3.0-ipm). The weld hit the process force limit of 45-kN (10,000-lbf) in the travel direction and the travel speed was thus reduced to 50-mm/min (2.0-ipm). This was done to protect the tool from fracturing or deforming due to poor parameters and is a common practice when developing a parameter set on any given material. The weld did appear to consolidate but a travel speed of 50-mm/min (2.0-ipm) is not practical. Due to the high weld forces, the RPM was increase to 110-RPM for Weld 2 to generate more heat in the weld and the travel speed remained at 76.2-mm/min (3-ipm). This greatly reduced the weld forces and created a visually acceptable weld. Weld 2 can be seen in Figure 14. Weld 3 was simply a repeat of Weld 2 to ensure repeatable of the weld and the weld was also visually acceptable.

Welds 4-8 were attempts to increase the travel speed of the weld and were made using T01. Weld 4 was made at 140-RPM and 127-mm/min (5-ipm), and appeared to
consolidate based upon visual inspection. The weld forces were high in weld 4, therefore weld 5 increased the rotational velocity to 150-RPM to generate more heat and the travel speed remained at 127-mm/min. A wormhole appeared on the advancing side of the weld, and thus these parameters are unacceptable. This wormhole can be seen in Figure 15. The small size of this wormhole suggests that this parameter is just outside of the parameter window. Thus 150-RPM is likely too high of a spindle speed for combination of this material, thickness, and tool design. Weld 6 was attempted at 135-RPM and 101.6-mm/min (4-ipm) and passed visual inspection. Weld 7 was made at 140-RPM and 114.3-mm/min. A defect near the bottom of the weld corresponding to the pin tip was seen in the exit hole and may be seen in Figure 16. After Weld 7, the tool was redressed to prepare for a butt joint. Weld 8 was a butt joint made at 135-RPM and 101.6-mm/min (4-ipm). The butt joint produced a defect near the pin tip. Weld 9 was a butt joint made at the same conditions as Weld 2 and 3 at 110-RPM and 76.2-mm/min. This weld also produced a wormhole near the bottom of the weld. The tool design T01 seems to be unable to consolidate material near the bottom of the weld in a full thickness butt joint. Thus a new tool design was need.

T02 was similar to the original tool design, with two main changes. The pin tip diameter was increased, but in order to do this, the included angle was decreased. Weld 10 was made with the new tool at 110-RPM and 76.2-mm/min as this parameter appeared to make the “best” weld based upon weld forces and visual appearances. Weld 10 was visually acceptable, and the parameters were then used to make a butt joint. The butt joint
was X-rayed and the inspection did not show any indications or defects. The weld can be seen in Figure 17 and the cross section of the weld can be seen in Figure 18. The cross section did reveal a slight amount of banding near the top of the weld. The weld was consolidated and defect free. A 1.5-m (5-ft) test run, Weld 12, was made to ensure the T02 design and selected parameter set were durable and able to still produce an acceptable weld. A wormhole defect was seen after this test pass and this design was eliminated.

The T03 design was the last and final tool design iteration. The tool was midway between T01 and T02 in both the included angle and pin tip diameter. This was done in order to have a large enough included angle to deal the onset of a wormhole defect, and have a large enough pin tip diameter to delay the formation of a pin tip defect. A butt joint, Weld 13, was made at 110-RPM and 76.2-mm/min. The weld was defect free and passed radiographic inspection. A 1.5-m test was also run as Weld 14 and the weld passed visual and radiographic inspection. These welds were subsequently labeled XT-1 and XT-2 as these welds also symbolized the start of the tool life trials. Weld 14/ XT-1 can been seen in Figure 19. The radiographic test results can be seen in Figure 20, and the metallographic cross section can be seen in Figure 21. The cross section does not show any tool material banding, however the compositional band present in the base metal can be seen across the weld metal.
4.2 Tool Life Trials

The tool life trials consisted of 0.3-m long mechanical test plate welds at the beginning of each trial followed by subsequent 1.5-m passes until a defect was formed within the weld. Each weld underwent radiographic inspection upon completion of the plate. A completed tool life plate can be seen in Figure 22 and a compiled X-ray can be seen in Figure 23. The results of the tool life trial are summarized in Table T3 for the extruded material and Table T4 for the recrystallized material. The tables shown each pass length, the adjusted weld length, which accounts for defects in the weld, the total weld length on the tools, and the tool’s current cross sectional area and current height. The tool cross sectional was calculated using integration over the area shown in Figure 24. The entire pin profile and 5-mm below the initial shoulder were chosen to encompass shoulder wear in addition to the wear on the pin. The weld length data is summarized in Table T5. The extruded material produced a total weld length of 46.33-m and when accounting for defect areas, the tool produced 41.30-m of defect free welds. The extruded material was able to produce about 10 more meters of weld length and about 1.5-m more weld length per redress as opposed to the recrystallized material and showed less variation in weld length per redress than the recrystallized material.

Two main defects were seen throughout the tool life weld trials, a wormhole on the advancing side and a small continuous pore near the pin tip. These are the same defects seen during the parameter development. A typical pin tip defect can be seen in the exit.
hole of Figure 25. Notice the small size of the defect. The pin tip defect was not detected by radiographic inspection. Thus it is reasonable to say that the defect was less than 2% of the material thickness (0.38-mm) as this is the sensitivity of the radiograph. The larger and more frequent defect, the advancing side wormhole defect did readily appear in the radiograph as the defect is much more sizable. A cross section of a wormhole defect can be seen in Figure 26.

