Built-up Edge Mechanisms in the Machining of

Duplex Stainless Steels

By

Junior Nomani, BEng, MEngSt

Submitted in fulfilment of the requirements for the degree of

Doctor of Philosophy

Deakin University

October, 2014
I am the author of the thesis entitled

submitted for the degree of

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Nomenclature

\( \varepsilon \) strain

\( \Delta \varepsilon \) strain difference

\( \alpha \) ferrite

\( \gamma \) austenite

\( \dot{\sigma} \) martensite

L Liquid state

\( PREN \) pitting resistance equivalent number

\( \bar{\sigma} \) average stress

\( \bar{\varepsilon} \) average strain

V Volume fraction

d distance

\( (\alpha) \) rake angle

\( \phi \) shear angle

\( \beta \) friction angle

k shear strength vs compressive strength gradient

to depth of cut

tc chip thickness

r chip ratio

u displacement

F force
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<tr>
<td>$t$</td>
<td>time</td>
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<tr>
<td>$E$</td>
<td>elastic modulus</td>
</tr>
<tr>
<td>$K$</td>
<td>material hardening constant</td>
</tr>
<tr>
<td>$n$</td>
<td>material hardening constant</td>
</tr>
<tr>
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<td>total strain</td>
</tr>
<tr>
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<td>stress</td>
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<td>$\sigma_y$</td>
<td>yield stress</td>
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</tr>
<tr>
<td>$\varepsilon_f$</td>
<td>strain at fracture, full damage variable</td>
</tr>
<tr>
<td>$D$</td>
<td>damage variable</td>
</tr>
<tr>
<td>$\tau_p$</td>
<td>shear stress</td>
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<tr>
<td>$M_z$</td>
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<td>thrust force</td>
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<tr>
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</tr>
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</tr>
<tr>
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<td>velocity of workpiece</td>
</tr>
<tr>
<td>$f$</td>
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</tr>
<tr>
<td>$\varnothing$</td>
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<tr>
<td>$A_o$</td>
<td>original grain size</td>
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</tr>
<tr>
<td>( \mu )</td>
<td>friction coefficient</td>
</tr>
<tr>
<td>( \dot{\alpha} )</td>
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<tr>
<td>( \beta )</td>
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<tr>
<td>( n )</td>
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<tr>
<td>( \sigma_{0.2} )</td>
<td>0.2% proof stress</td>
</tr>
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</tr>
<tr>
<td>( \varepsilon_{max} )</td>
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<td>( G )</td>
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<td>( \sigma_{1.5} )</td>
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<tr>
<td>( BUE )</td>
<td>built-up edge</td>
</tr>
<tr>
<td>DoC</td>
<td>depth of cut</td>
</tr>
<tr>
<td>( T_{ss} )</td>
<td>steady state temperature</td>
</tr>
<tr>
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<td>temperature</td>
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Abstract

This thesis investigates the mechanisms that trigger the formation of built-up edge during the machining of duplex stainless steel alloys. The work was aimed at attempting to resolve long-established machining issues with difficult-to-machine materials. Duplex stainless steel alloys are a two phase microstructure that machine (cut) with a high tendency to develop built-up edge (BUE) formation – a “sticking” type characteristic. This undesired phenomenon triggers machinability issues such as poor machined surface texture and accelerated tool wear, subsequently leading to increased manufacturing costs. The present literature does not fully explain the natural sticking behaviour of duplex stainless steels. If the triggering mechanisms causing BUE (built-up edge) could be more fully understood, then solutions could be developed to minimise or possibly avoid these mechanisms. The research in this thesis comprises of three main stages.

Firstly, an observational study was conducted with the aim of discovering the natural machining behaviour of duplex stainless steels. Machinability drilling trials were performed and indicated that overall, duplex stainless steel alloys 2507 and 2205 produced poorer machinability characteristics than Austenitic Stainless Alloy 316L - in terms of tool wear, cutting forces and machined surface finish. Tool wear observations found adhesion wear caused by BUE was the dominant wear mechanism in machining duplex.

A second part of the study focused on chip formation using an experimental 'quickstop' method in a turning operation to produce “frozen chip-root” samples. Scanning electron microscope (SEM) images of chemically etched chip-root samples uncovered a concentration of the ferrite phase in the stagnation zone, in a pre-developed built-up layer. Micro-cracks which act as BUE initiators were identified in this region of ferrite build-up.

The second stage sought to characterise the phases present in the stagnation zone using electron backscatter diffraction (EBSD) analysis. Phase mapping of the stagnation zone detected 65-85% more ferrite than austenite in the stagnation zone. This confirmed ferrite was collecting at the stagnation zone, and established the ferrite build-up was triggering the formation of BUE.
The third stage developed a finite element (FE) model to explain how the ferrite build-up was occurring. The FE model and grain boundary mapping data from EBSD analysis concluded the ferrite build-up was triggered by austenite softening under high strain. The model revealed, under high strain conditions \( \varepsilon > 1.5 \), austenite softened and plastically deformed to a greater degree than ferrite, prompting a ferrite build-up. A relation was identified with the amount of ferrite build-up and the strain difference \( \Delta \varepsilon \) between the two phases. The greater the strain difference, the greater the ferrite build-up. Based on the corroborating data, a relationship was established with (BUE):

\[
\text{The occurrence of built-up edge is proportional to the strain difference} \\
i.e. \Delta \varepsilon \propto \text{BUE}.
\]

The main significance of the thesis outputs is firstly, the academic community would have a better understanding of how BUE formation develops during machining of duplex stainless steel alloys; and secondly, for the first time, a duplex finite element cutting model has been developed, which is able to predict the plastic behaviour of individual phase microstructure during simulated chip formation.
Machining remains a prominent sector of the manufacturing industry, showing continual rise globally since the recent 2009 global recession. The ‘2014 Metalworking Capital Spending Survey’ by Gardner Research, forecasted an estimated spending on new metal-cutting equipment in the US at $7.44 billion dollars (USD); an 18% increase from the previous year 2013. Metal-cutting Job-Shop industries are reported to make up 35% of this total expenditure alone. One reason stated for the reported growth, is the increasing demand for improved quality. The incentive for higher quality standards has not only attracted industries to the purchases of new equipment, but also contributed to the ‘Reshoring’ of outsourced work to be manufactured locally within the US.

With regards to machining research in industry, history has shown advances in fundamental theory and practices of machine processes led to substantial industrial growth and major economic productivity. Before the establishment of tool-life principles by Fredrick W. Taylor, machining practice was considered simple-minded, using ‘brute-force’ or ‘common-sense’ type methodology. Those who were engineering-minded at the time, knew it was about selecting the right cutting speed or the right feed-rate for the particular process, etc. Nevertheless, it took Taylor eight years to determine that a cutting speed leading to a tool life of 20 minutes, gave a cutting tool its most optimal use. Taylor’s published paper [1] documenting his inaugural research of 26 years, communicated a number of ground-breaking achievements. One pioneering milestone was establishing the relationship between which engineering variables governed the machining process. Early companies that adopted Taylor’s principles and methodologies saw a 200-300% increase in productivity [2]. Taylor’s concepts are considered a basis for machining practice and are still widely used in the machining industry today.

Presently, the demand for high-performance materials is ever increasing. New alloys are continually emerging and being developed with the focus on becoming stronger, lighter, cheaper, more heat or corrosion resistant, etc. Also, with the commercialisation of these materials brings about the challenges associated with their manufacturing, in particular, their machinability aspects.
Modern duplex stainless steels are a material that is the result of continuous attempts to develop new engineering alloys with improved properties. An equal two-phase microstructure, duplex stainless combine the inherent benefits of both $\alpha$-ferrite and $\gamma$-austenite phases. The $\alpha$-ferrite consists of a body-centred cubic (BCC) structure which promote duplex stainless steels excellent pitting and corrosion resistant properties. While the $\gamma$-austenite, a face centred cubic (FCC) structure, promotes superior strength and toughness [3]. Duplex stainless steels are also less expensive than the more popular austenitic stainless grades, requiring lower amount of alloying nickel content. The applications of duplex stainless steel alloys are continually expanding, ranging from oil piping, to structural uses including on-land or off-shore such as oil platforms. Applications are also well-established in water treatment, milk and chemical processing. Duplex stainless are regarded for their ability to provide good material strength and toughness under harsh environments e.g. thermal or corrosive. Despite their wide applications and continually new emerging products, the machining aspects of duplex stainless steel alloys still remains a continual challenge. The existence of unresolved issues in machining these alloys becomes apparent when considering their combination of high strength, toughness, low thermal conductivity and small amount of non-metallic inclusions.

A primary long-established issue with the machining of duplex stainless steel alloys is they have a high tendency to generate built-up edge (BUE), an undesired phenomenon that describes the workpiece material adhering to the cutting tool during machining. The presence of BUE (built-up edge) can cause machining issues such as poor machined surface finish, wider dimensional tolerancing and more importantly, accelerated tool wear. This natural ‘sticking’ behaviour of duplex remains largely in the literature. Researchers, including cutting tool manufacturers, can only anecdotally note the machining conditions where BUE can be mostly avoided [4].

At this present stage, no research work has attempted to address the BUE issue in any significant detail, particularly in understanding how or why BUE occurs frequently when machining duplex stainless steels. The majority of the body of research reported in this thesis, focuses on the duplex alloy’s BUE phenomenon, in particular observing the BUE behaviour and complexity, in the attempt to uncover the mechanisms surrounding its development. In similarity to the development of Taylor’s tool life principles, if the BUE mechanism could be fully understood from a fundamental
aspect, then the proper strategies could be developed to minimise, or potentially eliminate BUE in machining duplex stainless steel alloys.

1.1 Thesis objectives

The main purpose of this thesis was to acquire a better understanding of how BUE is triggered in the machining of duplex stainless steel alloys. This purpose formed the basis of the primary research question which was to identify:

“What are the mechanisms triggering frequent built-up edge (BUE) during the machining of duplex stainless steel alloys?”

The three main objectives based on the research question were as listed below.

1. Observe the machinability of duplex stainless steel alloys and consequently determine what influence the formation of BUE had on the machining performance and furthermore, identify the most damaging forms of wear that occurs during the machining of these alloys.

2. Critically characterise mechanisms that trigger the formation of BUE in duplex stainless steel alloys and identify what material properties are responsible for triggering the behaviour and determine those process variables that could aid in controlling its development.

3. Develop an holistic model to explain how these mechanisms are occurring. Moreover build an FEM to simulate these mechanisms and validate their authenticity.
1.2 Thesis layout

This thesis comprises of seven main chapters and Figure 1.1 displays a general summary of the framework. Descriptions of the chapters are listed below.

Chapters 2 comprises of the main literature analysis and is divided into 3 sections, namely: an introduction to duplex stainless steel alloys, from brief history to physical metallurgy, including current trends and applications; a background on machining theory and associated literature, including machinability of duplex stainless steels; and the theory and background on finite element modelling in relation to metal cutting.

Chapter 3 contains two separate observational studies, the first assesses the machinability of duplex stainless steel alloys, and observes what impact the formation BUE had on the machining process via an experimental drilling process. The second study observed the chip formation behaviour by use of an explosive quick-stop method, where frozen ‘chip-root’ samples were created and observed under scanning electron microscope. The microstructural plastic flow of the material is reported.

Chapter 4 continues on from Chapter 5 and details a characterisation study on ‘chip root’ samples. Material phases are characterised in the stagnation zone using Electron Backscatter Diffraction (EBSD) technique. The plastic behaviour of individual phases is observed through grain boundary mapping. A brief summary detailing the collected data is conveyed in the discussion section of this chapter.

Chapter 5 concludes the experimental chapters, detailing the development of a finite element metal cutting model based on the duplex microstructure. Metal cutting data is acquired through simulations - such as probing of strain values of individual phases - and compared with experimental data. A stagnation zone is also simulated in the model, where strain data of the material phases in this region was collected.

Chapter 6 combines the thesis elements from the experimental chapters into a series of discussion sections. The triggering mechanisms to BUE and its relationships are identified and discussed with supporting experimental data and literature. A brief discussion on cutting temperature effects is included in this chapter, including how the thesis contributes towards machining literature and industry. The concluding chapter, Chapter 7 provides a complete summary of the thesis and discusses future work information.
Figure 1.1 General outline of thesis framework
CHAPTER TWO

2.0 Background

The aims of the following literature survey was to provide a broad, but clear and concise review of the most relevant literature related to the objectives of the thesis. Most importantly, identifying the gaps and discrepancies in the present literature. Each of the three sections of this chapter are also aimed at providing background literature and related theory on the subsequent experimental chapters. The main topics of the three sections include an introduction into duplex stainless steels, machining theory and literature on BUE edge formation, and lastly, finite element modelling in relation to metal cutting.

2.1 Introduction to duplex stainless steels

This first section serves as an introduction to the duplex alloy, providing detail on its past history, the general standards and classification, including the physical metallurgy of the alloy, and its applications and current application trends. The concluding two sections discuss the present literature on machining theory and finite element modelling.

2.1.1 History

Duplex stainless steels first emerged in 1927, when Bain and Griffiths [5] published in a paper what was described as austenite-ferrite alloys. Despite its given potential over existing stainless steels, which includes retaining higher strength and improved stress corrosion cracking resistance, duplex grades were restricted in usage. Early grades were used mainly for forgings in industries such as food processing, pulp and paper, and oil refinement [6] and also used in heat exchanger tubing. Early limitations in usage were primarily due to duplex offering poor weldability. Weld joins were weak, due to poor toughness and corrosion resistance, caused by ferrite forming in the heat affected zone. This issue would not be resolved until the late 1960’s with the development of cleaner steel producing technology. The vacuum and argon oxygen decarburisation processes (VOD and AOD) were able produce cleaner grades by controlling the carbon content. A nickel shortage in the 1970’s combined increasing
offshore oil and gas expeditions in the North Sea, led to the demand for both a cheaper stainless steel and a stainless steel to withstand harsh environments [3]. These events led to the development of a new generation of duplex grades, such as the 2205 grade released in the 1980’s and still the most widely used grade today.

