

**FRICTION STIR PROCESSING EFFECTS ON FATIGUE
CRACK PROPAGATION IN SUPER DUPLEX
STAINLESS STEEL**

BY

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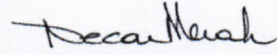
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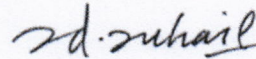
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[Dedication]

To my beloved parents, my wife, my Colleagues, and my well-wisher with love.

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[

All the praises and thanks be to Almighty ALLAH, the gracious, the Lord of the worlds, the most merciful, who gave me the perseverance, strength, and knowledge to accomplish this work. May peace and blessings be upon the prophet Muhammad (PBUH), his family and his companions.

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LIST OF ABBREVIATIONS

ASTM	:	American Society for Testing and Materials
CT	:	Compact tension
FEM	:	Finite element method
EDM	:	Electrical discharge machining
EBS	:	Electron Backscatter Diffraction
FCP	:	Fatigue crack propagation
FSPed	:	Friction Stir Processed
FSP	:	Friction Stir Processing
FSW	:	Friction Stir welding
FCG	:	Fatigue crack growth
HAZ	:	Heat affected zone
HV	:	Vickers Hardness
LEFM	:	Linear-elastic fracture mechanics
OM	:	Optical microscope
SDSS	:	Super Duplex stainless steel
SEM	:	Scanning electron microscope
SIF	:	Stress intensity factor
SZ	:	Stir Zone
XRD	:	X-Ray Diffraction

]

ABSTRACT

Full Name : [MD. EYASIN ARAFAT]

Thesis Title : [Friction Stir Processing Effects on Fatigue Crack Propagation in Super Duplex Stainless Steel]

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The use of duplex stainless steels (DSSs) is increasing in several industries because of its better performances compared to traditional austenitic and ferritic steel grades. DSSs and mainly super DSS (UNS S32750 (SAF 2507)) are considered excellent materials for desalination plants, power plants, and petrochemical plants. Therefore, understanding the fatigue behavior of emerging materials such as super duplex steel (SDSS) is an essential element in the design of structures and machine components as this type of loading is known to represent 80 to 90% of the total failures in the industry. Friction stir processing (FSP) is a grain refinement technique that improves metals mechanical properties (i.e. enhance metals fatigue and fracture properties). The goal of this study is to investigate the effects of friction stir processing (FSP) of UNS S32750 (SAF 2507) on the fatigue crack propagation (FCP) properties . Optimal FSP parameters are employed to process SAF 2507 plates in the rolling direction. Compact tension specimens (CT) are fatigue tested under constant amplitude load, at a stress ratio (R) of 0.1 and frequency of 10 Hz. Crack length is measured by a high-resolution camera. The developed residual stresses after FSP are determined by the drill-hole technique and the temperature is measured by infrared camera. The measured parameters (such temperature, residual stress etc.) are used to validate the specially developed finite element model for the friction stir process and to explain some

of the effects of FSP on fatigue crack growth behavior of the material. The results of this study show that friction stir processed (FSPed) specimens have higher fatigue crack growth (FCG) resistance and fatigue life compared to the base material. This mainly is due to grain refinement and reorientation in the stir zone (SZ), through which the crack grows. Scanning electron microscopic analyses of the fracture surfaces revealed a significant change in fracture mechanisms and micro-mechanisms in SDSS (SAF-2507) due to FSP. FSPed specimens fracture surfaces are characterized by numerous crack initiations site and complicated crack propagation paths, which contributed to the retardation of the crack propagation.

]

ملخص الرسالة

الاسم الكامل: محمد عياسين عرفات

عنوان الرسالة: آثار طريقة الاحتكاك التحركي على انتشار إجهاد الشقوق في الفولاذ المقاوم للصدأ سوبر دوبلكس
التخصص: الهندسة الميكانيكية

تاريخ الدرجة العلمية: ديسمبر 2020

يتزايد استخدام الفولاذ المزدوج المقاوم للصدأ (DSSs) في عدة صناعات نسبةً لأدائه الأفضل مقارنةً بالدرجات التقليدية من الفولاذ المارتنسيطي و الفريتي المقاومة للصدأ. الفولاذ المزدوج المقاوم للصدأ ، وبصورة رئيسة الفولاذ المزدوج الفائق المقاوم للصدأ (DSS (SAF 2507) UNS S32750)) ، تعتبر مواد ممتازة لمحطات توليد المياه ، محطات توليد الطاقة الكهربائية ، ومصانع البتروكيماويات. لذا فإن معرفة سلوك الإجهاد للمواد الناشئة كالفولاذ المزدوج الفائق المقاوم للصدأ (SDSS) هو عنصر أساسي في تصميم هياكل ومكونات الآلات لأن من المعلوم أن هذا النوع من الأحمال يتسبب في 80- إلى 90٪ من جملة الأعطال في مجال الصناعة. المعالجة بطريقة الإحتكاك التحركي (FSP) هي تقنية لتصغير الحبيبات تحسن من الخواص الميكانيكية للمعادن (أي أنها تحسن من خصائص الإجهاد والكسر). الهدف من هذا الدراسة هو تقصي آثار المعالجة بطريقة الإحتكاك التحركي (FSP) على الفولاذ المزدوج الفائق المقاوم للصدأ (SAF 2507) UNS S32750)) وتحديدًا على خصائص تمدد الشق الناتج عن الإجهاد (FCP). تم تطبيق معايير المعالجة بطريقة الإحتكاك التحركي (FSP) المثلى لمعالجة صفائح من الفولاذ المزدوج الفائق المقاوم للصدأ (SAF 2507) في اتجاه الدرفلة. تم اختبار الإجهاد لعينات شد صغيرة الحجم (CT) تحت مقدار تحميل ثابت النطاق ، ذو نسبة إجهاد (R) مقدارها 0.1 وتردد قدره 10 هيرتز. تم قياس طول الشق باستخدام كاميرا عالية الدقة. تم تحديد الإجهادات المتبقية الناتجة عن المعالجة بطريقة الإحتكاك التحركي باستخدام تقنية فتحة المثقاب وقيست درجة الحرارة عن طريق كاميرا تصوير بالأشعة تحت الحمراء. تم استخدام المتغيرات التي تم قياسها (كدرجة الحرارة والإجهاد المتبقي ... إلخ) لتأكيد صحة نموذج العناصر المنتهية الذي تم تطويره خصيصاً لأجل المعالجة بتحريك الإحتكاك ولتوضيح بعض آثار المعالجة بطريقة الإحتكاك التحركي على سلوك نمو الشق الناتج عن الإجهاد في الفولاذ المزدوج الفائق المقاوم للصدأ (SAF 2507). أظهرت نتائج الدراسة

أن العينات التي تمت معالجتها باستخدام الإحتكاك التحركي للصدأ (FSPed) ذات مقاومة أكبر لنمو الشق الناتج عن الإجهاد (FCG) وعمر كل أطول مقارنةً بالمادة الأساسية. يعزى ذلك بصورة أساسية لتعرض الحبيبات للتصغير و إعادة تموضعها داخل منطقة التحريك (SZ) التي ينمو خلالها الشق. أظهرت تحليلات المسح المجهرى الإلكتروني لأسطح الكسور تغيراً كبيراً في الآليات والآليات المصغرة لحدوث الكسر في الفولاذ المزدوج الفائق المقاوم للصدأ (SAF 2507) نتيجة المعالجة بطريقة الإحتكاك التحركي. أسطح الكسور في العينات المعالجة بطريقة الإحتكاك التحركي تميزت بالعديد من مواضع بدايات الشقوق ومسارات تمدد شقوق معقدة ، مما ساهم في إعاقه تمدد الشقوق.

CHAPTER 1

INTRODUCTION

1.1 Background

1.1.1 Friction stir porcessing

Friction stir processing (FSP) and Friction stir welding (FSW) technologies are believed to be the most important and most popular solid-state processing methods that have developed over the last few decades. The FSW method was developed in 1991 and patented in 1994 by Technical Welding Institute (TWI) [1]. Several potential applications of this process are available in the domain of shipbuilding, automotive, oil and gas, military fields, and aerospace industries.

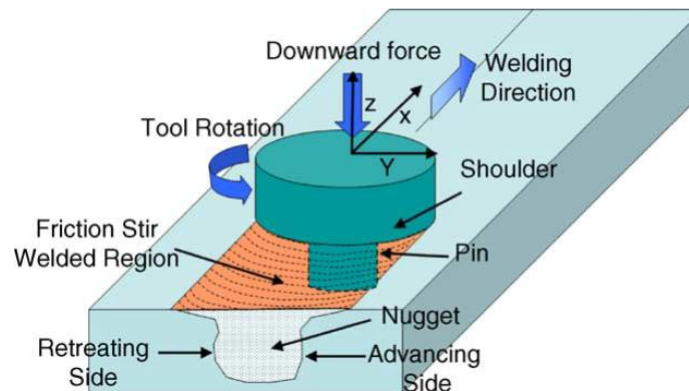


Figure 1-1: Schematic model of friction stir welding (FSW)[2].

FSW includes three main steps: tool plunging, tool traversing and tool retracting. FSP which is used mainly for material surface processing, goes through the same steps. The

schematic model of FSW is shown in Figure 1-1. FSW has many benefits over the conventional welding methods, including no filler wire, excellent mechanical properties, use of the non-consumable tools, no fume, and low distortion even in long welds [3]. On the other hand, it has some demerits, too, including special attention required while clamping the workpiece.

1.1.2 Duplex stainless steel

The duplex stainless steel (DSS) concept was first developed in the early 1920s but the first production of duplex alloys has not started until the 1930s. In the beginning, these alloys were limited to cast production and had some specific applications due to their high amount of carbon content [4]. The DSS contains 50/50 per cent of austenite and ferrite, enabling the amalgamation of excellent corrosion resistance and mechanical properties. Ferritic stainless steels and DSS have similar corrosion resistance. The strength of the DSS is higher than ferric stainless steel and lower than that of austenitic steel, while its mechanical strength is higher than that of austenitic stainless steel. These characteristics, therefore, suggest that DSS is an excellent material for applications in the gas, oil, electrical and chemical industries, desalination plants, heat exchangers, process and service water systems, fire-fighting systems, offshore platforms, pipes and ducts [5]. Stainless steels can be separated into the duplex family, martensitic, ferritic, and austenitic. Duplex stainless steel is the two-phase microstructure of austenitic (Cr-Ni) and ferritic (Cr only), which includes a large amount of Cr and low amount of Ni. Due to the combination of two phases, it is called “Duplex”. A right combination of toughness, economy, and weldability make

duplex family the alloys of choice when corrosion and strength properties or both are critical [6]. The standard austenitic stainless-steel grades are less popular than DSSs, for instance, type 316 and 304. For maintaining the balance of austenite and ferrite, it is necessary to avoid the secondary phases throughout welding and fabrication. The development of duplex steels increased the potential for structural applications owing to their excellent mechanical properties, higher corrosion resistance [7, 8], and reduced cost for construction purposes when using lower alloyed steels. The chemical composition of common DSS grades is given in Table 1 [9, 10].

Table 1-1: The chemical composition of DSS grades in (wt.%).

DSS grade	Cr	Mo	N	Mn	Ni
2304 [9]	21.5-24.5	0.05-0.6	0.05-0.2	2.5	3.0-5.5
LDX 2101 [8]	21.5	0.3	0.22	5	1.5
2205 [9]	22.0-23.0	3.0-3.5	0.14-0.20	2	4.5-6.5
2507 [9]	24.0-26.0	3.0-5.0	0.24-0.32	1.2	6.0-8.0

The amount of Nickel contents in duplex stainless steels is typically 1-7%, whereas the austenitic grades have 8-25% [11]. The oil and gas industry first utilized the DSS for the application in the North Sea [12]. However, the enormous structures of DSS required welding, where metallurgical challenges introduced undesirable phase changes that the structure of the materials reformed after melting in a matter of seconds. The equilibrium phase diagram in Figure 1-2 shows the transformation that can occur in DSS. From melting to solidification, the first ferrite formed, then in the solid-state, some of this ferrite changes to austenite. In actual practice, the welding can produce a fully ferritic weld if very rapid

cooling occurred. Therefore, the austenite has a short time to alter the phase. The Nitrogen gas shielding used during welding, may partition the austenite phase which may trap the ferrite to form nitrides, and that happens when the heat input is low.

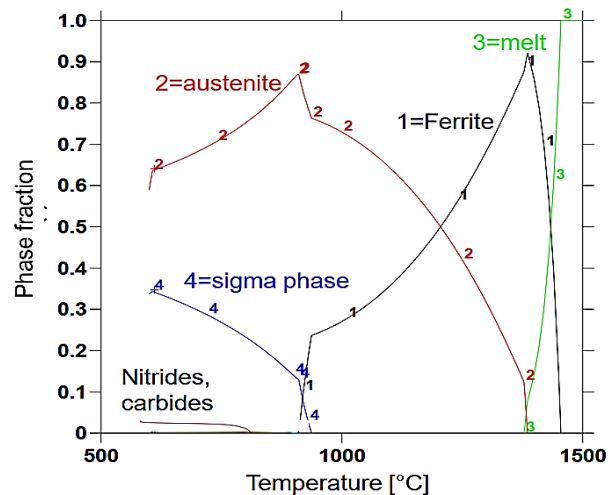


Figure 1-2: Equilibrium phases in a 22Cr duplex stainless steel as a function of temperature [11].

1.1.3 Super duplex stainless steels

The word “Super-Duplex” was first utilized in the 1980s to represent high-performance, highly alloyed duplex steel with a pitting resistance equivalent number which is more than 40 (based on $\%Cr + 16\%N + 3.3\%Mo$). The super duplex stainless steel (SAF 2507) chemical structure is shown in Table 1-1. Super-duplex steels (SDSS) are ternary systems alloys (Fe-Cr-Ni), characterized by a two-phase structure with nearly equivalent fractions of austenite and ferrite. They may present variations of 30% to 70% of ferrite proportions. SDSS has high mechanical fatigue resistance, and resistance to corrosion and pitting in several environments. The Pitting Resistance Equivalent (PRE) number on super duplex stainless steel is higher than 40 [13], which is a higher number than those of the 300

families of austenitic steels. For that reason, the super-duplex steel is widely used in chemical and petrochemical industries, especially in offshore environments, in centrifugal pumps, flow control valves, and seamless tubes under chloride concentrations up to 120.000 ppm [13]. The increase in chrome, molybdenum, and nitrogen contents promotes the achievement of high resistance to corrosion, and it has a very favorable effect on the mechanical strength when used in the nitrogen's environment. SDSSs are often produced with equal portions of ferrite and austenite in their structure. To that end, great control in cooling speed and when adding the alloying elements during their production is necessary [14]. A great variety of phases can be formed on super-duplex stainless steels in the 400 – 1000°C temperature interval, during the isothermal aging (processing or use) or welding [15]. Figure 1-3 illustrates a diagram of temperature-time transformations (TTT) for SAF 2507. The diagram shows also the different phases found in different thermal exposure times. Microstructural alterations that occur on the duplex steel families are due to the high amount of alloying elements. They constitute thermodynamically metastable systems insolubilized state at room temperature. The heat adsorption they achieve facilitates a more robust energetic state obtained by the precipitation of carbides and intermetallic phases [16, 17]. Intermetallic phase precipitation is typically associated with undesirable consequences such as matrix depletion of alloy elements such as chromium, molybdenum, and niobium, and therefore loss of strength, durability, and corrosion resistance, especially pitting corrosion [18].

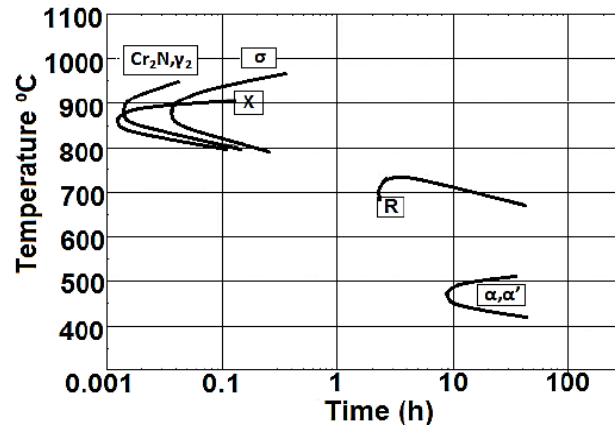


Figure 1-3: TTT Diagram showing the formation of intermetallic phases on steel SAF 2507 [19].

1.1.4 Friction Stir Processing and fatigue crack growth

FSP is considered as an emerging and prominent solid-state surface modification technology that can refine the microstructures, remove casting defects, and associated mechanical property enhancements in the processed area. Thus, developing ductility and strength, enhance formability, and improve resistance against fatigue and corrosion [20-23]. The principles of solid-state processing method FSP and FSW are invariant [24] except that it is applied onto one piece of metal. The process alters the local microstructure to obtain the anticipated properties by surface adapting rather than accumulating the sample (Figure 1-1). Figure 1-1 shows the FSP process; the tool (rod) has a flat shoulder and there are threads on the lateral surface of the cone-shaped probe attached below. The mechanism first drives the high-speed spinning rod during operation to pass into the set work piece. When the contact is completely established by the tool shoulder, it travels around the work piece. The tool produces a considerable amount of heat from the friction between the tool and the work piece as it rotates and crosses the work piece. The matrix surrounding the tool softens when the temperature is $\sim 0.7T_{mp}$ and can undergo substantial deformation [26].

The region affected is strongly deformed and is in an unstable state and has accumulated energy from the stirring and friction. The stored energy is the driving force for the dynamic recrystallization, leading to grain microstructure refinement [25-27].

The crack growth owing to cyclic loading is known as fatigue crack growth, and its estimation becomes necessary and important due to the wide application and acceptance of the damage tolerance concept in the engineering sector. It is known that the $\frac{da}{dN} - \Delta K$ log-log curve has three regions: region I (threshold), region II (Paris region), and Region III (high ΔK region), which are also called characteristic regions of that particular material. It has been observed by many researchers [28-31] that the relationships between Fatigue crack growth rate (da/dN) and stress intensity factor (SIF) range (ΔK) are not rectilinear even in the Paris region. Moreover, the effects of stress ratio on $da/dN - \Delta K$ curves are different in different materials. For example, the effects of stress ratio are near about similar for aluminum alloy (7055-T7511 and 2324-T39) [32] and titanium alloy (Ti10V2Fe3Al) [33] in the three regions. Namely, the $da/dN - \Delta K$ curves are almost “parallel” with various stress ratios. However, for other materials, the effects of stress ratio are completely diverse in the three areas that means these materials are ΔK dependent, for instance, ADB610 steel [34] and graphite cast iron [30]. Thus, there is a complex nonlinear relationship between the fatigue crack growth rates and SIF range.

1.2 Objectives

The principal goal of this research is to investigate the effects of FSP on the fatigue crack propagation of SDSS (UNS S32750; SAF 2507). The specific objectives are:

1. To optimize the friction stir process of super duplex stainless steel (SAF 2507)
2. To develop a finite element model to simulate FSP of SAF 2507 and estimate developed residual stresses, deformations and temperatures.
3. To determine the residual stress state induced by FSP by the drill-hole technique and use these experimental results along with measured temperature to validate the finite element model.
4. To investigate the effects of FSP on Fatigue Crack Propagation (FCP) characteristics using linear elastic fracture Mechanics (LEFM)
5. To determine the mechanisms of fracture controlling FCG in BM and FSPed SAF 2507

1.3 Research Methodology

In this work, the previously mentioned objectives will be addressed through the following approach:

1. State of the art literature review.

This research begins with the state of the art literature review that is related to duplex stainless steel, friction stir processing, fatigue crack growth, finite element modelling, residual stress determination procedure, fractography and the required analytical equation for calculating the generation of heat from the FSP tool.

2. Friction Stir Processing SAF 2507

Friction Stir Processing of SAF 2507 was performed using an RM-1 friction stir welder. Various pin depths and traveling speeds were applied at a constant rotation speed to

optimize the friction stir process. The processing details like (rotational and processing speed, clamping system) are discussed in chapter 3.

3. FE modeling and analysis of Friction stir process

The FE model was build and deveopled in ABAQUS 2019 environment. Where the real tool is replaced by Gaussian moving heat source. The simplified real tool geometry was used to calculate the generation of heat flux for the simiplicity of heat calculation through analytical equation. The measured pressure and calculated heat was introduce in the FE model using user subroutine code of DLOAD and DFLUX. The problem idealization and the outcome from the FE modeling discussed elaborately in chapter 4 and 5.

4. Materials characterization after FSP

Material characterization was performed after FSP by determining the residual stress, fatigue crack growth rate, and fractography and study the mechanical and microstructural properties of SDSS SAF 2507. To determine the residual stress state induced by FSP a semi-destructive technique called hole drilling was used. The reason to choose this method and the details experimental setup are discussed in chapter 3. The fatigue crack growth test was perform on INSTRON 8801 computer regulated servohydraulic machine with 100kN capacity. Both optical and SEM analysis were conducted for expalning the fracture mechanism.

5. Fatigue crack propagation (FCP) behavior of SDSS before and after FSP

The fatigue crack propagation rate of SDSS SAF 2507 has been determine for both BM and Friction stir processed (FSPed) material using linear elastic fracture mechanics

(LEFM) with application of Paris law. First the crack length with number of cycles were determined using high magnification camera and plotted the crack length vs number of cycles. The crack growth constant C and m were determined from the fatigue crack growth rate vs stress intensity curves. The details results are discussed in chapter 5.

6. Fractography

An Opto-digital microscope (OLYMPUS DSX510) used to investigate the physical micromechanisms and identifies the crack colonies. After that Scanning Electron Microscope (SEM) was conducted to depicts the fracture regions (initiation, propagation, and final fracture) of BM and FSPed material after fatigue testing without any surface treatment.

1.4 Motivation

The demand for Stainless Steel, especially duplex grade, has increased dramatically over the last decade, creating a new market. Duplex stainless steels (DSS) are sought after for their strength, corrosion resistance, attractive appearance, and low cost of maintenance. DSS is viable commercially also because they do not need protective coating and painting. In addition to that, DSS has outstanding mechanical properties, which make it a suitable candidate to replace the austenitic stainless steels, exclusively for the petrochemical and marine industry. Few studies have been carried out on the fatigue crack growth behavior of SAF 2507. Friction stir processing and its effect on the fatigue crack behavior of DSS and SDSS has not been addressed. Thus, the present work is designed to study the fatigue crack growth behavior of SDSS SAF 2507 and how it is affected FSP. The investigation starts with the optimization of the friction stir process for this high temperature material.

The state of residual stresses after FSP is then determined experimentally and by a specially developed finite element model. Finally, fatigue crack propagation characteristics and mechanisms are determined revealing the influence of friction stir processing on the FCP behavior of this important material.

1.5 Thesis Summary

This thesis comprises six chapters. Chapter one includes: an introduction, objectives, and motivations. Chapter two is the state of art literature review that covers the issues related to Friction stir processing, fatigue crack growth, fracture analysis, FEM, and heat source model. The experimental procedure is detailed in chapter three. It includes all the experimental setups and procedures for fatigue crack growth testing and fractography as well as the method for determining the residual stresses. Chapter four explains how the finite element model was developed to simulate the friction stir processing using a heat source with the help of user subroutine. In this chapter, problem idealization, heat generation equation, loads and boundary conditions, and meshing are discussed elaborately. Chapter five details the experimental and numerical results along with validations and discussion of all the outcomes. Finally, chapter six summarizes the research conclusions and recommendations for future work.]

[

CHAPTER 2

LITERATURE REVIEW

This chapter is a thorough literature review of the research works associated with duplex stainless steels, their friction stir welding and processing, fatigue crack growth, FEM, and heat source modeling. It also includes the effects of process parameters and FE modeling of the FSP/FSW process, characterization techniques, and fracture analysis. The importance of fatigue crack growth analysis in engineering is discussed. The linear elastic fracture mechanics (LEFM) approach to fatigue crack growth in metals, in general, and duplex stainless steels, in particular, are reviewed. The review of how researchers dealt with the plastic zone size effects on the variation of the fatigue crack growth rate (da/dN) with the stress intensity factor range (ΔK) is performed. The limits of the theory of linear-elastic fracture mechanics are also discussed.

2.1 FSW and FSP of duplex stainless steels

Stainless steel with a combination of two-phase microstructure austenite and ferrite is known as duplex stainless steel, and one of the members of this family is the super duplex stainless steel (SDSS) [35]. The oil and gas industry and nuclear reactors use DSS and SDSS for their attractive mechanical properties and corrosion resistance. These properties can be improved by grain refinement techniques [36].

Researchers [37,38] showed that the material properties of SDSS could be improved by decreasing the size of grains. There are numerous ways to refine the size of grains, for

example, equal channel angular pressing (ECAP), friction stir processing (FSP), and thermo-mechanical controlled processing (TMCP) [38]. FSP is a unique solid-state metal-working process, similar to Friction stir welding (FSW). In this process, a rotating tool is dipped into the workpiece and travel along a focused direction. In the beginning, this novel process was used for soft materials like magnesium alloy, aluminum etc. However, today, with the current advancement of instruments and equipment, it is feasible to perform FSP on high strength materials [39].

FSP is a metal-working method which is similar to FSW [40]. The schematic representation in Figure 2-1 illustrates the friction stir processing operation.

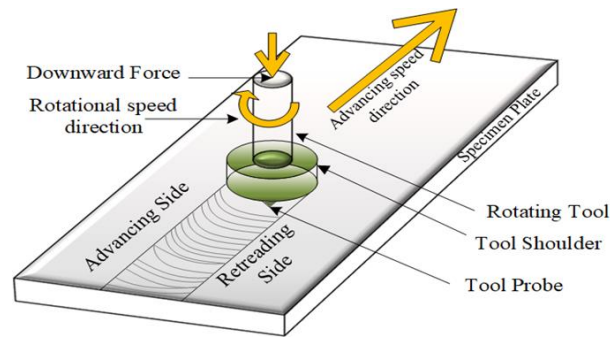


Figure 2-1: FSP schematic representation [41]

The movement of the materials around the friction processing tool may be substantially different and tremendously intricate from one tool to another due to the different geometrical structures of the tools. Three metallurgical zones develop because of FSP; Stir Zone (SZ), also referred to as Nugget Zone in comparison to conventional welding processes or (NZ), Thermo-Mechanical Affected Zone (TMAZ) and Heat Affected Zone (HAZ). Generally, the stir zone consists of fine grains structure because of dynamic recrystallization and simple plastic deformation [42].