4.3 Mechanical Testing

Mechanical test plates were fabricated after each redress to address the effects of redressing on the mechanical properties of the weld. The results are summarized in Figure 27. The base metal exhibited a yield strength of 497-MPa (72.2-ksi) in the traverse direction and 510-MPa (74-ksi) in the longitudinal/rolling direction. The ultimate tensile strength exhibited by the base metal was 613-MPa (89-ksi) in the longitudinal direction and 596-MPa (86.5-ksi) in the traverse direction with elongations of 41.79% and 42.09% respectively. All the tensile were traverse tensile tests and thus the traverse material properties will be used for comparison purposes. On average, the friction stir welds exhibited slightly overmatching ultimate tensile strengths of 610-MPa (89-ksi), and about a 9% loss in yield strength. The elongation is also around 35% in the majority of the tensile tests, representing a 7% loss in ductility. Each of the tensile test failed within the stir zone. The reason for this failure location is demonstrated in the stress strain curves in appendix B. The base metal exhibit much more ductility and is stronger than the weld.
metal at strains above 20%. Thus at 32% strain, where the welds failed, the base metal is over 100 MPa stronger under the given strain. In general, the failure was always on the retreating side of the weld, however; the welds that exhibited a loss in mechanical properties (RX-25 and RX-28) failed on the retreating side through a wormhole on the advancing side. The defect size which affected mechanical properties can be seen in Figure 26 and Figure 28. The strength was the welds were not greatly affected but the defects did greatly reduce the elongation of the material. Other welds failed visual inspection but showed no loss in mechanical properties and still failed in the retreating side of the weld. Bend test were also conducted. All the welds passed the bend test with the exception of RX-10, which exhibited one crack greater than 3.175-mm.

CVN test results can be seen in Table T6. The test was done at -20°C and the notch was placed in TL position with the notch running through the thickness and positioned at the weld centerline. The weld metal exhibited low toughness values in the range of 19.50-36.06 J, which is only about 6.5-11.5% of the base metal toughness at -20°C. It is fairly obvious that the ductile to brittle temperature was raised in the stir zone, as the toughness values are still very high in the base metal at this temperature. The weld metal is likely exhibiting lower shelf behavior whereas the base metal is likely exhibiting upper shelf behavior. This increase in ductile to brittle transition temperature is exhibited by previous research by Thompson et. al. and Fairchild et. al.\textsuperscript{(12, 33)} Although the loss in toughness is likely due to a change in the ductile to brittle transition temperature, more work is needed as these tests were done solely at -20°C. Additionally, the recrystallized material showed
much more volatile results, and thus the standard deviation of the recrystallized material was double that of the extruded material.

4.4 Tool Degradation and Degradation Mechanisms

The tool wear was captured by use of line scans to create a two dimensional profile of the tool. Line scans were taken after each weld and redress. A typical line scan for a tool throughout a trial can be seen in Figure 29. This is indicative for the wear seen throughout the trials. The majority of the wear occurred in the taper of the pin and a small amount of wear occurred at the shoulder. To quantitatively capture the wear, cross sectional areas were calculated after each pass and redress. The cross sectional area results after each weld can be seen in Table T3 and T4, which show the cross sectional area as calculated in Figure 24. Figure 30 shows the entire cross sectional area of the entire tool stock throughout the life of each tool. The large vertical losses in Figure 30, such as the large loss in cross sectional area at 8-m for the recrystallized tool and 10-m for the extruded tool, represent the points at which redresses occurred throughout the trial. In this experiment, redressing accounted for the majority of the material lost as opposed to tool degradation itself.

The wear of the welding portion of the tool which occurred in each trail is displayed in Figure 31. The chart shows a general reduction in cross sectional area with each linear meter of welding. The increase in cross sectional area present at 0.3-m in some trials is
due to material sticking to the tip of the pin after the mechanical test plate. The material was left on the pin to prevent damaging the tool during the removal of the steel. The sticking did not occur in the 1.5-m passes. The data is scattered in Figure 31 due to scatter of the redressed tool profile. The cross sectional areas varied from 629.88-mm$^2$ to 657.4-mm$^2$. Figure 32 is histogram of the cross sectional of the tool after each redress. The scatter of the redressed profile is normally distributed as shown by Figure 33. The probability plot in Figure 33 has two curved lines. 95% of normal distributed data will fall between these lines, and thus it is very likely this data is normally distributed. To account for the scatter, the cross sectional area data was normalized based upon the redressed cross sectional area. Thus every trial started at 1 and wore as a fraction of the original volume. The normalized results are plotted in Figure 34. Both tool materials exhibit linear normalized wear. The extruded material typically lost .69% of its original cross sectional area per linear meter, while the recrystallized material lost .88% of its original cross sectional area per linear meter. Based upon the normalized wear data, the recrystallized tool material degraded 27.5% more over a given length.

4.4.1 Extruded Tool Material Characterization

The mechanisms by which tool degradation occurred in each tool material were investigated using optical microscopy, scanning electron microscopy, and hardness testing. The pre weld microstructure can be seen in Figures 35-37. The pre weld microstructure was characterizing by elongated grains running parallel to the length of
the tool. These grains are nominally 150 μm x 30 μm. The microstructure is relatively uniform throughout the sample, with the exception of some minor twinning on the outside of the bar and a slightly more worked structure near the edge of the bar, both of which can be seen in Figure 37. The twinning is likely a result of the extrusion process and would be removed during the machining of the tool.