2.1.2 Standards and classification

In the classification for duplex wrought grades, some standards and numbering systems incorporate the chromium and nickel content. For example, the AISI and UNS uses similar numbering system, AISI 2205 and UNS S31803/S32205 is the same common 2205 grade. Both systems incorporate 22% Chromium and 05% Nickel content as part of the naming. The European standard ‘EN’ differs by generically classifying duplex stainless steels according to either its intended purpose or its chemical composition (% mass). Table 2.1 displays a list of 2nd generation duplex grades and their typical chemical composition.

### Table 2.1. General composition of common Duplex stainless steel alloys

<table>
<thead>
<tr>
<th>Type</th>
<th>AISI</th>
<th>UNS / ASTM</th>
<th>EN</th>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>N</th>
<th>Mn</th>
<th>Cu</th>
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<tbody>
<tr>
<td>Duplex (2nd gen)</td>
<td>2205</td>
<td>S31803</td>
<td>1.4462</td>
<td>0.03</td>
<td>22.0-23.0</td>
<td>4.5-6.5</td>
<td>3.0-3.5</td>
<td>0.14-0.20</td>
<td>2</td>
<td>-</td>
</tr>
<tr>
<td>Duplex (2nd gen)</td>
<td>2304</td>
<td>S32304</td>
<td>1.4362</td>
<td>0.03</td>
<td>21.5-24.5</td>
<td>3.0-5.5</td>
<td>0.05-0.6</td>
<td>0.05-0.20</td>
<td>2.5</td>
<td>0.05-0.6</td>
</tr>
<tr>
<td>Duplex (2nd gen)</td>
<td>2507</td>
<td>S32750</td>
<td>1.4410</td>
<td>0.03</td>
<td>24.0-26.0</td>
<td>6.0-8.0</td>
<td>3.0-5.0</td>
<td>0.24-0.32</td>
<td>1.2</td>
<td>0.5</td>
</tr>
<tr>
<td>Austenite</td>
<td>316L</td>
<td>S31603</td>
<td>1.4435</td>
<td>0.03</td>
<td>16.0-18.0</td>
<td>10.0-14.0</td>
<td>2.0-3.0</td>
<td>0.1</td>
<td>2</td>
<td>-</td>
</tr>
<tr>
<td>Duplex (1st gen)</td>
<td>329</td>
<td>S32900</td>
<td>1.4460</td>
<td>0.08</td>
<td>23.0-28.0</td>
<td>2.5-5.0</td>
<td>1.0-2.0</td>
<td>-</td>
<td>1.00</td>
<td>-</td>
</tr>
</tbody>
</table>

2.1.3 Physical metallurgy

In the family of stainless steels, duplex exists between austenite and ferrite grades as shown in the Schaeffler diagram - Figure 2.1. The chromium content is higher in duplex than austenitic grades, but the nickel content is lower, making duplex less expensive. Duplex alloy’s distinct banded microstructure originates from a 1:1 matrix of γ-austenite and α-ferrite, as shown in Figure 2.2. Both phases exist in relatively large separate volumes in approximately equal fractions rather than an inclusion phase embedded in the matrix formed by the other [7]. The α phase contains a BCC crystal structure. It is responsible for high pitting and crevice corrosion resistance properties.
While the $\gamma$ phase, a FCC crystal structure, is responsible for relative ductility and strength, and also for its resistance to uniform corrosion [8].

![Figure 2.1. Schaeffler diagram displaying the four groups of stainless steels](image1)

Figure 2.1. Schaeffler diagram displaying the four groups of stainless steels

![Figure 2.2. Typical ‘banded’ duplex stainless steel microstructure](image2)

Figure 2.2. Typical ‘banded’ duplex stainless steel microstructure [9]

The formation of the banded microstructure can be described in the iron chromium nickel binary phase diagram, Figure 2.3. Duplex stainless steels solidify to 100% complete ferrite. The $\alpha \rightarrow \gamma$ transformation initiates upon cooling to lower temperatures, around 1000°C, depending on the chemical composition. During cooling, some of the ferrite is transformed to austenite, and at lower temperatures there is little further change to the equilibrium balance. The amount of retained ferrite is largely dependent on the chemical composition makeup. Chromium acts as the primary ferrite stabiliser, additions of chromium increases the formation of BCC structure of iron, while nickel is the main austenite former promoting the FCC structure. Other elements, molybdenum, nitrogen, copper, carbon, support chromium
and nickel in the stabilisation of $\alpha$ and $\gamma$ phases. Their influence on phase stabilisation is shown in the following set of equations. A multi-variable linear regression predicting the amount of retained ferrite.

$$Cr_{eq} = \%Cr + 1.73\%Si + 0.88\%Mo$$  \hspace{1cm} \text{Equation 2.1}

$$Ni_{eq} = \%Ni + 24\%C + 21.75\%N + 0.4\%Cu$$  \hspace{1cm} \text{Equation 2.2}

$$\%Ferrite = -20.93 + 4.01Cr_{eq} - 5.6Ni_{eq} + 0.016T$$  \hspace{1cm} \text{Equation 2.3}

Where $T$ in degrees Celsius is the annealing temperature ranging from 1050 – 1150°C and the element compositions are in wt.%.

The Schaeffler equations however, do not factor in cooling rates [10], as cooling rates from high temperatures also influence the amount of retained ferrite. Quicker cooling rates lead to a higher percentage of ferrite retention. Ultimately, both the chemical composition and the cooling rates are strictly controlled during material’s processing.

![Section of the Fe-Cr-Ni Binary phase diagram at 68% iron](image)

**Figure 2.3 Section of the Fe-Cr-Ni Binary phase diagram at 68% iron**

### 2.1.4 Material properties

A main function of a stainless steel is to be corrosion resistant. The pitting resistance equivalent number (PREN) is considered a common measure of corrosion resistance that is defined according to Equation 2.4. A high PREN number indicates a high resistance to pitting and crevice corrosion. Duplex stainless steels are reported having a higher PREN number than austenite 316L, $\text{PREN} > 40$ [11]. This is due to the higher alloying of Cr Mo and N, according to Equation 2.4.
Equation 2.4

\[ PREN = \%Cr + 3.3(\%Mo + 0.5\%W) + 16\%N \]

Duplex stainless steels are well known for their strength among stainless steels, obtaining higher yield and ultimate tensile strength compared to austenitic grades. Table 2.2 displays strength and hardness comparisons. Tseng [12] showed that increasing nitrogen content has a good correlation with increasing strength and hardness in duplex. This would correlate to increasing austenite presence, since nitrogen is an austenite former. The gain in material strength and hardness in duplex stainless steels is balanced by a lower total elongation. Duplex grades also exhibit lower levels of work-hardening compared to austenitic grades due to lower metastability [4, 13].

### Table 2.2 Duplex tensile strength and hardness figures

<table>
<thead>
<tr>
<th>Alloy (AISI)</th>
<th>Yield (MPa)</th>
<th>UTS (MPa)</th>
<th>Hardness HV$_{100g}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>2507</td>
<td>570</td>
<td>866</td>
<td>285</td>
</tr>
<tr>
<td>2205</td>
<td>556</td>
<td>777</td>
<td>279</td>
</tr>
<tr>
<td>Austenite 316L</td>
<td>326</td>
<td>640</td>
<td>254</td>
</tr>
</tbody>
</table>

#### 2.1.5 Two-phase system properties

As a two-phase system the mechanical properties of the bulk material in a duplex alloy exist from combining the constituent properties of austenite and ferrite. Both $\alpha$ and $\gamma$ phases obtain a complete different set properties, where the final bulk property is based on two factors, namely: the present amount of the constituents; and the interaction between them. The classical law of mixtures provides a generalised model of how volume fraction influences the makeup of the bulk material property. An example is shown in Figure 2.4. Illustrating the influence of volume interaction in regards to yield strength.
In terms of load sharing, the two phases will respond differently to applied load. An element of the reason is due to differences in mechanical properties, such as yield strength. The other element, is due to phases being under different pre-strain conditions (austenite under residual tension and ferrite under residual compression) caused by dissimilar coefficient of thermal expansion [15]. Cho and Gurland [14] reported the load partitioning of $\alpha$ and $\gamma$ phases in duplex could be modelled by a modified law of mixture equations, first introduced by Tamura [16]. Similar to the law of mixtures, Equations 2.5 and 2.6 show that the average stress $\bar{\sigma}$ and strain $\bar{\varepsilon}$ partitioned to each phase is dependent on the volume fraction ($V$) of each phase present.

\[
\bar{\sigma}_C = \sigma_\alpha V_\alpha + \sigma_\beta V_\beta \quad \text{Equation 2.5}
\]

\[
\bar{\varepsilon}_C = \varepsilon_\alpha V_\alpha + \varepsilon_\beta V_\beta \quad \text{Equation 2.6}
\]

Cho [14] reported $\alpha$ and $\gamma$ phases in duplex stainless steels to be in good agreement with the modified law, Equations 2.5 and 2.6, in the observed range of strain between 1-6%. However, there is still a nonlinear deformation element to consider at higher strain ranges.
2.1.6 Current trends and applications

Duplex stainless steel alloys have gained acceptance in industries that operate under harsh environments. The most popular 2205 grade provides corrosion resistance in areas which would supersede austenitic grades AISI 304, 316 and 317 [6]. Today, duplex grades are common place in the chemical, petrochemical, water desalination, dairy, pulp, power, oil and gas industries in products ranging from piping to mechanical seals, including structural members [3, 17]. The utilisation of duplex alloy’s mechanical strength, can also offer cost savings. A 10% cost reduction is reported in high pressure piping when compared to austenitic grades [18].

![Figure 2.5. Europe half-yearly monthly alloy surcharge for flat stainless products in 2014, (Source Outokumpu)](image)

Though duplex stainless steel alloys share of the stainless steel market share remains relatively small, less than 1% in 2007, its annual production growth increased over 100% in that decade [19]. The demand for duplex is growing and they are slowly replacing austenitic grades because of their benefits in corrosion and mechanical properties, and cost reductions [20]. One of the main current trends is the growth and development of duplex “Lean” grades, e.g. LDX 2101 & LDX 2404 [21]. The lean grades maintain their attractive duplex mechanical and corrosion properties, but were developed with lower nickel content making them less expensive than regular duplex. Apart from ferritic grades, such as AISI 439, LDX 2101 is the cheapest alloy according to current cost figures shown in Figure 2.5.

There has also been developing growth with duplex used in the construction sector. Although the initial cost of stainless steel structural products is four times greater than...
carbon steel [22], the initial high price would be justified by factoring the additional immediate costs and longer term costs such as corrosion protection, fire proofing, maintenance, etc. Duplex stainless steels in construction require very low maintenance and have a long lasting appearance which is why they are becoming the material of choice for bridges [23], such as the Stonecutters Bridge Tower opened in 2009 in Hong Kong, Figure 2.6. The bridges entire outer stainless steel skin is made from duplex alloy 2205. It was chosen after standard molybdenum alloyed austenitic grades failed the design corrosion and strength requirements, and higher alloyed austenitic grades could not meet the cost objectives. The bridge includes a total length of 1596m and has a main span of 1018m.

![Figure 2.6 Stonecutters bridge constructed of duplex 2205 stainless steel cabling [24]](image_url)
2.2 Machining and machinability

The ‘machinability’ of a material is a complex definition and at present cannot be fully defined due to the physics and complexity of variables involved. Instead, the machinability term takes on various measures to describe the relative ease a material is able to be machined to its final shape or form. Such measures range from analysing the cutting force required during the machining process, to observing the surface texture finish quality of the machined surface. These measures rely solely on comparative data to evaluate the performance of machining the material. Free-machining carbon steel UNS G12120 is generally used as a base material for comparisons [25, 26]. Machinability studies relating to tool wear, cutting forces, surface integrity, etc. have been a common approach to investigate how a material responds to machining. They are particularly useful in identifying problematic issues faced with machining difficult materials. However, as a measure of performance, machinability tests are limited in that they are often unable to identify the associated triggering mechanisms in machining issues. Furthermore, detailed studies such as investigating chip formation are necessary to expand on these mechanisms.

This section focuses on machining theory and research into the understanding of machinability of duplex stainless steels.

2.2.1 Metal cutting theory

The metal cutting process can be described as high strain deformation. In a simplified model, it is the transfer of kinetic energy from a rigid tool body to the workpiece material. During the transfer, the workpiece material is strained beyond its elastic, and subsequently plastic region to form a chip. This energy is referred to as cutting energy. Most of the cutting energy is focused around shear planes, where the main principal shearing occurs [27]. Shear planes was first introduced by Usachev [28] and the term coined by Ernst [29]. Shear planes are considered as boundary-lines which separate the deformed and undeformed material. Figure 2.7 shows an ideal cutting model first introduced by Piispanen based on shear plane theory [30]. Piispanen’s model describes the material being sheared as a sliding deck of cards, stacking on one another as it passes through the shear plane ‘AB’. Although Piispanen’s model does not take into account complete cutting variables such as chip curling, plastic contact friction and
BUE, its concepts are still valid, which is why it is still found in today’s textbooks as an easily understood interpretation of the chip forming process.