2.2 Effect of process conditions on grain size on duplex stainless steels

A numerous research has been conducted on the impact of FSP on the material properties using different advance and rotational speeds. The effects of FSP on DSS 2205's micro and mechanical actions were investigated by Emami et al. [43,44]. FSP was performed using a tungsten carbide tool at speeds of 50 mm/min and a rotation speed of 400 rpm. Fine grains were formed due to high frictional heat and dynamic recrystallization in the stir zone (SZ) . Esmailzadeh et al. [45] also conducted a similar analysis on the microstructure of the material using a tungsten carbide test on Lean DSS. They were spinning at 800 rpm and at three different traverse rates of 50, 100 and 150 mm/min. They showed that with the increase in the traverse rate increase, the grain size decreased, and the strength and hardness were increased. Saeid et al. [46] performed FSW in 2205 SDSS using traverse speed ranging from 50-250 mm/min at fixed tool rotational speed of 600 rpm using tungsten carbide tool. The grain structure was found to become more simplified in the processed zone by increasing the advancing speed from 50 mm/min to 200 mm/min. Lack of fill defect was produced at a speed of 250 mm/min, due to lower heat input. The authors reveal that higher hardness and strength of the SZ when the grain size is small. Further analysis into 2205 SDSS [47] reveals that the sizes of the grain in the SZ are not similar, and the grains in the AS are smaller in comparison to ones at RS . Mishra et al. [48] investigated the influence of the microstructure of 2507 SDSS on various progressive speeds (10-175 mm/min). The grain size was found decreasing upto tool traverse speed of 100 mm/min and subsequently increased again when the speed was increased. The fact that FSP created a sharp border on the AS of the processed area was also revealed. On the other

hand, the scale of the grain from base metal (BM) to the RS SZ was steadily decreased. Sato et al.[49] has documented similar results using Polycrystalline Cubic Boron Nitride (PCBN) at 60 mm/min and a rotational speed of 450 rpm, for friction-stir-processed (FSPed) 2507 SDSS. Grain refinement has increased DSS strength and hardness in the SZ. Santos et al. [50] explored the impact of the microstructure and strength of DSS on rotational and processing speeds. The processing speeds for 200-600 rpm and 50-150 mm/min with a PCBN stir tool is considered. The AS had higher stresses than the RS, contributing to more grain refining at the axial load of 22 and 37 kN. The SZ was also found to be more tensile and strong compared to BM.

Figure 2-2 displays the base metal (a), one pass SZ (b) two pass SZ , and (c) phase map by EBSD. For the sample of SDSS (grade 2507). Base metal comprises lengthened grains of austenite and ferrite. One pass, SZ sample grains size reduced significantly. However, with the increase in pass number, grain size is reduced. The dynamic recrystallization occurs due to the refinement of grain size [40]. The austenite and ferrite restoration process is the dynamic recrystallization and active recovery correspondingly [41]. Santos et al. illustrated that for simple deformation of DSS, from dynamic recovery the restoration mechanism is shifted to dynamic recrystallization [42]

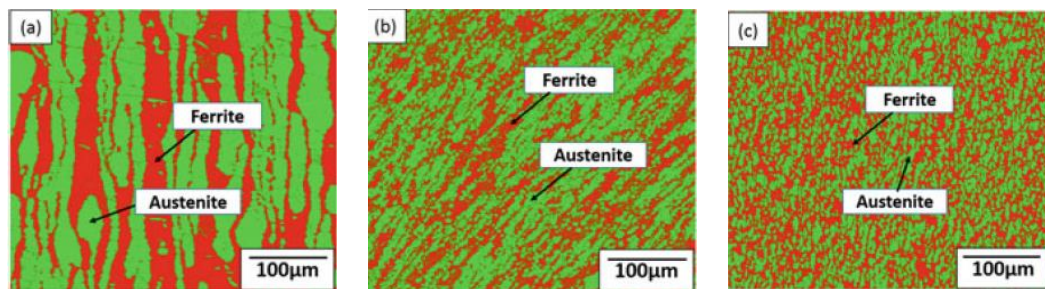


Figure 2-2: Base metal (a), one pass SZ (b), and two pass SZ (c) phase map by EBSD [40].

Takahara et al. [40] have shown that among the reasons for improved mechanical properties are as the presence of favorable residual stress distribution induced by the friction stir process. Abubaker et al. [41] observed the changes in the microstructure of SDSS SAF 2507 from the base material to the treated area in the various points (a, b, c, d) in Figure 2-3. The line intercept technique was used to calculate the average grain size within the BM and SZ, as 160 microns in the BM and 2 to 30 microns in the SZ. The refining of the grains in the SZ was mainly due to the nuclear dynamic recrystallization. In Figure 2-3, the grain sizes of the SZ can also be observed. The author results was similar to the findings published on FSWs for similar materials from Mishra et al [48] and Santos et al [50].

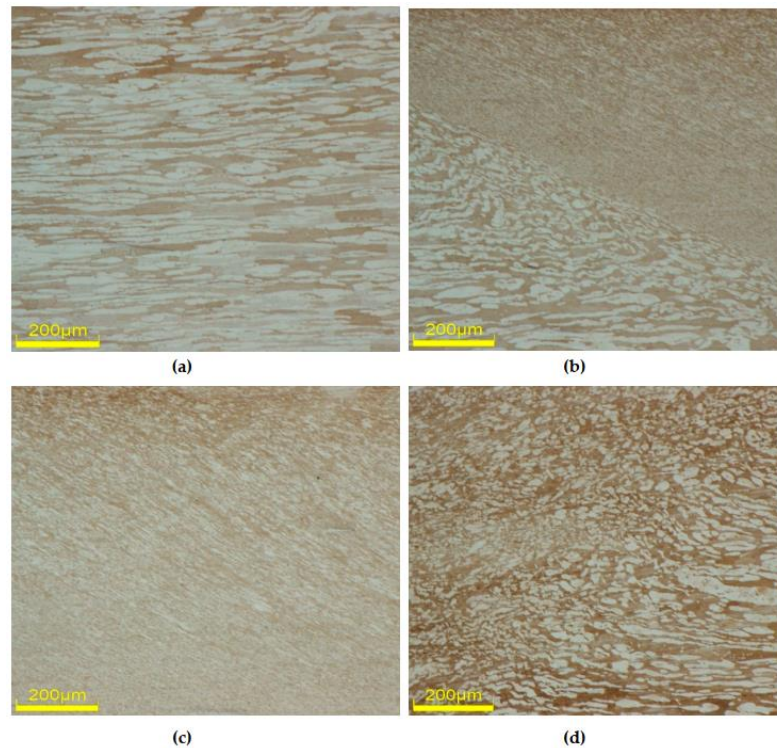


Figure 2-3: Processed material (SAF 2507) microstructure (a) Base Metal, (b) Boundary of TMAZ-AS, (c) SZ, and (d) Boundary of TMAZ-RS [41].

2.3 Residual Stress in FSW

Residual stresses (RS) are locked in stresses present in the engineering components, even when there is no external load. They may be formed throughout the life of a mechanical component or may be initiated in objects due to previous loading or technological processes like forging, thermal processing and rolling, etc. Residual stresses (RS) mainly develop due to the non-uniform volumetric change in the materials. This change can take place during the manufacturing of the component due to the application of heat like heat treatment and welding processes and deformation control based approach like friction stir processing.

Further, the residual stresses can be of either tensile type or the compressive type, depending upon the location and the nature of the non-uniform volumetric change, which is taking place. Most of the time, tensile residual stress is found harmful for the performance of the weld joints as compared to the compressive residual stresses because many times, compressive residual stresses are intentionally developed to enhance the performance. Compressive residual stresses are generally useful because they retard/stop initiation and fatigue crack propagation and increase corrosion resistance and wear. The residual stresses nature can vary depending upon how the contraction after the heating is taking place and how, the mechanical forces are being applied in a very localized manner [51]. Operations that generate destructive tensile stresses are wire or rod drawing, machining, welding, and grinding [52].

In the last decades, numerous methods were established by the researchers to measure the residual stresses of various components. Residual stresses measuring techniques are of

three types, namely non-destructive, destructive, and semi-destructive, as summarized in Figure 2-4. Non-destructive techniques contain magnetic methods, neutron or X-ray diffraction, and ultrasonic methods. Deep-hole, sectioning, ring-core, contour, and hole-drilling are the principal's mechanical methods that are used to measure the structural member's residual stresses. All the methods typically measure a parameter related to the stress. It is done to assess the fatigue-related damage so that we can take the initiative to prevent the damage [53].

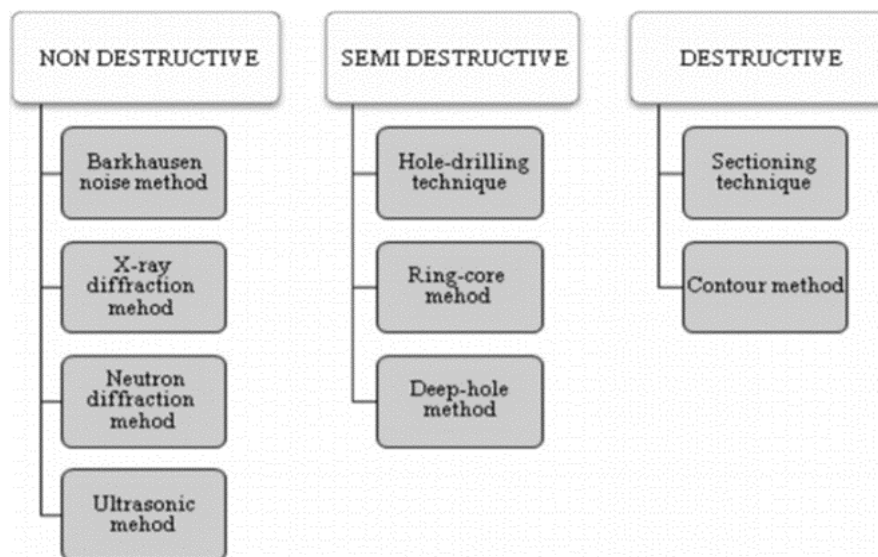


Figure 2-4: Measuring techniques of Residual stresses [53].

The hole drilling technique described in ASTM E837 standard [54] is a semi-destructive and straightforward which is also suitable for industrial and laboratory conditions (Figure 2-5). If the in-plane stress gradients are small, then this test method is applicable for the determinations of residual stress profiles. The workpiece used for the determination of residual stresses can be “thin” or “thick”. If the thickness is significantly less than the drilled hole diameter D , it is considered as “thin,” and the opposite is regarded as “thick.”

Thin workpieces are recommended only for uniform stress measurements, whereas thick workpieces are suitable for the measures of both uniform and non-uniform stress. The hole-drilling process measures the relieved strains by drilling a small hole of a component surface, at the strain gauge rosette center. The maximum allowable hole depth is about 0.4 D. The average values of surface strain are obtained through the single measurements in the grid area initiated by the stress relaxation, and the more sensitive relaxation is observed close to the surface. The sensitivity decreases with the increase in depth and goes until zero. The measured strain values at the hole location are then used to calculate the residual stresses. The released strains rely on the stresses that are initially present at the drilled hole boundaries, and the residual stresses around the strain gauge rosette are presumed to act consistently over the in-plane region and to vary with the material thickness. The released strains drop quickly with distance from the edge of the hole, and the stresses measured in the hole vicinity by the strain gauges represent about 25% to 40% of the original residual stresses.

There are some limitations of this technique as the maximum residual stresses of the materials do not exceed 50% of yield stress. If the residual stresses exceed 50% of yield stress, a correction method is defined, and that can be applied wherever stresses remain the same with depth. ASTM standard reported the limitation related to the thickness of the component, and that is the approximate results will be considered if the thickness is between 0.4 D and 1.2 D [51-56].

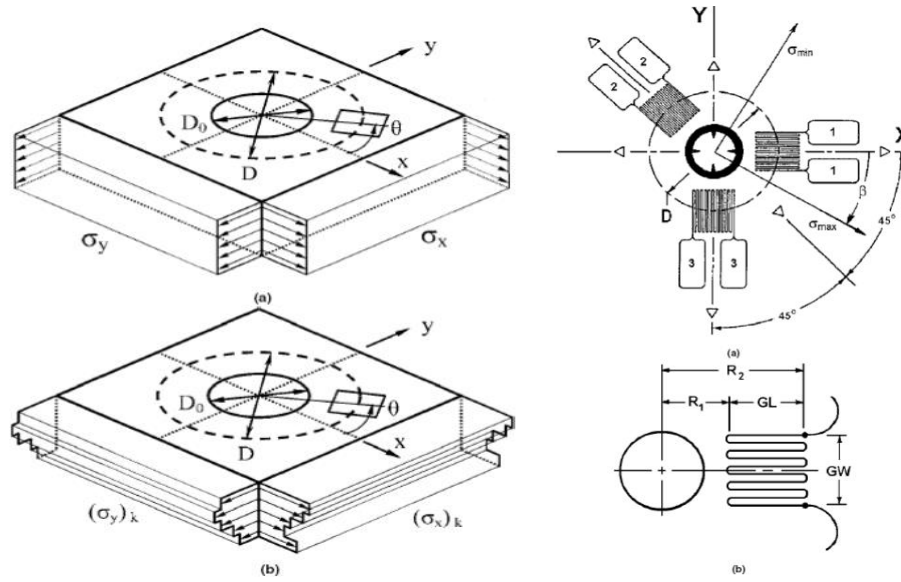


Figure 2-5: Drill hole technique (a, b) State of stress and (c, d) geometry of strain gauge rosettes [54].

2.4 Fatigue crack growth

The present section concentrates on Linear Elastic fracture mechanics (LEFM) application to fatigue crack growth (FCG) in metals, in general and DSS, in particular. The limitations which arise from the actual metallic behavior of the material are discussed as well.

Fatigue is a situation in which component is subjected to cyclic loading to failing the materials by repetitively applied loads. Design stress that is endurance strength used in fatigue loading is much less than yield strength and ultimate strength of the material. Almost 90 percent of machine components fail due to fatigue. If structural components of, for example, aircrafts, rail vehicles, wind turbines, or ships, are exposed to cyclic loading, microscopic damage can occur even below the material's ultimate strength. This damage can develop into a macroscopic crack and grow until brittle fracture. There are many reasons behind these failures, such as, inadequacies in the design, defects in the materials,

deficiencies in construction, and uncertainties in the loading. However, design against failure is a very active area of the research. Fracture mechanics is a powerful technique that is widely employed to assess the structural integrity and predict the fail-safe life of structures and components with cracks [57, 58]. In fatigue loading, life is predicted from the crack propagation rate (da/dN) versus the stress intensity factor (SIF) range ΔK .

SIF is the most common fracture mechanics parameter used when the structure behavior is linear elastic; i.e. when the plastic zone at the crack tip is small compared the crack length and thickness of the specimen.

2.4.1 Stress intensity factor (K)

The loading forces affect the stress field intensity at the crack tip. Depending on the mode of each crack surface displacement (opening, sliding shearing), a specific form of the elastic stress field can be defined by the stress intensity factor (K). The stress intensity factors for opening (K_I), sliding (K_{II}) and shearing (K_{III}) mode cracks in infinite sheets with crack length, a , and under applied stress, σ , are determined from equations 2.1, 2.2 and 2.3, respectively [59]:

$$K_I = \sigma \times \sqrt{\pi a} \quad (2.1)$$

$$K_{II} = \sigma \times \sqrt{\pi a} \quad (2.2)$$

$$K_{III} = \tau \times \sqrt{\pi a} \quad (2.3)$$

The opening mode is the most common mode of failure and the corresponding stress intensity factor, K_I , can be obtained from the applied load as follows:

$$K_I = F_p \frac{P}{t\sqrt{b}} \quad (2.4)$$

Where F_p is a geometry function, P the applied load, t the thickness of the member, and b is wide as shown in Figure 2-6.

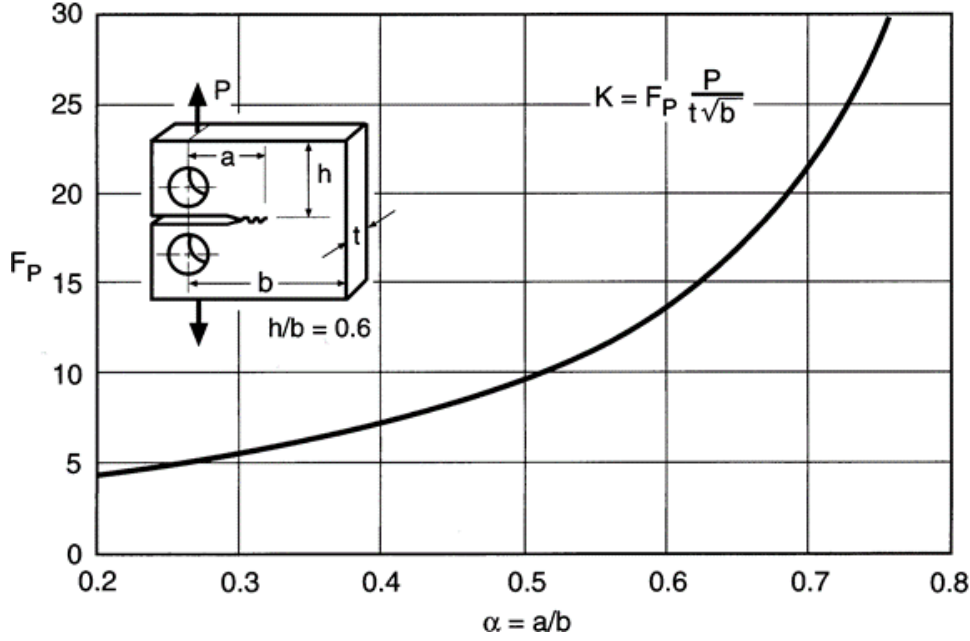


Figure 2-6: Geometry function F_P for the compact (tension) specimen CT [59].

The geometry function F_p for the CT specimen with crack length, a , and specimen width w , can be estimated from the curve of Figure 2-6 or calculated using the following equation given by Srawley [60]

$$F_p = f(a/w) = \frac{2 + \frac{a}{w}}{\left(1 - \frac{a}{w}\right)^{3/2}} \left[0.866 + 4.64 \left(\frac{a}{w}\right) - 13.32 \left(\frac{a}{w}\right)^2 + 14.72 \left(\frac{a}{w}\right)^3 - 5.60 \left(\frac{a}{w}\right)^4 \right] \quad (2.5)$$

2.4.2 Effect of plastic zone on FCG

A specimen is considered to be thick when the thickness (B) is at least $2.5 \times \left(\frac{K}{\sigma_y}\right)^2$ where, σ_y denotes the yield stress [59]. Under static loading, material yielding near to the crack tip is limited and does not affect the thick plate situation (plane strain). On the contrary, if the thickness (B) is smaller than $2.5 \times \left(\frac{K}{\sigma_y}\right)^2$ (plane stress case), the plastic zone size at the crack tip becomes important and should be dealt with by modifying the crack length. In the case of cyclic loading, the plastic zone at the crack tip is usually smaller than that for monotonic loading (Figure 2-7). If the plastic zone is insignificant compared with the crack length, stress intensity factor offers a sensible stress field description at the crack tip when the total section stress does not exceed $0.8\sigma_y$. In the case of ductile metals, under static loading, the plastic zone size radius r_p at the crack tip, can be estimated by $r_p = \frac{1}{2\pi} \times \left(\frac{K}{\sigma_y}\right)^2$ for plane stress [59]. Under cyclic loading, reverse yielding may occur during unloading; in that case, the correction for plastic zone size can be made by replacing K by ΔK and σ_y to $2\sigma_o$ and the cyclic plastic zone radius (r_c) is calculated as shown in equation 2.6:

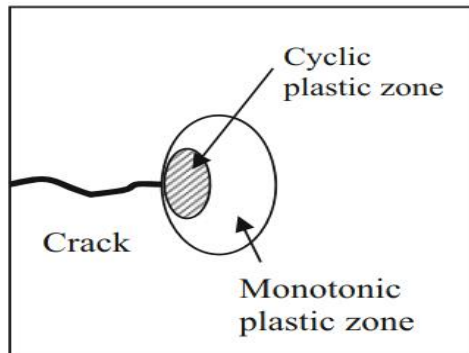


Figure 2-7: Monotonic zones and cyclic plastic at the tip of a fatigue crack [59].

$$r_c = \frac{1}{8\pi} \left(\frac{\Delta K}{\sigma_0} \right)^2 \quad (2.6)$$

Where the amplitude of the stress intensity factor is ΔK and σ_0 symbolizes the material flow strength.

2.4.3 Fatigue crack growth behavior

During fatigue crack growth testing, the measured crack length, a , is represented in terms of the number of loading cycles as shown in Figure 2-8. The crack growth rate is obtained from the slope of the a - N curve.

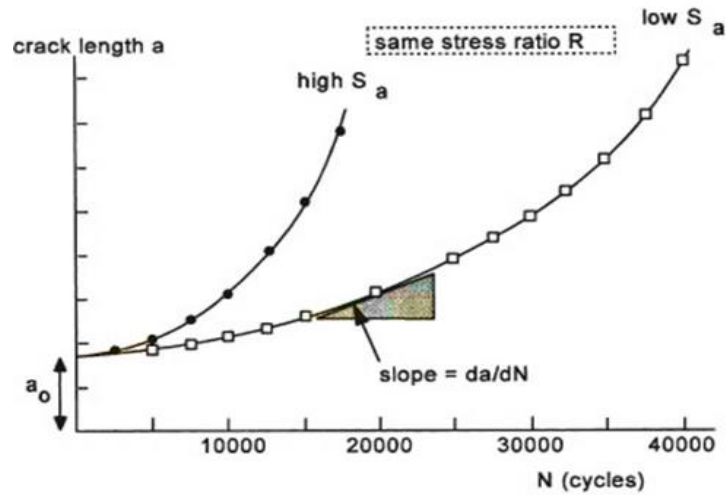


Figure 2-8: Crack length an over several cycles N at two different stress levels [61].

The relationship between the crack growth rate (da/dN) and the stress intensity factor range, ΔK , is commonly used to describe the behavior of crack growth for a given material.

Figure 2-9 shows an example of the da/dN - ΔK curve on a log-log scale.

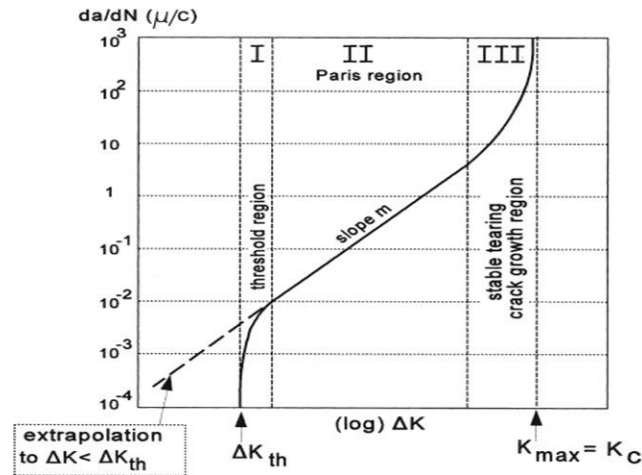


Figure 2-9: Typical da/dN - ΔK curve

With such curves obtained from experiments with standard specimen geometries (like CT specimen), the crack growth rate in a real component can be predicted using the similarity principle and life estimation can be made [59,61]. Figure 2-10 illustrates, qualitatively, the steps for obtaining fatigue crack growth data and their engineering application.

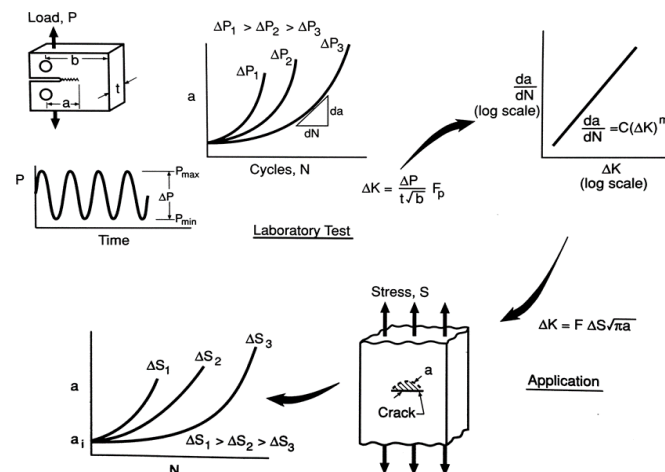


Figure 2-10: Qualitative steps on how to obtain crack growth rate curves from experiments and how to use it for an engineering application [59].

The da/dN - ΔK curve can be divided into three different regions I, II, and III. Region I is called the threshold ΔK -region. In this region at low crack growth rates, the curve is approaching the fatigue crack growth threshold ΔK_{th} which is a vertical asymptote. Below this value, no crack propagation occurs. Region III is the unstable tearing crack propagation region where the crack growth rate increases rapidly until fracture. The curve approaches the maximum stress intensity factor K_{max} which is equal to the fracture toughness K_C . In engineering, region III is of lower interest because a component crack growth life spent in this region is very low. In region II, the crack growth is stable and the correlation between da/dN and ΔK can be described by the Paris equation,

$$\frac{da}{dN} = C(\Delta K)^m \quad (2.7)$$

Where m is the slope of the curve and C is a constant with unit $\frac{mm/cycles}{MPa\sqrt{m}}$. The unit of C depends on the units of da/dN and ΔK . Both, C and m are material constants. The Paris equation yields the following linear function in the double logarithmic plot

$$\log \frac{da}{dN} = \log(C) + m \log(\Delta K) \quad (2.8)$$

By knowing C and m , the number of cycles needed to grow a crack from the initial length to its final length can be determined.

The da/dN - ΔK curve shown in Figure 2-8 is obtained when the behavior of the crack growth is not affected by the plastic zone at the crack tip, causing crack closure. It is known that, in the presence large plastic zone size, the fatigue crack closure phenomenon may occur and LEFM may not be valid. In these cases, ΔK is corrected to include the plastic

zone effect or the elasto-plastic fracture mechanics parameters such as the J-Integral range (ΔJ) is used instead of ΔK . The J-integral range is a path independent method that is used to characterize FCG for materials with important elastic-plastic behavior. Unlike SIF, J-integral considers both elasticity and plasticity at the crack tip region. A number of researchers have investigated the effect of the plastic zone size on the FCG behavior on the short cracks. Polak et al. [63] established the law of short crack growth. He found that plastic strain amplitude and J-integral methods can illustrate the short crack growth kinetics, and he also identified the first fatigue cracks initiation that occurs in the ferrite grains persistent slip bands. Chai et al. [62] demonstrated a very high closure threshold value for DSS (SAF 2507) grade that promotes the dislocation glide to very low SIF on the slip systems. The formation and growth of crack can be affected by the heterogeneity of slip activity. The rate of crack propagation reduces because of the initiation of branches, which contribute to the retardation of crack development in the key path where branching behavior is responsible for the high fatigue threshold value.

2.4.4 Fatigue crack propagation behavior of DSS

Branching and crack deflection were detected by Chai et al. [62] through a backscatter electron image (Figure 2-11) that exposes the cracking geometry. The first crack (A) is initiated after 500 cycles, the second crack (B) is after 3800 cycles, but after the 980 cycles, significant deflection (C) was observed. The test was stopped after 5100 cycles in the in-situ fatigue test. The DSS microstructure is shown in Figure 2-11b before the crack initiation through fatigue by an Electron Backscatter Diffraction (EBSD) image. The grain boundaries high angle indicated by the tiny black lines in the EBSD image are higher than

10 degrees. From the microstructure, it is clearly shown that the red mark is the fine austenitic grains, and the blue mark is the large ferritic grain. The dark line expresses the pre-fatigue cracking. [62].