The post weld microstructure of the extruded material shows no fundamental changes. The center of the pin (Figure 38 and 39) still exhibits the same morphology and grain size of 150 μm x 30 μm as the pre weld material. This is true throughout the tool material and can be seen in Figures 38-47. Hardness maps were taken of both pre weld tool samples and the post welding tool material. The maps are displayed in Figures 48- 50. XT-1 was the cap of the extrusion and thus was not indicative of the microstructure throughout the bar. XT-2 was used to characterize the pre-weld microstructure. The average hardness for the pre welding sample XT-2 was 524 HV and the post welding hardness average was also 524 HV. The samples show similar hardness values for the pre weld and post weld samples, suggesting the sample did not experience work hardening due to the high weld forces or softening due to the high welding temperatures. The tool itself did experience some fundamental wear. Along the edges of the pin and shoulder, which is the welding surface, fundamental signs of abrasive wear are seen throughout the sample. Figure 40 shows sheared grains near the edge of the sample. The raised area with two separate ridges is nearly the same morphology seen in Figure 1 depicting abrasive wear. Figure 41 depicts a worn edge on the right edge of the pin along the taper. The focus of the
image is the depressed surface near the edge of the tool. This surface clearly depicts a worn surface which may have been sheared from the main body of the tool. The shoulder of the tool also shows signs of abrasive wear. The rolling surface seen in Figure 42 is a sign of abrasive wear and a similar surface can be seen in the diagram in Figure 1.

Small amounts of cracking were seen throughout the tool shank but not near the shoulder or pin. The cracks ran parallel to the tool length and were found between the grain boundaries. Figure 44 shows optical images of the cracks in the shank and Figure 45 displays an SEM of the cracks in the shank. Figure 46 displays the intergranular nature of the shank cracks. The shank cracks are only present in the area of the shank with the cooling hole. Additionally cracks were found near corners of the bottom of the cooling hole (the portion of the cooling hole nearest to the pin) on both side of the tool. The crack appears along the small radius and appears to have originated from the stress concentration present at the corner. The crack is displayed in Figure 47, and appears to be a transgranular crack. This area undergoes a large amount of stress due to the thermal gradient present due to chilling water flowing through the cooling hole. The reduction in area cause by the cooling hole also places a larger torsional load near the area which may contribute the cracking.

4.4.2 Recrystallized Tool Material Characterization
The pre-weld microstructure of the recrystallized microstructure was characterized as equiaxed grains with an ASTM grain size of about 4.5 at the center (grain diameter of about 75 μm) of the material and 5.0 (grain diameter of about 65 μm) near the edge. The pre-weld microstructure can be seen in Figures 51–53. The grains exhibit no signs of deformation or wear and the pre-weld microstructure is free of cracks and voids. The post-weld microstructure has the same grain size as the pre-weld microstructure, and the post-welding microstructures can be seen in Figures 54-61. The post-welding microstructure did exhibit a much harder microstructure than the pre-weld microstructure and the results of the hardness maps can be seen in Figures 63-65. The hardness in the pre-weld microstructure averages 450 HV, as opposed to an average hardness in the post-welding tool sample of 502 HV. This hardening represents a fundamental change in the microstructure, although the grain size remained relatively unchanged. Using SEM, some twinning was identified throughout the post-weld microstructure. Figure 56 displays twinning in the center of the pin, and some twinning was also seen intermittently throughout the microstructure such as in Figure 57.

Deformation, abrasive wear, and intergranular failure were observed in the recrystallized tool material. Abrasive wear similar to the wear seen in the extruded material was identified on the pin tip, pine sides/tapers, and along the shoulders. This abrasive wear is displayed in Figures 57-59. This wear can once again be related to the abrasive wear illustration in Figure 1. Intergranular failure was observed at the corners of the tools such as the point where the pin meets the shoulder (Figure 60) and the points at which the pin
taper meets the pin tip (Figure 61). It is extremely clear that entire grains are missing from these locations. Throughout the remainder of the tool, other signs of intergranular failure were observed, albeit on the pin sides/tapers the effects were not as pronounced. Figure 62 displays a grain located on the side of the pin, which is slightly offset from the other grains along the surface. It appears as though a grain was removed from the region which caused the offset. A smaller void similar to the one in Figure 60 can also be seen in the top left portion of the picture. The small amounts of twinning seen throughout the samples were a result of deformation. Thus the tool did undergo some critical amount of stress and deformation to cause the twinning seen within the microstructure. Overall the recrystallized material showed a similar surface to that seen in the recrystallized material with the addition of twinning and some intergranular failure.

Cracking was also seen in the recrystallized tool material near the bottom of the cooling hole and throughout the tool. Figure 66 shows a crack originating from the cooling hole. Figure 67 shows a higher magnification view of the crack and the crack appears to be intergranular. Figure 68 is a shot into the cooling hole itself. The image clearly shows grain failing intergranularly and the crack runs throughout the entire circumference of the tool. Smaller intergranular cracks can be seen throughout the shank of the tool, similar to the extruded material. Images of the cracks are displayed in Figure 69. Additionally micro cracking was observed in the pin itself and is displayed in Figure 55. It should also be noted that large amounts of twinning were present at the edges of the shank material.
The tool life trials produced 14 mechanical test plates and macro cross sections were cut from each 0.3-m long test plate. Thus by examining these samples, the evolution of the weld microstructure throughout the tool life can be observed. The base metal microstructure is displayed in Figure 70. The microstructure is characterized by polygonal ferrite (PF(I)), and ferrite with a mixture of martensite (M) and bainite (FS(B)). This is typical of most pipeline steels as the thermo-mechanical processing creates a very small grain size with a mixture of soft constituents such as ferrite with hard constituents such as martensite and bainite, All microstructural analysis was done by optical microscopy and may have inaccuracies due to the nature of the investigation and small grain size. A typical weld microstructure is displayed in Figures 71-73. The weld metal microstructure appears to consist of ferrite with an aligned second phase and a larger second phase of polygonal ferrite. The ferrite grain size within the stir zone is larger than that of the base metal. The microstructure does not show any changes between the advancing and retreating side, but does show a difference in the through thickness microstructure. As the distance from the weld surface increases the ferrite grain size decreases, and the second phase alignment becomes finer. Figure 73 shows the microstructure near the root of the stir zone and the ferrite grain size is still larger than the base metal, but smaller than the grain size near the top of the weld. The extruded tool produced a nearly identical microstructure throughout the weld as the recrystallized material. Weld metal from the top of the weld from both the advancing and retreating
side are displayed in Figures 74-77. The final weld cross sections showed the polygonal ferrite grain size reduced, and martensite began to form in the weld metal. The cross sections were examined and one major trend was observed, as the tool life trials progressed, the polygonal ferrite grain size in the stir zone decreased steadily with each redress in both tool materials, and martensite/bainite began to form in the weld metal. It is extremely to distinguish the two optically and further investigation is needed to understand which morphology is truly present in the welds. Photos of the last two mechanical test plate welds are displayed in Figures 77 and 78. The primary ferrite grain size near the top of the weld in the final welds is comparable to primary ferrite grain size near the bottom of the original welds at the start of the tool life. The grain size is reduced throughout the entire stir zone and the grain size in the HAZ is also reduced. This reduction is once again, incremental with each redress of the tool. The only major change between the final welds and the initial welds would be the distance between the cooling hole and the pin itself, as this distance decreases with each redress. The water flow rate through the cooling hole was held constant throughout welding, and thus the grain size in the later welds suggest that the welds themselves either experienced a lower peak temperature, and/or a faster cooling rate.