Figure 2.7 Chip formation flow diagram based on shear plane theory, After Piispanen [31]

Unknown to Piispanen, Merchant and Ernst [32] developed similar shear plane concepts. They justified the use of shear planes by stating the area where shear was occurring was so small that it could be approximated to a single plane. Merchant initially concluded that shear lies on a plane which the shear stress is a maximum and equals the shear stress of the workpiece material. Using this relation and including the rake and friction angles ($\alpha$) and $\beta$, Merchant arithmetically derived the following relation to describe the shear plane, in terms of a shear angle $\phi$.

$$\phi = 45^\circ + \frac{\alpha}{2} - \frac{\beta}{2}$$

Equation 2.7
Figure 2.8 Influence of shear plane angle on shear strain [33]

Equation 2.7 shows the importance of the shear angle $\phi$ having a direct influence on friction forces i.e. the friction angle $\beta$. Figure 2.8 shows how the variation of shear plane angle influences the value of shear strain, for three constant rake angles. Higher shear angles and subsequently rake angles correspond to a lower shear strain. For smaller shear angles $\phi < 5^\circ$, the shear force can be more than five times than the minimum where $\phi = 45^\circ$. Merchant however, later realised from his initial conclusion [32], the sum of Equation 2.8 would only total $45^\circ$ for an ideal plastic material. Merchant later adjusted the relation to account for the shear and compressive behaviour of the material, by introducing a $k$ value which equals the slope shear strength vs compressive strength curve of the metal, Equation 2.9.

$$2\phi = \arccot k + \alpha - \beta$$  \hspace{1cm} \text{Equation 2.8}

Rearranging Equation 2.8 and substituting terms in relation to chip formation geometry gives a general form of the equation, Equations 2.9 and 2.10, commonly displayed in today’s textbooks.
CHAPTER TWO

\[
\tan \phi = \frac{r \cos \alpha}{1 - r \sin \phi}
\]
Equation 2.9

\[
r = \frac{t_o}{t_c} = \frac{\sin \phi}{\cos(\phi - \alpha)}
\]
Equation 2.10

Where \(t_o = \text{depth of cut}\)

\(t_c = \text{chip thickness}\)

Furthermore, Merchant’s [34] observation of shear planes led to his ground breaking work for defining friction in cutting, where there was no mechanism in existence at the time. Known today as ‘Merchant’s Circle’, the analysis allowed for engineering quantities to be calculated for the first time and with good approximation, such as force, stress and energy. It was here that orthogonal cutting was also derived by Merchant, by viewing two-dimensional cutting as a static free-body system, Figure 2.9.

Figure 2.9 Merchant’s Circle analysis showing the relationship between component forces and chip formation [2]

A drawback to the use of Merchant’s circle and shear plane theory is that it is not valid in the presence of BUE formation. Built-up edge (BUE) is when the workpiece material adheres to the cutting tool, usually at the tool nose. This phenomenon is described in further detail later in this section, Chapter 2.2.6.
Another drawback to Merchant’s circle is the free-body cutting analysis described by orthogonal cutting is only valid for unbounding free flowing continuous chips and discontinuous chip formations. These different chip types are discussed in the next section. Predicting the actual size and shape of the shear plane still remains today as one of the main challenges in machining science [35], as there is not yet an accurate model to determine these.

![Figure 2.10 Chip formation showing primary shear zones](image)

**Figure 2.10 Chip formation showing primary shear zones**

In reality, the shearing and chip forming of most metals can be described by Figure 2.10. In this representation first introduced by Lee and Shaffer [36] using slip line theory, the shear planes are referred to as zones, since the deformation the material exhibits, develops along a region or a field rather than a single plane [37]. There can be up to three shear zones depending on the material and cutting conditions, these are the primary, secondary and tertiary shear zones. These shear zones act as entry points where the workpiece enters and yields before transitioning into the chip. A stagnation zone can occur in the shear zones located at the tip point of the tool. The material in this region remains stagnant, commonly referred to as dead material (or zone). The formation of a stagnation zone generally precedes the development of a BUE.
2.2.2 Chip types

The machine chip will vary in shape, size and geometry (morphology) depending on the type of cutting and conditions involved. A chip is formed from the moment it begins to flow along the tool rake face. In this region of material separation, machined chips can be classified into four groups.

Continuous chips

Continuous chip formation is considered steady state cutting. It involves continuous plastic flow from shear without fracture. Continuous chips generally form as a result of high cutting speed or large rake angle during cutting. Experimental and finite element work by Rosa et. al [38] showed continuous chips also form at low cutting speeds, in the orthogonal cutting of pure lead 99.9%. Continuous chips tend to deform along a narrow primary shear zone with a probability of forming a secondary shear zone [37]. Most ductile materials tend to form continuous chips – known as swarf. Continuous chips can also form across a wide primary shear with a curved boundary and include lower tertiary shear boundary acting below the machined workpiece, see Figure 2.11. These will occur in softer materials [37].

Continuous chips with a built-up edge

Built-up edge (BUE) is a condition when residue from the chip workpiece material is deposited by adhesion onto the workpiece material and is prevalent in the machining of ductile materials. BUE can severely affect cutting conditions. Its presence can create a positive tool rake angle, becoming part of the cutting edge. This can lead to poor machined surface finish and accelerated tool wear. Mechanisms leading to the formation of BUE are discussed in chapter section 2.2.6.

Discontinuous chips

Discontinuous chips occur when materials are unable to withstand high shear. The produced chip segments are either loosely attached or completely fragmented as they are cut. These tend to occur in materials that are brittle or contain hard particles and impurities. Discontinuous chips will occur at low rake angles and large depth of cut. These chips are more desirable in machining since chip breakup results in reduced friction between the chip-tool interfaces, giving better surface finish. Chip collection
and disposal is also more convenient. Discontinuous chip formation is generally common in machining cast irons.

**Serrated chips**

Most stainless steels, including duplex stainless steel alloys, machine with a serrated chip profile. These are sometimes referred to as homogeneous chips or segmented chips. These chips display a ‘sawtooth’ like profile, in which high and low shear strain areas are visible, Figure 2.11. A continuous thermal cycle of fracture and re-welding is thought to be responsible for creating the sawtooth profile [37]. Rhim and Oh [39] defined these chips as macroscopic continuous chips consisting of narrow bands of heavily deformed material alternating with larger regions of un-deformed material. This chip profile is also common in machining brittle materials such as ceramics and some hardened fine grain steels.

![Figure 2.11 Main types of chips](image)

2.2.3 Plastic instability in chip forming

In proper detail, the saw tooth profile of serrated chips are a product of more than just the fracture and re-welding model proposed by Shaw [40]. Ramalingam and Black [41] originated the notion that the formation of lamella structures was due to plastic instability occurring during high strain deformation. This plastic instability is inevitable and will even occur even during steady-state continuous chip formation [41], as shown by Barry et. al [42] in Figure 2.12. Barry et. al. observed micro lamellae
folds occurring on the free surface of a continuous chip formation in an orthogonal turning operation. Ramalingam and Black [41] concluded these lamella structures form as product of adiabatic shear. The deformation process involved in adiabatic shear, details to large strain concentrations in small volumes [43]. Doyle and Aghan [43] suggested an alternative mechanism to plastic instability involving fine grain sizes. However, as mentioned by Doyle and Aghan [43] the adiabatic shear mechanism is a complex process which is yet to be fully understood.

![Micro lamellae folds on the free surface of a continuous chip, orthogonal cutting of Ti-6Al-4V alloy](image)

**Figure 2.12** Micro lamellae folds on the free surface of a continuous chip, orthogonal cutting of Ti-6Al-4V alloy [42]

### 2.2.3 Machinability of duplex stainless steels

Stainless steels in general are regarded as difficult to machine materials due to their ability to work harden, their toughness and relatively low thermal conductivity [44-47]. Other problems stem from their high fracture toughness, which increases the tool-chip interface temperatures leading to poor surface finish and poor chip breaking. Furthermore, BUE formation is present even at elevated cutting speeds [4]. This deteriorates the finish of the machined surface and increases the cutting forces [48]. In light of this, the duplex stainless steel alloys are more difficult to machine than most austenitic grades.

Figure 2.12 shows machinability data produced by stainless steel manufacturer Outokumpu. The data show duplex stainless steel alloys 2304, 2205 and 2507 have a poorer machinability index than austenite 316L when machined with carbide tools. Meanwhile, lean duplex S32101 has a better machinability index, compared to all measured grades in the study. Jin et. al. [25] reported similar findings in a more comprehensive database study. The machinability of duplex 2507 was ranked poorer
than all austenitic grades apart from two higher alloyed austenitic grades UNS S32654 and S32154.

**Figure 2.12** Relative machinability index of duplex stainless steel alloys compared with austenite stainless 316L in turning (Source Outokumpu) [3]

### 2.2.4 Correlating machinability

In the past two decades, researchers have stipulated what influences the machinability of duplex stainless steel alloys. Carlsson [49] suggested increasing the volume fraction of austenite causes a reduction in machinability in duplex, due to austenite being more ductile and ‘stealing’ carbon from neighbouring ferrite. However, no supporting data had been disclosed, and no further research at present has expanded on this theory.

Ostlund [50] and Paro [51] presented work showing a relation between tool life and the pitting resistance equivalent number (PREN). The graph displayed in Figure 2.13 showed there is good correlation between the PREN value and tool life. The lower the PREN value the greater the tool life. However, both studies neglected to elaborate on the mechanics as to why these two parameters are linked.
There is an obvious connection between resulting microstructure governed by the additions of alloying elements in the PREN equation, Equation 2.4. Studies have shown a more machinable microstructure can be designed and created based on alloying. Jeon [52] and Renaudot [53] recently showed additions of sulphur content could improve the machinability of a duplex microstructure. The formation of manganese sulphides improved chip breaking and lubrication at the chip-tool interface. However, the resistance to pitting corrosion decreases with the increase of sulphur content [53].

2.2.5 Tool wear modes

Tool wear, including the rate of tool wear, is an important aspect of machining since it heavily influences manufacturing costs and product dimensional tolerance accuracy. Tool wear is mainly concentrated in two regions in a cutting tool, the flank face in relation to ‘flank wear’ and the rake face in relation to ‘crater wear’, shown in Figure 2.14. The performance of a cutting tool is directly related to the condition of the flank and rake surfaces. Extensive flank or crater wear will result in either tool failure or poor cutting and dimensional inaccuracies caused sometimes by tool drifts [37]. The type of wear which occurs will often fall into two categories, abrasive or non-abrasive, these are known as wear modes. There can be any number of wear modes actively
occurring in triggering flank wear or crater wear. This is dependent on the material and machining environment.

![Figure 2.14 Regions of tool wear in orthogonal cutting](image)

Abrasional wear will naturally occur during sliding contact between two surfaces. The material removal is often in the form of scratching. Standards [54] define abrasion wear as wear due to hard particles or protuberances “teeth-like” moving across or forced along a solid surface. High-stress abrasion occurs when stress is sufficiently high enough to cause fracture, illustrated in Figure 2.15(a). This leads to scratching and indentations appearing along interacting surfaces.

Adhesion wear as its name suggests, is associated with bonding. It is defined as localised bonding between contacting surfaces, causing either material loss or transfer along interacting surfaces [54], highlighted in Figure 2.15(b). The manner of material removal is often described as a ‘plucking’ type action [40, 55] with the localised bond being strong enough to generate a fracture in the weaker surface.

Figure 2.16 shows an example of both wear modes abrasion and adhesion wear appearing on the rake face of a solid carbide drill bit, from drilling duplex stainless steel alloy 2205. High stress abrasion typically occurs along the rake face in cutting tools. This is caused by sliding contact between the chip and tool rake face during metal cutting. Also, since the tool is generally harder than the work material, abrasion typically leads to metal transfer of the work material onto the tool, as shown in Figure 2.16(b). This figure also highlights frittering or flaking, which is damage to the tool coating caused by abrasion. The arrow points for adhesion highlight cavity region
caused by adhesion wear and regions along the cutting edge where work material i.e. BUE is presently adhered.

![Schematic diagram of wear modes](image)

**Figure 2.15** Schematic diagram of wear modes (a) abrasion wear (b) adhesion wear in two-body system

In the machining of duplex stainless steel alloys, Paro et. al, [51] reported adhesion wear triggered by BUE, was the dominant wear mechanism in their machinability study of cast duplex. In a related duplex machinability study, Pellegrini et. al, [4] indicated tool life and cutting speed was being limited to a greater extent by the presence of BUE. As previously mentioned in the introduction of this chapter, machinability and tool wear tests are limited in their observations by only being able to detect what the related machining issues are. Such tests generally cannot detail the mechanics involved in the evolution of these issues such as BUE. More specialised tests are necessary to determine the triggering mechanisms involved.

![Mixed wear modes](image)

**Figure 2.16** Mixed wear modes adhesion and abrasion wear located on (a) rake face of solid carbide drill tool (b) magnified image
2.2.6 The built-up edge structure

Built-up edge formation (BUE) is an undesirable phenomenon that occurs when the workpiece material bonds onto the cutting edge. The presence of BUE (built-up edge) can cause detrimental and rapid breakdown of the cutting edge, and it is also known to cause tool insert failure [37]. Other problems related to BUE stem from its negative effect on tool geometry causing poor surface finish and machining tolerances.