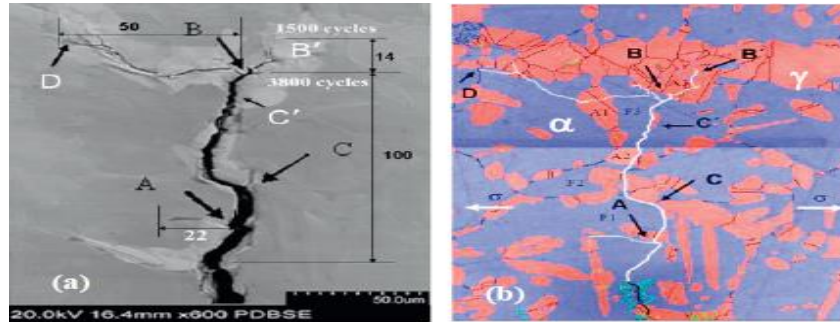


Figure 2-11: (a) The terminated and unloaded in-situ fatigue test backscatter electron image; (b) the EBSD image indicated the path of crack growth where the main crack specified by the dense white line and the tinny white lines defining the branches [62].

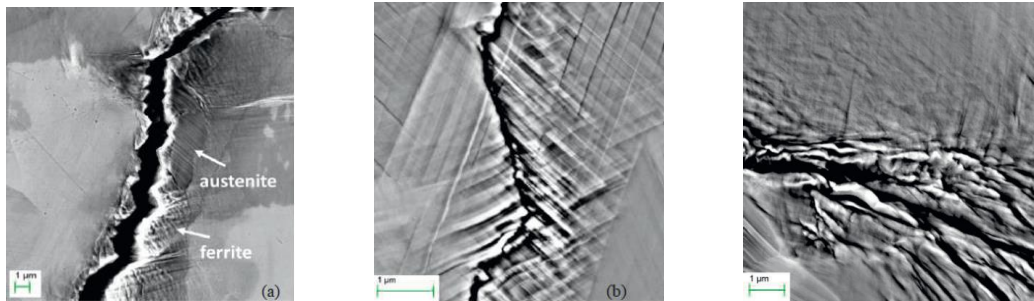


Figure 2-12: DSS crack propagation behavior (a). ferritic and austenitic phase crack, (b). austenitic phase crack, (c). ferritic phase crack [62].

Figure 2-12 shows the heterogeneous features for crack propagation and dislocation slipping of DSS in the ferritic and the phase of austenite. Planar dislocation slipping is mainly observed for the austenitic phase (Figure 2-12a) or cross planar slipping (Figure 2-12b). River shaped cracks can be seen in the crack front (Figure 15c) owing to the diverse directions of slipping dislocation interactions. Figure 2-13 indicates the fatigue crack growth (FCG) rate (da/dN) versus SIF range (ΔK) curves using a CT specimen and

traditional fatigue crack propagation measurements. The study reveal the Paris constants values C and m is 5.35×10^{-10} mm/cycles and 3.52, respectively.

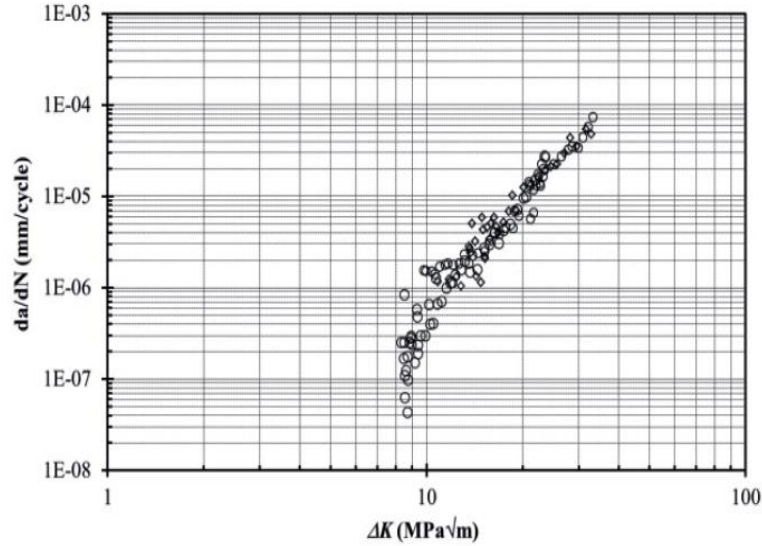


Figure 2-13: Conventional fatigue crack growth (FCG) rate (da/dN) versus SIF range (ΔK) [62].

2.4.5 Fatigue of FSW/FSP high melting temperature materials

In the beginning, the novel solid-state friction stir welding and processing (FSW/FSP) were used for soft materials like magnesium and aluminum alloys. However, today, with the current advancement of instruments and equipment, it is feasible to perform FSP on high strength materials [64]. A number of studies on FCG of friction stir welded high melting temperature materials can be found in the open literature. However, a limited number of research is reported on fatigue life, fatigue crack propagation, and fractography of SDSS before and after FSW/FSP.

Lakshminarayanan et al. [65] studied the behavior of fatigue crack growth (FCG) and fatigue life of AISI 409M ferritic stainless steel joints by performing friction stir welding

(FSW) and showed that FSW resulted in superior FCG resistance and fatigue life compared to the base material. They concluded that the enhancement of fatigue properties was mostly caused by the synergetic effect of the twin phase microstructure of DSS, favorable residual stresses, and improved tensile properties. Toumpis et al. [66] have undertaken a comprehensive assessment of fatigue performance of DH36 steel after friction stir welding and concluded that slow friction stir welding speed produces highly refined and flawless microstructure with higher fatigue life. Similar results were obtained by Lakshminarayanan and Balasubramanian [65]. They found that friction stir welded AISI 409M showed higher fatigue crack propagation (FCP) resistance and fatigue life compared to that of the base metal (BM) because of finer microstructure and compressive residual stress distribution.

2.4.6 Dual Phase (DP) steels fracture behavior

The brittle fracture of DP steels occurs due to the existence of a coarse microstructure [67-69] and interrelated martensite islands [70,71]. In the ferrite grain, a cleavage crack is triggered owing to stress building up at the tip of crack when breaks the martensite. With the increase of volume fraction, crack is growing up as the martensite phase size increase. Martensite connectivity is enhanced for large volume fraction, which is the main reason for cleavage occurrence in the ferrite. For the QT-700-37%, dimples cover the central part of the rupture surface, and there is no uniform distribution of cleavage on the fracture surface. Under the uniaxial tension, failure happens due to the fracture, as a yielding rupture is a primary mechanism to pledge in the middle of the samples [72,73].

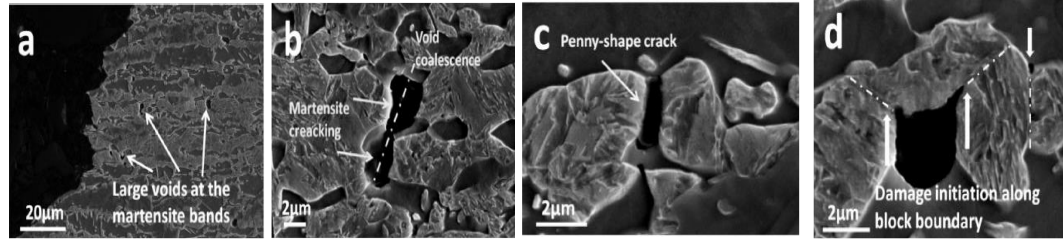


Figure 2-14: Damage Mechanism of QT-700-37% and fracture surface void observation [72].

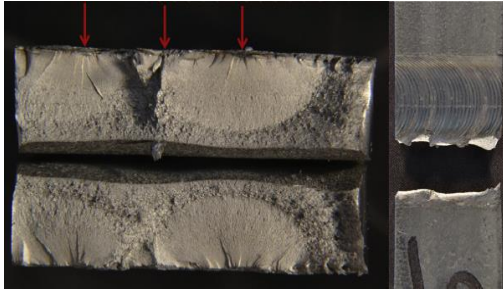
Figure 2-14 display the damage mechanism of QT-700-37%, which relates to large volume fraction for dual-phase steel. The Large void at the martensite bands (Figure 2-14a), martensite fracture was observed (Figure 2-14b). Between the combination of two adjacent large voids. The formation of Penny-shape voids by the block boundary is shown in Figure 2-14d while martensite fracture in Figure 2-14c .

2.4.7 Fracture surface analysis

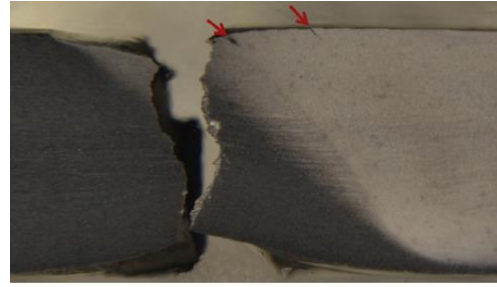
In ship structures, particularly in stiffened panel DH36 steel is commonly used. The steel plates with 6 mm thickness was friction stir welded using Megastir Q70 pcBN-WRe tool with a scrolled shoulder diameter of 36.8 mm and stepped spiral probe of 5.7 mm length. As shown in Table 2-1, FSW was carried out at various traverse and rotational speeds.

Table 2-1: Sets of welding parameters [74].

Welding speed group	Slow	Intermediate	Fast
Traverse speed (mm/min)	100	250	500
Rotational speed (rpm)	200	300	700



(a)



(b)

Figure 2-15: intermediate weld fracture surface specimens and Side view of fracture path Macrograph [74]

The observation of crack initiation and propagation for the recurrent pattern of a fracture surface specimen is shown in Figure 2-15. Which exhibits real crack initiation and lap defect on the top of the weld surface. The arrow indicates the crack initiation of multiple sites. In the parent materials, the fracture path propagated perpendicular to the top surface shown in Figure 2-15. From the SEM observation, it can be explained that after coalescence (MVC), microvoid, and crack-tip blunting fracture, the specimens failed with trans crystalline cleavage below 80 1C. Unstable fracture is happening of ductile fracture till the distance is ordered of 200 μm ; see Figure 2-16a and 2-16b. The study on interstitial free (IF) steel plates shows an ultimate tensile strength (451 MPa) and higher yield strength (325 MPa) and after FSP than Base materials (192 MPa and 314 MPa).

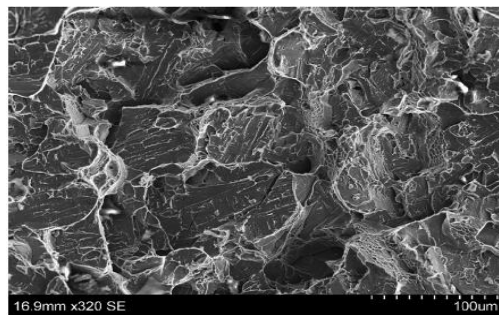
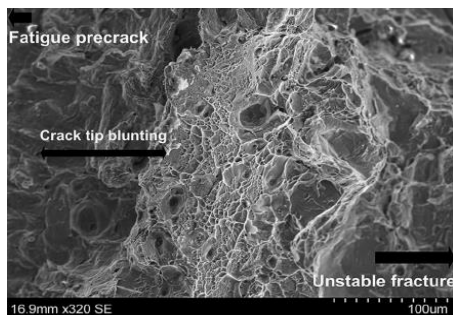


Figure 2-16: Base metal fracture surface SEM images: (a) ductile fracture and crack-tip blunting and (b) cleavage fracture of trans crystalline [75].

Moreover, in the NaCl and air solution, the Friction stir processing of interstitial free steel revealed higher fatigue life than the base materials. Still, the elongation remains the same for both cases, which are 67.5% to 66.2%, respectively. The processed steel displayed a transgranular fracture, but the base materials offered an intergranular fracture. The fatigue cracks growth and increased resistance of nucleation are credited due to the enhanced corrosion fatigue performance [74]. Homogenization [74,75] microstructure refinement [76] are severe plastic deformation method for FSP [77] which happens instantaneously. The fabricated materials by FSP are capable of improving the ductility and strength [78,79] with small dislocation densities and a larger ratio of high-angle grain boundaries [80,81].

2.5 Finite Element modeling

Al-Badour et al. [82] developed a localized 3-D FEM for the prediction of defect generation of friction stir welding. They used Eulerian formulation for modeling the workpiece and frictional contact of the FE model. Lagrangian formulation was used for modeling the tool while the interaction between the workpiece and the tool was described using Coulomb's frictional model. The inflow and outflow velocities of the material define the speed of welding. They showed that the significant effect for void formation is due to the coefficient of Friction (see Figure 2-17) and lower the coefficient of Friction more considerable the void and also lower the speed of weld, smaller the size of the void. Thus, welding at a lower rate creates a smaller void and broader plastic zone and gives high-quality weld. Therefore, the effect of welding speed is phenomenal.

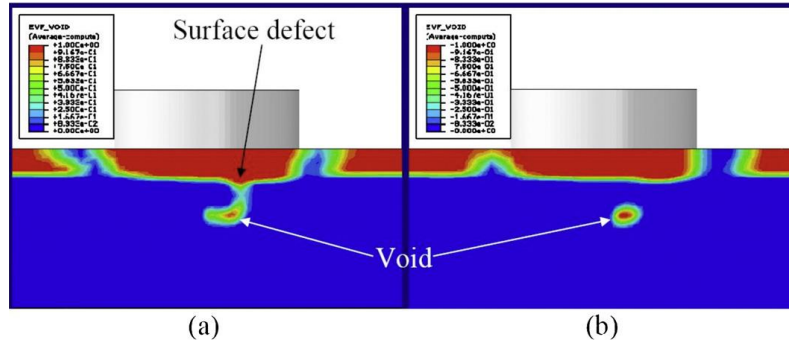


Figure 2-17: Void formation at friction coefficient 0.58(a), and 0.8(b) [82].

Veljić et al. [83] developed a coupled thermo-mechanical model to study the temperature fields, the plastic deformations, and the plunging force during FSW process of 2024-T351 Al alloy under various rotating speeds (350, 400 and 450 rpm). The model was built in ABAQUS/Explicit by using Lagrangian-Eulerian formulation with Johnson Cook material law, and Coulomb's friction law. The study of the FSW process depicted that the matrix temperature is smaller than that of the melting value. The temperature distribution along the welding line is nearly symmetrical. The force of plunge will be minimized if the speed is increased. Near the welding tool on the trailing side on the lower plastic strain area can be identified. The low plastic strain area retreats with an increasing speed of rotation. In relation to the returning side, an equal plastic strain is raised on the advancing side. The temperature field and time dependent force are shown in Figure 2-18 and 2-19 respectively.

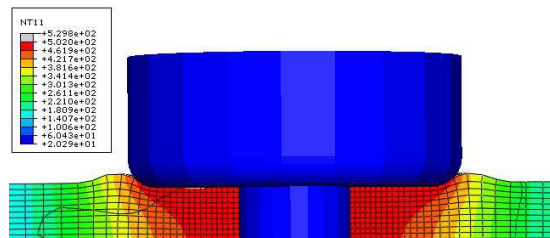


Figure 2-18: Temperature fields at 350 rpm rotation speeds [83].

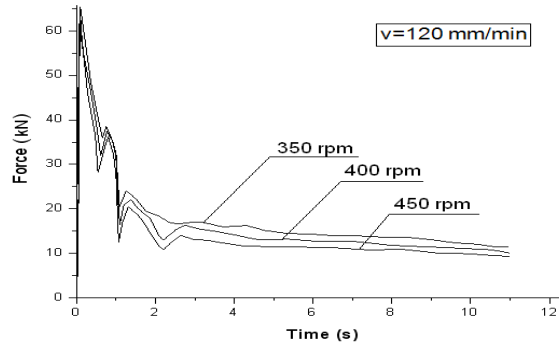


Figure 2-19: The time-dependent force [83].

de Jesus et al. [84] analyzed the effect of numerous tool geometries on the friction stir welding morphology, like microstructure, processed regions hardness, and material flow. The study discovered that tool geometry influenced processed regions' quality. The Convex shoulder geometry of the tool allows the nugget to be placed on the weld toe but cause weld defects such as cavities or plate thickness reduction. FSP in both the alloys (AA 6082-T651 and AA 5083-H111) examined induces substantial grain refinements, increases the fatigue strength, and eliminates defects, such as pores and weight loss, preexisting in the MIG welding ring. The improvement of fatigue strength occurs for both alloys only for the cylindrical threaded pin and rounded and concave edge shoulder tools. Yu et al. [85] used a coupled thermomechanical 3-D FEM for friction stir welding in the ABAQUS environment using the failure model and material law of Johnson Cook. He investigated the effect of heat conduction and temperature evolution of friction stir welded aluminum alloy 7050 for the moving stages, dwell, and plunge. Results showed the symmetrical distribution of temperature across the cross-section of the plate, and V type shape presents in the weld nugget zone. The frictional heat which conducts in the dwelling stage, preheat the plate. The heat was steadily accumulated during the moving stage until a quasi-stable

temperature field is established. The temperature field surprisingly shrinks with the growing heat convective coefficient of the backplate. Hattel et al. [86] focused on the classifying and presenting residual stress models for aluminium alloys FSW according to their application, background, and numerical framework. It is because the residual stresses affect the component's final mechanical performance. They divided the FSW model into flow models, microstructural models, thermal models, and residual stress models.

Buffa et al. [87] examined the residual stress field of AA7075-T6 due to the mechanical and thermal actions during FSW. This process has used the Lagrangian implicit code to highlights the induced residual stresses and metallurgical phenomena, both experimentally and mathematically. They made the comparison of residual stress profiles with the numerical model through the cut-compliance methodology and have shown the residual stress prediction in the FSW joint by the mechanical tool action. Zhang et al. [88] explored the temperature distributions and the axial pressure effects on the behavior of the AL 6061-T6 material during the FSW process for a fully coupled thermomechanical model. They found that the critical factor is axial pressure for successful FSW, and failure occurs for excessive or insufficient axial force.

2.6 FE heat source modeling

Heat source modeling for the welding process is the main intricacy of the thermal field simulation. The foundation of analytical heat distribution for welding was commenced by the Rosenthal [89-91] during the late nineteenth century. His heat flows analytic solution made feasible for the first time to analyze the process parameters, i.e., the weld geometry, voltage, current, and welding speed. Zaeh et al. [92] employed an experimental uniform

thermal load to heat 20 powder layers to calculate the residual stress fields and temperature of a cantilever beam with a support on the substrate. Chao et al. [93] established the FSW process heat transfer using two boundary value problems (BVP): (a) transient for the workpiece and (b) steady-state BVP for the tool and. They revealed about 95% heat efficiency in FSW, which is exceptionally high compared to the conventional fusion welding as the traditional efficiency of heat is about (60-80)%. The author considered model are shown in Figure 2-20.

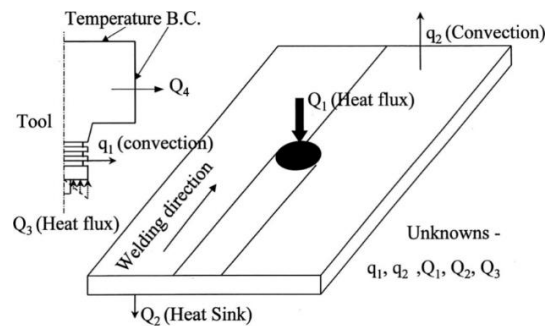


Figure 2-20: Illustration of heat transfer in tool and workpiece in FSW by Chao et al. [93] (tool half model due to symmetry).

Heat conduction mathematical models through moving heat sources have real applications in several manufacturing processes like grinding, welding, and drilling of metal and metal cutting. Few researchers have taken this problem as a subject of publications and used analytical techniques to simulate such manufacturing processes [94-96]. However, in regular domains, this type of model has limited applications, but it is very profitable in process analysis. Araya et al. [94] presented a temperature distributed analytical solution for the 3D finite domain. Gaussian distribution was used to model the heat source, and the authors were concerned about the temperature of a workpiece due to boundary effects. Cheng et al. [95] investigated 3D temperature fields in a finite thick plate using an analytical model. A moving heat source heated the plate with Gaussian distribution. The author presumed that the heat source stirred with a fixed velocity along the line. Manca et

al. [96] assumed a circular Gaussian heat source stirring at a fixed speed over a solid surface with finite width and depth. In 1998, Chao et al. [97] published a 3-D heat transfer FE model, where a constant heat flux input was assumed from the tool shoulder/workpiece interface, this paper adjusted the heat input with the measured temperature on the trial-and-error basis. Al-Badour et al. [98] developed a 3-D thermomechanical FE model to investigate the residual stresses of FSSealW of tube-tube sheet joint using DFLUX and DLOAD and validated the outcome by experimental result. In that investigation, the author used the temperature-dependent properties of the materials. Frigaard et al. [99] developed an FSW model, where the tool shoulder frictional heat was assumed as heat input. Based on the calculated temperature, the friction coefficient; used to calculate heat input, throughout the welding process was adjusted in order not to exceed the melting point of the material. This FSW model used the Rosenthal equation for thin plate heat transfer. Recently, the researcher focused on FSW thermal analysis and heat transfer. Few authors [100-102] recommended a simple model of heat transfer to forecast the workpiece temperature distributions. FE formulation for a moving heat source applicable for welding applications was developed in [103]. Several FSW processes were examined for coupled thermo-mechanical modeling [93,104]. The amount of heat energy generated through FSW obtained numerically was compared with experimental data. Khandkar et al. [105] introduced a torque-based heat input model, where heat source expression used the innovative torque/power. Similarly, Shi et al. [106] used experimental mechanical power as an input in a 3D thermo-mechanical FE model and observed the effect of torque and plunge force on the residual stresses.

CHAPTER 3

Experimental Investigation

In this chapter, a short description of processed material, as well as experimental setup and the calculation procedures for fatigue crack growth, have been considered. The residual stress determination procedure, types, and the selection of hole drilled methods for measuring the residual stress have been reviewed in detail. The hole drilled method measurement principle, scope, and experimental setup have been presented elaborately.

3.1 Base materials

The material utilized in this study is SDSS grade UNS S32760 (SAF2705) manufactured by ArcelorMittal [107]. The as-received base material chemical composition is shown in Table 3-1, while the BM and FSPed materials mechanical properties are listed in Table 3-2. Typical optical micrograph of the as-received base material in the rolling direction is shown in Figure 3-1. The microstructure consists of a ferrite matrix with austenite islands.

Table 3-1: Chemical composition (wt. %) of as-received SAF 2507 [107].

%Ni	%Si	%C	%Cu	%S	%P	%Mo	%Cr	%Mn	%N	%Fe	PRE
6.681	0.357	0.026	0.114	0.0002	0.027	3.729	25.299	0.847	0.28	Balance	42

Table 3-2: Average Mechanical Properties of as received and FSPed SAF 2507.

Material	Tensile Strength (MPa)	Yield Strength (MPa)	Vickers Hardness (VH)
BM [41]	793.6	655.6	275.6
FSPed [41]	890.2	749.1	314.3

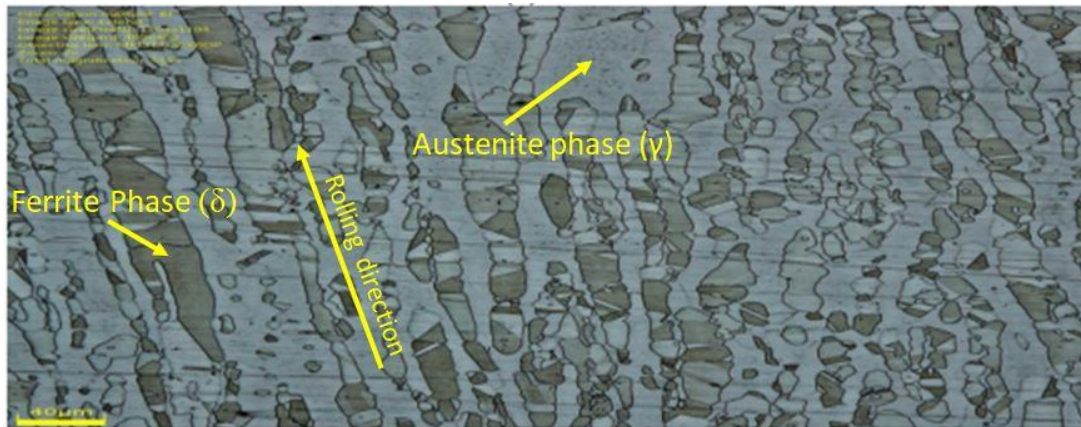


Figure 3-1: Optical microscope image of as-received SAF 2507 ferrite and austenite grains along the rolling direction.

Figure 3-1 shows the optical microstructure image of as-received SAF 2507 ferrite and austenite grains along the rolling direction. The as received SDSS SAF 2507 have two-phase microstructure comprising of grains of austenitic and ferritic stainless steel. The picture depicts the bright phase as austenitic islands” surrounded by the dark ferritic phase. The result is a microstructure of approximately equal volume fractions of ferrite (δ) and austenite (γ).

3.2 Friction Stir Processing

FSP is considered as an emerging and prominent solid-state surface modification technology that can refine the microstructures, remove casting defects, and enhance the mechanical properties of the processed area. FSP enhances properties such strength, formability and resistance against fatigue and corrosion. In the current study, SDSS SAF2507, a high-strength material, was subjected to FSP to manipulate the microstructure locally and achieve grain refinement to enhance fatigue crack propagation resistance.

3.2.1 Experimental Setup for FSP

Research-based friction stir welder RM-1 (MTI-USA) was used to process SDSS SAF2507 plates, with dimensions (210x100x6.5 mm). A 25 mm convex-scrolled shoulder PCBN tool Q70 (Megastir-USA), with conical pin; 5.2 mm length, was used in this work (Figure 3-2). A convex shoulder can monitor the footprint of the tool and the footprint relies on the force of the tool or the depth of penetration. A water-cooled tool holder (Megastir-USA) was used to maintain tool assembly within the acceptable operating temperature. Generally, in FSW processing, relative movement of the workpiece for the backing plate is not allowed, as this results in defect bond formation or shifting tool path. Therefore, the workpiece must be clamped rigidly to the backing plate. Conventional clamps used in milling operations were utilized to rigidly fix the workpiece to a steel backing plate. FSP area temperatures were monitored using an infrared camera (FLIR T-660-USA). Figure 3-3 presents the experimental setup.

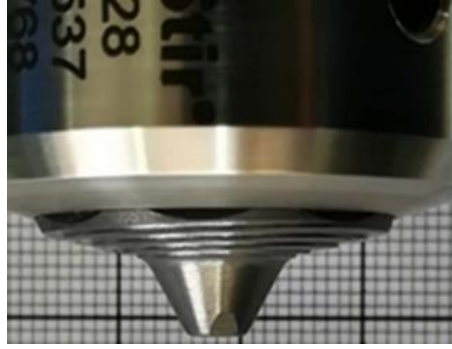


Figure 3-2: Real tool

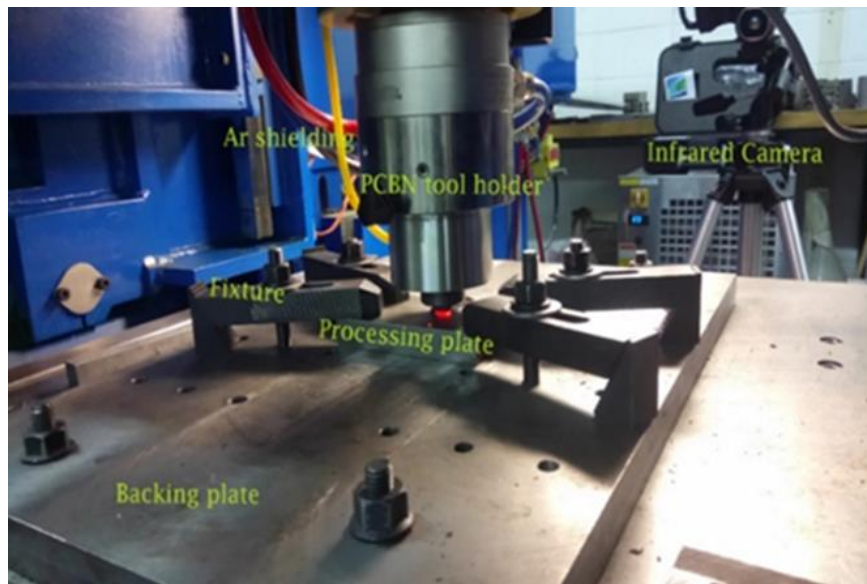


Figure 3-3: Experimental setup of FSP.