Examination of the weld defects themselves seen in the cross sections revealed tungsten alloying near each defect. The microstructure in the surrounding matrix remained unchanged and did not display any noticeable differences from the remainder of the weld metal. Immediately adjacent to each of the defects seen during the experiment a region
appeared which was not etched by the nital solution. The region was not always seen on both sides of the defect and an indicative sample is displayed in Figure 79. The sample was placed in the SEM and backscattering imaging (Figure 80) clearly displayed a change in material density near the defect. EDS was conducted near the interface at several location to determine the composition of the denser region. The EDS results are displayed in Figure 81 and it clearly displays W-Re in the steel. Regions similar to the ones displayed here were seen throughout the tool life, and the nital etchant did not etch out the regions. Banding was seen in consolidated cross sections but was much more random and not seen in locations corresponding to the defect locations.
5.0 Discussion

5.1 Challenges of Process Parameter Optimization

FSW of thick section steel is not readily understood by open literature and presents many unique challenges. In order to perform a proper tool life study, a design and parameter set with a large process window was required. Currently, there is no baseline for parameter development to streamline the process. The most effective method is trial and error. The trial and error relies heavily upon past knowledge and industrial experience of the individual. This method has its pros and cons, but each alloy demands a new tool design (or testing of a previous design) and parameter development. For any change in thickness or alloy the process outlined in this work would need to be repeated to accommodate the process for the new material. This method can become very costly and time consuming. During parameter development, the risk of breaking the tool is high when establishing welding parameters. The tool material costs as least 1,000 USD/ in depending upon the diameter. The process can take a person anywhere between 20-100 hours to also develop the proper parameters and this time does not include metallography. Additionally, developed parameters by this method can never truly become optimized as there is an infinite number of designs and parameter sets. A model of FSW steel would greatly accelerate the process and lead to more idealized results, but a greater understanding of the tool material, tool design, and process itself is needed to achieve a properly
functioning model. With this in mind, the iterative process was used in this work as current methods are not readily available.

In this work the parameter were developed by using the extruded material and a baseline VPT design commonly used to weld steel, characterized by a large tapered pin and a small shoulder to provide material constraint. For the given design, two defects could be directly related to the design of the tool. The wormhole defect appeared after a short weld length when the included pin angle was less than $60^\circ$. When the pin angle was increased to $64.7^\circ$, the wormhole defect was healed. However, after a short weld length, a defect arose near the pin tip. It appears that a large pin angle is needed to consolidate material on the advancing side. However, when the angle was increased the pin tip diameter inherently had to decrease. The formation of the pin tip defect is likely due to insufficient heat generation near the base. As torque is closely related to heat input, the smaller pin tip diameter would have a lower local torque and thus create less heat near the root of the weld. A simple solution would be to simply increase the diameter of the tool, allowing for a larger included angle and pin tip. Doing so would not only create a higher torsional load on the tool material, but put more heat input into the weld. Equation 1 shows a simplified relationship between welding parameters and heat input. To apply this equation, take the shoulder diameter should be replaced with the pin diameter at a given location within the weld. Without additional cooling methods, the increased tool size would likely cause a loss in mechanical properties. Thus a balance is needed between the pin angle and the pin tip diameter. The final tool design incorporated a large included
angle of 61.8° and a pin tip diameter of 13.21-mm. In the tool life, both pin tip and wormhole defects were seen. Based upon the defect formation seen in the parameter development, the T03 was the proper balance between the two variables as both appeared throughout the tool life. This suggests the defects seen were the product of wearing in different locations upon the tool as opposed to the defects seen due to the design during parameter development.

5.2 Tool Degradation

Both materials experienced degradation in the forms of abrasive wear and likely some adhesive wear, although very little evidence was seen suggesting that this wear occurred. The recrystallized material also experienced tool degradation in the forms of intergranular failure and deformation, which were not observed in the extruded tool material. The deformation was not a large form of wear as the only signs were some twins seen in the pin of the tool. Additionally, no macroscopic geometric changes, such as the mushrooming observed by Wei Gan et al., were observed in the line scan of the material to suggest deformation. Intergranular failure however, did play a large role in the recrystallized tool material. In areas which did not experience large amount of material loss, such as the pin tip and the intersection of the shoulder and pin, entire grains were removed from the tool material. In the pin taper, which experienced the largest loss of material, evidence of intergranular failure was limited. No areas showing entire grains removed from the matrix were found, despite some area suggesting intergranular failure
A possible explanation for this is the presence of abrasive wear. Abrasive wear was seen throughout the surfaces of both tool materials, and found readily in the areas experiencing the most material loss. Thus both intergranular failure and abrasive wear occur currently along the taper/side of the pin. Thus when intergranular failure occurs it creates a local stress concentration as the surface in longer smooth. In this location abrasive wear would be accelerated due to the stress concentration causing more material loss in this region. Hence once intergranular failure occurs, abrasive wear will quickly wear the surrounding material removing evidence of the intergranular failure. Thus intergranular failure may have regularly occurred within this area, but the evidence would quickly be erased by abrasive wear.