The occurrence of BUE while machining duplex stainless steel alloys is significant. The high tendency is not only noticed in drilling but also in other machining operations. Studies by Carlborg [56], Caprio [57] and more recently Królczyk [58] reported BUE to be an issue in turning. Most machining related studies have made comment on BUE when machining duplex. Even Bouzid [59] reported BUE to be an issue in milling duplex, leading to increased surface roughness. These studies give testimony to duplex stainless steel alloy’s natural sticking ability. Most machining related studies mainly focus on the machinability aspect of duplex. At present, there are no machining studies which detail the cause or triggering mechanisms to the presence of BUE when machining duplex.

2.2.6.1 Formation of the built-up edge

The formation of BUE is a complex mechanism which would be difficult to describe using a single model due to its complexity. Tribological contact conditions at the tool-work interface will play a role to BUE generation. While observing BUE occurring in low carbon steel, Trent [60] found BUE to be a sequence of strain-hardened layers. While observing seizure at the chip-tool interface, Trent made comment on BUE as a dynamic mass of continuous accretion of strain hardened layers and fractures, “stick-slip” process. This sliding and seizure would occur simultaneously at different positions on the tool-work interface, as mentioned by Trent [60].

The BUE is found to develop from the formation of micro-cracks. Iwata [61] observed the formation of BUE using video capture during the turning of low carbon steel inside a scanning electron microscope (SEM) chamber. In his findings, Iwata reported two micro-cracks play a significant role to the formation of BUE. The first crack appears along the flank face region, and subsequent growth extends in the direction of maximum shear i.e. the primary shear zone, shown in Figure 2.17. A developed stagnation zone or dead material precedes the formation of the first crack. This is
generally located at the tip region of the tool. The second crack forms along the rake face of the chip. The two cracks propagate and subsequently join, forming a fully developed BUE structure. Figure 2.18 shows a fully developed BUE structure in a chip root sample from an earlier investigation by Williams [62], who also reported micro-cracks triggering the formation of BUE. Williams suggested the formation of the second crack was due to limited ductility in the second phase, in the case of machining two-phase materials.

Figure 2.17. Progression of micro-cracks triggering the formation of BUE. After Iwata [61]

A fully developed BUE eventually breaks away from the chip to either remain embedded on the machined surface as fragments or adhere to the cutting tool material, depending on bonding strength between the workpiece and the tool material [37].

Figure 2.18 Formation of a BUE structure, material spheroidized steel [62]
2.2.6.2 Identifying triggering mechanisms

An important field of research is the identification of the mechanisms that trigger the formation of BUE. Some researchers rely on empirical studies and interpret data on parameters that trigger BUE, while others observe chip formation and gather information on the BUE. Much of the known parameters and tool development used in the industry to avoid BUE formation e.g. cutting speed, feed, tool geometry, fluids, etc. are in recognition of the research work in this field. Having said this, there remains discrepancies with some well-established mechanisms stated in literature that do not fully explain the BUE behaviour occurring with the machining of duplex stainless steel alloys. The work-hardening mechanism is the first example.

There is agreement in literature regarding work-hardening as an important mechanism to the formation of BUE. Takeyama [63] suggested the two correlate with each other, the higher the degree of work-hardening the greater the tendency to BUE. Much of this belief arises from hardness measurements made on chip root samples. Zlatin and Merchant [64] found a 300% bulk hardness increase in the built-up region compared to the original hardness in the workpiece. However, there is contradiction with the work-hardening and BUE tendency when machining duplex. In a preliminary study [9], both duplex stainless steel alloys SAF 2205 and SAF 2507 were found to machine with a higher tendency to the formation of BUE compared to stainless steel Austenite 316L. However, austenite stainless steels have higher work-hardening rates compared to duplex [13, 51], which is likely due to having a higher metastability [13]. Williams [65] observed similar inconsistency in his analysis of machining of pure and single-phase metals. Williams noticed the pure and single phase metals in his study, machined without a BUE despite these materials possessing high work-hardening rates.

Another discrepancy with BUE in duplex stainless steel alloys is that it occurs even at high cutting speeds. Pellegrini [4] reported this during turning duplex 2205 and finding BUE marks even at high speeds 160-180m/min. This is counter intuitive to literature as BUE is knowingly avoided under the increase of cutting speeds [27, 66].
BUE has been described as temperature dependant and subsequently dependant on cutting speed [27]. They reportedly thrive at low cutting speeds, due to the created low temperature environment combined with high pressure, which leads to pressure welding of the chip onto the tool. Another theory suggests the triggering conditions are opposite, that BUE forms as a result of higher cutting temperatures triggering a ductile reaction which causes BUE [67]. In light of both views, the report by Pellegrini [4] in finding BUE in both high and low cutting speeds, contradicts both cases. Ultimately, there is a significant need for clarification in this field of literature.

The discrepancy between BUE theories and reported literature on BUE behaviour in machining duplex, suggests there are other mechanisms contributing to the triggering of BUE in these alloys. Literature provides other possible mechanisms to consider, for example ductility. A more ductile material is considered to be more adhesive [67, 68] and will therefore generate more BUE. Such can be explained by high tendency of BUE found when machining aluminium. Another possible mechanism is material bonding compatibility, as BUE is thought to be a function of the affinity between the workpiece and the tool material [37, 63].

A suggestion was made by Carlborg [56] in a related duplex machinability study, where he concluded the amount of BUE is triggered by the ferrite content in duplex. Carlborg reported higher ferrite content in duplex stainless steel alloys led to increase BUE presence in machining. However, no evidence or mechanism to show how this was occurring was mentioned or has been presented since. In a similar study Williams and Rollason [62] reported two-phase materials promote an additional fracture point occurring along the chip-tool rake face during BUE formation, while single phase materials maintain only one fracture point. Williams and Rollason also commented the fracturing was a result of reduced ductility in the second phase, but could not show how this was occurring.

2.2.7 Summary

This section has highlighted the machinability of the common duplex grades such as 2205 and 2507 are poorer than general austenite 316L stainless grade. It also highlighted BUE is an issue in machining duplex that remains to be addressed. Reiterating the main research question “What are the mechanisms triggering frequent built-up edge (BUE) during the machining of duplex stainless steel
alloys?”. The present literature indicates there is a significant gap in the explanation of BUE occurring in these alloys. Despite the presence of existing theories such as Carlborg’s [56] ferrite hypothesis. These literature works fail to provide any supporting evidence or mechanism detail to collaborate with their theories. Well-established mechanisms to BUE described in text-books such as work-hardening and temperature, fail to fully explain the BUE behaviour in duplex stainless steel alloys. All of these above indicate, that there could exist an unidentified mechanism triggering BUE in duplex stainless steel alloys under machining. The uncovering of such a mechanism would aid in the development and better understanding of how to machine these alloys.
2.3 Finite element modelling

Finite element analysis (FEA) or Finite element method (FEM) is an effective computational tool in providing numerical solutions in analyses that would generally prove more difficult to determine or measure physically. This difficulty can be due to costs or time constraints or even due to the complex nature of the process, such as the case in metal cutting. For over 40 years, since the first applications applied to metal cutting in the early 70’s [35], there has been substantial growth in finite element analysis in machining. Much is aided by development in technology, with increased processing power and reduced computing costs, and also with current market saturation of modelling packages such as ABAQUS and ANSYS.

Despite the growth in FE development and applications, FE analysis for metal cutting has been considered an uncommon application in the machining industry [69]. Much of this transpires due to the physical complexity of the machining process which questions the reliability and verification of finite element results. The continual demand to build an accurate finite element solution for such a high non-linearity process remains one of the main challenges today.

This section provides a review of the finite element analyses and how it is applied to metal cutting, outlining the strengths and current limitations in present cutting models.

![Figure 2.19 Basic concept of a finite element representing a continuum thin-wall plate](image)
2.3.1 Describing a mechanical problem

The main concept behind all finite element analyses is to replace a continuum solid body of material or fluid or a thin-wall structure with an assembly of finite elements. The continuum is represented by a connecting number of set points, referred as nodes. An example of this is illustrated in Figure 2.19 showing a thin-wall plate of thickness $t$ that is being represented by a triangular set of finite elements.

Rather than determining an exact solution of the continuum, these elements such as highlighted element ‘e’ allow numerical solutions to be calculated by interpolation, based on values and quantities such as displacement $u$ and force $F$ occurring in the surrounding nodal points $i, j, k$. These calculations are described in detail in Appendix A.1 providing an example calculation of determining strain $\varepsilon$ at element e using variables at the highlighted node points.

2.3.2 Finite element approaches to metal cutting

Simulating metal cutting can be thought of as modelling the plastic flow of material as it transitions from the workpiece region into the chip. There are two well-established approaches to modelling metal cutting, they are Eulerian and the Lagrangian method. A schematic representation displayed in Figure 2.20 illustrates each methods different approach to modelling chip formation.

2.3.2.1 Eulerian method

In the Eulerian method, the finite elements remain as fixed points in space where material is allowed to flow through either into the chip region or across to the machined workpiece section. Originally derived from fluid mechanics, the Eulerian method has the advantage of element shape remaining constant. This reduces computing time as computation which is dependent on the element shape therefore the $B$-matrix in Equation A1.3b only needs to be calculated once, refer to Appendix A.1. In the Eulerian method, attention is placed in tracking the viewpoint between each element during the same time e.g. tracking velocity $\dot{u}$ of elements ① and ② during time $t$ in Figure 2.20.
CHAPTER TWO

Figure 2.20 Eulerian and Lagrangian approach to chip formation, after Childs [35]

The Eulerian approach is useful for studies observing steady state parameters as boundary conditions create a controlled volume flow rate environment while computations may progress over extensive time while remaining stable since shape of the elements do not change. A main disadvantage of the Eulerian method is that the chip formation is an input factor and is required to be pre-determined. But in a machining problem simulation, determining the chip size and the formation is part of the problem needing to be solved. Another disadvantage to the Eulerian method is material properties remain constant and does not allow for changes between elements during flow. This case would be ideal for fluid flow however plastic deformation of a solid body would result in changes in the material properties due to factors such as work-hardening where the tensile strength of the material is affected by increasing strain.

2.3.2.2 Lagrangian method and ALE method

The Lagrangian method allows for changes in the material properties during flow of the material, as the state of the material property is fixed within each element and can change with the state of the element. In the case of the Lagrangian method, the output variables of the elements are viewed individually and attention is drawn to how these individual elements vary with time, between $t_1$ and $t_2$, see Figure 2.20. The main contrast between the Lagrangian and the Eulerian method is rather than having elements fixed in space, they are fixed to the flowing material and allowed to deform along with the material. The Lagrangian method does require added computations as the $B$-matrix in Equation A1.3b requires continual updating, due to both geometric and material non-linearities. An advantage the Lagrangian method holds is no prior assumptions or inputs are required for the chip geometry. The chip shape will form
accordingly to computations and input variables. This makes the Lagrangian method ideal for chip formation studies. A main disadvantage however, is finite elements become heavily distorted such that it may require mesh regeneration. Also, a node separation criterion needs to be established similar to a self-crack propagating in fracture mechanics simulation. While there are now a wide range of node separation criteria in existence, the main disadvantage is the selected criterion may have an influence on the results.

There is also a third approach to modelling metal cutting which combines the benefits of both the Eulerian and Lagrangian methods, known as the Arbitrary Lagrangian Eulerian (ALE) method. In the ALE method the finite elements are neither fixed to space nor the material. The mesh itself has a motion that is independent of the material. There are many advantages to the setup of the ALE method, the main is that frequent re-meshing and interpolation is avoided as it becomes part of the solution procedure. There is also no requirement for any node separation criterion. The chip formation occurs by continuous plastic of material around the tool [69]. But like the Eulerian method, the ALE method does require some initial input knowledge on the chip geometry shape. In summary, the choice of finite element approach when modelling machining mainly relies on the type of study to be undertaken or simulated.

2.3.3 Material properties

A finite element (FE) model is only as good as its material properties. The only materials used in this thesis were general duplex stainless steel alloys whose material properties are well-established in literature and manufacturer’s data. The finite element work in this thesis did however require strength curves of the phase constituents to be mapped. As there are a number of well-established models regarding the stress-strain relationship, the Ramberg-Osgood model was selected specifically for being an established fit for stainless alloys. First developed in 1943 [70] the Ramberg-Osgood model in its original form is described in Equation 2.11

\[ \epsilon = \frac{\sigma}{E} + K \left( \frac{\sigma}{E} \right)^n \]  

Equation 2.11

Where \( K \) and \( n \) are hardening constants and \( E \) is the elastic modulus.

The Ramberg-Osgood equation describes the full range of the strength curve by representing total strain as a linear and non-linear combination. The first term on the
right-hand side of Equation 2.11 represents the elastic strain and the second term the plastic strain.

\[ \varepsilon = \varepsilon_{el} + \varepsilon_{pl} \]  \hspace{1cm} \text{Equation 2.12}

The \( K \) constant is usually replaced with an \( \alpha \) expression as \( K \) is related to \( \alpha = K \left( \frac{\sigma_y}{E} \right)^{n-1} \). Then the equation becomes

\[ \varepsilon = \frac{\sigma}{E} + \alpha \frac{\sigma_y}{E} \left( \frac{\sigma}{\sigma_y} \right)^n \]  \hspace{1cm} \text{Equation 2.13}

The \( \alpha \frac{\sigma_y}{E} \) term is approximated to \( \approx 0.002 \) as the term represents the proof strain. This is highlighted in Figure 2.21 showing the representation of the yield point. Therefore the reduced common form of the Ramberg-Osgood equation is shown in the following equation.

\[ \varepsilon = \frac{\sigma}{E} + 0.002 \left( \frac{\sigma}{\sigma_y} \right)^n \]  \hspace{1cm} \text{Equation 2.14}

Figure 2.21 Ramberg-Osgood representation of the stress-strain curve

Although originally developed for aluminium, the Ramberg-Osgood model has proven itself with the stainless curves such that most standards tend to use the Ramberg-Osgood expression, Equation 2.14 as a constitutive representation of stainless steel [71].
2.3.4 Material separation

To achieve chip formation or element separation, damage and failure models are commonly applied to model material separation. There are an extended range and combination of models which can be used to model chip formation. A material ‘ductile damage law model’ based on plastic strain was used as part of the FE model in this thesis. A ductile failure model was used previously by Liu to simulate chip formation in 304 stainless steel [72]. Figure 2.22 show the implemented damage law affects the material strength. The curve is a magnified representation of the elastic region. The solid blue line and dashed line outlines the path of the normal material curve ‘undamaged’.