3.2.2 Optimization of FSP

In this study, experiments were directed to investigate the effects of FSP parameters for various transverse speeds 50, 100 and 150 mm/min and a rotational speed of 400 rpm on the weld quality of super duplex stainless steel (SDSS) as can be seen in Table 3-3, the best weld quality (smooth surface with little to no flash) resulted for 100 mm/min traverse speed

and 400 rpm rotational speeds. The processing parameters concerning the rotational speed, tilt angle and plunging depths were optimized in earlier work by the research team [41]. The optimal parameters were obtained by performing a single-pass bead under position control at 400 rpm and a tilt angle of 2° with a tool plunging depth of 5.8 mm. After the processing, specimens were cooled at room temperature. Samples were prepared effectively, and mechanical tests were carried out to validate the specimens such as visual inspection, microstructural observation of the welded surface, hardness test, and FCG test.

Table 3-3: Optimization of Friction stirrs processing (FSP).

FSP No.	Depth of Plunging in mm	Travelling speed mm/min with rotation in rpm	Processed surface finishing	Remarks
1	5.2	50 @400		Rough surface with lateral flash
2	5.8	100@400		Smooth surface with no flash
3	6.2	150@400		Smooth surface with lateral flash

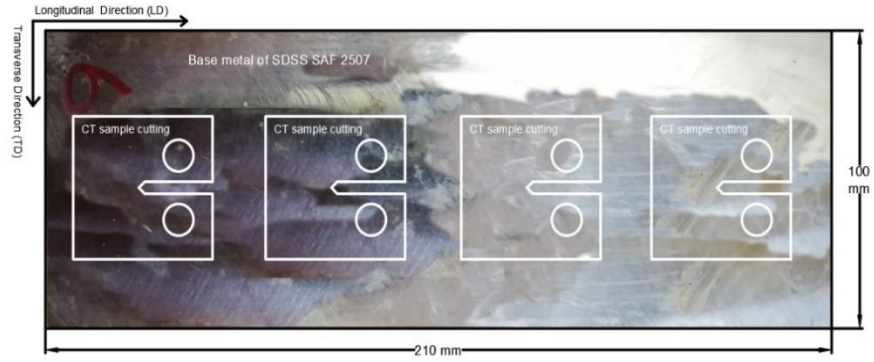
3.3 Fatigue Crack Growth Testing

The behaviour of crack propagation under cyclic loading is known as fatigue crack growth. The propagation of the crack occurs on the atomic level breakage owing to the imperfections in the atomic structure. Hence, new surfaces are created in the solid as the crack nucleates and continue to grow. The stable or unstable crack growth arises on the attributes of the driving force dependence versus crack extension and the material resistance to continuous growth. Nevertheless, the component will fail at some point due to lack of load-bearing capacity by assuming a strictly increasing load. The details of the fatigue crack growth investigation of material SAF 2507 is given below.

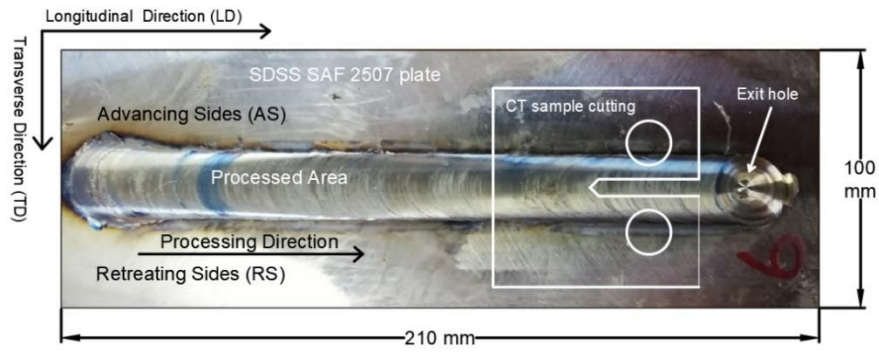
3.3.1 Specimens preparation

The as-received, 20 mm thick hot rolled plates were cut into strips of 6.5 mm thickness for FSP and specimen preparation. Compact-Tension (CT) specimens were cut from the machined plates, in the rolling direction, using electric discharge machining (EDM). The crack propagation is the easiest along longitudinal direction (rolling direction) because in this direction, the fine and the coarse grains is most aligned and elongated. On the contrary, the crack propagation preference is lower in transverse directions [108]. The present study is considered the worst case (highest possibility) of fatigue crack growth rate. Therefore, this study cut the samples in the rolling direction to investigate the changes in the microstructure within the processed zone and the adjacent ones. Figures 3-4-a and 3-4-b show how the CT specimens extracted from the base metal plate and Friction stir processed plate, respectively. The specimen geometry and dimensions; prepared according to the

ASTM E647 standard [109], are presented in Figure 3-5. The CT specimens surfaces were grounded and polished by abrasive paper before performing the fatigue test.



(a) CT samples cutting from the base metal plate



(b) CT samples cutting from FSPed plate

Figure 3-4: (a) and (b) Cutting of CT specimens for BM and FSPed plate, respectively.

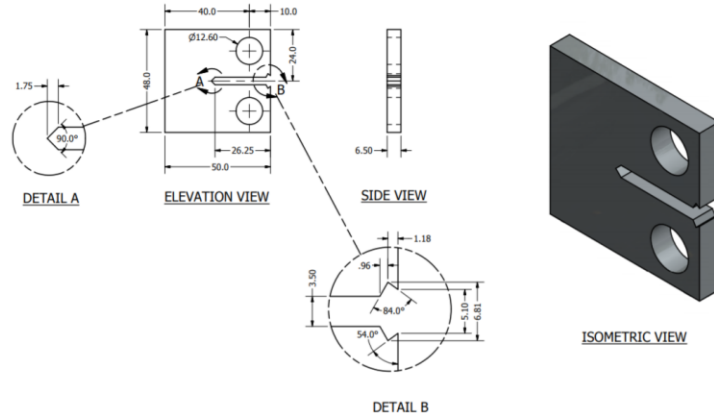


Figure 3-5:Designed C (T) specimen.

3.3.2 FCG testing

In this present study, fatigue crack propagation tests were performed on the CT samples at three different stress range levels, i.e., 66, 94, and 113 MPa. The tests were conducted using INSTRON 8801 computer-regulated Servo-Hydraulic Machine with a capacity of 100 kN. All experiments were performed in room temperature conditions, and at a constant load amplitude, a sinusoidal waveform, a constant stress ratio ($R=0.1$), and at 10 Hz frequency. The experimental setup in Figure 3-6 shows the specimen as mounted on the testing frame. The crack length (a) was measured using a high-resolution camera with a magnification factor up to 200x.

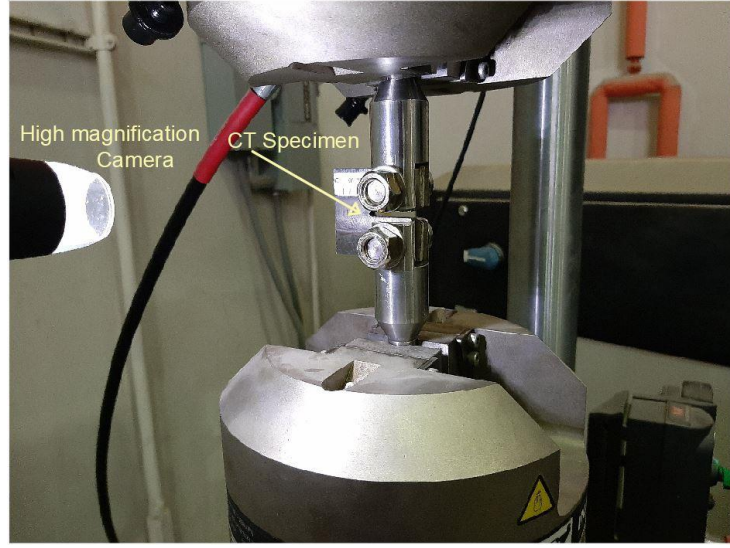


Figure 3-6: Experimental setup of FCP.

The experimental results were analyzed using fracture mechanics based empirical Paris power equation [110] proposed by Paris-Erdogan which is given below

$$\frac{\Delta a}{\Delta N} = C \times (\Delta K)^m \quad (3.1)$$

Where, $\frac{da}{dN}$ indicates the crack propagation rate, ΔK dignify the range of stress intensity factor (SIF), and C and m are the material constants that have to be calculated. For the diverse values of crack length, the SIF values were determined with the following expression [111].

$$\Delta K = f(a/w) \times \Delta \sigma \times \sqrt{\pi a} \quad (3.2)$$

However, the factor of geometry can be calculated using equation 3.3 [60]

$$f(a/w) = \frac{2 + \frac{a}{w}}{\left(1 - \frac{a}{w}\right)^{3/2}} \left[0.866 + 4.64 \left(\frac{a}{w}\right) - 13.32 \left(\frac{a}{w}\right)^2 + 14.72 \left(\frac{a}{w}\right)^3 - 5.60 \left(\frac{a}{w}\right)^4 \right] \quad (3.3)$$

The crack growth was recorded from the beginning to the final failure of the specimen. Similar crack propagation tests were carried out for several specimens and recorded the tested data.

3.3.3 Data analysis

A simple method was applied to obtain the FCG rate (da/dN) from crack length vs. the number of cycles data by determining the slope between successive points as illustrated in Figure 3-7. The numbering of data points is 1, 2, 3 . . . j; then the ending point is j for the segment of the growth rate.

$$\frac{da}{dN} \approx \frac{\Delta a}{\Delta N} = \frac{a_j - a_{j-1}}{N_j - N_{j-1}} \quad (3.4)$$

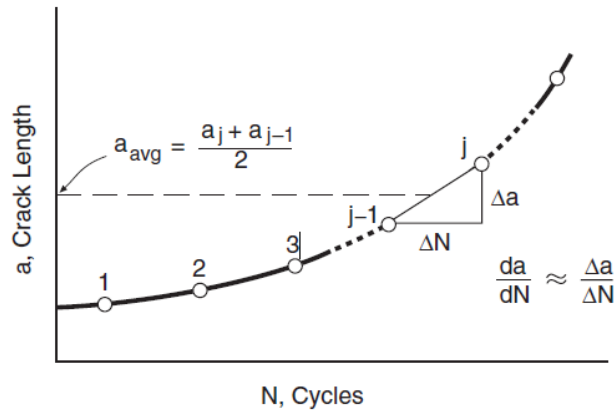


Figure 3-7: Crack length vs. N data points adjacent pairs to obtaining the crack growth rates.

The modified stress intensity factor for CT specimen was calculated using

$$\Delta K_j = \frac{\Delta P f(a/w)}{B\sqrt{w}} \quad (3.5)$$

where: B = thickness [m], ΔP = load range [N], W = width [m], and $f(a/W)$ = the compact tension specimen geometric correction factor, where $f(a/W) = f(\alpha)$ and $\alpha = a/W$, is evaluated at the same average crack length, using

$$\alpha_{avg} = \frac{a_{avg}}{W} = \frac{a_j - a_{j-1}}{2W} \quad (3.6)$$

Unfortunately, the result obtained from equations 3.4 and 3.5 cannot give ready-made values of the desired constants. Therefore, the crack growth rate da/dN vs stress intensity factor range (ΔK) are plotted on the log-log scale. The slope of the plot would give us the "m" and the Y intercept will give us the value of C. The Paris Law constants (C and m) for the SAF 2507 material before and after friction stir processing are calculated using the equation 3.1.

3.4 Residual Stress Measurement

The hole-drilling (HD) method is a standardized technique and widely used to determine the residual stresses adjacent to the surface of a component [112]. Typically, a hole is drilled in a small depth increment. After each depth increment, the cutting is paused, and the deformation near the hole is measured through strain gauges that measure the released stresses due to the generated hole. This method can also be applied to determine the distribution of residual stresses through the specimen depth. Following ASTM E837 [113], residual stresses near the surface were measured. For that purpose, Vishay RS-200, which consists of a milling guide, depth setting gauge, and a high-speed turbine (Figure 3-8) was used to perform the hole drilling and 3-element EA-06-062RE-120 strain gage rosettes from the Vishay-micro-measurement group were mounted on the test plate near advancing and retreating sides to measure developed residual stresses after FSP. Before mounting,

the sample surface was prepared according to the manufacturer's manual. After the sample preparation, the strain gage rosettes were bonded to the surface using the M-bond 200 adhesive and the lead wires were soldered to the strain gage terminals. The drilling of small holes was performed with a carbide-tipped cutter drill bit with 1.8 mm diameter. Hole drilling position was performed perpendicular to the sample in the center of strain gauge rosettes. The measured strain data from blind hole drilling was fed to the H-drill v.2⁺ software and the built-in integral method was selected to estimate the principal stresses.

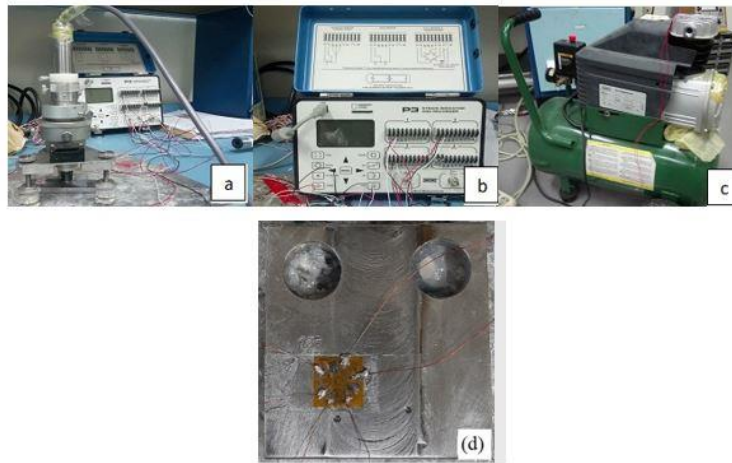


Figure 3-8: Experimental setup for residual stress determination (a) Hole drilling setup, (b) P3 Strain gauge indicator, (c) An air turbine, (d) Bonded strain gauge on the processed sample.

The experimental setup for the determination of residual stresses, using drill hole method is shown in Figure 3-8. The hole-drill (HD) device (Figure 3-8 (a)) was installed and fixed accurately above the strain gauge rosette (Figure 3-8 (d)) as stated by the RS-200 Milling Guide. The alignment of the drill device crosshairs with the strain gauge center is the most challenging and most important task to do. For the accurate measurement, it is mandatory to ensure that the point of interaction of the crosshairs coincides exactly with the center of

the strain gauge circle. That was achieved with the help of the microscope attached to the HD setup.

The model P3 strain indicator data recorder (Figure 3-8(b)) was first connected to a PC for data acquisition. Due to the high cutting speed requirements (around 200,000 rpm), an air turbine (Figure 3-8(c)) was used in the drilling operation with a diamond-coated tungsten carbide cutter. Each new experiment required a new drill bit to minimize the errors associated with the shape of a drilled hole. After establishing the zero-depth and setting the strain readings to zero, the hole drilling was performed in increments of 0.064 mm until a depth of 1.0 mm. After reaching each increment, the strain readings were stabilized for a couple of minutes before recording the values. Following hole drilling (HD) completion, the drill hole diameter was measured with an optical microscope that is attached to the setup.

CHAPTER 4

FINITE ELEMENT METHOD

In this chapter, friction stir processing has been established by FEM heat source modeling in a commercial Software ABAQUS 2019 using user subroutine. The details about heat source modeling like problem idealization, assumptions, materials model, heat generation equation, loads and boundary condition, and meshing have been discussed thoroughly.

4.1 Friction Stir Process

Friction stir processing (FSP) is an emerging material enhancement technique developed from the solid-state welding technique known as friction stir welding (FSW). FSW was invented by the welding institute in the 1990s [114] with a significant number of rewards like improved mechanical properties and reduced material distortion compared to other fusion welding processes. FSP is accomplished using a specially designed rotating tool that exerts heat through friction on the alloy, causing the material to soften while being stirred as a result of combined rotational and translational motion along the desired path as illustrated in Figure 4-1. The mechanical properties enhancement is accomplished by altering the microstructure and elimination of internal voids or defects in the case of cast structures [115]. In FSP, the workpiece never reaches a melting point because of the solid-state nature of the process; therefore, associate defects from melting-solidification phases are avoided, and processed material obtains higher mechanical properties in comparison to

the conventional methods. The stirring is considered a plastic deformation under a relatively low temperature was causing dynamic recrystallization as well as grain growth. To be optimized as a material enhancement or grain refinement technique, the grain growth must be minimized by altering the rotational speed, feed rate, and welding medium [116,117]. The plastic deformation occurs due to the pin, whereas temperature rise and frictional deformation happen due to the shoulder effect. The fined grain microstructure generates through the pin stirring action.

Depending on the necessity of the plate surface to be modified, a multi-pass or single-pass friction stir process can be applied [118]. During the processing, the material becomes softer due to localized heating around the pin; then, the material moves from front to back of the probe because of the combined translation and rotation motion of the tool. The tool shoulder restricts the flow of metal underneath the lowest shoulder surface. The movement of the materials around the probe may be substantially different and tremendously intricate from one tool to another due to the different geometrical structures of the tools. Three metallurgical zones develop because of FSP; Stir Zone (SZ), also referred to as Nugget Zone in comparison to conventional welding processes or (NZ), Thermo-Mechanical Affected Zone (TMAZ) and Heat Affected Zone (HAZ). Generally, the stir zone consists of fine grain structure because of dynamic recrystallization and simple plastic deformation [119]. Transverse and rotational speed evaluate the quantity of heat input in the stir zone, which affects the materials properties and the microstructure [120]. As the heat input decrease, the grain refinement increases and vice-versa, and the heat input must be adequate to soften or plasticize the material; if it is not enough, the processed zone will have defected.

To understanding the problem and minimize experimental work, finite element models capable of simulating the FSP are deemed necessary to predict the resulting thermal profile and large deformation at different rotational and translational speeds as well as residual stresses. Due to the tremendous distortion occurring owing to material stirring, researchers used different techniques in FEA to minimize mesh distortion; these include Arbitrary Lagrangian-Eulerian (ALE) [121,122] and Coupled Eulerian-Lagrangian CEL [123,124]. Schmidt et al. [125] developed a thermo-mechanical model using Abaqus/Explicit to investigate the conditions where FSW takes place flawlessly, and on the weld line, no voids occur. Zhang et al. also developed a thermo-mechanical model using Abaqus/Explicit to predict the changes in material deformation and thermal histories in FSW using different welding parameters and ALE formulation. The FE model revealed that with the increase in rotational speed increases the welding temperature, power consumption, and the sample residual stress [126]. One problem associated with the flow-based simulation of FSW is the high computational time.

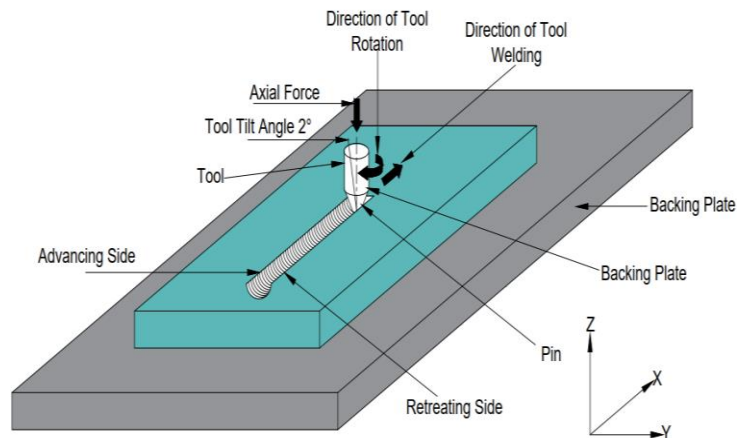


Figure 4-1: A typical Diagram of FSP/FSW

Song et al. [127] investigated the influence of the dwell period on the temperature fields. In both studies, they used experimental plunge force and an effective friction coefficient in the expression of the heat source by assuming pure sliding conditions. In general, a moving heat source problem required mostly four data for thermal analysis [101]:

1. The heat source strength and distribution.
2. Cooling media for convection to reflects the coolant effect.
3. The work materials thermal properties and
4. The heat source moving speed.

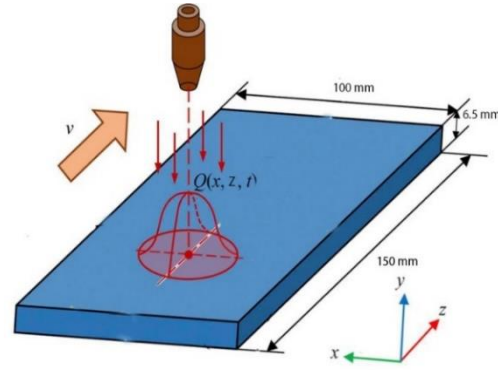
The cooling rate and temperature gradient near the surface of the material can affect the hardness distribution, thermal shrinkage, residual stresses, microstructures, and heat-affected zone. The prediction of thermal fields through numerical models induced by a moving heat source is a handy tool to investigate the above-stated problems. By knowing the distribution of temperature due to friction, it is feasible to minimize issues that are related to temperature. Thus, in the transient and stationary case, the friction between two mediums are often modeled by the concept of moving heat source. This study aims to predict the temperature gradient generated during FSP of SDSS SAF 2507 induced by FSP tool shoulder and probe considering both sliding/sticking conditions. Moreover, estimating the developed residual stresses in the workpiece after FSP. Residual stresses develop in the material due to the variation of heating and cooling condition as well as material deformation during FSP. The nature of the residual stresses substantially affects the processing quality and fatigue crack growth propagation of the material. The present study used the SDSS grade sheet because of its enormous commercial applications nowadays in the field of oil and gas as well as desalination industry.

4.1.1 Problem idealization:

FSP is a fully coupled thermo-mechanical solid-state process without the use of any filler metal. This process consists of a tool and one or more plates. The present study uses a single plate of super duplex stainless steel (SAF 2507) for friction stir processing.



(a)



(b)

Figure 4-2: (a) Real FSP, (b) Idealize FEM model for FSP using a heat source (Q), and distributed pressure (P).

As the FSP is a fully coupled thermo-mechanical process, the generation of heat is related to frictional or contact conditions as well as material flow. Without a pre-knowledge of the heat generation it cannot be feasible to predict the temperature history by the thermal model that is why several analytical expressions are available in the literature for the generation of heat for a weld as a function of welding or processing parameters (e.g., rotational speed) and tool geometry (e.g., tool radius). In this work, the tool heat effect was introduced as Gaussian moving heat source (see equation 4.1 & 4.2). The amount of heat production by the tool shoulder and the pin can be calculated using sticking and sliding conditions (see equation 4.2 to 18) of the tool throughout the processing and applied as a moving surface heat flux through user subroutine. Figure 4-2(a) shows the real experimental setup, and

Figure 4-2(b) illustrates the idealized model using heat source replacing the tool. However, modeling of FSW/FSP is tackled using the two forms; non-flow based modeling, where the tool is replaced by a heat source and distributed pressure, while the other technique is using flow-based simulation (based on ALE or CEL); which are computationally heavy and the details of these modeling will be found [128-130]. The authors have utilized the non-flow-based method due to its low computational cost; also, the residual stresses within the HAZ are expected to be identical, using flow or non-flow simulation. Although researchers carried out various types of heat source models, the limitations of this model have not been specified clearly. As the Gaussian heat source model generated heat over a surface, it may not be applicable for thick plates. However, several researchers [131,132] proposed and established heat generation analytical equation for the friction stir tool because of the highest amount of heat production by the workpiece and tool shoulder interaction. The present study applied those simplified equations for FSP. The present study also specifies the limitations for this model as incapability of simulating the asymmetry of FSP.

4.1.2 Assumption during heat source modeling

The following consideration is taken into account during the heat source modeling of FSP.

1. The isotropic tool is considered as Gaussian moving heat source.
2. Material properties of the workpiece are temperature dependent.
3. Workpiece as Lagrangian domain
4. The tool is considered as a cylindrical taper tool during heat calculation.

4.1.3 Heat generation:

Three conditions for the analytical heat estimations can be considered due to tool-workpiece interaction: (a) sticking condition, $\delta = 1$ (full sticking), (b) pure sliding condition in the interface, $\delta = 0$ (full sliding) and (c) partial sticking or sliding condition [131]. Where δ is a contact state variable, which connects the velocity of the contact points relative to the tool point at the material surface. For the sticking condition, it is assumed that the shear occurs in a layer very close to the interface and is limited to processed material shear strength, on the other hand, for the sliding condition, it takes place at the contact interface and is governed by Coulomb's friction model. The FSP tool used in this study has multifaceted geometries (see Figure 4-3). For simplification of the analytical estimation, a simple tool design with a flat shoulder and pin tip surfaces and tapered pin surface were presumed. Figure 4-4 shows the location of different regions for heat generation during friction stir processing. The heat generated by tool shoulder, side of the tool pin and tip of the tool pin is Q_1 , Q_2 , and Q_3 respectively and can be calculated based on equations (4.1).