The differences in material hardness did not drive the wear in this study. The hardness of the extruded material post welding was 524 HV as opposed to the recrystallized material 500HV Overall, The hardneses of the materials are relatively similar. Research conducted by Gates et al. concluded harder materials were indeed more resistant to abrasive materials than softer material. The study did show that alloys with similar hardness values and composition, did experience varying amounts of wear. Thus it is apparent that hardness alone cannot account for differences in wear characteristics of any given material.

The recrystallized tungsten material did experience a hardening of approximately 50 HV as opposed to the pre welding hardness. The uniformity of the hardening in the post
welded recrystallized tool material (Figure 65) suggests the hardening occurred throughout the entirety of the tool material. This hardening likely occurred early in the tool life as a result of high temperatures reached by the tool material. If the material hardening was a result of deformation, the hardening would not appear as uniform and greater amount of hardening would be observed at high strained areas. This is an area which would require further study to determine when this hardening occurs and the nature of the hardening.

The extruded material proved to be more resistant to tool degradation and have a longer tool life than the recrystallized tool material. The fundamental reason behind the differentiation between the performances of the two materials was the microstructure. The recrystallized material is much more susceptible to intergranular failure as stated within the previous research on tungsten by Briant et al. The cracks seen in the cooling hole vividly display how a crack propagates through each material. The crack was transgranular in the extruded material and much shorter in length. The cooling hole cracks in the recrystallized material were long, visible by the naked eye, and intergranular. In the extruded microstructure the fracture path is much more tortuous and this path is displayed in Figure 2. Thus once a crack begin it is extremely difficult to remove and entire grain. Therefore, intergranular failure was not observed in this material. The crack propagation path within a recrystallized microstructure is much less tortuous and thus is easier for a crack to propagate. The relative ease for crack propagation explains the intergranular propagation observed within the recrystallized
sample. The intergranular nature of the cracking is displayed in Figure 68. The intergranular cracking not only caused material loss, but created preferential locations for accelerated abrasive wear. Thus due to the recrystallized material’s susceptibility to intergranular failure, the extruded material is the more suitable microstructure for steel welding applications in thick section steel.

5.3 Steel microstructure and tool wear

The effects of tool wear on the steel itself was not the focus of this study, however several notable observations were made throughout the tool life trials. The microstructure observed in the steel, generally agreed with the temperature models presented in open literature. The ferrite grain size near the top of the weld was generally larger than the ferrite grain size seen near the bottom of the weld. This confirms that the weld is generally a lower temperature (~650°C) near the bottom of the weld than the surface of the weld (1200-1300°C). The tensile tests failed on the retreating side of the weld, with two exceptions due to large defects present on the advancing side. A differentiation between the advancing side and retreating was not able to be distinguished optically as the grain sizes and morphologies appeared identical for any given weld. Further investigation is needed to understand why the retreating side was the initiation point for failure during tensile testing.
The steel microstructure is noticeably affected by the proximity of the cooling hole. The stir zone ferrite grain size decreased as the distance between the cooling hole and pin decreased. The reduction in grain size suggests that either a lower peak temperature was reached within the stir zone or the cooling rate within the stir zone was increased. The change in microstructure is likely a result of a faster cooling rate, as martensite is observed in the final welds, which is a cooling rate dependent phase. Despite the fundamental change in weld microstructure due to the increased cooling rate, the mechanical properties remained relatively unchanged in the welds. The final and initial welds demonstrated similar tensile strength and ductility. The only welds with degraded properties throughout the entire tool life trials can be attributed to defects within the weld. The CVN tests showed failure right along the weld centerline and the larger ferrite grain sized welds had similar impact toughness properties as the smaller ferrite grain size welds with martensite present in the weld. This suggests that even with a small increase in cooling rate, the change in toughness in the friction stir welds was not significant. The microstructure could affect other properties not tested such as corrosion, fatigue life, and formability. Thus based upon the microstructures and mechanical tests, redressing did not affect the tool’s ability to create repeatable welds with uniform properties. Further investigation is needed to determine the effects of microstructure and processing on the resulting toughness in thick section steel.

Inclusions of W-Re were found in several cross sections throughout the tool life trails. It bands of W-Re were observed throughout the life as seen in Figure 18. W-Re bands in
the weld metal were explained by Thompson and Babu as a diffusion phenomenon, which occurs during the stick phase of friction stir welding. During sticking, a given amount of tungsten rhenium is diffused into the steel stuck to the tool. Upon release from the tool material, the stuck steel is deposited into the weld and results in the bands of W-Re enriched areas often seen in cross sections.\(^{42}\) During the tool life trials, defect regions often demonstrated an area of W-Re enrichment right along the defect/opening. It is currently unclear which of two mechanism resulted in the W-Re enrichment. The first possible explanation is that wear is localized in these regions, and worn particles are mechanically deposited into the steel and then chemically diffuse into the steel. The additions of the refractory metals would raise the flow stress of steel at temperature and require a larger forging force to consolidate the material. Thus, once the strength at this location was greater the forging force, a defect would form. The material when consolidated would then then be stirred throughout the entire weld but the lack of consolidation would leave the W-Re enriched area near the defect. The second explanation, and more likely scenario, is that the formation of a defect creates a condition which promotes more tool degradation. The change in the flow of material caused by the defect would cause localized wear near the defect and force more W-Re particles into the steel near the defect. The particles would then chemically diffuse into the steel in the regions near the defect. This would also lead to the W-Re enriched region and explain why the alloying is not seen in the defect locations prior to defect formation. The actual cause of the defects themselves is related to wear but the exact cause is not currently understood by open literature in steel. It appears that defect formation is related to tool
designs and tool degradation changing the geometry of the tool during welding. This work suggests that for the wormhole defect, the more wear seen along the pin taper, the more likely an advancing side wormhole defect was likely to form. More work is needed to understand the relationship between tool geometry and defect formation.
6.0 Future Work