![Stress-strain curve with progressive damage evolution](image)

Figure 2.22 Stress-strain curve with progressive damage evolution

The degradation or damage influence to the material is outlined by the solid red line showing the loss of strength in the material. The true stress $\sigma$ in this region is replaced by an effective stress $\bar{\sigma}$. This is governed by a scalar damage variable $D$ according the following equation.

$$\sigma = (1 - D)\bar{\sigma}$$

Equation 4.4

The plot in Figure 2.22 shows the damage at the yield point $\sigma_y$ is zero i.e. ($D = 0$). Damage initiates after the yield point $\sigma_y$ and equivalent plastic strain $\varepsilon_0$. According to the damage law, in the ductile region, increasing damage ($0 > D > 1$) degrades the material’s load carrying capacity. The material strength continues to decline to the
point of failure $\varepsilon_f$ where the material can no longer sustain loading i.e. when $D = 1$. The damage initiation point and the plastic strain failure point can be selected inputs in the plastic region of the material curves.

### 2.3.5 Recent development in FE metal cutting models

A literature survey revealed there has been no published work on any duplex finite element (FE) cutting model or development of any kind, highlighting a significant gap. Regardless, the last five years has seen on-going advancement in FE metal cutting with developed models continually closing the gap between simulated and experimental data.

Davim [73] reported good correlation in his FE analysis, comparing experimental data with simulations based on a commercial modelling package AdvantEdge. Commercialised software such as AdvantEdge and DEFORM developed specifically for metal cutting, have become widely available and have seen growing involvement with machining research [74-77]. In the simulated 2D cutting of AISI 1045 steel, Davim [73] found a 2.5% plastic strain and 1.6% strain rate error between experimental values in conventional milling, and for high speed machining (HSM), the plastic strain and strain rate error was 1.6% and 6.5%. The standard low errors highlight an achievement in accuracy in modern finite element cutting models.

There has also been development in 3D cutting models. Pittalà’s [78] 3D milling model based on DEFORM 3D commercial coding, reported good correlation between experimental and simulated data analysis. 3D models are not only useful for 3-axis analysis of cutting forces, stress, and temperature etc., they are also useful for the analysis of tool condition, in the simulation of tool wear. Ozel [79] utilised a 3D model to predict wear on four different coated carbide inserts, to analyse wear rates between different coatings during turning of Titanium. Ozel reported his model could predict which coated inserts would produce the least wear rate, which reported good agreement with experimental tool wear observation.

Recent models [65, 80] have also been able to predict the formation of a stagnation zone. Arrazola [65] and Muñoz [80] developed similar cutting models based on Arbitrary Lagrangian Eulerian (ALE) method, for different materials using ABAQUS software. Both models validated a stagnation zone was occurring through velocity
contour plots, as shown in Figure 2.23. Arrazola’s [65] model predicted the size of a stagnation zone was controlled by the shear stress value, where reduced shear stress triggered less of a stagnation region. Muñoz’s [80] model found the stagnation zone area was increasing with increasing tool nose radius. A disadvantage in Arrazola and Muñoz’s models is they are both ALE type models and are therefore developed from a preformed condition. Therefore, an FE analysis to predict or observe how a stagnation zone would form, could be limited.

Figure 2.23 An ALE model developed by Arrozola [65] simulating a stagnation zone under two shear stress $\tau_p$ conditions
CHAPTER THREE

3.0 Machinability and chip formation study

This experimental chapter presents two observational studies. The first, a machinability drilling study which was conducted to observe the natural machining behaviour of duplex stainless steel alloys. Tool wear, cutting forces and surface roughness of two common duplex grades was compared and analysed under similar machining conditions. The second study took a more detailed observation into the chip formation. A quick-stop device was used in a turning operation. Observing chip formation is generally difficult since metal cutting usually occurs at high speeds while the material is plastically deforming at an instantaneous rate. The quick-stop method is a proven and resourceful tool that provides a unique window of observation in studying the chip form. The method replicates the act of pausing the cutting process by rapidly accelerating (near instantaneously) the tool away from the workpiece while cutting. The result is an un-altered chip that remains attached to the material workpiece, referred in literature as a ‘frozen chip’ or ‘chip-root’.

The main drive behind all the experimental work in this thesis was to collect data that would aid to uncover the triggering mechanisms to long-established issues such as BUE. The aims of the two studies presented in this chapter was to collect data consists of the following:

- assess the machinability of duplex stainless steel alloys and observe what the practical issues at the machinability limits are;
- observing the behaviour of the two phases, on how γ-austenite and α-ferrite respond to plastic shear during transitioning into the chip region;
- monitoring of the stagnation zone, as it is an area prone to the formation of BUE; and
- to collect data based on the work-hardening behaviour, since is work-hardening has been indicated as a mechanism to initiation of BUE, as mentioned in the literature research.
3.1 Experimental design

The duplex grades used in all physical experiments in this thesis were two common grades, 2205 grade and higher strength 2507 ‘Super duplex’. These alloys were selected for their variability in strength and being the most common grades used in the industry [3]. Their chemical composition and mechanical properties are listed in Table 3.1. Austenite 316L was employed as a benchmark material in the machinability study. Materials were used in Ø20mm round-bar form and machined in As-Supplied condition. These were all manufactured by hot-rolling.

Table 3.1 – Chemical composition and mechanical properties of duplex stainless steel alloys

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Fe</th>
<th>UTS</th>
<th>σ_y</th>
<th>HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>SAF 2507</td>
<td>0.02</td>
<td>0.74</td>
<td>0.23</td>
<td>0.01</td>
<td>0.02</td>
<td>6.77</td>
<td>25.1</td>
<td>3.68</td>
<td>Bal</td>
<td>866</td>
<td>570</td>
<td>285</td>
</tr>
<tr>
<td>SAF 2205</td>
<td>0.02</td>
<td>0.8</td>
<td>0.4</td>
<td>0.01</td>
<td>0.02</td>
<td>5.2</td>
<td>22.4</td>
<td>3.05</td>
<td>Bal</td>
<td>777</td>
<td>556</td>
<td>279</td>
</tr>
<tr>
<td>AISI 316L</td>
<td>0.03</td>
<td>1.5</td>
<td>0.4</td>
<td>0.03</td>
<td>0.03</td>
<td>10.5</td>
<td>17</td>
<td>2.1</td>
<td>Bal</td>
<td>640</td>
<td>326</td>
<td>254</td>
</tr>
</tbody>
</table>

3.1.1 Machinability test setup

Drilling experiments were performed on a Haas (XYZ) Super VF-3 CNC vertical machining centre using Ø12 mm diameter SECO SD203A-M geometry drills. These were TiAIN + TiN coated solid carbide twist drills with internal coolant supply. General purpose emulsion type mineral oil based cutting fluid with a dilution concentration of 5% was supplied at a continuous flow rate of 9.9 l/min. Machining parameters for all the drilling trials comprised a cutting speed of 60 m/min; a penetration rate of 0.15 mm/rev; and a hole depth of 30 mm continuous. A Kistler 9257b cutting force dynamometer coupled with a Kistler 16-channel charge amplifier was used to measure the reaction forces as well as the torque during drilling. Readings were data-logged on computer using ‘Dynaware’ cutting force software. The experimental setup is depicted in Figure 3.1. The workpiece was mounted in a special fixture that was located rigidly on the centre of the dynamometer platform – i.e. equi-spaced between the four quartz crystals. Tool wear on the flank face was measured at regular intervals using an optical microscope. Drilling continued until a tool wear value $VB_{max} = 0.15$mm was reached or until tool failure. The surface roughness of

---

1 UTS and $\sigma_y$ units displayed in MPa, Vickers hardness measured with 100g load HV100g
machined surfaces of each workpiece was recorded using a stylus measurement device, namely a Talysurf Intra Series 50. The experimental setup apparatus for tool wear measurement and surface roughness is shown in Figure 3.2

Figure 3.1 (a) Drilling experimental setup (b) Schematic diagram of experimental setup

Figure 3.2 (a) Micrograph setup for tool wear observation and (b) stylus profilometer
3.1.2 Quick-stop experimental setup

An explosive type quick-stop device was mounted to a Colchester CNC 2000L turning lathe. The main components to the quick-stop device comprised of a tool holder fitted inside a secure housing, held in position by a pivoting rod and shearing pin designed to fracture at the moment of an applied downward force, shown in Figure 3.3. To effectively replicate the act of freezing the cutting process, the tool must accelerate to a velocity greater than cutting velocity i.e. $V_{\text{tool}} > V_{\text{workpiece}}$. The driving mechanism was therefore provided by a captive bolt stunner gun ‘Cash Special’. Within a secure housing, the gun fired a release bolt using 22” calibre gunpowder casings. Goldberg [81] previously analysed the performance of the same quick-stop device. Goldberg verified by calculation that the bolt speed was significantly faster than the spindle speed utilised in his quick-stop tests, at $V=146\text{m/min}$.

![Figure 3.3 (a) Explosive quick-stop device setup (b) schematic diagram of tool holder and workpiece](image)

Griffiths [82] assessed the use of explosive type quick-stop devices and reported them having a normal upper limit, operating to a maximum cutting velocity of 305m/min due to issues of deflection. The cutting trials conducted in this study were operated below the reported upper-limit by Griffiths. Table 3.2 displays the cutting parameters used in this study.

<table>
<thead>
<tr>
<th>Lathe Machine</th>
<th>Cutting speeds</th>
<th>Feed</th>
<th>Conditions</th>
</tr>
</thead>
<tbody>
<tr>
<td>Colchester CNC-2000L</td>
<td>94m/min</td>
<td>0.15mm/rev</td>
<td>Dry</td>
</tr>
<tr>
<td></td>
<td>65m/min</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
The insert type mounted on the tool holder was solid carbide ‘WNMG-TF’ type. These were 80° trigon shaped with 0° clearance. Inserts were coated with titanium nitride and aluminium oxide coating TiCN + αAl2O3 + TiN.

### 3.1.2.1 SEM sample preparation

The produced chip root sample was cut away from the workpiece using a wet cutting wheel at low RPM. Chips were then hot mounted in PolyFast resin, and wet grinding was applied to reach the chip root layer, Figure 3.4.

![Figure 3.4 Frozen sample preparation (a) attached to workpiece (b) sectioned and (c) hot mounted in PolyFast resin](image)

For scanning electron microscope (SEM) preparation, all samples were prepared by standard polishing procedures, using MD-Mol pads down to 1μm before finishing with OPS type MD-Chem pad. To further reveal phase microstructure under SEM, samples were etched using Beraha’s tint etchant, consisting 85ml of water, 15ml HCl, 1g K2S2O5.

SEM scans were taken on a FEI (Philips) XL30 S-FEG high resolution scanning electron microscope. Images were acquired under high current, 10mm working distance, 60μm aperture operating at 20kV accelerating voltage.

### 3.1.2.2 Hardness tests

Microhardness measurements were performed on mounted chip root quick-stop samples using a Knoop indenter. Samples were polished identical to SEM preparation, but tests were performed on un-etched surfaces. This allowed for grain boundaries to be visible while maintaining a clear contrast between indentation and microstructure.
for accurate measurements. A load weight of 300g was chosen in order for the size of indentation to be large enough to cross multiple grain boundaries as shown in Figure 3.5. The suited indentation size allowed for hardness readings to be less affected by an individual phase. Readings were taken at various locations along the frozen chip sample.

![Figure 3.5 Microhardness Knoop indentations on frozen chip root sample](image)

### 3.2 Machinability results

Different areas, such as, flank face, rake face, chisel edge and all over the flute, of the carbide drill tools were examined for wear/damage. Figure 3.6 displays the location of these areas on the solid carbide twist drills used in the study. The amount of wear of the tool was dependant on the degree of contact and interaction with the workpiece material. The most important wear of concern in this study was flank wear, as the degree of flank wear was being measured for the tool wear $V_{B_{max}}$ criterion.

![Figure 3.6 Solid carbide twist drill used in trials, type SD203A-M geometry drills](image)
3.2.1 Flank wear

Figure 3.7 shows the progression of maximum flank wear with the number of drilled holes for the three workpieces. The rate of flank wear was very high for the duplex 2507. It reached the set flank wear criterion after drilling 26 holes. The rate of flank wear development in drilling duplex 2205 was less than that of duplex 2507, from at the start until 40th hole – although the wear stabilised after that. Though the drill tool wear remained below the set criterion, the tool failed after drilling 69 holes. Severe damage in the flute was noted approximately 10 mm above the drill tip as shown in Figure 3.8. The resulted damage may be from higher cutting loads and poor chip evacuation. The rate of tool wear during machining of the austenite 316L was very low initially (until 15th hole) then it stabilised (until 35th hole). Thereafter, the wear rate increased (until 55th hole) and then stabilised again. In this case, the drill tool succeeded in maintaining acceptable flank wear and without tool failure in 75 holes of drilling.