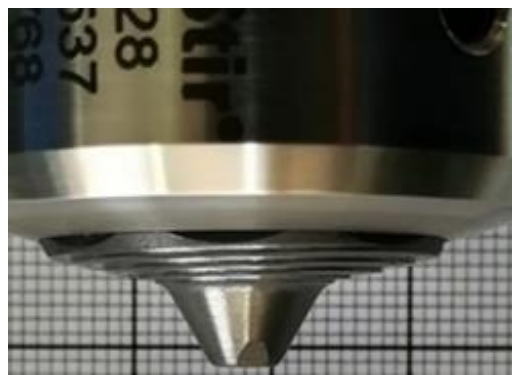


Figure 4-3: Real tool

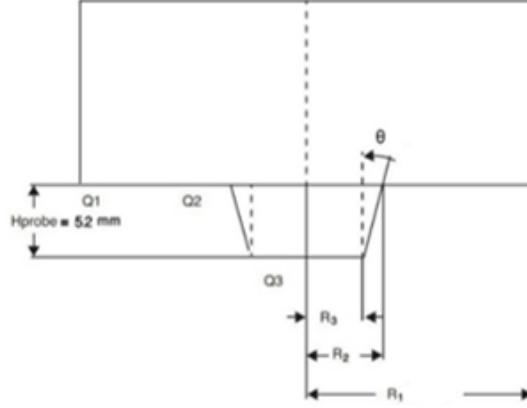


Figure 4-4: Location of different regions for heat generation from the tool.

$$\left. \begin{aligned} Q_1 &= \frac{2}{3} \pi \omega \tau_{contact} * (R_1^3 - R_2^3) \\ Q_2 &= \frac{\pi \omega \tau_{contact}}{2} * \frac{H_{probe}}{\cos \theta} * (2R_2 - H_{probe} * \tan \theta)^2 \\ Q_3 &= \frac{2}{3} \pi \omega \tau_{contact} (R_2 - H_{probe} * \tan \theta)^3 \end{aligned} \right\} \quad (4.1)$$

Where ω is the tool rotation in rad/s, $\tau_{contact}$ is the developed shear stress at the tool-workpiece interface in Pa, R_1 tool shoulder radius, R_2 tool pin root radius, and R_3 tool pin tip radius, θ pin taper angle, and H is the tool pin length. The total heat can be calculated for the taper pin FSP tool by adding the heat produced by each element, as shown in equation (4.2) [131]:

$$Q_{total} = \frac{2}{3} \pi \omega \tau_{contact} \left[R_1^3 - R_2^3 + \frac{3}{4} * \frac{H_{probe}}{\cos \theta} * (2 * R_2 - H_{probe} * \tan \theta)^2 + (R_2 - H_{probe} * \tan \theta)^3 \right] \dots \dots \dots (4.2)$$

In sticking condition, the contact shear stress $\tau_{contact}$ is equal to material shear yield strength. Considering distortion energy theorem, $\tau_{contact} = \frac{\sigma_{yield}}{\sqrt{3}}$ [132].

Considering pure sliding between the tool and workpiece, the contact shear stress is governed by Coulomb's frictional law, where the contact shear stress in equation (4.2) will equal $\tau_{contact} = \mu p$; μ is the kinematic coefficient of friction developed between the tool

and workpiece, and p is the contact pressures. P can be calculated based on average experimentally recorded tool forging force and the contact area, ignoring the conical pin surface. $\left(P = \frac{F}{A}\right)$

Assuming pure sticking or sliding is not feasible, simultaneous sticking slipping or partially stick-slip phenomenon is expected in FSP. Therefore, the heat generation during welding can be estimated based on contact state variable δ as shown in equation 4.3.

$$Q_{total} = \delta * Q_{total,sticking} + (1 - \delta) * Q_{total,sliding} \dots \dots \dots (4.3)$$

In this work, the adopted values in calculating heat generation are listed in Table 4-1.

Table 4-1: Data used in the analytical expression for the tool heat Generation.

Symbol	Magnitude	Reference
R_1	10 (mm)	-
R_2	5 (mm)	-
θ	30 (degree)	-
H_{probe}	5.2 (mm)	-
μ	0.2	[133]
p	88 MPa	measured forging force
δ	0.9	-

In this investigation, the heat source was considered as a 2D Gaussian distribution; surface flux, as earlier illustrated in Figure 4-2(b). Therefore, the distribution of heat flux on the solid surface is related to radial position $r^2 = x^2 + z^2$, for the tool center. The distribution of Gaussian heat flux is described in equations (4.4) and (4.5) [134]:

$$Q = Q_0 \text{Exp}^{-\left(\frac{r^2}{r_0^2}\right)} \dots \dots \dots (4.4)$$

Where Q_0 signifies the maximum power in Watt; r denotes the radial distance to the center, and r_0 is the radius of the heat source. Therefore, the Gaussian moving circular heat source intensity (I), can be calculated using equation (4.5) [135]:

$$I(x, z) = \frac{2 * Q_w}{\pi r_o^2} \text{Exp} \left[\frac{-2 * (x^2 + z^2)}{r_o^2} \right] \dots \dots \dots (4.5)$$

Where r_0 is the radius of the Gaussian moving heat source and Q_w , the amount of heat transferred from the tool to the workpiece. The amount of heat produced at the interface of the tool-workpiece and transferred to the tool can be estimated from the tool and workpiece material thermal properties by the analytical expression shown in (4.6), which is based on a point heat source between dissimilar materials.

$$f = \frac{Q_w}{Q_T} = \sqrt{\frac{(k \rho_* C_p)_W}{(k \rho_* C_p)_T}} \dots \dots \dots (4.6)$$

Where C_p is material specific heat, the density of the material is ρ , and κ is thermal conductivity, and T and W denote the tool and the workpiece, correspondingly [136].

Therefore, Q_w as a function of total heat produced can be presented

$$Q_w = \frac{f+1}{f} Q_{total} \dots \dots \dots (4.7)$$

Here, Q_{total} can be obtained from equation 4.3, as described in the above section.

4.1.4 Material model

The temperature-dependent thermal and mechanical properties are considered in this investigation. Elastic-perfectly plastic model was assumed to describe the material flow stress, while considering the effect of temperature. Figure 4-5 presents the temperature dependent properties considered in this work i.e. modulus of elasticity (E), yield strength (Sy), thermal expansion (α), thermal conductivity (k), and specific heat (Cp). Other physical parameters were considered constant (independent of temperature) such as Density (7800 kg/m³) and poisson's ratio (0.27) .

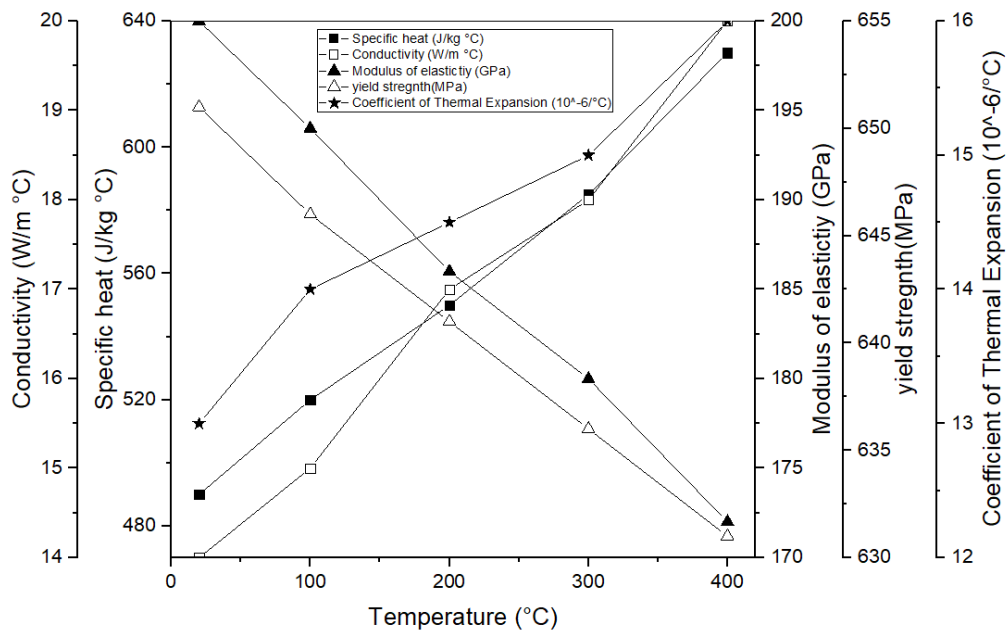


Figure 4-5: Temperature-dependent properties of SAF 2507 [41,137-139].

4.1.5 Loads and Boundary Conditions

The interaction between tool and workpiece was idealized through moving forging load and Gaussian heat, where the tool moves in a linear direction on the workpiece. The

movement of the pressure, i.e., forging load and heat source, was controlled using subroutines DLOAD and DFLUX, respectively. Both subroutines are built-in Abaqus (2019) that define the surface pressure distribution and surface heat flux as a function of position and time. For each time increment, the tool movement was simulated by changing the pressure and heat flux position. The pressure exerted by the tool, and the heat flux were applied to the surface area that is equivalent to the tool shoulder. The pressure was calculated using the average measured forging force during the experiment, which was 27.8 kN during the welding phase. Figure 4-6 presents the measured forging force vs tool traveled distance. Practically, the forging force is steady if the process conditions are within an optimum window, where process temperature maintains constant (steady state) also force can be constant.

Considering the efficiency of the process, where part of heat will transfer to the tool and the other into the workpiece, the amount of heat converted to the workpiece was calculated using equation (4.6), and the value f is 0.64; considering properties at room temperature. The total calculated heat flux is 3135.19 W, and the heat flux values were calculated using equations 4.3 and 4.7. Figure 4-7 presents the measured tool torque vs. tool traveled distance. Torque-traveled distance trend shows stable nature similar to forging force.

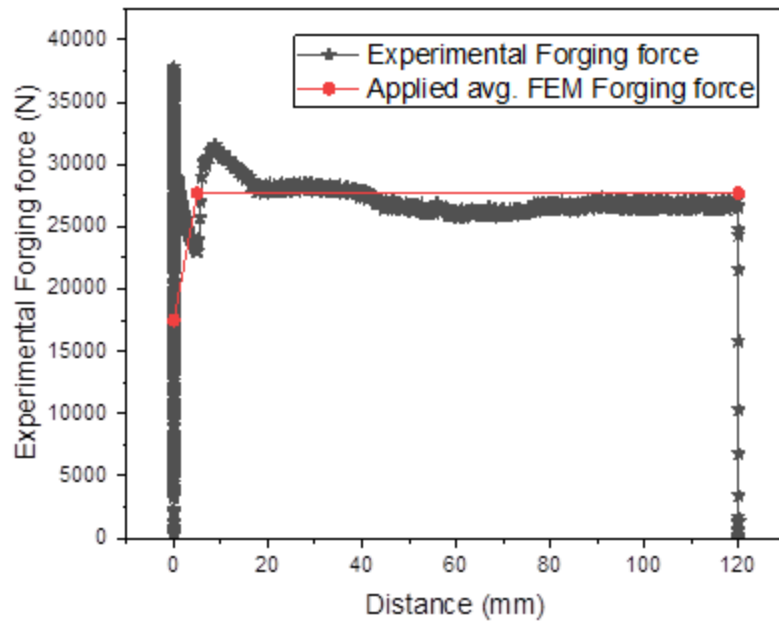


Figure 4-6: Forging force measured during FSP of SDSS.

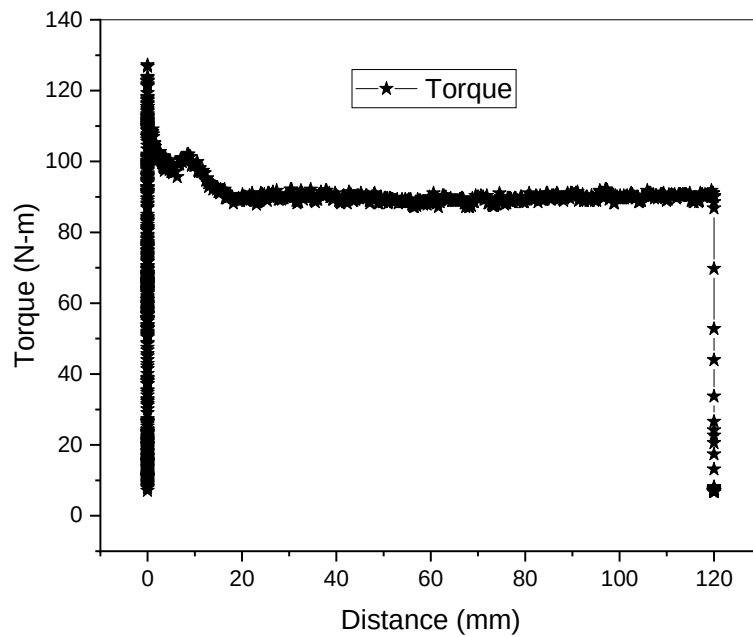


Figure 4-7: Load vs tool travel distance.

Heat dissipation to the surrounding considered being by natural convective heat transfer.

Researchers have used diverse values of h (convective heat transfer coefficients) for

FSW/FSP simulation, which is in the range of $25\text{-}30\text{ Wm}^{-2}\text{K}^{-1}$ [125, 140]. The current work used $25\text{ Wm}^{-2}\text{K}^{-1}$. A fictitious convective heat transfer was used at the back of the plate to mimic the heat conduction into the backing plate, which is affected by multiple factors, i.e., workpiece\backing plate materials and pressure at interface. The fictitious convective heat transfer coefficient was found in the literature to be $2000\text{ Wm}^{-2}\text{K}^{-1}$ [141]. Mechanically, displacement constraints were applied to simulate the action of the clamp.

4.1.6 Element selection and Mesh

An 8-node thermally coupled brick, trilinear displacement, and temperature control element C3D8T was used to mesh the plate. Bias seeding was applied to generate a fine mesh near the processing zone with a total number of elements equal to 23040. This method reduces the computational time and improves the solution accuracy. The size of the mesh around the tool plunge; side length is 1 mm, and the outside of the tool area is 2 mm. The plate was partitioned into smaller cells to obtain the graded mesh. Before selecting the final model, several mesh densities were tried, and the present one (Figure 4-8) was found to provide good accuracy with reasonable computational time. Table 4-2 lists the mesh density vs computational time.

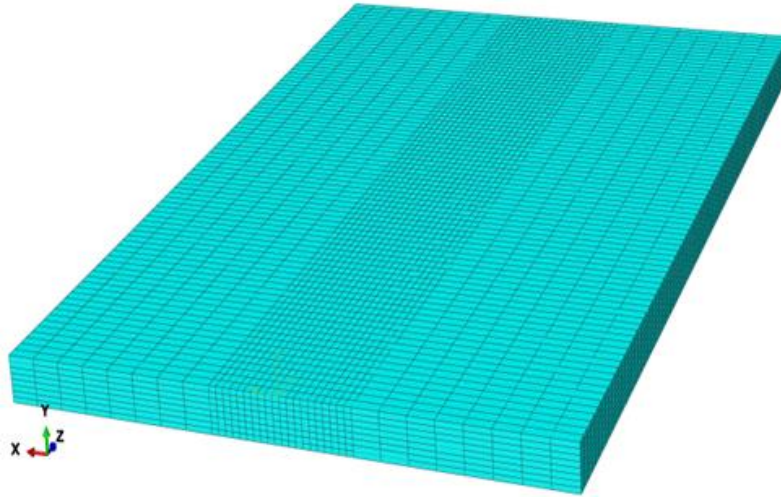


Figure 4-8: Meshing of the heat source model.

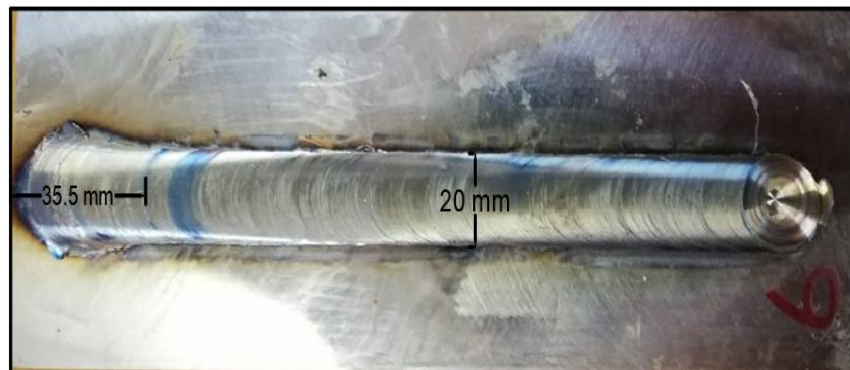
Table 4-2: Element size vs computational time.

Iteration no.	Element size (Side length of tool plunge area and outside of the tool plunge area)	Total no. of Element	Computational time* (days)	Max. temperature (°C)
1	0.5 mm and 1.5 mm	86400	16	1391
2	0.8 mm and 1.8 mm	54750	10	1377
3	1mm and 2 mm	23040	4	1351
*CPU: Core i7 (8th generation) RAM: 16 GB OS: Windows 10				

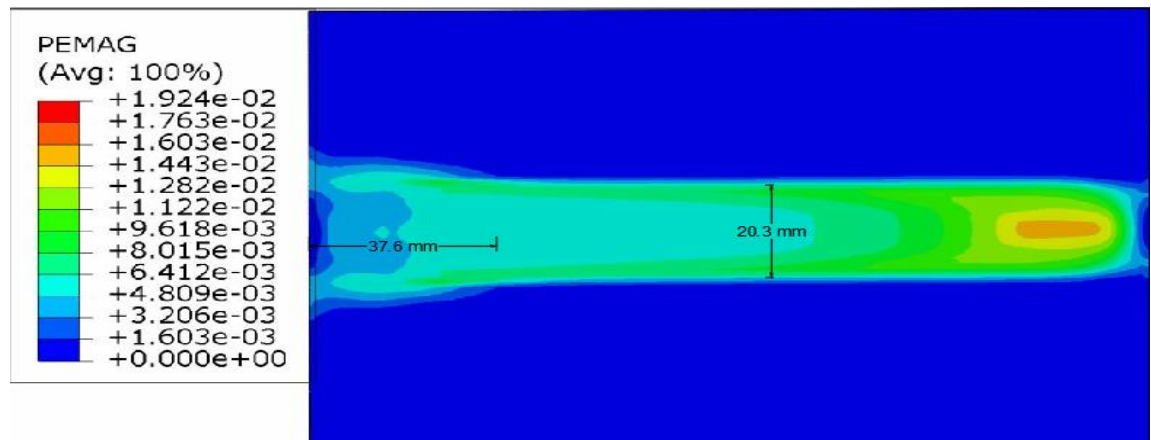
4.2 Validation of FEM

The experimentally measured loads (forging force) and material yield strength ($\sigma_y = 651 \text{ MPa}$) [41] were used to calculate the thermal and mechanical stresses and to validate the estimated FE model. Qualitative and quantitative validation of the model was achieved by considering the developed plasticized zone, temperature buildup, and residual stresses.

The bead developed by the FE heat source model (Figure 4-9 (b)) was compared with that obtained after FSP (Figure 4-9 (a)). Figure 4-9 (b) represents the contour plot of plastic strain magnitude estimated using the developed model. It can be easily noticed that the overall shape of the FEM developed bead is matching that of the real bead, with an average width of about 20 mm. FEM seems to have overestimated the length of the plunging zone by about 6%, 37.6 mm instead of 35.5 mm.



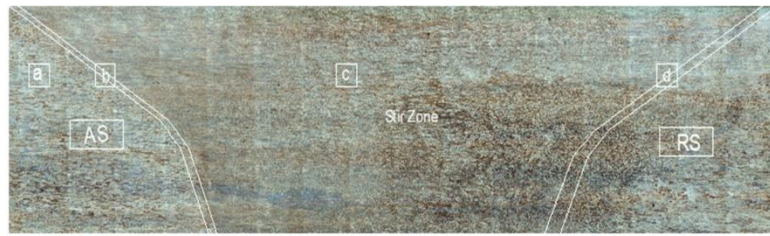
(a) The full shape of experimental FSP bead



(b) Full shape contour of heat source FE bead

Figure 4-9: Processed bead comparison (a) real bead and (d) contour of heat source model bead.

Figures 4-10 (a) and (b) display the optical microscope image of the bead cross-section and estimated plasticized zone using the developed FEM, respectively. The experimental optical cross-sectional image was taken from an earlier work by the team on the same material using the optimal friction stir processing parameters [41]. The symbols: a, b, c, and d indicate, respectively base metal, the advancing side (AS), the thermo-mechanically affected zone (TMAZ), the stir zone (SZ), and the retreating side (RS). The comparison between the optical microstructural image and the predicted plasticized zone implies that all of these four zones are present in the simulated plasticized zone. The difference in SZ is mainly because the real tool is tapered, whereas the heat source model is applied on the surface. Moreover, the assumption of the Gaussian heat source does not capture the asymmetry of FSP between AS and RS.



(a) Experimental optical microstructure



(b) Heat source microstructure

Figure 4-10: Microstructural comparison of SDSS SAF 2507 after FSP (a) Experimental optical microstructure [41], and (b) Heat source microstructure.

The estimated temperature profile across the processed line is shown in Figure 4-11-a. The figure shows that the temperature is dropping rapidly across the process line when the heat source travelling in the processing direction. Thus, the temperature gradient around the process zone is very high that denotes the key characteristics of FSP when compared to any other process. Figure 4-11-b presents the thermography captured experimentally during processing. It can be noted that the highest temperatures are almost identical. A good agreement is found between simulated and measured temperatures. The highest temperature estimated was 946.6°C, whereas the average temperature was about 677.8°C. The infrared camera measures the peak temperature value of 1059.5°C, but the average temperature is 727.8°C. Thus, the average temperature difference is about 7%. The error may be a result of the assumption of constant heat input in the finite element model or the artificially assumed conduction into backing plate. Although the outcome is in the acceptable range, other parameters such as thermal boundary conditions can be considered to yield better results.

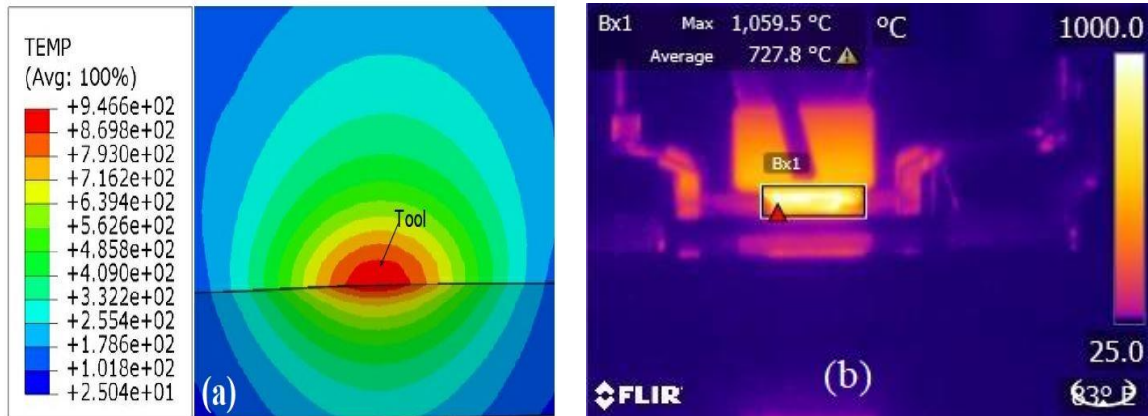


Figure 4-11: Temperature distribution during welding (a) heat source model, and (b) experimental infrared camera.

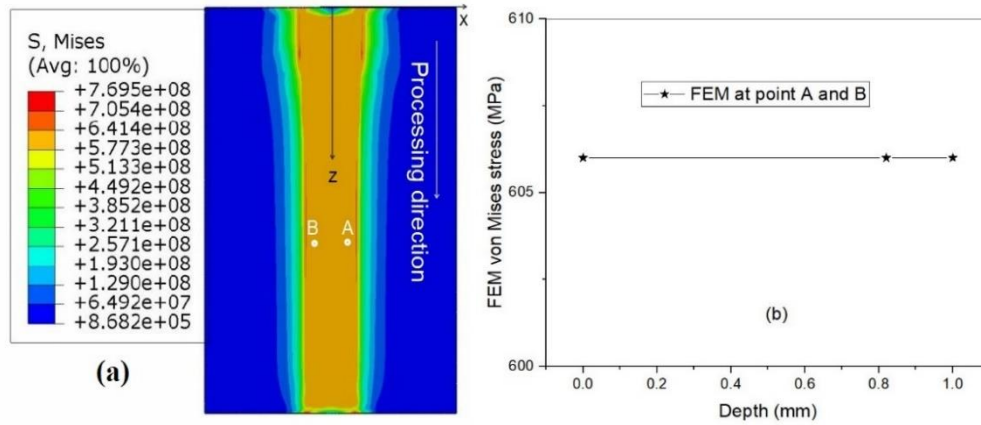


Figure 4-12: (a) Contours of FEM simulated Equivalent stress (von Mises) (b) Von Mises RS of FEM over depth.

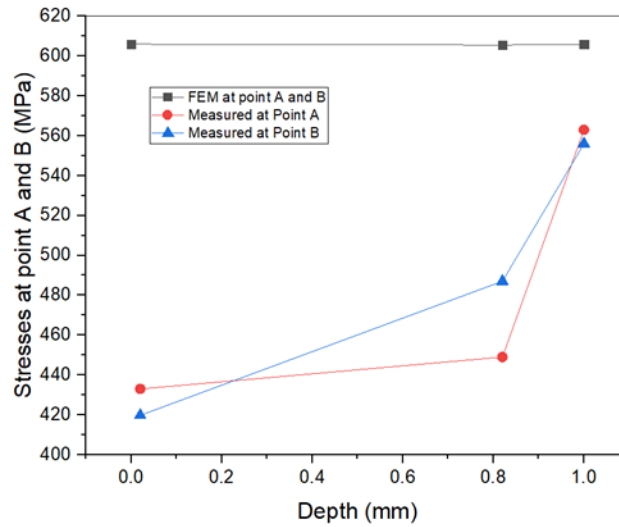


Figure 4-13: Comparisons of experimental and FE simulated residual stresses at each depth.

Figure 4-11-a illustrates the contours of FEM simulated von Mises residual stress profile using heat as shown above and Figure 4-11-b demonstrates the von Mises RS of FEM over depth for both the point A and B. The highest von misses residual stress was found 606 MPa at the transition between the processed zone up to the depth of 1.0 mm. The FE results show the identical von Mises stresses along the treated zone. It is because of the element size and the assumed constant heat input during processing. Therefore, the heat source

model does not capture the asymmetry between AS and RS, discussed earlier. The higher residual stresses were generated owing to both mechanical and thermal loads in the SZ and HAZ. However, the measured von Mises stress at 1 mm depth for A and B was about 563 MPa and 514 MPa consecutively (see Figure 5-2). The comparison of experimental and FE simulated residual stresses at each depth are shown in Figure 4-13. The difference between measured and FEM von Mises stresses is about 8%. These differences result from the limitation of the applied test method because, in the hole drilling method, the relaxation of the lattice is measured by removing the material from the surface to estimate the residual stress. Still, in a heat source FE simulation, the elimination of the exact volume of elements from the FE model and approximation of mesh relaxation is a cumbersome task because FE model ignore the real tool stir action.

4.3 Temperature history

Figure 4-14 depicts the FE approximated temperature distribution contours across the processing line of the plate, at plunging (Figure 4-14-a), in the middle of the processing Figure (4-14-b), and at the end of the FSP Figure (4-14-c). The maximum temperature obtained during plunging is about 1351 °C and then moving towards the traverse direction, the temperature is decreasing slowly up to 946 °C. It becomes consistent until the end of processing. The experimental temperature values reveal a close agreement with estimated temperature values. The variation of temperature from the beginning to the end of processing is demonstrated in Figure 4-15. The higher temperature buildup at the beginning is due to the plunging –dwelling phase, where the tool maintains in the same position for a period allowing the temperature to build up. Figure 4-15 demonstrates the cooling rate after

FSP and releasing the clamp. During cooling, secondary phases may be found in the form of the sigma phase (σ), which is the brittle and tough phase. This intermetallic compound can deteriorate the material mechanical properties like toughness and affect the properties of super duplex stainless steel.