The work clearly demonstrated that tool microstructure strongly influenced the degradation characteristics of the material. Although the susceptibility to cracking played a large role in the degradation, grain orientation may also play a role. Figures 40 and 57 show the difference in grain structure along the pin taper. The extruded microstructure likely has more similar orientations due to the extrusion, while recrystallized microstructures typically exhibit a more random orientation. Orientation imaging microscopy (OIM) would provide interesting information into the microstructure. It would also give insight into the deformation seen within the tool material.

Further work is also needed to understand the failure and general behavior of tungsten rhenium at temperature. The reason for the intergranular failure needs to also be investigated. It appears that this is the weakest point in the microstructure as cracks propagated preferentially through the grain boundaries. Torsional and bending tests at elevated temperatures would give excellent insight to the material behavior at temperature and the associated failure mechanisms.

Further wear studies should also be investigated. The formation of defects could not be identified during in process welding. During production, this would create a serious issue and demand NDE after each weld. A possible solution could be in situ force monitoring when using a position controlled machine. As the tool wears the torque or other forces
applied on the tool may decrease or show a pattern suggesting a defect was present in the weld.

Additionally work is needed to determine the effect of the plunge on tool wear. In this study, it was unclear as to if the wear mostly occurred in the weld length or during plunge. The effect of the plunge needs to be further investigated, in addition to varying plunge parameters. This study simply identified a plunge parameter which placed a minimal amount of force upon the tool. A repeated plunge study compared to extended weld lengths would clearly demonstrate the effects of the plunge on wear.

Further work is also needed to increase the efficiency and effectiveness of the parameter development and tool design for FSW applications. Currently, most development is done in an iterative fashion relying heavily upon the knowledge and judgment of the operator. Due to the infinite number of tool designs and parameter combination, a system model is needed. Through modeling, countless man hours and tool material could be conserved. Currently, it may take an engineer weeks or even months to achieve the desired properties using FSW. The model may not eliminate the need for parameter development but would streamline the process greatly.

The water cooling hole appeared to be a stress concentration for the tool design. Cracks appeared near the end of the cooling hole and a large amount of microcracks appear in the shank near the cooling hole. Twinning was also seen near the outer of diameter of the
tool shank. Alternate cooling methods and or different cooling hole designs should be employed in the tool design to minimize stress in the tool shank itself to prevent tool failure.

The machining of the tungsten-rhenium also proved to be a challenge in this study. The large variations in the redress dimensions affected the tool life results and were likely the cause of the variations in life from redress to redress. The tolerances on the pin dimensions were +/- .05 mm and these tolerances were not met on a regular basis. The tools in this study were single point turned in an ISO certified machine shop and these variations still occurred. The tools like should be CNC ground to maintain dimensions as single point turning appears to be difficult. This area needs investigation as to the best machining practice.

Lastly, the area of fracture toughness is an area which greatly needs addressed in the area of FSW of steels. It appears through this work and previous research that as thickness increases, the toughness of the welds decreases. The welds in this work exhibit very low toughness. This problem should be addressed at a fundamental level. Using OIM, the prior austenite grain size can be investigated and see if a correlation between this and the toughness can be made. Other topics such as microstructure and the relation of toughness and heat input should also be further investigated.
7.0 Conclusions

- **Microstructure has a large effect upon tool degradation:** Each microstructure demonstrated distinct degradation mechanism. The recrystallized microstructure experienced deformation and intergranular failure, which were not observed in the extruded material.

- **The extruded microstructure is more resistant to tool degradation:** The extruded material experienced 27.5% less material loss than the recrystallized material. The extruded material was also able to produce 31.5% more defect free weld length than the recrystallized material.

- **Abrasive wear was the primary wear mechanism for the extruded material:** The extruded microstructure wore primarily through abrasive wear. The recrystallized material also demonstrated large amounts of abrasive wear, but did also demonstrate intergranular failure.

- **Redressing is an effective method to increase the tool life of a given piece of material:** The tools were able to create repeatable welds throughout the tool life trials, without a loss of mechanical properties. In turn, redressing greater increased the usable life of a tool and the weld length which a single piece of tool stock may produce.
• **Included angle and pin tip diameter are crucial tool design variables:** A large included angle of 61.8° or greater is desirable in the tool design. A large pin tip diameter of 13.2-mm or greater is also desirable in a tool design. Often a balance between these two variables is needed as increasing one of these variables decreases the other due to tool stock diameter limitations.

• **Repeatable full thickness welds are achievable:** Full thickness welds in 19-mm thick X-70 steel was achieved. The welds were repeatable made throughout a tool life with similar mechanical properties.

• **Extended tool life is possible in thick section steel:** The total acceptable weld length for the extruded material was greater than 41-m (~135-ft.). Further development of the parameters and tool material/microstructure could increase the life.