![Figure 3.7 Maximum flank wear during drilling of duplex SAF 2507, SAF 2205 and Austenite 316L](image)

Figure 3.8 Flute damage after drilling 69 holes in SAF 2205
The micrographs of wear progression on the flank face are presented in Figure 3.9. After drilling 6 holes, frittering and flaking are visible under microscope for all the three materials. These are mainly caused by resulting abrasion wear generated from sliding between the cutting lip and the chip material flow. The overall damage caused by abrasion wear was seen as minor, despite its wide appearance to surround the cutting edge perimeter.

With the progression of drilling, the presence of BUE on the flank face became frequent for both duplex stainless steel alloys, shown around the mid-stage of trial images in Figure 3.9. In this figure, the flank wear for drilling of duplex 2507 is presented only at 6\textsuperscript{th}, 15\textsuperscript{th} and 26\textsuperscript{th} holes as the cutting tool damaged after drilling 26\textsuperscript{th} hole. Thus, the cutting tool performance was worst for drilling duplex 2507 and is not comparable to that of duplex SAF 2205 and austenite 316L. The presence of BUE triggers the adhesion wear, causing damage to the flank surface. In monitoring the flank wear of the duplex drills, it was found that the higher $VB_{\text{max}}$ were directly caused by adhesion wear. The $VB_{\text{max}}$ is highlighted on the flank face in the images in Figure 3.9 Micrographs of the flank face at different stages during drilling different materials.
3.9. Austenite 316L responded with less severity to BUE in comparison, resulting in minimal flank wear.

![Figure 3.10 BUE formation on the flank face of a drill tool, triggering adhesion wear in drilling](image)

(a) SAF 2205 and (b) SAF 2507

The micrographs in Figure 3.10 illustrate adhesion wear triggered by the formation of BUE that was observed during trials. Bonding occurs between the hardened built-up material layer and the drill cutting lip (carbide material). At a certain stage with further drilling, the shearing action removes the BUE layer while simultaneously detaching a region of the cutting lip by a plucking type action. This leaves behind cavities in the drill tool.

3.2.2 Chisel edge wear

The condition of the chisel edge is also important, since it is responsible for surface penetration and consumes over 50% of the total thrust load [83]. A damaged chisel edge can result to cutting load increase and lateral vibration at the entrance [84]. Micrographs in Figure 3.11 demonstrate that the chisel edges of tools for duplex stainless steel alloys were more affected by wear. Crater type cavities corresponding to fatigue wear appeared on both duplex drills, the largest appearing for duplex 2507, measured at 0.24 mm. SEM imaging of the fracture surface shown in Figure 3.12 shows visible ‘beachmark’ propagation lines, typically found in a fatigue fracture. For duplex 2205, the tool appeared to have suffered a fracture, first appearing after drilling 30 holes. After fracture, the chisel edge cavity was observed at regular intervals.
Figure 3.11 Wear on chisel edge of drill tools after cutting materials (a) 2507 after 26 holes (b) 2205 after 69 holes (c) 316L after 75 holes.

Figure 3.12 SEM of fracture surface along chisel edge showing ‘beachmark’ propagation lines

Figure 3.13 BUE layer imbedded in chisel edge cavity of 2205 drill tool after 46 holes of drilling
It was found that the fractured cavity promoted formation of BUE as shown in Figure 3.13. This resulted in further growth in the cavity by adhesion, as the BUE layer was removed from further cutting. The chisel edge on the austenite 316L tool mostly remained unaffected by abrasion and adhesion wear modes, at most, causing chipping and flaking on the chisel edge. Surprisingly, the thrust load experienced during the surface penetration by tool chisel edge was higher in the case of austenite 316L compared to both duplex stainless steel alloys, as highlighted in the cutting force plot in Figure 3.14.

### 3.2.3 Cutting forces

Mean thrust forces for different workpiece materials are presented in Figure 3.15a. Higher force is noted for duplex 2507 and indicates that this material is harder to cut, sequentially followed by austenite 316L and duplex 2205. The order did not correlate to material hardness or mechanical strength values as it would suggest. Jiang [45] reported a similar force comparison in the grinding of the same materials. Duplex 2205 had a lower grinding force compared to austenite 316L. The effect of tool wear on cutting force is shown by the increasing trend in average thrust force, particularly for duplex 2205 and austenite 316L.
Figure 3.15 Thrust forces for different materials (a) average thrust force (b) thrust force profiles for a single hole

Figure 3.16 Drilling torque for different materials (a) average drilling torque (b) drilling torque profiles for a single hole

Figure 3.15b shows the thrust force profiles for a single hole. The initial high peak displayed by austenite 316L shows stronger resistance to surface penetration, as mentioned in the previous section regarding chisel edge wear.

After surface penetration, when the main cutting lips initiate cutting, its thrust force reduces and remains constant, achieving steady state. Duplex 2507 displayed a higher steady state range after surface penetration, and remaining unchanged until the drilling depth was reached. In contrast, the thrust profile for duplex 2205 is shown to gradually decrease while in steady state cutting.

Measuring the torque provided a general indication of power consumption, since power in drilling is the product of torque and the rotational speed of the drill. Thus average torque and comparison of torque profiles in Figure 3.16 shows both duplex stainless steel alloys require more cutting power to drill, compared to austenite 316L. The sudden increase of mean torque displayed by duplex 2205 at the closing experimental stages, illustrated the severe effect of flute wear damage experienced by the tool.
3.2.4 Surface roughness

Surface profile measurements of the 25th machined hole for the three materials are presented in Figure 3.17. The profile of surface for austenite 316L is smoother compared to duplex stainless steel alloys. The poorest surface profile was obtained for duplex 2507. The surface roughness ($R_a$) for duplex 2507, duplex 2205 and austenite 316L are 2.06, 1.49 and 1.13 μm respectively. All these were contributed by higher tool wear, higher built-up-edge and worse chip removal process for duplex 2507.

Figure 3.17 plots the variation of surface roughness with the number of drilled holes for all the three materials. The spread of average surface roughness ($R_a$) ranged between 0.5 to 3μm. Austenite 316L maintained lower fluctuations of $R_a$ compared to the other materials. This represents a stable cutting process for austenite 316L. In contrast, for the case of duplex 2507 and 2205, a wider range of fluctuations of $R_a$ are noted. This shows less cutting stability.

![Figure 3.17 Surface profile of the 25th drilled hole for different materials](image)

Figure 3.17 Surface profile of the 25th drilled hole for different materials
The combination of tool wear and BUE is the most apparent cause for the fluctuation in $R_a$. It significantly influences the tolerance and precision machining of duplex materials. The moving average surface roughness plotted against number of drilled holes is shown in Figure 3.18b. It shows that the trend of $R_a$ for both duplex stainless steel alloys is increasing with the number of drilled holes. The surface textures produced from machining duplex appeared rougher by comparison, and weighed heavily on the condition of the drill tool.
3.3 Chip formation results

Long serrated chips were produced while turning both duplex stainless steel alloys SAF 2205 and 2507. The raw machining chips of all turning trials can be viewed in Appendix A.2. Figure 3.19 displays SEM images of a sectioned 2205 chip root sample interrupted at 94m/min cutting velocity, feed-rate 0.15mm/rev and 2mm undeformed chip thickness.

Figure 3.19 Scanning Electron Microscope (SEM) images of quick-stop specimen SAF 2205 frozen at speed 94 m/min, feed 0.15 mm/rev, undeformed chip thickness 2.0 mm, magnified at various locations α-ferrite, γ-austenite phase (a) overview of chip sample (b) primary shear plane & (c) secondary shear plane (d) stagnation zone with BUE developing at tip of cutting tool.
Two shear zones are clearly visible in the overview image Figure 3.19(a). The arrows in Figure 3.19(a) indicate the path and direction of the workpiece material flowing into the chip, through the primary and secondary shear zones. As both austenite and ferrite phases approach these entry points, they exhibit rapid deformation due to high strain. As a result, highly elongated grains develop, that skew in the direction of plastic flow, as shown in Figures 3.19(b) and (c). The highly deformed microstructure would also be an indication that work-hardening has occurred during this transition. The flow pattern of the material is typical of an orthogonal model.

3.3.1 Stagnation zone

The stagnation zone located at the tip point of the tool region is a common area where the material can remain stationary and does not experience plastic flow for a certain period. Figure 3.19(d) shows the stagnation zone for SAF 2205 at high magnification. The magnified images revealed the following.

- There is a dominant build-up of ferrite in this region. Although there appears to be visible traces of austenite, the initial built-up layer shows to be mostly comprised of ferrite. In the outer layer region, highly elongated austenite grains appear to be negating away from this region. This type of banding was also observed in the stagnation zone of a 2507 chip root sample produced with the same parameters, see Figure 3.20.

- Micro-cracking was found developing in the stagnation zone, highlighted in Figure 3.19(d). There were two types of micro cracking detected in this region. The first (i) intergranular cracking, that appears restricted mainly inside the ferrite phase boundaries. Cracking growth shows to border around the ferritic grains, but does not cross the interphase boundaries. Further higher resolution scans such as TEM imaging, would uncover the significance and influence of this type of micro-crack. Secondly (ii) transgranular cracking, this appeared more dominant in this region. Cracking growth extended beyond the ferrite phase and growth even showed to extend out to the chip-tool interface. It should be noted, there is a possibility this cracking may have also been caused by shock caused by tool suddenly accelerating away from the chip-tool interface.
3.3.2 Work-hardening behaviour

Figure 3.21 shows the measured hardness profiles of duplex 2205 and 2507 chip root samples, each at two different cutting velocities 94 m/min and 65 m/min. Greater hardness was evident in the chip region for all cases. The difference in hardness increase between the two regions ranged between 32 – 44% on average, shown in the bar graph of Figure 3.22. Johansson [85] reported work-hardening occurs greater in the austenite phase, and occurs due to planar dislocation movements causing pileup at the twin and grain boundaries in austenite. The bar graph in Figure 3.22 also showed 2205 displayed a higher sensitivity to work-hardening, indicated by the larger difference in hardness between the two speeds 94-65 m/min.

The regions of higher and lower hardness are separated by the primary shear plane. Plotting the hardness as a function of the distance to the primary shear plane revealed a general correlation as shown in Figure 3.23. It illustrates the location of a general transition zone behind the shear plane in the workpiece region.
Figure 3.21 Knoop microhardness HK300g measurements on quick-stop samples (a) SAF 2205 & (b) 2507 at speed 94 m/min, (c) SAF 2205 & (d) 2507 at speed 65 m/min

Figure 3.22 Increase in micro-hardness between material and chip region

Hardness distribution in this zone indicates that the hardness of workpiece material increases as it moves into the chip region through the primary shear plane. Once the workpiece has passed into the chip region, its hardness has matched that of the chip region. Previous studies have shown this correlation to be common in other metals [86, 87].
Figure 3.23 Correlation of hardness and distance from the primary shear plane for SAF 2205 quick-stop sample frozen at speed 94 m/min

3.4 Discussion

The literature survey [4, 25, 51, 56, 57, 59, 88] has mentioned the significant ties between the machining of duplex stainless steel alloys and the occurrence of BUE. Based on the present (chapter 3) machinability study, the machining data supports this relationship - indicating frequent BUE does occur while machining duplex stainless steel alloys.

Literature had also established BUE is caused by material work-hardening and the natural high-tendencies for adhesion. M'Saoubi [89] reported single phase materials including austenite 316L undergo higher concentrations of work-hardening compared to 2205 duplex in machining. This would suggest austenite 316L is exhibiting a higher rate of BUE compared to duplex, which was not the case. The hardness results in the chip formation study also showed disagreement with the work-hardening correlation. Duplex 2205 was found to undergo higher work-hardening compared to high strength 2507 duplex. Yet during drilling trials, duplex 2507 was shown to machine with a higher response to BUE compared to 2507. What the above statements indicate is, there should exist some other mechanism that is not work-hardening related, that is triggering a frequent BUE response when machining duplex stainless steels.
As mentioned previously in the literature survey in Chapter 2.2.6.2, researchers such as Carlborg [56] and Williams [62] hypothesised other possible mechanisms such as the amount of ferrite in duplex [56] or an increasing amount of secondary phase [62] that is triggering BUE. Yet these two hypothesis still remain to be validated. The ferrite build-up found in the stagnation zone in the present chip formation study. The observational data suggests this newly found behaviour could be a possible candidate responsible for the triggering of BUE. The evidence includes, the ferrite build-up was detected in a stagnation region of an initial built-up layer, combined with the detection of micro-cracking in the ferrite build-up, when micro-cracking is a well-known trigger to BUE. If correct, the ferrite build-up mechanism would lay support towards Carlborg’s [56] hypothesis with the ferrite content in duplex.

In terms of the adhesion behaviour of duplex, machinability tests revealed the adhesion properties of the BUE produced while drilling, was sufficient to adhere to the tool surface despite the tool containing anti-adhesion coating. Titanium aluminium nitride (TiAlN) is known to produce a protective aluminium oxide film (Al₂O₃) on the surface upon heating [90].

3.5 Summary

The aim of this chapter was to observe the machining behaviour of duplex stainless steel alloys. Machinability trials found these alloys to have a frequent BUE issue, more frequent than austenite 316L. This BUE issue triggered adhesion wear and was ultimately found responsible for a poorer machined surface finish.