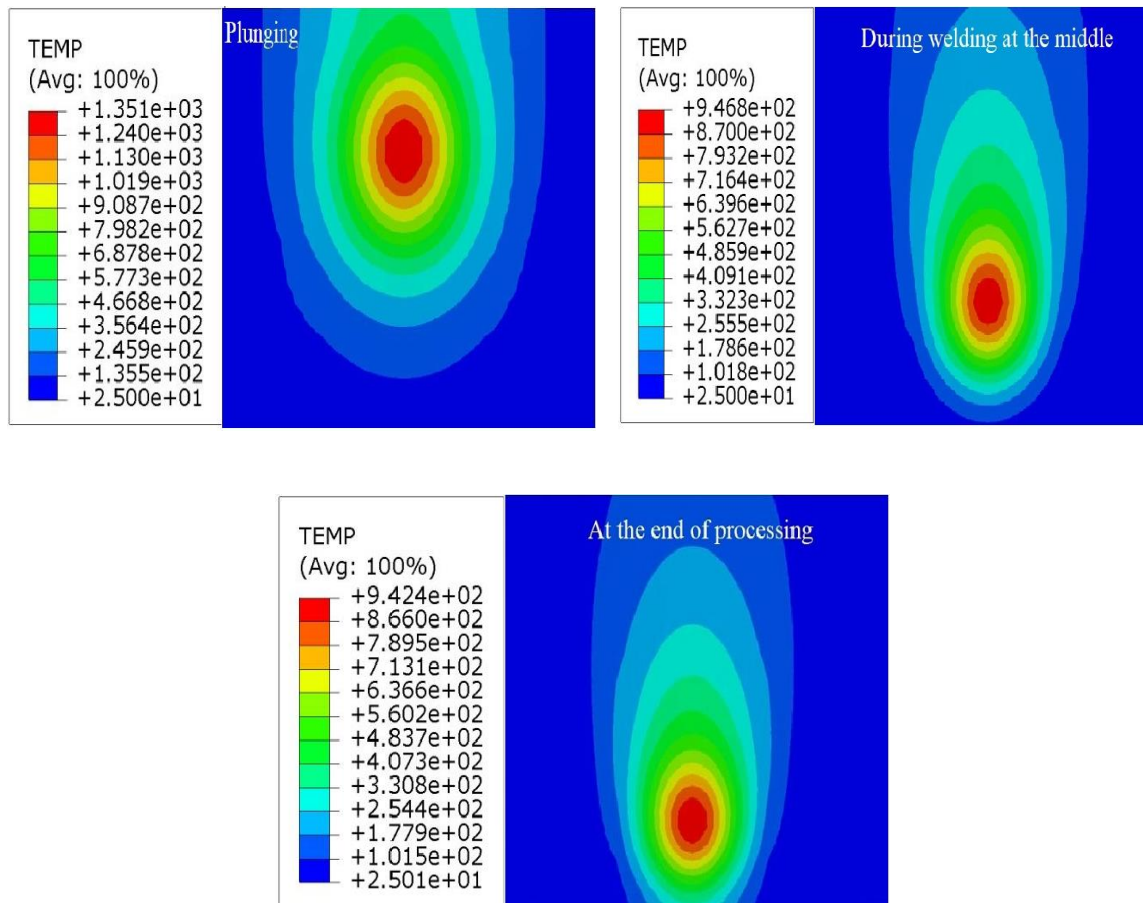


Figure 4-14: Temperature history of the plate during (a) Plunging (t = 50s), (b) at the middle of processing (t = 89.7s), and (c) at the end of processing (t = 129.3s).

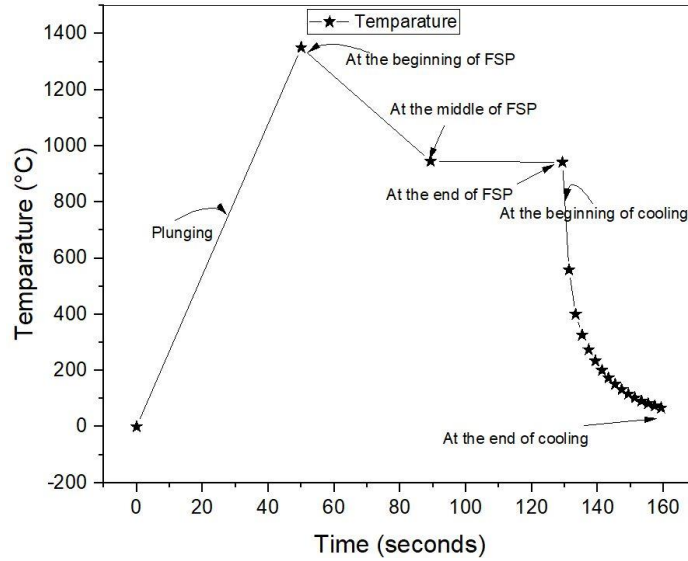


Figure 4- 15: The variation of temperature and cooling rate over time.

In DSS, the sigma phases are typically formed at the temperature range of (700–900) °C [142]. However, Li et al. [143] revealed that the sigma phase would not be formed at the temperature range of (700–900) °C for FSPed hyper DSS because it requires more time to develop. Sato et al. [144] also shown that about 52 min will be needed at temperature 827 °C for the formation of sigma phase in FSW of DSS, on the contrary, Elmer et al. [145] observed that 3.0 min would be needed at the temperature of 850 °C for sigma phase formation during FSW of SDSS 2205 grade. Cui et al. [146] found that for carbon steels treated at a 100 mm / min welding speed, the cooling rate was about 75 °C/s. Therefore, sigma phases are not formed in this present study due to the rapid cooling after FSP (see Figure 4-15), and similar findings are also obtained by previous authors [41]. X-ray diffraction analysis by the authors [41] has revealed that no secondary phases were present in the FSP sample as they did not find any new peaks from the XRD pattern of SAF 2507 when performed the XRD.

4.4 Estimated residual stresses and deformation

Figure 5-16 shows the contours of FEM estimated residual stresses in the FSPed SAF 2507 plate in the XZ plane. The average and maximum von mises stresses are approximately 602 MPa and 676 MPa, respectively, and the maximum stress is found to occur in the transition zones between processed and unprocessed material. The details distribution of RS over the distance for longitudinal and transverse direction are shown in Figure 5-17 at the XZ plane of the plate. The maximum longitudinal and transverse stresses are 647 MPa and 607 MPa, respectively. The stress distribution is seen to be symmetric about the FSP centerline (see Figure 5-17-b).

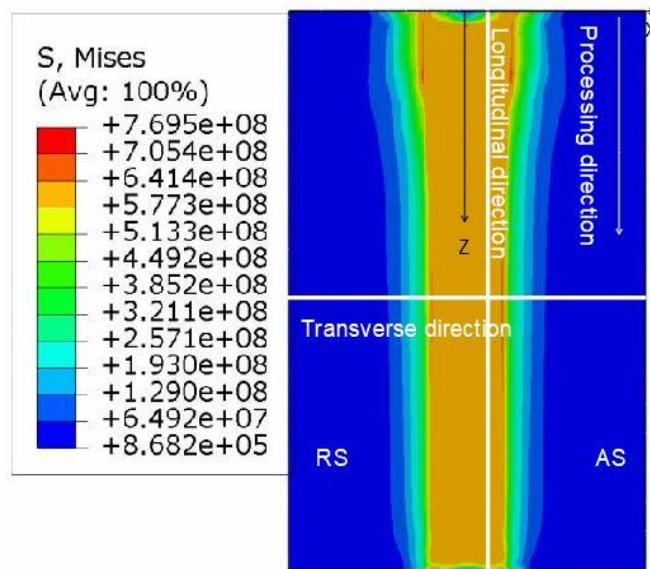


Figure 4-16: Contour of FEM estimated von Mises residual stresses.

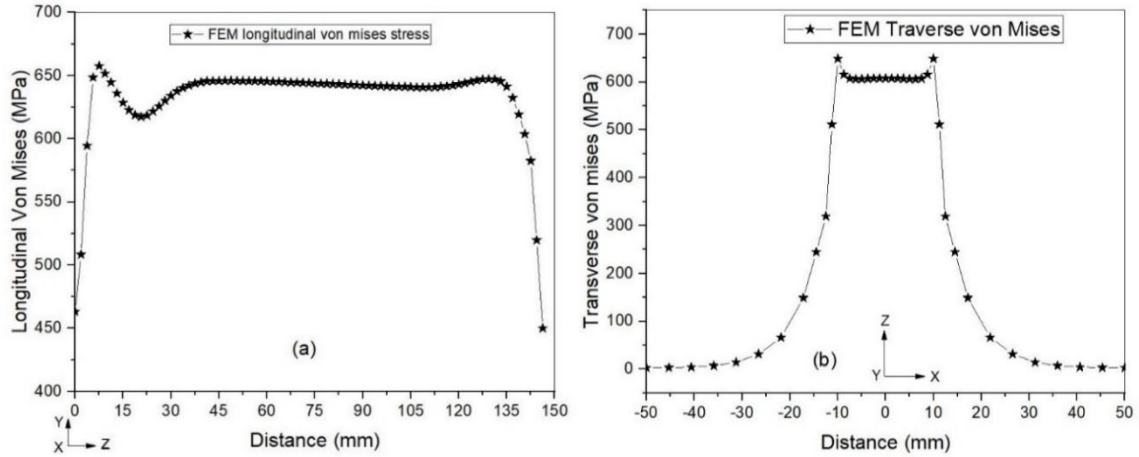


Figure 4-17: Contour of FEM von Mises residual stress distribution (a) longitudinal direction parallel to the processed line, and (b) transverse direction perpendicular to the processed line.

The Figure 4-18 shows the contour of compressive principal residual stresses develop in the plate. It is imperative to emphasize the fact that the residual stresses developed in friction stir processing are compressive. From the figure it can be observed that the compressive residual stresses are present on the depth of the processed zone. However, on the top of the plate the residual stress is positive because of the constant heat input, intensity used and shape of heat has a major effect. A high tensile stress was observed due to the balance between thermal expansion and rapid solidifications and becomes compressive when proceed along the processing line of the plate.

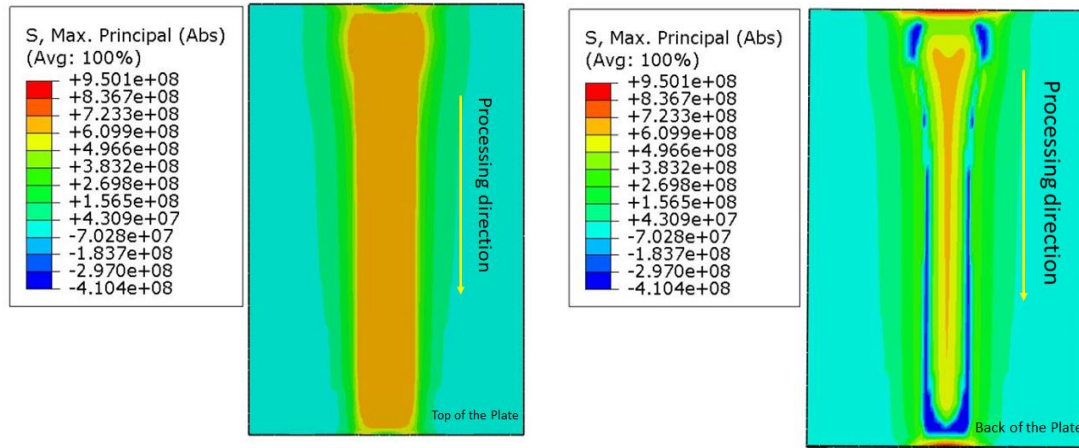


Figure 4-18: Principal residual stresses (a) at the top of the plate, and (b) at the back of the plate.

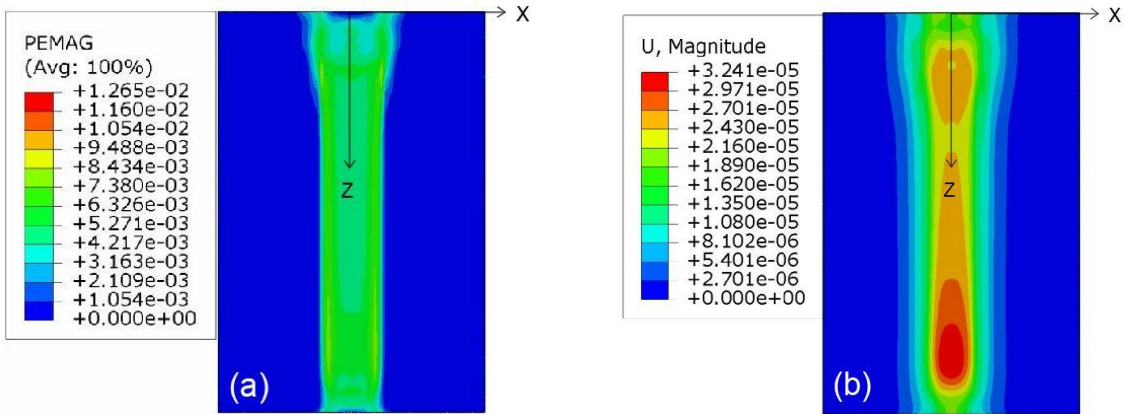


Figure 4-19: (a) Contour of plastic deformation, and (b) Contour of distortion.

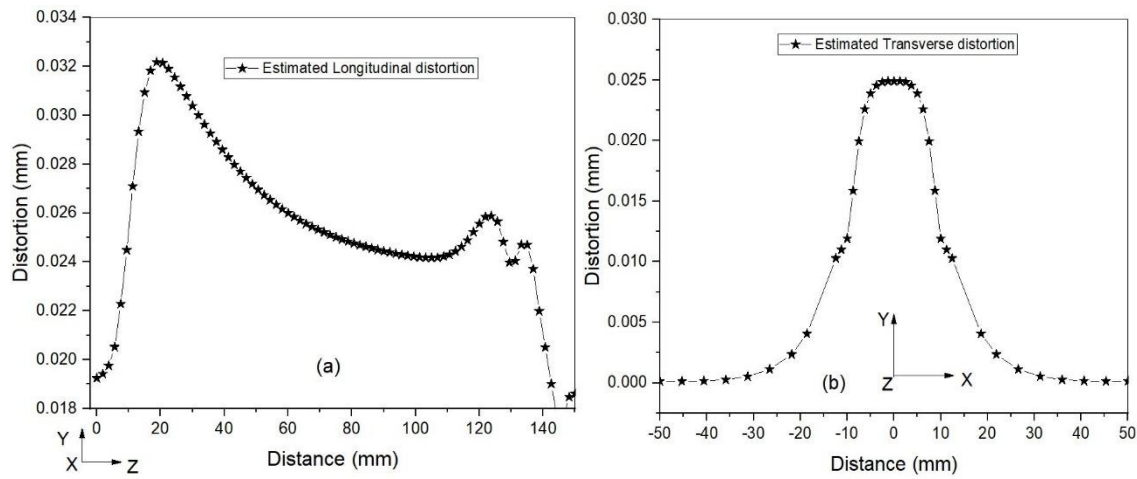


Figure 4-20: (a) Longitudinal distortion, and (b) Transverse distortion.

Figure 4-19-a illustrates the plastic strain distribution in the processed bead after releasing the clamp and cooling down the plate. The highest plastic strain in the processed bead area is about 7.3×10^{-3} mm/mm. In the FSP, the plate is processed with highly localized heating that causes a complicated state of stress because of the contraction and expansion of the heated material. The critical parameter for assessing the degree of plate distortion is the plastic strain. The plastic strain distribution of the processed bead in both the longitudinal and transverse direction at all point are positive and non-uniform, which indicates that tensile stress caused by thermal stress cannot be entirely overcome by cooling shrinkage.

Moreover, Figure 4-19(b) provides the distortions of the FSP plate under 100 mm/min processing speeds. It can be observed that the processing plate depicts a saddle shape distortion (see Figure 4-20), and the similar shape distortion was observed by Guo et al. [147] when investigating the effect of thermal input on welding distortion. The maximum distortions in length are 0.034 mm, and in width is 0.026 mm. Moreover, the distortion produced by the FSP is much smaller than conventional FSW at the same welding speed due to the heat source uniform heating. Both longitudinal and transverse distortion curves in the middle of the plate are shown in Figure 4-20.

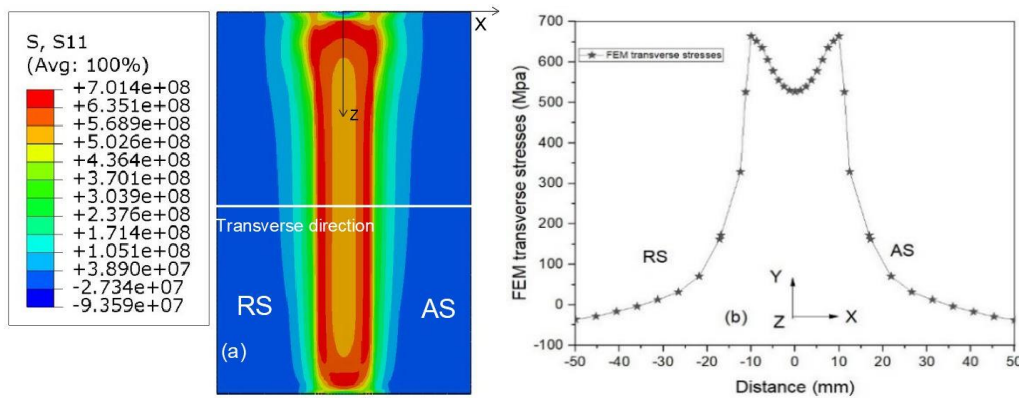


Figure 4-21: (a) Contour of residual stress distribution in the X-direction, and (b) FEM transverse stress distribution in X-direction perpendicular to the processed line.

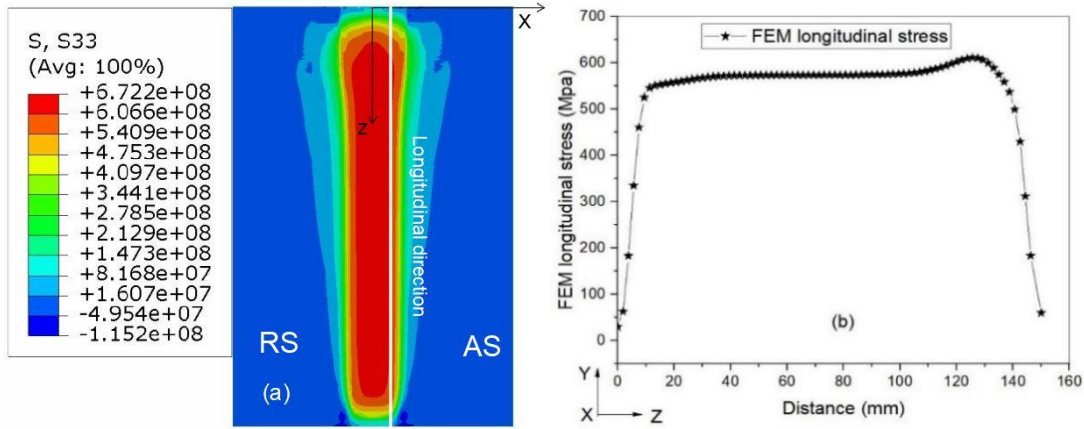


Figure 4-22: (a) Contour of residual stress distribution in the Z-direction and (b) FEM Longitudinal stress distribution in the Z-direction parallel to the processed line.

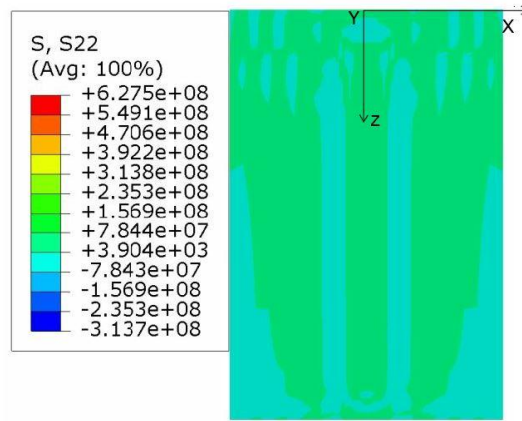


Figure 4-23: Contour of residual stress distribution in the Y-direction (normal stress).

Figures 4-21 to 4-23 shows the maps of residual stresses for X-direction (transverse), Z-direction (longitudinal), and Y-direction (normal). Because of constant heat input throughout the processing, all stresses are symmetric about the processed centerline. The maximum residual stresses estimated in X and Z direction are 664 MPa and 611 MPa, respectively. These maximum stresses are near to the shoulder edge of the FSP plate. The longitudinal stress is higher than transverse stress, and there is no noticeable effect of normal stress, which is in the Y direction.

CHAPTER 5

RESULTS AND DISCUSSION

5.1 Experimental Residual stresses in the Stir Zone

The metallurgical transformations and thermal cycles during FSP induced stresses in the processed and heat affected area due to the misfits between the various phases of the processing and inhomogeneous thermal strain. The residual stresses were measured 2.2 mm away from the bead edge of both advancing sides (Point A) and retreating side (Point B) in the processed zone, as shown in Figure 5-1. The measured von Mises residual stresses at points A and B are then plotted as a function of drilled hole depth in Figure 5-2.

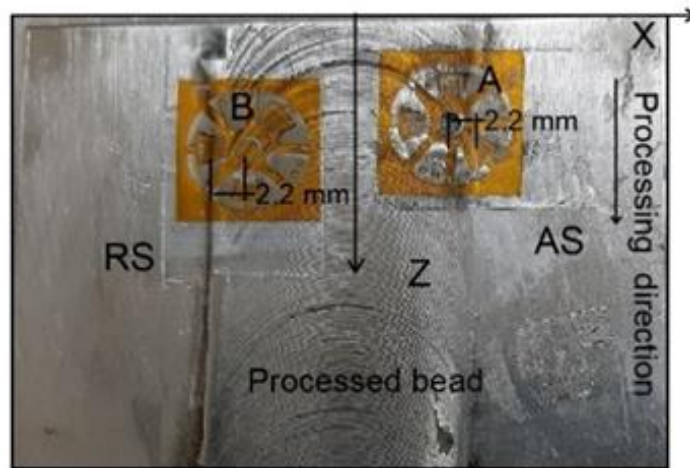


Figure 5-1: RS measurement points A and B.

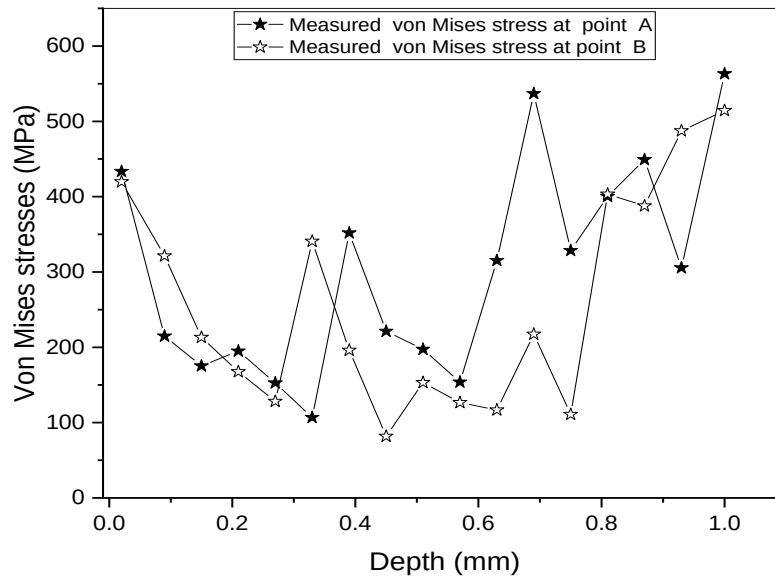


Figure 5-2: Measurement RS at points A and B concerning depth.

Figure 5-2 describes the variation of von Mises residual stresses over 1.0 mm depth for the two points (A and B). From the figure, it can be seen that von Mises stress on the bead near the surface is around 430 MPa at a depth of 0.064 mm, before starting to decline to an average value of 140 MPa. It can be seen that ignoring the variations, which may be due to measurement errors, the average value of von Mises residual stress seems is about 250 MPa within down to a depth of 0.7 mm. A significant rise in residual stresses is observed in the 0.7 to 1.0 mm in depth. The average value within this range is about 500 MPa. The measured maximum von Mises stress at point A and B was 563 MPa and 514.3 respectively at a thickness of 1 mm from the top surface because of the rapid cooling of inner surface compared to the outer surface [148]. While heating, the expansion of steel is higher due to the higher coefficient of thermal expansion and the material undergoes compressive stress during heating. As cooling is the last phase of the thermal process, the precipitation-

hardening mechanism of the material takes place from high temperatures to room temperature, which moves in the solubility field with sudden/fast cooling [149,150]. An asymmetric distribution can be observed; indeed, the advancing side of the FSP processed bead has the maximum stress value than the retreating side (RS). This is expected owing to the higher temperature at the advancing side as compared to the retreating side [151], which is also supported by the thermography digital photo. The FSP bead and the thermally modified area cool after the processing, but their contraction to the original positions prevented via rigid grips and the cooler and natural material surrounding them. These elements generate residual stresses and the asymmetric residual stress distribution in the workpiece is caused by the asymmetric temperature distribution as well as by the material flow in the stirring zone. The material flows from the retreating side, across the centerline of the FSP bead to the forward side. The tool tilted two degrees from a vertical position to boost the forging action. However, the forward side of the processed area is usually the hotter side because the material flows from the receding side, across the centerline of the FSP bead to the forward side.

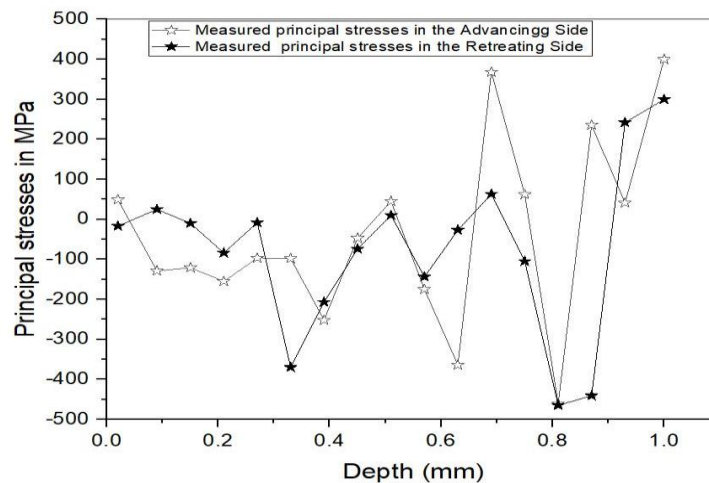


Figure 5-3: Measured principal stresses in the advancing and retreating sides processed area.

Figure 5-3 shows the favorable residual stress distribution that is present in the processed area of the plate, both in the advancing and retreating sides. The measurement also demonstrates that friction stir processing has resulted in mainly compressive principal stresses with a rough average close to -50 MPa. While it is expected that redistribution and relaxation of these residual stresses would occur, in the FSPed zone, during specimen preparation, a fair amount of the residual stresses can still be present [65].