• **Using the extruded tool material would result in an approximately 10-15% cost savings in offshore pipeline construction:** The life of 41-m, coupled with the 19-mm wall thickness would result in an approximate cost savings of 10-15% according to Kumar et al. The portions of this work which adversely affected the cost saving were the welding speed and tool material cost. Increasing the tool life further would also increase the cost savings.
References


Appendix A: Figures and Tables

\[ q = \frac{2\pi}{3S} \cdot \mu \cdot p \cdot \omega \cdot Rs \cdot \eta \]

Equation 1 - Heat Input of a FSW Weld (From 57)

Where \( \mu \) is the coefficient of friction, \( p \) is the normal force in kN, \( \omega \) is the rotational velocity in rev/s, \( Rs \) is the shoulder radius, \( \eta \) is the efficiency, and \( S \) is the travel speed in mm/s.

Fig. 13. Schematic diagrams (longitudinal and transverse sectional views) illustrating four mechanisms of abrasion. (a) Microploughing; (b) side-fin formation; (c) wedge formation; (d) microcutting. The less widely recognized mechanisms of side-fin formation and wedge-formation represent intermediate cases between the extremes of microploughing (little or no primary material removal) and microcutting (large proportion of groove volume removed as primary debris).

Figure 1- Abrasive Wear Schematic From Source 41
Figure 2- Effect Of Microstructure on Crack Path (schematic)

From Ref. 48

Figure 3- W-Re Binary Phase Diagram (From Reference 50)

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Note the labeling on this diagram as this will be used throughout work. Pin side are also called pin tapers.

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<th>Element</th>
<th>Results %</th>
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<td>Boron</td>
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<td>Vanadium</td>
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Table 1 - Material Composition determined by OES Laser
Figure 5- Tool Life Plate Set Up

Figure 6- 0.3-m Mechanical Test Plate Set Up
Figure 7- In Process Weld

Figure 8- Parameter Development Tool Designs

<table>
<thead>
<tr>
<th>Tool Design</th>
<th>Starting Cross Sectional Area (mm²)</th>
<th>(A) Included angle °</th>
<th>(B) Pin Tip (mm)</th>
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<td>T01</td>
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<td>T02</td>
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T03 was selected for the tool life study.
Figure 9- Tool in Holder with Shielding Shroud

Figure 10- Laser Scan Set Up
Figure 11- Profile Scan of a Tool

Figure 12- Tool Life Plate X-Ray Set Up
Figure 13- Flowchart of Tool Life Procedure
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<thead>
<tr>
<th>Weld Trial</th>
<th>Tool Design</th>
<th>Travel RPM</th>
<th>Travel Speed (mm/min)</th>
<th>Water Cooling (LPM)</th>
<th>Weld Length (m)</th>
<th>Defect</th>
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Table 2- Parameter Development Table

Figure 14- Weld 2 made using T01 at 110 RPM and 76.2mm/min
Defect is circled in red. Weld 5 was made at 150 RPM and 127 mm/min using T01. Rotation direction marked by arrow.

Defect circled in red. Weld 7. 140 RPM 114.3 mm/min using T01. Rotation direction marked by arrow.
Figure 17- Weld 14. 110 RPM and 76.2 mm/min using T02

Acceptable Weld

Figure 18- Cross Section of Weld 11

Note the banding of the tungsten tool material present in the weld
Figure 19- Weld 13/ XT-1 Made at 110 RPM and 76.2 mm/min using T03

Minor oxidation due to flash affecting the shielding.

Figure 20- Radiograph of Weld 13/ XT-1

This is a defect free weld and passed radiographic inspection
Composition band present from the rolling and material processing is pulled into weld.

The 1.5-m passes produce a large amount of flash, which was removed before radiographic inspection.
Pin tip defects were seen within welds RX-29 and XT-37. These defects are less than 2% of material thickness and thus are not seen during the radiographs.

Shaded area is the calculated cross sectional area calculated for analysis
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| Total Adjust Weld Length | 41.30 |

Table 3- Extruded Tool Material Tool Life Results
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| 31.18 | Total Adjusted Weld Length |

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<td>Recrystallized (RX)</td>
<td>31.18</td>
<td>9.45</td>
<td>7</td>
<td>4.45</td>
<td>2.83</td>
</tr>
</tbody>
</table>

Table 5- Summary of Weld Lengths during Tool Life

Figure 25- Typical Pin Tip Defect.

Note the small size and location. The arrow represents the direction of the rotation
Figure 26- Cross Section of Typical Wormhole Defect.

Defect spans approximately 5.8-mm through the thickness and has a cross sectional area of approximately 3-mm$^2$

Figure 27- Tensile Test Results.

Note: Defects Shown in Cross Section of RX-25, RX-28, XT-9, XT-14, and XT-33. Only RX-25 and RX-28 showed a reduction of properties due to the presence of a defect.
Figure 28- Macro of RX-28

Defect size of 3.6 mm through the thickness with a volume of approximately 1-mm$^3$

<table>
<thead>
<tr>
<th>BM</th>
<th>XT-1</th>
<th>XT-5</th>
<th>XT-21</th>
<th>XT-25</th>
<th>XT Avg.</th>
<th>RX-1</th>
<th>RX-10</th>
<th>RX-13</th>
<th>RX-22</th>
<th>RX Avg.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Absorbed Energy (J)</td>
<td>309</td>
<td>25.48</td>
<td>22.48</td>
<td>22.50</td>
<td>19.52</td>
<td>22.50</td>
<td>18.70</td>
<td>36.06</td>
<td>15.20</td>
<td>29.84</td>
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<tr>
<td>Shear %</td>
<td>100</td>
<td>17</td>
<td>17</td>
<td>17</td>
<td>17</td>
<td>17</td>
<td>15</td>
<td>21</td>
<td>21</td>
<td>21</td>
</tr>
</tbody>
</table>

Table 6- CVN Test Results

Average of 5 tests each at -20°C. The XT Tool material had a Standard Deviation of 6.22 J, while the RX tool material had a standard deviation of 12.06 J. Welds with defects were not tested.
Figure 29- Typical Line Scan.