The SEM images of frozen chip-root samples revealed what appeared as a build-up of α-ferrite collecting at the stagnation zone. The images tend to perceive austenite as flowing away from the stagnation zone in the advancement of the cutting tool. This observation could hold major significance as the build-up of ferrite could be a triggering mechanism to the formation of BUE. However, there is a present degree of uncertainty with the application of chemical etching to reveal microstructure, for phase identification in the stagnation zone as shown in Figure 3.24.

The results from the chip formation study recognised that a more comprehensive investigation was required at the stagnation zone, to acquire definitive data in the highly deformed stagnation region.
Figure 3.24 Optical micrograph of chip root samples at the stagnation region after etching.
4.0 Phase characterisation of the stagnation zone

The chip formation study in chapter 3 detected a build-up of the α-ferrite phase collecting at the stagnation zone that could be relevant to the mechanism of BUE formation. The work presented in this chapter is a continuation of the chip formation study with the aim of uncovering the mechanisms surrounding the plastic flow of the microstructure in the stagnation zone. The objective also includes obtaining a definitive identification of the phases present in this region.

A degree of uncertainty is present when dealing with metallographic etching techniques to reveal microstructure, as used in the previous study on frozen chip-root samples. Two main difficulties are presented in obtaining the required data to meet the objectives. The first is that the sample size, the stagnation zone is very small i.e. in the order of 10-100μm. Secondly, the stagnation zone consists of highly deformed structures. Because of these requirements, attention was drawn to the electron backscatter diffraction (EBSD) technique as a suitable method for obtaining the required data.

The EBSD technique has been utilised in many studies relating chip formation. M’Saoubi [91] observed the deformation behaviour of chip root samples of various alloys. Wroński [92] used EBSD to investigate the plastic behaviour of the individual phases in a duplex alloy during a tensile test. The EBSD method sends foreshattered
or backscattered electrons that have reflected from a heavily tilted sample specimen onto a phosphor screen, shown in Figure 4.1.

![Image of Kikuchi bands]

**Figure 4.2 Phase mapping of Kikuchi bands from ferrite phase in duplex structure using Channel 5 HKL software**

These generate electron backscatter diffraction patterns (EBSP) also known as Kikuchi bands based on the atomic arrangement of the sample microstructure. These bands are mapped using computing software and matched with existing databases to identify the microstructure. Figure 4.2 shows the Kikuchi pattern mapping of the ferrite phase in a duplex microstructure.

### 4.1 Experimental design

Quickstop samples were produced on a Hafco Metalmaster CL-38 centre lathe using the same explosive type quick-stop device in the previous chapter, refer to Chapter 3.1.2 for the experimental quick-stop procedure. The machining parameters are listed in Table 4.1. Solid carbide inserts type WNMG-TF were used in the experiments.

<table>
<thead>
<tr>
<th>Lathe Machine</th>
<th>Cutting speeds</th>
<th>Feed</th>
<th>Conditions</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hafco Metalmaster CL-38</td>
<td>74m/min</td>
<td>0.20mm/rev</td>
<td>Dry</td>
</tr>
<tr>
<td></td>
<td>48m/min</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
4.1.1 EBSD preparation and setup

Chip root samples were sectioned and prepared similar to the previous study. Hot mounted sample sizes were reduced to Ø15mm to allow for ease of manoeuvring in SEM chamber during EBSD positioning. Samples were prepared by standard polishing procedures, finishing with OPS polishing before placed into the SEM chamber un-etched.

EBSD scans with were taken under high current in a LEO 1530 FEG-SEM high resolution scanning electron microscope, operating at 20kV using a Nordlys S high resolution CCD detector, at a 176mm insertion distance, with the sample at a 70° tilt angle, shown in Figure 4.3. Working distances ranged between 8-12mm, and a 60μm aperture size.

The stagnation zone was phase-mapped, using forward scanning detector (FSD) images in Figure 4.4(a)-(d). The highly distorted elongated grains reduced the electron backscatter diffraction pattern (EBSP) quality. Indexing became more difficult as scanning drew nearer towards the tool-chip interface. The use of optimal beam parameters and appropriate data clean-up assisted in obtaining optimal results. Maps were acquired with AZtecHKL software and processed using Channel 5 HKL software. All maps were cleaned at 3x zero solutions at level 5.
Figure 4.4 FSD images used for phase mapping. (a) SAF 2205, $V=74\text{m/min}$ (b) SAF 2507, $V=74\text{m/min}$ (c) SAF 2205, $V=48\text{m/min}$ and (d) SAF 2507, $V=48\text{m/min}$; feed $=0.2\text{mm/rev}$ (e) location of FSD scans on chip root sample, the stagnation zone region
4.2 Phase mapping the stagnation region

Phase map images shown in Figure 4.5 indicated a collection of ferrite build-up in the stagnation zone. EBSD scans revealed a distinguished build-up of ferrite bands leading towards the chip-tool interface. The size of the cluster of ferrite bands was found to be larger with chip-root samples produced at 48m/min, shown in Figures 4.5(c) and (d). These regions appear compact with ferrite grains and substructures, and also appear to accumulate in size during tool advancement.

Figure 4.5 Phase map of the stagnation zone on chip root samples (a) SAF 2205, \( V = 74 \text{m/min} \) (b) SAF 2507, \( V = 74 \text{m/min} \) (c) SAF 2205, \( V = 48 \text{m/min} \) and (d) SAF 2507, \( V = 48 \text{m/min} \); feed = 0.2mm/rev for all, (colour map: ferrite red, austenite blue)

Figure 4.6 shows the phase count percentage heavily skewed towards the ferrite phase. These values are based on the population count of phases in the stagnation zone region only. The subset data and mapping of original as-supplied microstructure can be viewed in Appendix A.3. Phase maps and the population counts conclusively shows the stagnation zone is saturated with the ferrite phase.
4.3 Strain in the stagnation zone

Average strain values were determined by comparing the geometric change in grain size to an unstrained average grain. Long [93] and Chen [94] showed strain values could be approximated by analysing geometric changes in grain structures that were highly deformed during a friction stir welding process. This approach is similar to how strain contouring and intergranular misorientation map algorithms calculate strained areas, by contouring according to grain size comparisons to a determined average grain size.

Table 4.2 Sampled grain sizes of austenite and ferrite phases in as-received condition

<table>
<thead>
<tr>
<th>Material</th>
<th>Grain Size</th>
<th>Sample Population</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>ferrite (µm²)</td>
<td>austenite (µm²)</td>
</tr>
<tr>
<td>SAF 2205</td>
<td>11.218</td>
<td>10.695</td>
</tr>
<tr>
<td>SAF 2507</td>
<td>10.563</td>
<td>8.672</td>
</tr>
</tbody>
</table>
An average grain size was determined based on a sample population size of over 1000 grains, recommended by ASTM standards. Average grain size values were generated using Channel 5 HKL software grain statistics, shown in Table 4.2.

Using the average grain sizes for each phase as the original grain size $A_o$, then using Channel 5 software to determine the average grain size of each phase in the stagnation zone $A_f$, the average true strain $\bar{\varepsilon}$ was then determined by the following.

$$\bar{\varepsilon} = \ln \left( \frac{A_o}{A_f} \right)$$  \hspace{1cm} \text{Equation 4.1}

These calculated average values in Figure 4.7 correlated to the strain profile of an FEM model observing the same region for austenite stainless steel 316L [95]. Based on these strain values, cutting conditions appeared favourable during higher speed 74m/min, with lower strain values. According to Figure 4.7 ferrite exhibited higher strain levels in comparison to austenite, under both cutting speeds, 74m/min and 48m/min. This appears consistent, since strain would normally partition towards the softer phase.

![Figure 4.7 Average strain values of phases in the stagnation region](image)

**4.4 Plastic behaviour at the stagnation region**

Grain boundary mapping is an effective method used to highlight areas subject to deformation and strain so that plastic behaviour of the microstructure can be observed. This method involves the mapping of misorientation between grains. Highly deformed regions will have high concentrations of low angle grain boundaries (LAGB). The detection of higher angled grain boundaries (HAGB) generally indicate the detection
of interphase boundaries. Annealing twin boundaries can also be identified as having a 60° misorientation around the <111> plane.

Figure 4.8 Grain boundary map of chip root sample material 2507 produced at \( V = 74 \text{m/min}, \ f = 0.20 \text{mm/rev}, \ \text{depth of cut}=2\text{mm} \) (a) Overview image (b) stagnation zone (austenite, LAGB green 2-10°, HAGB blue > 10°) (ferrite, LAGB red 2-10°, HAGB yellow > 10°)

The grain boundary maps shown in Figure 4.8 reveals a significant change in microstructure in terms of grain size and misorientation. For the first time, what is
revealed in the overview image Figure 4.8(a), is the development of sequential stages of strain loading in the duplex microstructure induced by the cutting tool.

Annealing twins are a common occurrence in the austenite phase. These are described to form as a result of ‘accidental’ grain growth during processing. In an unstrained duplex structure, annealing twins have a dominating appearance in austenite phase [96, 97]. This is true for location (i). Here the microstructure is in an unstrained state. Most of the blue lines HAGBs in this region are annealing twins in the austenite phase.

At location (ii) grains begin showing indications of strain loading. With no indications of grain size distortion. But the strain has become large enough to trigger small dislocations in sub-grain structures at the grain boundaries. This is shown by the green and red LAGBs gathering along the grain boundary lines. Located somewhere in location (ii) is the transition point between the elastic and plastic region. The accumulation of LAGBs indicates the level of strain increasing with the distance moving towards the tool interface.

Location (iii) shows the microstructure begin to evolve into heterogeneous structures as a result of high strain. Also referred as lamellar boundaries [98], these dense structures are more suited to handling high strain. They are a combination of high and low angle grain boundaries, compacted together forming an intricate network of grains and substructures. These are shown in more detail in Figure 4.8(b). Heterogeneous structures typically form at strain levels $\varepsilon > 1$ [98], this agrees with estimated strain figures previously shown in Figure 4.7.

Approaching the stagnation zone, heterogeneous structures become fully developed. The location highlighted (iv) in Figure 4.8(b) is fully saturated with heterogeneous structures. These show to reduce in cross-sectional area significantly upon reaching the stagnation zone (v), showing a large banded collection of highly deformed structures. Figure 4.9 shows a larger build-up of heterogeneous structures had developed in slower cutting speed chip root sample. The grain boundary map also detected a large cluster of HAGB in the region, particularly for the ferrite phase, highlighted in yellow. Plotted frequency distribution graphs of misorientation shown in Figure 4.10, does show the overall count to be greater for the number of LAGB in the stagnation zone. This would be distinguishing for regions of high strain. The grain
boundary misorientation distribution profile was found to be identical in all measured chip root samples at the stagnation zone.

Figure 4.9 Grain boundary map of chip root sample material 2507 produced at $V=48\text{m/min}$, $f=0.20\text{mm/rev}$, depth of cut=2mm (ferrite, LAGB red 2-10°, HAGB yellow > 10°) (austenite, LAGB green 2-10°, HAGB blue > 10°)
Figure 4.10 Misorientation distribution count of the (a) ferrite phase and (b) austenite phase, distribution colouring (ferrite, LAGB red 2-10°, HAGB yellow > 10°) (austenite, LAGB green 2-10°, HAGB blue > 10°)

4.5 Evolving twinning structures

What was apparent in the stagnation region was a large decline in the presence of annealing twins in the austenite phase relative to its original state. Figure 4.9(b) shows a large population decline in the number of high misorientation angles ranging 57.5°-60.5°. This reduction was noticed in all chip root samples, shown in Figure 4.11. Wróński [92] commented on the disappearance of annealing twins in the austenite phase in a URN45N duplex tensile sample due to deformation. He also showed that the number of missing twins increased with increasing deformation, though did not disclose what was occurring to the twin boundaries.
Annealing twins have a $60^\circ$ misorientation around the $<111>$ plane. Figure 4.12(a) highlights twining boundaries in the overview chip root image. It shows the amount of annealing twins becoming fewer in austenite as the microstructure moves closer to the tool interface. The special-boundary map also detected a region of high-angle boundaries which is common under influence of medium to high strain [98], and known to misorientating along the $<111>$ plane [99]. These high angle boundaries appeared in the form of primary slip systems, particularly planar character slip which is common in the austenite phase at high strain and are sometimes referred to as ladders [100] from its distinct progression of parallel slip lines. These can be seen in Figure 4.12(b). Their appearance are a visible indication of work-hardening. The special-boundary mapping of the stagnation zone, Figures 4.12(b) and (c), revealed both annealing and planar slip lines are ideally not a part of the stagnation zone.
Figure 4.12. Twin boundary map of 2507 chip root samples (a) overview image of sample frozen at $V=74m/min$, (b) stagnation zone of sample frozen at $V=74m/min$ (c) stagnation zone of sample frozen at $V=48m/min$ (mapping detection, blue at $\theta=60^\circ$ misorientation, at $<111>$ plane)
4.6 Discussion

This section covers the discussion of two topics. The first discusses the data from this study in collaboration with the chip formation study of Chapter 3, in view of providing an explanation for what is causing built-up in duplex stainless steel alloys. The second discussion involves discussing what grain boundary mapping data represents, to analyse and conclude what is occurring to the austenite annealing twins which was highlighted in the previous section.

4.6.1 Ferritic bands triggering the formation of built-up edge

The findings from the EBSD phase maps collaborate with SEM scans of chemically etched chip root samples from the chip formation study in the previous chapter. Both EBSD phase maps chemical etching techniques confirm the ferrite phase is collecting at the stagnation region, and as Figure 4.5 displayed, is collecting in the form of ferritic bands. The micro-cracking detected in the ferrite phase from SEM images provides significant evidence to support the hypothesis that ferritic bands at the stagnation zone is triggering the formation of BUE. These transgranular micro-crack patterns displayed previously in Figure 3.19d, are similar to micro-cracks observed by Wallbank [101].