5.2 Effect of FSP on Fatigue Crack Growth Behavior

Figure 5-4 illustrates the variation of fatigue crack length (a) with the number of cycles (N) for the base material (BM) and the friction stir processed (FSPed) material of Super Duplex stainless steel SAF2507 at different stress levels. It can be seen from the a-N curves that, irrespective of stress level, the initial slope of a-N curve starts low and increases with increasing number of cycles. The crack initiation life is lower at higher stress levels for both base and FSPed materials. The effect of FSP is translated by higher initiation and total fatigue life for the same stress level. For example, the total fatigue life for BM at the applied stress range of 94 MPa is 56,093 cycles with an initiation life of 10,002 cycles, while those of FSPed material are respectively, 68,546 and 26,100 cycles. This is an indication that friction stir processing improves both crack initiation and FCG resistance. The main reasons for this fatigue life enhancement could be due to the higher material strength and fracture toughness [41] as well as the presence favorable residual stress distribution as shown in Figure 5-3. These residual stress measurements show that friction stir processing has resulted in mainly compressive principal stresses with a rough average close to -50 MPa. While it is expected that redistribution and relaxation of these residual stresses would

occur, in the FSPed zone, during specimen preparation, a good amount of the residual stresses can still be present [65].

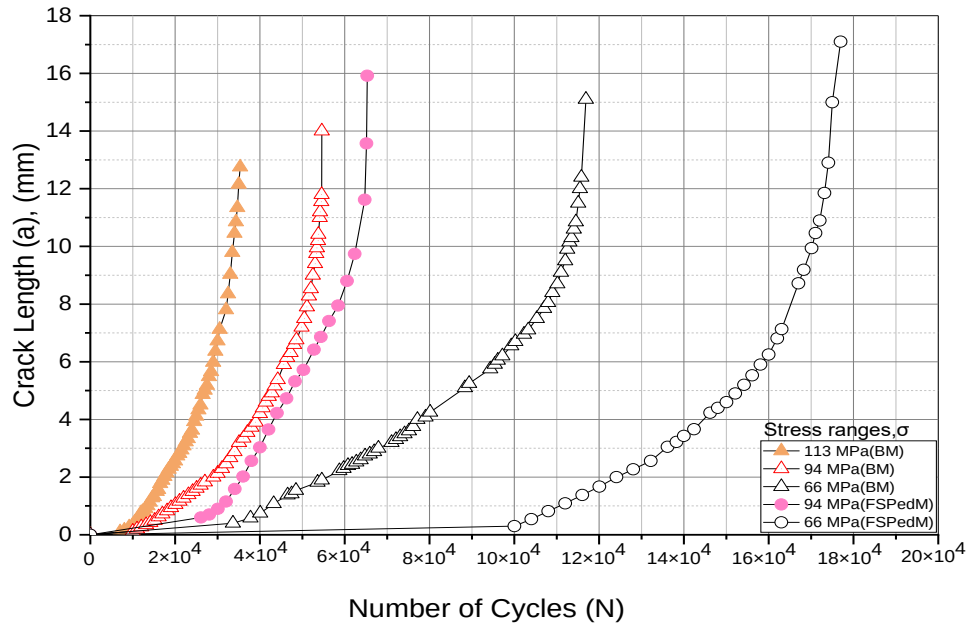


Figure 5-4: Fatigue crack growth curves for base metal and FSPed SAF2507.

Figure 5-5 shows low magnification micrographs for the fracture surfaces of BM and FSPed samples, tested at a stress range of 66.0 MPa. The total fatigue lives of BM and FSPed specimens are 120,558 cycles and 178,482 cycles, respectively. This is corresponding to 32% of life improvement due to material friction stir processing. It can be seen that for FSPed SDSS, the crack propagated for a distance 13% longer than that for BM.



Figure 5-5: SDSS SAF 2507 fracture surface of CT specimens, (a) BM and (b) FSPed.

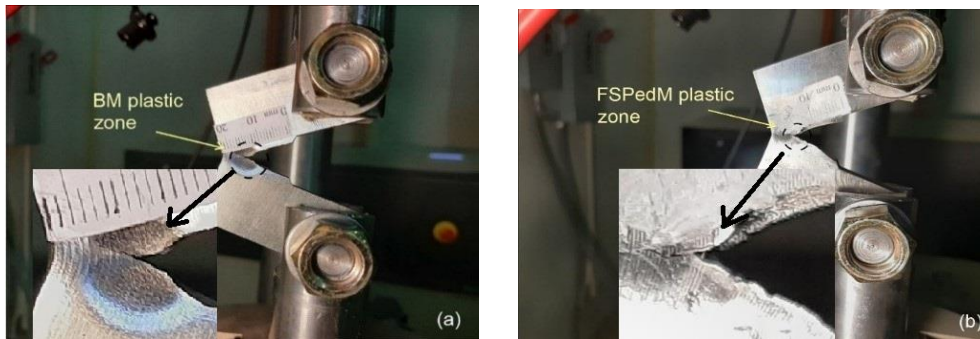


Figure 5-6: C (T) specimen's plastic zone observed during the experiment (a) BM specimen and (b) FSPed specimen.

The images of Figure 5-5 show also that the fracture surface of the FSPed specimen is smoother as compared to that of the BM specimen. Significant thickness reductions, before final rupture, are observed for both specimens, indicating mostly ductile failure. It should be noted that due to its lower yield strength, the base metal sample necking is more pronounced than that of FSPed specimen. Figure 5-6 reveals that because of lower yield strength, the plastic zone size at the crack tip of the base metal is more significant than that of the friction stir processed metal.

The material's performance is influenced greatly by microstructural characteristics such as grain texture [153], grain size [154], phase distribution [155], particularly interface character distributions [156]. There are two interface groups in DSSs with component ferritic (α) and austenite(γ) phases shown in Figure 3-1.

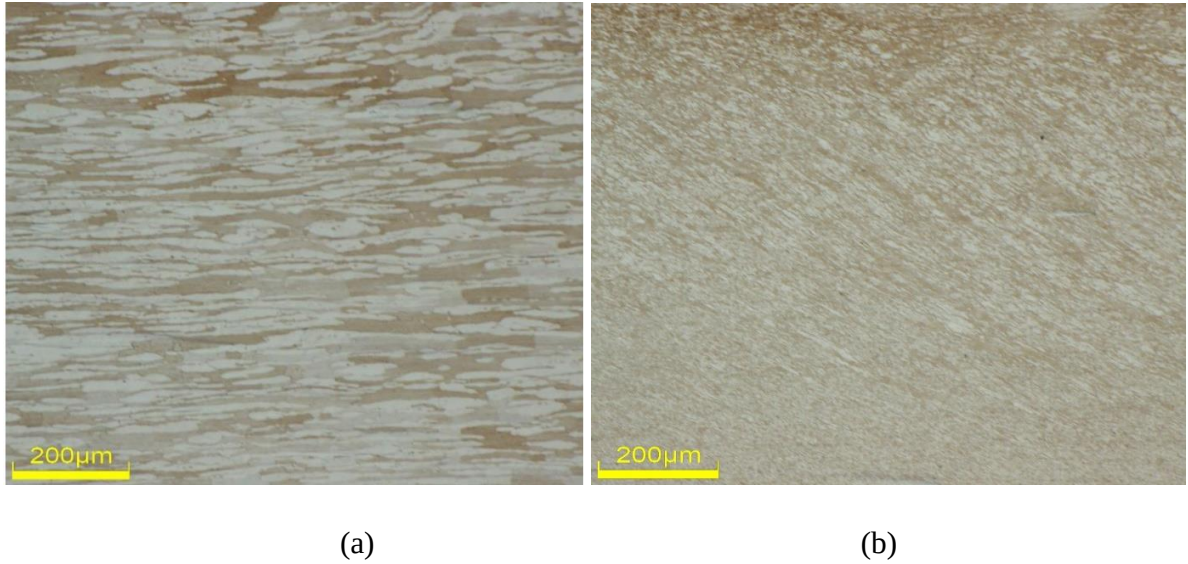


Figure 5-7: Microstructure of base metal and processed material in the stir zone [105].

Figures 5-7-a and 5-7-b, respectively show the distribution of ferrite and austenite grains in the parent material and the stir zone along the transverse direction (perpendicular to the crack tip). Figure 5-7-a reveals that both ferrite and austenite grains were refined by friction stir processing. The grain sizes were measured using the line intercept procedure resulting in average sizes of about $160\text{ }\mu\text{m}$ for BM while the grain size in the stir zone was found to vary between 2 and $30\text{ }\mu\text{m}$ [41]. It can also be seen that besides the grain refinement, the orientation of the austenite grains was changed by friction stir processing. The change in orientation of the grains and the increase in the number of grain boundaries are expected to affect the crack growth behavior. However, owing to the existence of phase boundaries when exposed to thermal mechanical treatment, DSSs naturally have a refined microstructure and thus show stronger mechanical properties and corrosion resistance compared to single-phase alloys [157] when two constituent phases are well-distributed. It is understood that material with larger grain sizes has a limited number of grain boundaries

and is weaker than material with smaller grain sizes and a substantial number of grain boundaries. Since the dislocation motion needs to proceed through a smaller number of grains boundaries in greater grain content [158,159]. Various models, such as the Hall-Petch relationship, are available in the literature, which describe that grain size is inversely related to the hardness and strength of the material, and this increase in strength and hardness is compensated for by a decrease in ductility of the material[160,161]. The grain boundary is defined by the angle of misorientation to the grain boundaries. Grain boundaries are usually classified as low-angle grain boundaries (LAGBs) and high-angle grain boundaries (HAGBs). The microstructure of FSPed has large fractions of HAGBs that are beneficial during deformation for strain hardening, grain boundary sliding, and blocking slipping dislocations [162]. Therefore, large fractions of HAGBs lead to high tensile strength and elongation in the SZ microstructure.

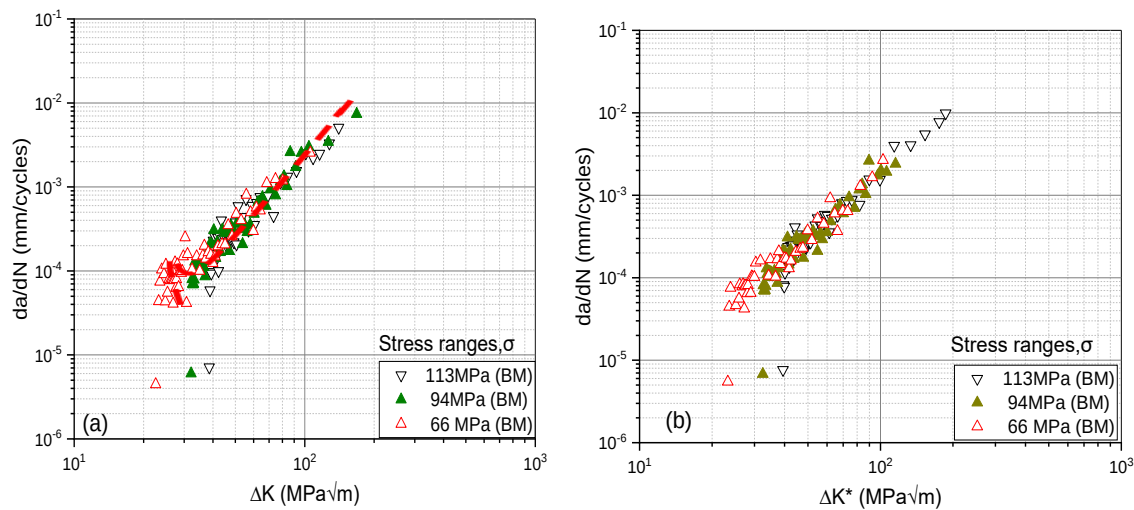


Figure 5-8: FCG rate variation with SIF for as-received SDSS SAF-2507 (a) without plastic zone correction and (b) with plastic zone correction.

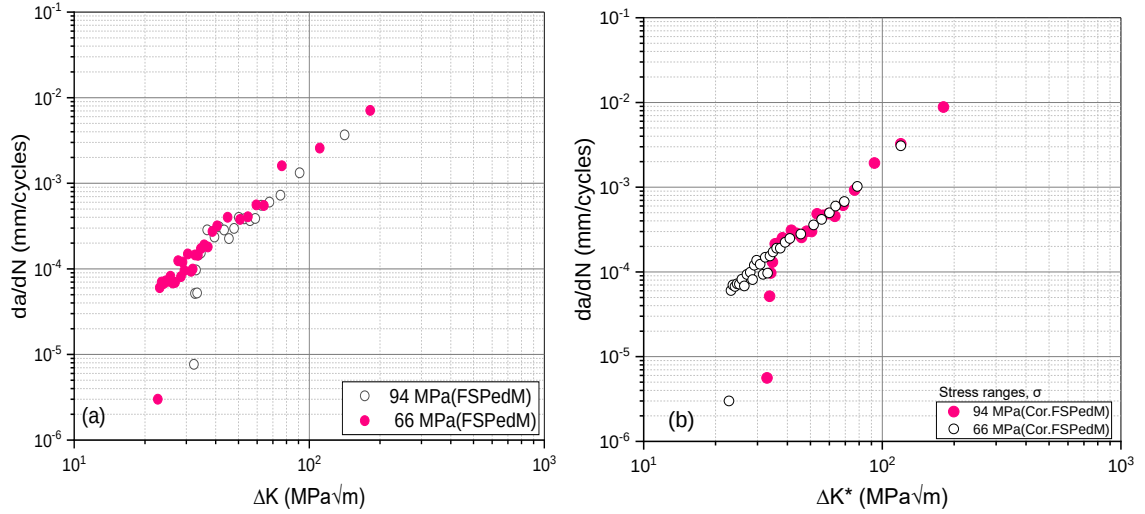


Figure 5-9: FCG rate variation with SIF for FSPed SDSS SAF-2507 (a) without plastic zone correction and (b) with plastic zone correction.

The fatigue crack propagation rate (da/dN) was computed from the a-N data, for all three stress levels, and represented in terms of SIF range ΔK (Eq. 3.1 & 3.5) in Figures 5-8 and 5-9 for BM and FSPed materials, respectively. The da/dN - ΔK curves obtained for SAF 2507 are in good agreement with those reported by Di Cocco for super duplex 25 Cr 7 Ni stainless steel at the stress ratios between 0.1 and 0.75 in the air [163]. The correlation between da/dN and ΔK for base metal shows a complex behavior at low propagation rates, as shown by the red line in Figure 5-8-a. It is usually an indication of plasticity-induced crack closure under plane stress conditions. The presence of crack closure phenomenon, which may occur even at a positive stress ratio, is usually due to the occurrence of crack tip plasticity during unloading [164,165]. Figure 5-8-b reveals that a better correlation of FCG rate with SIF corrected for the plastic zone, ΔK^* , was obtained. The correction for plasticity was performed by adding the plastic zone size r_p , $r_p = \frac{1}{\pi} \times \left(\frac{\Delta K}{2\sigma_y} \right)^2$ for plane stress [59], to the measured crack length. Even though the plastic zone size for FSPed SAF

2507 was smaller than that of BM, this correction was applied to the da/dN - ΔK data of Figure 5-9-a, resulting in the correlation shown in Figure 5-9-b. The typical effect of friction stir processing on the fatigue crack growth behavior of SAF 2507 is illustrated in Figure 5-10 with the best-fit lines for the stress range of 94 MPa.

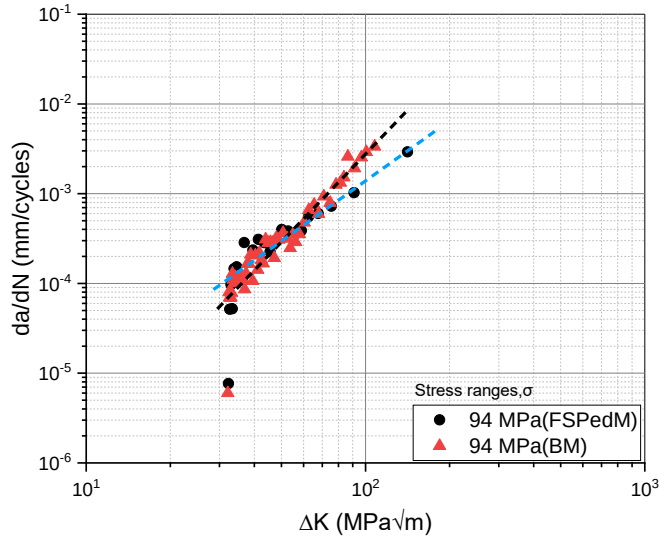


Figure 5-10: Comparison of FCG-SIF for BM and FSPed for the stress range of 94 MPa.

The estimated values of slope (m) and the intercept (C) were respectively, 3.49 and 4.23×10^{-10} for the base material and 3.09 and 19.9×10^{-10} , respectively, for the FSPed material. The average values of (m) and (C) obtained from the data of Figures 5-8 and 5-9 are listed in Table 5-1. These values are within the range of 3 to 5 for m and 10^{-8} to 10^{-11} mm/cycle for C [59]. The values of the intercept, C , and exponent, m , for the present base metal are close to those reported by Idris et al. [166] for dual-phase steel ($C = 4.33 \times 10^{-10}$ mm/cycle and $m = 3.82$). It can be seen, from Table 5-1, that the exponent (m) for processed material is smaller than that for base material, while the intercept (C) is higher. This is an indication of higher crack growth resistance in friction stir processed material. As

mentioned above, grain refinement and multiplication of grain boundaries (see Figure 5-7) along with FSP induced residual stresses (see Figure 5-3) are the main reasons behind the improvement in FCG resistance. Similar observations were made by Lakshminarayanan and Balasubramanian for friction stir welded AISI 409M ferritic stainless steel [65].

Table 5-1: Plasticity correction effects on FCG constants C and m.

Present Material (SAF 2507)	C (mm/cycle)	m
BM before plasticity correction	6.02×10^{-10}	3.56
BM after plasticity correction	2.71×10^{-10}	3.73
FSPedM before plasticity correction	2.08×10^{-09}	3.05
FSPedM after plasticity correction	1.36×10^{-09}	3.17

5.3 Fractography

Figure 5-11 depicts the fracture regions (initiation, propagation, and final fracture) of BM and FSPed material after fatigue crack propagation testing. The fracture surfaces of failed CT samples were examined by an Opto-digital microscope (OLYMPUS DSX510) without any surface treatment to investigate the physical micromechanisms of the crack growth. It can be observed that each region has a distinct microstructure for both BM and FSPed specimen. The FSPed specimen is showing multiple possible initiation sites and a smoother stable crack propagation region whereas BM specimen showing a single crack source. It indicates that FSP has improved the material fatigue crack growth rates. More details of SDSS 2507 material crack propagation behavior has been discussed through SEM analysis for both base material and FSPed materials which has shown in Figure 5-12.

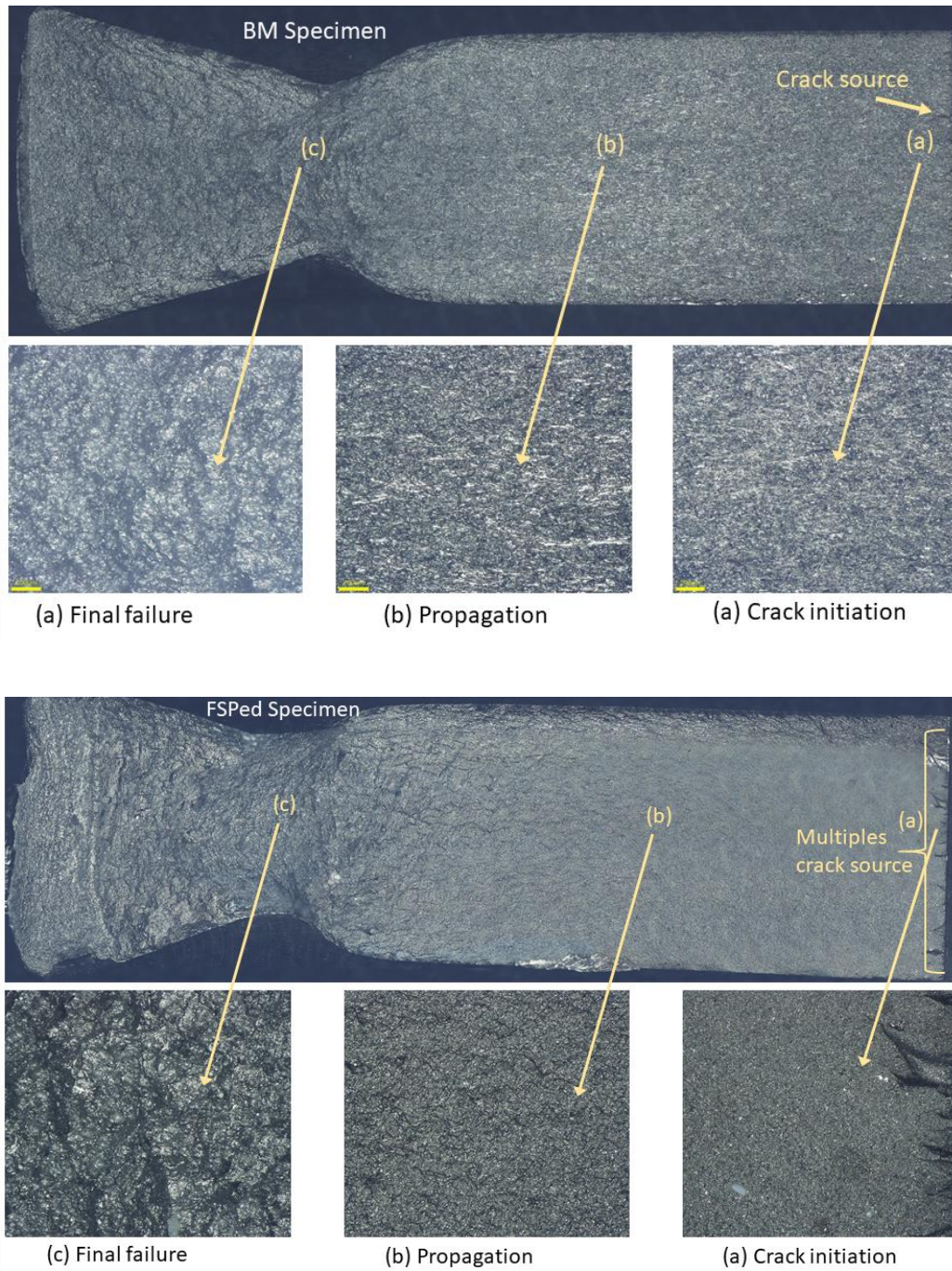


Figure 5-11: Optical microscopic observation of CT specimens for both the BM and FSPedM.

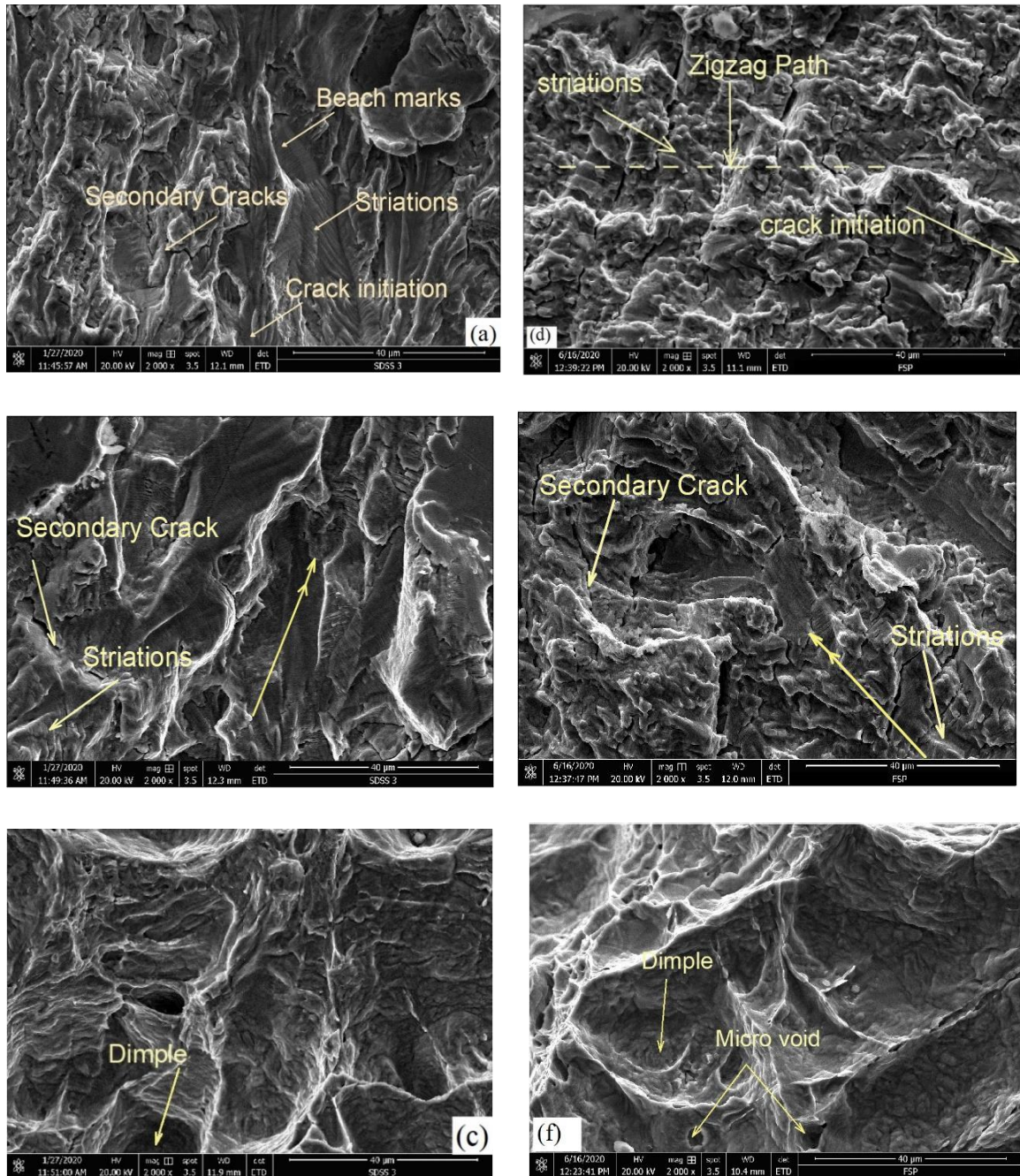


Figure 5-12: SEM fractographs for both BM and FSPed SAF 2507, (a) initiation (BM), (b) propagation (BM), (c) final failure (BM), (d) initiation (FSPed), (e) propagation (FSPed), and (f) final failure (FSPed). (The double arrow indicates the main crack direction.)

Figure 5-12 reveals the details at three different regions of the fracture surfaces. The first three SEM fractographs (a, b and c) correspond, respectively, to the initiation, stable crack

growth, and final fracture of the base material, whereas fractographs d, e, and f correspond to similar regions in FSPed material. Clear fatigue striations and few secondary cracks can be seen on the fractograph of Figure 5-12-a for BM, while multiple crack initiation sites and a larger number of secondary cracks are present on the FSPed material fractograph (Figure 5-12-d). Fewer ductile striations are visible near the initiation site of the FSPed specimen. This is an indication of reduced transgranular crack propagations caused, probably by the presence of refined grains, resulting in more intergranular crack growth. It indicates that the crack propagation rate was slowing down, and the path of crack was significantly complicated, resulting in zigzag advance. Local cleavage is observed on the stable crack growth zone for both specimens (Figure 5-12-b and 5-12-e). The small cleavage areas occur mostly in the ferrite matrix [167]. The secondary cracks are due mainly to stress concentrations caused by dislocations pile up at the grain boundaries [167, 168]. The final fracture region of both BM and FSPed specimens is characterized by typical ductile failure topographies evidenced by numerous cone and cup dimples.