Notice the majority of the wear occurs throughout the tapered portion of the tool. Some shoulder wear occurs, but is minimal compared to the wear along the pin taper.

Figure 30- Loss of tool material over entire tool life
Figure 31- Degradation per Trial

Note the large difference in the initial cross sectional areas at 0-m of weld length.

Figure 32- Histogram of starting Tool Cross Sectional Areas

Note the large scatter. The desired cross sectional area is 637.8-mm$^2$ according to the model.
Figure 33- Starting Profile Probability Plot

Note that the distribution fits a normal distribution. This suggests normal error in the machining.

Figure 34- Normalized Degradation Data

The data is a linear fit. The XT tool wore about 0.69%/m, while the RX wore 0.88%/m.
Figure 35- Optical microscopy of XT pre Weld

Figure 36- Pre XT Microstructure - middle of sample
Figure 37- Edge of XT Material

Note small amount of twinning on the right edge of the sample

Figure 38- XT Optical Center of Pin Microstructure
The difference in shading is not significant. Note the extruding and sheared region suggesting abrasive wear.
Figure 41- Wear face XT Pin Taper SEM Photo

Figure 42- XT Shoulder Wear SEM Photo

Rolling pattern suggests abrasive wear.
Figure 43- XT Shoulder Wear SEM Photo

Figure 44- XT Cracks in the Shank Optical Photo
Figure 45 - XT Cracking seen in shank of tool material SEM Photo

Figure 46 - XT Cracking seen in shank SEM Photo
Figure 47- XT Cooling Hole Tip Crack SEM Photo

The material filling the crack could not be determined by EDS or SEM analysis.

Figure 48- Hardness Map of Pre-Weld Material XT-1

Cap of the extrusion, hardness is not indicative of sample

Figure 49- Hardness Map of Pre-Weld Material XT-2

Core is softer than outer region
Figure 50 - Hardness Map of Post-Weld XT Pin

Core Region is softer than outer region

Figure 51 - RX Optical Pre Weld Microscopy
Figure 52 - RX Pre Weld Microstructure middle SEM Photo

Figure 53 - RX Pre Weld Edge Microstructure SEM Photo
Figure 54 - RX Post Weld Pin Center Optical Photo

No noticeable change from pre-weld microstructure

Figure 55 - RX Pin Center Microstructure SEM Photo

small intergranular cracking in center
Figure 56- RX Pin Center SEM Photo

Note Some twinning

Figure 57- RX Pin Taper Abrasive Wear SEM Photo

Smeared material suggests abrasive wear
Figure 58 - RX Shoulder Wear

Close up of grooved region which suggests abrasive wear

Figure 59 - RX Pin Tip Abrasive Wear SEM Photo

Close up of grooved region which suggests abrasive wear
Figure 60 - RX Pin Shoulder Interface Intergranular Failure SEM Photo

An entire grain is removed from the tool in this location.

Figure 61 - RX Pin Tip left edge SEM Photo

Shows area of missing grains, demonstrates intergranular failure in RX material.
The protrusion of the lowest grain and the groove between the top and middle grain suggest a grain was removed from the outer surface.

Figure 62- RX Pin Taper Intergranular failure SEM Photo

Figure 63- Hardness Map of Pre-Weld Material RX-1

Figure 64- Hardness Map of Pre-Weld Material RX-2
Figure 65 - Hardness Map of Post-Weld RX Pin

Uniform hardening occurred during welding

Figure 66 - RX Cooling Hole Tip Crack SEM Photo

Intergranular cracking within tool near cooling hole. Likely due to stress concentration from cooling hole.
Figure 67- RX Cooling hole Cracking SEM Photo

Close up of Figure 66.

Figure 68- RX Cooling Hole Cracking SEM Photo

Shows intergranular nature of cracking
Figure 69: RX Micro Cracks in Shank Optical Photo

Figure 70: X-70 Base Metal Microstructure

Mixture of Ferrite/pearlite and likely bainite/martensite
near the cap of the weld. Note the large size of the polygonal ferrite and the prior austenite grain boundary.

near cap. Similar to Advancing side with large prior austenite grain size and large polygonal ferrite.
The weld metal near the bottom of the weld has smaller grains of polygonal ferrite and difficult to identify prior austenite grain boundaries.

near the top of the weld. Similar to RX-1 microstructure with larger prior austenite grain size, and large polygonal ferrite present.
Figure 75- Weld XT-1 Weld Metal Microstructure on the Retreating Side

Near the top of the weld, similar to advancing side in morphology

Figure 76- Weld XT-1 Weld Metal Microstructure Near the Root of the Weld

Similar to RX-1. Some polygonal ferrite, with ferrite with aligned phase
Near the top of the weld. Weld near the end of the tool life trial. Mixture of grain boundary ferrite, ferrite with aligned second phase, and martensite. Note the large reduction in ferrite grain size.

Figure 78- Weld RX-28 Advancing Side Weld Metal near the top of the weld

Weld near the end of the tool life trial. Mixture of grain boundary ferrite, ferrite with aligned second phase, and martensite. Note the large reduction in ferrite grain size.
Figure 79- Optical Image of Weld 10 Pin tip Defect
Note the areas in which the nital did not etch as these are tungsten enriched areas

Figure 80- Backscatter SEM image of Weld 10 Defect
Lighter areas show higher tungsten and rhenium concentrations
Notice the W and Re contents. Disregard the high C concentration as EDS cannot properly detect carbon. Also disregard the letters K and M.
Appendix B - Tensile Test Stress-Strain Curves

**Tensile Test Stress-Strain Curves**

**TL-1 Curve**

**Curve 1 - Longitudinal Base Metal**

**TT-1 Curve**

**Curve 2 - Traverse Base Metal Sample**
Curve 3- Typical RX Weld Stress Strain Curve

Curve 4- Typical XT Weld Stress Strain Curve