5.4 Discussion

In this study, the thermo-mechanical effects on the FSP process are addressed. In particular, the distributions of the highly relevant field variables were assessed for the processed plate. Figure 5.2 shows the variation of measured von Mises residual stresses and while and Figure 4-15 presents the estimated temperature profiles through FEM. An asymmetric distribution can be observed in measured residual stresses; indeed, the advancing side

(A.S.) of the FSP processed bead has the maximum stress value than the retreating side (R.S.) in Figure 5-2. At the same time, the temperature history is less evident in Figure 4-15 and 4-16). The current model considers both mechanical and thermal effects and offers a detailed representation of the FSP process for numerical analysis, as already described in the previous section, that this model adopts the material elastic-plastic behavior and related mechanical behavior at varying strains and temperatures. The final stress of the plate was raised after the release of the clamping constraints. It was then highlighted as longitudinal stress, i.e., in the direction of the processing, and transverse stress, i.e., in the direction perpendicular to the processing. The map of these residual stress is shown in Figures 4-16 and 4-17. As anticipated based on the asymmetric material flow due to tool action, skewed distributions were observed experimentally (Figure 5-2). Still, the FE model residual stresses are symmetric due to constant heat input during processing. The measured stress profile on the processed bead is not uniform because the stress is directly associated with the temperature. The peak of residual stress occurs in the processing hot zone, which is produced during the plunging stage and is required to expand the heat at the opening of the FSP cycle. The maximum stress was found near to the tool shoulder edge. However, the FEM von Mises stresses for both advancing side, and retreating side is identical because the heat source model did not capture the asymmetry of the tool and applied constant heat input while processing the plate. However, it can be observed that (Figure 4-18) the predicted RS of FEM are in acceptable limit with measured one through drill hole technique by ignoring the real tool stirring action. The developed FE heat source model can also deliver active distributions of the local values of all relevant field variables along the process.

In addition, in FSP, heat is very critical because it affects the cooling rate, residual stresses, and the Stir Zone microstructure, and thus affects the mechanical properties of the material. If the cooling rate is slow then there is a feasibility to form the sigma phases which has the detrimental effect on the mechanical properties of material like toughness. This is because sigma phase is the tough phase and brittle in nature. However, the present study reveals that the cooling rate is very fast for both the experimental FSP and heat source model (Figure 4-11 and 4-15). Thus, there is no chance to form the sigma phase. Because it requires at least 3.0 min at the temperature of 850 °C for sigma phase formation [147] whereas the present work estimated cooling rate is more than 80 °C/s with the average temperature 727 °C. The higher the heat input, the more harmful ferritic phase formation can be formed. The opposite is true for austenitic phase. A successful processing technique would be to minimize heat input in order to prevent unnecessary ferrite phase from occurring. In addition to that the high temperature produces high residual stresses. This study revealed that the advancing side has higher residual stresses (Figure 5-1 and 5-2) than retreating side because the advancing side is hotter than the retreating side. The reason for the hotter advancing side is the material moving from the retreating side to advancing side. The principal compressive residual stresses; about -50 MPa, were found on the stir zone (Figure 5-3) which has helped to retard the crack propagation rate of the SDSS SAF 2507 material. Figure 5-7 shows the grain refinement of processed material compare to the base material. Abubaker et al. [41] reveals that the base material grain size is about 160 micrometers whereas for the processed material, it varies between 2 to 30 micrometers using the line intercept method for the SAF 2507

material for the similar parameters of this current study. Thus, the FSPed material is more refined than base material. From Figure 5-7-b it can also be noticed that the grain orientation has changed for the processed material due to temperature gradient. The change in orientation has produced a large number of grain boundaries [156,163]. Therefore, the grain refinement and multiplication of grain boundaries retarded the crack propagation rate for the processed material. The material fails due to persistent slip band (PSB) stress levels (with respect to dislocation motion). These PSBs are capable of traversing low-angle grain boundaries, thereby belonging to grain clusters. The PSBs develop and connect with high-angle GBs as a consequence of the ongoing cyclic slip process, resulting in dislocation pile-ups, static extrusions at the grain boundaries (GB) in the form of ledges/steps. Therefore, a large number of grain boundaries retard the crack growth. The fractography of both base and processed material fracture specimens reveal that the base material has a single crack source and the processed material has multiple crack sources and the crack propagation path is complex, resulting in zigzag advance. It is observed that the number of secondary cracks on the stable crack growth region for FSPed fracture surface is higher than that of BM fracture surface. Secondary cracks drained energy and decreased the drive for primary crack development locally. Thus, the presence of secondary cracks not only increased energy consumed in fractures but also led to the stabilization of main crack propagation. Thus, multiple crack initiation and complex crack path of FSPedM has improved the crack growth resistance. It can be concluded from the study that the three main reasons for the low fatigue crack propagation of FSPed SDSS SAF 2507 are (a) Grain refinement, (b) Multiplication of grain boundaries and change in orientation of austenite, and (c) Favorable compressive residual stress.

CHAPTER 6

CONCLUSIONS AND FUTURE WORK

6.1 CONCLUSIONS

The current study has been conducted to observe the effect of friction stir processing on fatigue crack propagation behavior of a super duplex stainless (SAF-2507). Fatigue tests were performed at several stress levels with a stress ratio of 0.1 on CT specimens prepared from base metal and FSPed metal. $da/dN-\Delta K$ curves were obtained for both types of samples, and optical microscopy and SEM fracture surface analysis are performed to investigate the different failure mechanisms. In addition to that, friction stir processing of SDSS SAF 2507 was modeled using 3D fully coupled thermo-mechanical finite element techniques where tool-workpiece interaction was idealized as a moving heat and pressure distribution. Based on the obtained results, the following conclusions can be drawn:

1. Friction stir processing of SDSS SAF 2507 enhanced the total fatigue life by more than 30%.
2. Significant thickness reductions, before final rupture, were observed for both base material and FSPed specimens, indicating mostly ductile failure.
3. Because of possible plasticity induced crack closure, $da/dN-\Delta K$ curves displayed complex behavior at low ΔK . Plastic zone corrected ΔK resulted in improved FCG rate correlation.

4. Friction stir processing of SDSS SAF 2507 increased crack growth resistance of the material and resulted in higher Paris constant, C and lower exponent, m .
5. The SEM fracture surface analyses revealed multiple crack initiation sites and secondary cracks for FSPed specimens with a significantly complicated crack path caused mainly by the more significant number of grain boundaries.
6. Finite Element modeling of FSP is very challenging, owing to its multi-physics characteristics. However, the idealized numerical simulation of FSP was able to give a nice prediction of the FSP bead geometry and shape when compared to the experimental bead.
7. The predicted temperature distribution during processing was in good agreement with experimental results obtained by an infrared camera.
8. FE residual stress estimation was in agreement with that measured using drill hole residual stress measurement technique.
9. The FSP heat source modeling of SDSS SAF 2507 can be used to visualize and analyze the temperature, residual stress, and deformation profiles in the processed plate without doing any costly experimental work.

6.2 Future work

This study represents the effects of FSP for a single pass and similar metal only so we can recommend the following for future work.

1. The work can be expanded by involving other parameters of welding (such as tool geometry) and using different DSS steel grade (e.g., LDSS, HDSS, etc.)
2. The weld quality of the SDSS can be studied concerning the plunging rate and dwell time.
3. Instead of a single pass on the workpiece of SDSS, multi-pass processing can be used to see the effect of residual stresses.
4. The fatigue crack propagation can also be determined using various stress ratios, as the present study is with constant stress ratio 0.1.
5. Different processing parameters effects on fatigue crack growth rate can be investigated.

APPENDIX

Stress calculation and data table

Stress Calculation:

$$f(a/w) = \frac{2 + \frac{a}{w}}{(1 - \frac{a}{w})^{3/2}} \left[0.866 + 4.64 \left(\frac{a}{w} \right) - 13.32 \left(\frac{a}{w} \right)^2 + 14.72 \left(\frac{a}{w} \right)^3 - 5.60 \left(\frac{a}{w} \right)^4 \right]$$

$$f(0.5) = \frac{2 + 0.5}{(1 - 0.5)^{3/2}} [0.866 + 4.64(0.5) - 13.32(0.5)^2 + 14.72(0.5)^3 - 5.60(0.5)^4] = 9.66$$

$$\Delta K_j = \frac{5000 * 9.66}{0.0065 \sqrt{0.05}}$$

$$\Delta K_j = 33.23$$

$$\Delta K_j = 33.23 \text{ Mpa}\sqrt{m}$$

$$\Delta K_j = \frac{6000 * 9.66}{0.0065 \sqrt{0.05}}$$

$$\Delta K_j = 39.88$$

$$[\Delta K_j = \Delta S \sqrt{\pi a}]$$

$$39.88 = \Delta S \sqrt{\pi * 0.040}$$

$$39.88 = \Delta S * .354$$

$$\Delta S = 113 \text{ Mpa}$$

$$[\Delta K_j = \Delta S \sqrt{\pi a}]$$

$$33.23 = \Delta S \sqrt{\pi * 0.040}$$

$$33.23 = \Delta S * .354$$

$$\Delta S = 94 \text{ Mpa}]$$

$$\Delta K_j = 39.88 \text{ Mpa}\sqrt{m}$$

$$\Delta K_j = \frac{3500 * 9.66}{0.0065 \sqrt{0.05}}$$

$$\Delta K_j = 23.26 \text{ Mpa}\sqrt{m}$$

$$[\Delta K_j = \Delta S \sqrt{\pi a}]$$

$$23.26 = \Delta S \sqrt{\pi * 0.040}$$

$$23.26 = \Delta S * .354$$

$$\Delta S = 66 \text{ Mpa}]$$

Table 0-1: Data of crack length (a) versus Number of Cycles (N) for both BM and FSPedM of SDSS.

BM(113 MPa)		BM(94 MPa)		BM(66 MPa)		FSP(94 MPa)		FSP(66 MPa)	
Cycles	crack length	Cycles	Crack length	Cycles	Crack length	Cycles	Crack length	Cycles	Crack length
0	18	0	18	0	18	0	18	0	18
7015	18.1	10002	18.12	33653	18.4	26100	18.6	100035	18.3
8032	18.16	11003	18.2	37793	18.58	28041	18.7	104017	18.54
9163	18.27	12003	18.27	40082	18.75	30104	18.9	108012	18.82
10019	18.35	13165	18.35	43263	19.08	32020	19.15	112020	19.09
10503	18.44	14165	18.42	46550	19.38	34074	19.59	116026	19.38
11065	18.53	15181	18.5	47476	19.42	36033	20.02	120067	19.67
11500	18.63	16114	18.622	48493	19.54	38018	20.56	124092	20
12028	18.77	17021	18.72	53601	19.82	40012	21.03	128044	20.27
12500	18.83	18112	18.83	54609	19.9	42015	21.65	132218	20.56
13000	18.9	19046	18.94	58744	20.22	44018	22.22	136156	21.05
13511	19	20080	19.05	59752	20.3	46280	22.73	138249	21.22
14000	19.13	21029	19.16	60760	20.4	48266	23.32	140000	21.43
14600	19.28	22003	19.27	61745	20.44	50269	23.72	142409	21.66
15000	19.324	23127	19.39	62753	20.52	52737	24.42	146226	22.23
15680	19.501	24133	19.5	63777	20.6	54396	24.86	148040	22.4
16000	19.57	25049	19.61	64809	20.73	56312	25.41	150057	22.6
16500	19.773	26039	19.72	65927	20.8	58440	25.95	152124	22.9
17000	19.887	27031	19.83	66934	20.88	60560	26.8	154206	23.2
17500	20.02	29010	20	67941	21	62348	27.74	156120	23.53
18000	20.14	30017	20.13	71070	21.2	64720	29.62	158062	23.9
18500	20.27	31039	20.3	72071	21.3	65133	31.57	160001	24.25
19000	20.37	32002	20.46	73079	21.4	65330	33.92	162049	24.81
19500	20.46	33002	20.67	74137	21.5			163056	25.13
20000	20.554	34018	20.88	75144	21.6			167032	26.72
20500	20.73	35067	21.2	76152	21.75			168275	27.19
21000	20.82	36123	21.35	77150	22			170123	27.94
21500	20.94	37038	21.55	79081	22.08			171054	28.46
22000	21.08	38021	21.73	80157	22.25			172037	28.9
22500	21.22	39038	21.9	88351	23.1			173163	29.85
23000	21.33	40006	22.2	89361	23.25			174096	30.9
23500	21.516	40712	22.4	94362	23.75			175000	33
24000	21.624	41401	22.6	95412	23.91			176961	35.1
24500	21.92	42107	22.8	96117	24.05				
25000	22.11	42804	22.92	97194	24.2				
25500	22.33	43504	23.16	99297	24.55				
26000	22.5	44219	23.38	100265	24.7				
26500	22.86	45642	23.9	102283	24.95				
27000	23.012	46433	24.15	103294	25.1				

27500	23.181	47249	24.32	105340	25.5				
28000	23.48	48047	24.6	107030	25.85				
28500	23.657	48601	24.76	108000	26.05				
29000	23.98	49845	25.2	109001	26.4				
29500	24.36	50474	25.5	110163	26.7				
30000	24.72	51082	25.9	111003	27.1				
30509	25.12	51587	26.28	112003	27.5				
32013	25.8	52008	26.53	112500	27.9				
32500	26.352	52512	27	113002	28.15				
33011	27.03	53018	27.4	113506	28.3				
33500	27.79	53320	27.75	114036	28.6				
34000	28.452	53470	27.95	114514	28.85				
34330	28.85	53715	28.2	115101	29.5				
34660	29.343	53800	28.42	115505	30				
35009	30.15	54138	29	115841	30.4				
35330	30.75	54206	29.2	116900	33.1				
		54474	29.55						
		54573	29.8						
		54600	32						

Table 0-2: Data of da/dN and ΔK for BM and FSPedM of SDSS before plasticity correction.

BM (113 MPa)		BM(94MPa)		BM(66MPa)		FSP(94 MPa)		FSP(66 MPa)	
ΔK	da/dN	ΔK	da/dN	ΔK	da/dN	ΔK	da/dN	ΔK	da/dN
38.56	7.13E-06	32.17	6.00E-06	22.57	1.19E-06	32.30	7.66E-06	22.69	3.00E-06
38.84	5.90E-05	32.44	7.99E-05	23.25	4.35E-05	32.65	5.15E-05	23.13	6.03E-05
39.08	9.73E-05	32.61	7.00E-05	23.54	7.43E-05	33.00	9.69E-05	23.57	7.01E-05
39.34	9.35E-05	32.79	6.88E-05	23.97	1.04E-04	33.35	5.22E-05	24.04	6.74E-05
39.58	1.86E-04	32.96	7.00E-05	24.54	9.13E-05	33.84	1.46E-04	24.55	7.24E-05
39.84	1.60E-04	33.14	7.87E-05	24.85	4.32E-05	34.59	1.53E-04	25.08	7.18E-05
40.11	2.30E-04	33.38	1.29E-04	25.00	1.19E-05	36.81	2.85E-04	25.68	8.20E-05
40.46	2.65E-04	33.64	1.10E-04	25.38	4.35E-05	39.54	2.36E-04	26.28	6.83E-05
40.76	1.27E-04	33.90	1.01E-04	25.73	7.43E-05	41.34	3.10E-04	26.86	6.95E-05
40.95	1.40E-04	34.17	1.18E-04	26.13	1.04E-04	43.47	2.85E-04	27.70	1.24E-04
41.20	1.96E-04	34.45	1.06E-04	26.53	9.13E-05	45.56	2.25E-04	28.45	8.12E-05
41.55	2.66E-04	34.73	1.16E-04	26.72	4.32E-05	47.88	2.97E-04	28.90	1.20E-04
41.98	2.50E-04	35.01	1.13E-04	26.87	1.18E-04	50.14	4.00E-04	29.44	9.55E-05
42.28	1.00E-04	35.31	1.07E-04	27.00	5.48E-05	52.88	3.84E-04	30.45	1.49E-04
42.63	2.65E-04	35.62	1.09E-04	27.17	7.94E-05	55.99	3.65E-04	31.45	9.37E-05
43.03	2.19E-04	35.91	1.20E-04	27.40	7.74E-05	58.96	3.87E-04	31.96	9.92E-05
43.47	4.00E-04	36.21	1.11E-04	27.61	7.94E-05	62.55	5.54E-04	32.69	1.45E-04
44.00	2.40E-04	36.52	1.11E-04	27.78	9.92E-05	67.69	6.01E-04	33.59	1.44E-04

44.43	2.60E-04	36.91	8.59E-05	28.01	4.06E-05	75.43	7.26E-04	34.59	1.72E-04
44.86	2.40E-04	37.34	1.29E-04	28.37	7.94E-05	91.00	0.00133	35.76	1.91E-04
45.29	2.60E-04	37.77	1.66E-04	28.72	7.81E-05	141.49	0.00367	37.03	1.81E-04
45.70	2.00E-04	38.26	1.66E-04	28.96	1.26E-04			38.76	2.73E-04
46.12	2.60E-04	38.83	2.10E-04	29.20	6.26E-05			40.56	3.18E-04
46.54	2.00E-04	39.48	2.07E-04	29.45	7.94E-05			45.00	4.00E-04
46.91	2.00E-04	40.34	3.05E-04	29.76	1.19E-04			50.78	3.78E-04
47.32	2.40E-04	41.12	1.42E-04	30.27	6.39E-05			54.81	4.06E-04
47.78	2.40E-04	41.72	2.19E-04	30.70	9.99E-05			59.60	5.59E-04
48.28	2.80E-04	42.39	1.83E-04	31.04	9.92E-05			64.15	5.49E-04
48.84	2.80E-04	43.02	1.67E-04	32.47	9.45E-05			76.46	0.0016
49.34	2.20E-04	43.90	3.10E-04	33.98	9.93E-05			110.92	0.00257
49.96	3.80E-04	44.86	2.83E-04	35.03	1.49E-04			181.82	0.0071
50.58	2.08E-04	45.66	2.90E-04	36.16	2.51E-04				
51.44	5.92E-04	46.48	2.83E-04	36.69	4.14E-05				
52.52	3.80E-04	47.16	1.72E-04	37.22	1.58E-04				
53.46	4.40E-04	47.94	3.43E-04	38.16	1.04E-04				
54.38	3.40E-04	48.98	3.08E-04	39.14	1.49E-04				
55.68	7.20E-04	50.72	3.65E-04	39.95	1.00E-04				
56.99	3.04E-04	52.64	3.16E-04	40.80	1.52E-04				
57.83	3.38E-04	53.75	2.08E-04	42.01	1.99E-04				
59.10	5.98E-04	54.98	3.51E-04	43.76	1.39E-04				
60.45	3.54E-04	56.23	2.89E-04	45.13	1.66E-04				
61.91	6.46E-04	58.01	3.54E-04	46.57	1.55E-04				
64.08	7.60E-04	60.34	4.77E-04	48.37	1.24E-04				
66.51	7.20E-04	62.69	6.58E-04	50.44	1.48E-04				
69.17	7.86E-04	65.50	7.52E-04	53.00	1.96E-04				
73.27	4.52E-04	67.93	5.94E-04	55.78	2.07E-04				
78.48	0.00113	70.89	9.33E-04	58.22	2.06E-04				
84.32	0.00133	74.77	7.91E-04	59.81	3.50E-04				
92.13	0.00155	78.41	0.00116	61.68	2.58E-04				
101.10	0.00252	81.26	0.00133	64.11	4.76E-04				
108.76	0.00221	83.73	0.00102	68.43	4.00E-04				
115.96	0.00249	86.44	0.00259	74.68	8.05E-04				
127.95	0.00331	91.39	0.00172	80.24	4.98E-04				
140.41	0.00509	96.68	0.00254	105.62	2.98E-04				
		104.23	0.00299						
		126.59	0.00342						
		168.03	0.00737						

Table 0-3: Data of σ_a/dN and ΔK for BM and FSPedM of SDSS after plasticity correction.

BM (113 MPa)		BM(94MPa)		BM(66MPa)		FSP(94 MPa)		FSP(66 MPa)	
ΔK^*	da/dN	ΔK^*	da/dN	ΔK^*	da/dN	ΔK^*	da/dN	ΔK^*	da/dN
39.39	7.65E-06	32.49	6.80E-06	23.26	5.47E-06	32.89	5.61E-06	22.78	3.00E-06
39.62	8.17E-05	32.67	8.06E-05	23.57	4.45E-05	33.73	5.15E-05	23.22	6.03E-05
39.88	7.81E-05	32.84	7.05E-05	23.87	7.54E-05	34.11	9.69E-05	23.67	7.01E-05
40.16	1.15E-04	33.03	6.93E-05	25.10	4.63E-05	34.68	1.30E-04	24.15	6.74E-05
40.42	1.83E-04	33.19	7.05E-05	25.88	5.55E-05	35.58	2.14E-04	24.66	7.24E-05
40.69	1.67E-04	33.39	7.92E-05	26.05	8.26E-05	36.78	2.20E-04	25.20	7.18E-05
40.99	2.28E-04	33.68	1.29E-04	26.71	7.83E-05	38.19	2.52E-04	25.80	8.20E-05
41.36	2.37E-04	33.92	1.11E-04	26.89	8.33E-05	39.77	2.36E-04	26.41	6.83E-05
41.68	2.21E-04	34.20	1.01E-04	27.20	4.21E-05	41.60	3.10E-04	27.11	9.34E-05
41.89	1.36E-04	34.48	1.19E-04	27.37	8.06E-05	43.76	2.85E-04	27.97	9.90E-05
42.17	1.74E-04	34.76	1.07E-04	27.55	7.98E-05	45.89	2.55E-04	28.62	8.12E-05
42.55	2.46E-04	35.05	1.17E-04	28.01	6.46E-05	48.26	3.02E-04	29.08	1.20E-04
43.02	2.44E-04	35.34	1.14E-04	28.20	8.11E-05	50.57	3.00E-04	29.75	1.37E-04
43.34	2.49E-04	35.66	1.07E-04	28.96	6.51E-05	53.37	4.84E-04	30.79	1.23E-04
43.72	1.70E-04	35.95	1.10E-04	29.21	1.04E-04	56.57	4.65E-04	31.67	9.37E-05
44.16	4.11E-04	36.26	1.21E-04	29.46	1.02E-04	59.61	4.87E-04	32.34	1.49E-04
44.65	2.84E-04	36.56	1.12E-04	29.98	1.02E-04	63.30	4.54E-04	33.09	9.68E-05
45.23	3.37E-04	36.87	1.12E-04	30.37	1.52E-04	68.59	6.11E-04	33.86	1.54E-04
45.70	2.64E-04	37.36	8.64E-05	31.76	1.62E-04	76.59	9.26E-04	34.88	1.72E-04
46.18	2.64E-04	37.74	1.30E-04	34.36	1.06E-04	92.75	1.93E-03	36.07	1.91E-04
46.66	2.65E-04	38.25	1.67E-04	34.91	1.67E-04	119.65	3.24E-03	37.38	1.91E-04
47.12	2.44E-04	38.74	1.67E-04	36.68	1.03E-04	180.54	8.82E-03	39.15	2.23E-04
47.58	2.44E-04	39.40	2.11E-04	37.32	1.67E-04			41.00	2.48E-04
48.05	2.44E-04	40.07	2.08E-04	37.87	2.09E-04			45.56	2.80E-04
48.47	2.13E-04	41.14	3.07E-04	38.47	1.46E-04			51.52	3.58E-04
48.94	2.34E-04	41.66	1.44E-04	39.93	1.73E-04			55.69	4.16E-04
49.46	2.56E-04	42.37	2.20E-04	40.62	1.69E-04			60.04	4.97E-04
50.03	2.78E-04	43.03	1.85E-04	41.76	1.30E-04			63.77	5.98E-04
50.66	3.00E-04	43.66	1.69E-04	42.50	1.61E-04			69.33	6.77E-04
51.24	2.68E-04	44.82	3.12E-04	44.51	2.05E-04			78.57	1.02E-03
51.95	3.23E-04	45.63	2.86E-04	46.46	2.24E-04			119.55	3.08E-03
52.66	3.17E-04	46.46	2.93E-04	47.67	2.29E-04				
53.66	4.33E-04	47.31	2.86E-04	49.84	3.74E-04				
54.92	5.28E-04	47.84	1.74E-04	51.91	2.85E-04				
56.03	4.47E-04	48.92	3.45E-04	54.86	5.21E-04				
57.13	4.27E-04	49.95	3.10E-04	58.18	4.49E-04				
58.68	5.83E-04	52.53	3.68E-04	61.91	9.17E-04				
60.27	5.66E-04	53.87	3.20E-04	64.59	6.01E-04				
61.30	3.57E-04	54.80	2.11E-04	66.33	3.67E-04				
62.88	5.23E-04	56.40	3.53E-04	69.76	6.46E-04				
64.56	5.35E-04	57.36	2.92E-04	73.08	6.41E-04				

66.43	5.67E-04	60.11	3.56E-04	82.48	0.00129				
69.25	8.04E-04	62.13	4.82E-04	92.47	0.00165				
72.47	8.57E-04	65.00	6.63E-04	102.48	0.00267				
76.12	8.78E-04	67.94	7.60E-04						
81.98	7.71E-04	70.00	6.01E-04						
89.91	0.00157	74.15	9.40E-04						
99.57	0.00155	78.02	6.99E-04						
114.02	0.00399	81.70	0.00117						
133.46	0.00406	83.94	0.00135						
153.26	0.00551	86.86	0.00103						
175.48	0.00776	89.58	0.00262						
186.55	0.00988	97.43	0.00173						
		100.44	0.002						
		106.02	0.00192						
		115.47	0.0024						

Table 0-4: Principal and von Mises residual stresses at point A.

Depth in mm	Principal stress in the X direction	Principal stress in the Z direction	Von Mises in Mpa
0.02	456	49	433.6
0.09	119	-129	214.8
0.15	80	-121	175.3
0.21	64	-155	195.0
0.27	79	-97	152.7
0.33	16	-98	106.9
0.39	150	-252	351.9
0.45	194	-47	221.3
0.51	216	45	197.4
0.57	-62	-175	153.7
0.63	-178	-364	315.3
0.69	616	368	536.8
0.75	355	62	328.4
0.81	-203	-461	400.2
0.87	518	236	449.2
0.93	324	41	305.6
1	644	400	563.1

Table 0-5: Principal and von Mises residual stresses at point B.

Depth in mm	Principal stress in X direction	Principal stress in Z direction	Von Mises in Mpa
0.02	411	-17	419.76
0.09	333	25	321.23
0.15	208	-10	213.18
0.21	109	-84	167.61
0.27	124	-8	128.19
0.33	-69	-370	340.78
0.39	-24	-207	196.10
0.45	14	-74	81.90
0.51	158	10	153.24
0.57	-50	-144	126.63
0.63	101	-27	116.86
0.69	242	63	217.46
0.75	11	-105	110.91
0.81	-210	-465	403.33
0.87	-154	-441	387.66
0.93	561	242	487.36
1	593	300	513.56

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