In his BUE study, Wallbank [101] traced the origin of shear between the chip and the built-up layer and reported it originating from micro-cracking initiating in the ferrite phase in 0.1%C and 0.4%C grade steels. A more recent study by Dönges [102] suggests these micro-cracks could be triggered from high-cyclic loading. While observing the plastic behaviour of 2205 duplex under cyclic loading, Dönges reported fatigue cracks frequently initiate transgranular in ferritic slip bands or intergranular at the ferritic phase boundaries. Given that material is potentially not moving in the stagnation zone relative to the cutting tool, the strain paths would still be highly active in the region [95]. The induced loading on the stagnant ferrite bands inhibited by the neighbouring flowing material would similarly generate a high cyclic loading environment. Subsequently triggering micro-cracking, initiating the first stage in the formation of BUE.

The mechanism of collecting ferrite bands at the stagnation region, does support Carlborg’s [56] earlier hypothesis suggesting a higher content of ferrite in duplex would cause more frequent BUE.
4.6.2 Absence of annealing twins

It is most likely the high deformation occurring has caused the annealing twin boundaries to misorientating beyond their ideal $60^\circ$ misorientation. Given that $60^\circ$ is the maximum misorientation angle of detection in the $\langle 111 \rangle$ plane since the lowest angle representation is always determined, see Figure 4.13. Therefore, the twin boundaries have structurally deformed and was now being detected at lower misorientation angles. These twin boundaries are shown to be migrating according to the distribution plot in Figure 4.10 showing to no longer lie on their usual $60^\circ$ misorientation axis.

![Figure 4.13 Diagram showing maximum possible detection angle of misorientation between grains in crystal structure, $\langle 111 \rangle$ crystal rotational axis](image)

To show this was occurring, grain boundary maps detecting misorientation between $(20-60^\circ)$ in the austenite phase, was observed for possible evolved twins. Four possible candidates were found, shown in Figure 4.14. These were seen most likely to have been annealing twins, formed prior to deformation. In each case, parts of the grain boundary was detected as a twin i.e. matching special boundary features. Also, one of the candidate grain boundaries (c), was located parallel to an adjacent grain boundary line, similar to the formation of a twin boundary. Mechanical twins were ruled out since they are more likely produced in larger grain sizes [103], which would not be the case in the stagnation region.
Figure 4.14 Tracking evolved austenite twin boundaries; mapping colouring; (austenite, pink > 20°, yellow >30°, green > 40°, aqua 50° > 0, blue special boundary <111>, 60°, red – ferrite phase)

Figure 4.15 Misorientation profiles plotted along evolved twin boundaries at the stagnation zone at locations shown in Figure 4.14

Misorientation profiles were plotted in a straight line path from distance x = 0, to intersect these suspected twin boundaries. These plots shown in Figure 4.15 highlights the change in orientation. Boundaries detected between 20-60° indicates points of
intersection, along the suspected evolved twin. The variation from $60^\circ$, at these intersecting points, highlights the orientation of the twin boundary has evolved, and is varying at different locations along the boundary line. Dislocations by edge or screw dislocation, would be dislocation mechanisms causing this re-orientation effect.

### 4.7 Summary

This study has provided definitive characterisation of the phases present in the stagnation zone. This study has also acquired significant data regarding the plastic behaviour of the duplex microstructure leading up to the stagnation zone, which may underlie the cause of the collection of ferritic bands. The collaborative findings from the chip formation study in Chapter 3 and the present study, has created an essential need to explain how this phenomenon is occurring by uncovering the mechanisms causing ferrite to gather in the stagnation region. In conclusion, the findings from the study established a need to develop a model to fully explain how these phases, austenite and ferrite, are plastically deforming, and subsequently in such a way that ferrite is accumulating at the stagnation region.
5.0 Modelling two-phase metal cutting

The main objective of this chapter was to generate a finite element (FE) model to simulate the metal cutting of duplex stainless steel alloys, for the collection of data detailing the plastic flow of the microstructure during chip formation. The development of an FE model would greatly contribute to the collaboration of findings acquired from the previous two experimental chapters. Such a model would aid in the further understanding of how the individual phases behave during chip formation, particularly how the phases behave as they approach the chip-tool interface as shown in Figure 5.1, a phase map of a chip root sample highlighting the phases, austenite and ferrite experiencing heavy deformation when approaching the tool interface.

![Figure 5.1 EBSD phase map of chip root sample generated using AZtecHKL software](image)

It was anticipated such a model for this study would need to achieve the following objective:

1. To model the plastic behaviour of two phases during chip formation, to probe the magnitude of the strains present in each phase, and to collect data at various parameters to compare with experimental data.
To achieve this objective, an FE model was created based on the physical duplex microstructure. Johansson [15] developed a similar model based on the phase microstructure to study the residual stress behaviour of duplex under applied loading. At present, there is currently no published literature regarding an FE cutting model of any form involving duplex stainless steel alloys. This chapter covers the development of a 2D finite element orthogonal cutting model for duplex stainless steels, and summarises the simulated results.

5.1 Model development

An FE mesh was generated based on the physical duplex microstructure using OOF2 software V2.1.11 developed by the National Institute of Standards and Technology (NIST). OOF2 is an ‘object oriented’ program that is able to generate FE meshes based on data imagery. The program is often used to model microstructure under mechanical or thermal loading [104, 105], and more recently has been applied to model cast iron microstructure under machining[106].

![Figure 5.2 Finite element mesh of a 2205 duplex microstructure using OOF2 software, (a) EBSD map (b) quad mesh, (ferrite=red, austenite=blue)](image)

Finite element meshes were created based on Electron Backscatter Diffraction (EBSD) images of the duplex microstructure in its original As-supplied condition, as shown in Figure 5.2(a). Quadrilateral elements shown in Figure 5.2(b) were selected for being more stable during heavy element distortion than triangular elements. These elements were type-CPS4R, four node bilinear plane stress elements with reduced integration and hour glass control.
An element size of 130 X-direction and 70 Y-direction ratio was selected based on EBSD image size, creating a rectangular array of elements. This generated a workpiece element mesh size of 9100 elements. The X and Y sizes was chosen based on an initial convergence study which found increasing the X and Y elements numbers beyond the selected values, had an insignificant influence on output variables but significantly increased computation time. Generated meshes were submitted to ABAQUS 6.12.3 for analysis.

5.1.1 Model setup

The created model was based on the Lagrangian method which was more suitable for studying the phase behaviour during actual chip formation. The Eulerian method and the Augmented Lagrangian Eulerian (ALE) method were less suitable since these approaches involved simulating metal cutting of a pre-formed or fully formed chip.

The imported mesh from OOF2 was mirrored to create an extended workpiece length, shown in Figure 5.3 detailing the model setup in ABAQUS. Also a 0.1mm thickness was assigned to the workpiece plane strain elements. The workpiece was fixed at the bottom and left-hand wall.

The tool was modelled as a 2D planar discrete rigid body. Tool geometry was based on manufacturer’s data and the cutting profile of chip root samples observed under SEM imagery. This comprised of a °0 clearance, °12 rake angle, and a 0.02mm tool.
nose radius. The tool was positioned to make a 0.2mm depth of cut, placed initially at 0.01mm away from the workpiece. Both clearance and rake face were allocated as the master surface in ‘surface to surface’ contact interaction while the workpiece mesh was made the slave region. The penalty contact method was enforced together with interaction properties ‘Tangential’ and ‘Normal’ behaviour, [107] applying a constant friction coefficient $\mu = 0.2$ and ‘Hard’ contact pressure over-closure.

Studies [65, 72, 74, 106-108] involving metal cutting models generally applied friction coefficient values between $\mu \approx 0 – 0.7$. Friction is described as one of the hardest phenomena to simulate in machining [80]. It is more difficult to measure physically as it relates to tool and work material pairing, which would require measuring stress distributions along the contact surfaces [65]. Arrazola [65] reported a friction model would have more influence over the thrust force than it would the cutting force. However, Arrazola [65] also indicated, simple friction models does not limit the simulated maximum shear stress or unrealistic values which could occur for high contact pressures. The selected $\mu = 0.2$ friction coefficient was based on a study by Nasr [95]. Nasr applied a constant friction coefficient $\mu = 0.2$ in his cutting model for 316L which was based on simulated experimental test conducted by M’Saoubi [109].

Ductile damage initiation and damage evolution with element deletion was used to model chip separation. An equivalent fracture strain value $\varepsilon = 0.2$ was selected for damage initiation, when the scalar damage variable $D = 0$, and displacement at failure set to $\varepsilon_f = 0.0055$ for when $D = 1$. The fracture strain value was selected based on preliminary tensile data. According tensile data for duplex 2205, $\varepsilon = 0.2$ is 59% of the total elongation $\varepsilon_{total} = 0.34$. Under the ductile damage model and free micro-crack conditions [72] it is assumed fracture would typically occur in this plastic region after UTS $\varepsilon_{UTS} = 0.185$. The displacement at failure value $\varepsilon_f = 0.0055$, takes into account the physical element size to govern the rate of loss in strength from $D = 0$, to complete element failure $D = 1$. Under this condition the material element fails and loses its ability to support loading. Based on element size, an acceptable range was determined $0.009 < \varepsilon_f < 0.004$. The upper limit would impose a state where elements would fail to shear, triggering computation error. Meanwhile, the lower limit imposed the opposite effect where elements would fail instantly under the fracture strain value, and display negligible strain. The acceptable $\varepsilon_f$ range produced good fracture and chip predictability based on experimental comparison.
5.1.2 Material properties

Plastic material curves for phases were created using a modified expression of the Osgood Ramberg model previously discussed in Chapter 2.3.3. Rasmussen [110] developed a modified version of the original to accommodate a better transition between the elastic and plastic regions, see Equation 5.1. Rasmussen also showed this modified expression fit well with duplex stainless steel material curves [110]. Also incorporated in the solution of the elastic region (i.e. for \( \sigma \leq \sigma_y \)) are \( \alpha \) and \( \beta \) parameters as shown in Equation 5.2 displaying the full modified solution in terms of stress in the elastic and plastic regions. Both \( \alpha \) and \( \beta \) are introduced to match the stress and slope at elastic to plastic interface [111].

\[
\varepsilon = \begin{cases} 
\frac{\sigma}{E} + 0.002 \left( \frac{\sigma}{\sigma_{0.2}} \right)^n & \text{for } \sigma \leq \sigma_y \\
\frac{\sigma - \sigma_{0.2}}{E_{0.2}} + \varepsilon_u \left( \frac{\sigma - \sigma_{0.2}}{\sigma_u - \sigma_{0.2}} \right)^m + \varepsilon_{0.2} & \text{for } \sigma > \sigma_y 
\end{cases} \quad \text{Equation 5.1}
\]

\[
\sigma = \begin{cases} 
E \varepsilon - \alpha \varepsilon^\beta & \text{for } \sigma \leq \sigma_y \\
(\sigma_u - \sigma_{0.2})e^{\ln(\varepsilon_u)/m} + \sigma_{0.2} & \text{for } \sigma > \sigma_y 
\end{cases} \quad \text{Equation 5.2}
\]

Where \( \alpha \) and \( \beta \) are equal to

\[
\alpha = \frac{E \varepsilon_{0.2} - \sigma_0}{\varepsilon_{0.2}} \quad \text{Equation 5.3}
\]

\[
\beta = \frac{E \varepsilon_{0.2} - \sigma_0}{\varepsilon_{0.2}} \quad \text{Equation 5.4}
\]

A full list of input parameters for austenite and ferrite curves can be viewed in Appendix A.4. Figure 5.4 displays a plot of the modified Ramberg-Osgood model compared with the original Ramberg-Osgood model displayed previously in Equation 4.3. Both plotted to model the measured tensile curve of duplex 2507. The plot confirms the modified version of the Ramberg-Osgood model to be a more suited match to the experimental tensile curve.
The material curves for austenite and ferrite phases was fitted to the experimental curve of the bulk material, as show in Figure 5.5. A similar approach was implemented by Johansson [15] in his material model for duplex phases. The different tensile yields and UTS of the two phases was selected based on characterisation work using Nano-indentation. Nano-hardness is an effective method for determining local mechanical
properties of the material [112]. Figure 5.6 shows a sequence of nano-indentation measurements performed on duplex 2205 using a Berkovich tip indenter.

![Nano indentation on etched duplex sample](image)

**Figure 5.6 Nano indentation on etched duplex sample, (ferrite=dark, austenite=light)**

<table>
<thead>
<tr>
<th>Phase / Material alloy</th>
<th>2205</th>
<th>2507</th>
<th>2205</th>
<th>2507</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ferrite (MPa)</td>
<td>28.5</td>
<td>48.0</td>
<td>85.7</td>
<td>138.4</td>
</tr>
<tr>
<td>Austenite (MPa)</td>
<td>31.4</td>
<td>53.8</td>
<td>93.9</td>
<td>150.9</td>
</tr>
<tr>
<td>Difference (%)</td>
<td>9%</td>
<td>11%</td>
<td>9%</td>
<td>8%</td>
</tr>
</tbody>
</table>

Table 5.1 Mechanical data based on Nano-indentation characterisation

As a result of the characterisation work, Table 5.1 displays averages of compressive yield and elastic modulus which was calculated based on methods reported by Giannakopoulos [113]. These calculated values were not used in the material model, as they are generally assumed to be a representative of the bulk material [112]. Instead, the percentage difference % was incorporated into the material curve for values of tensile yield and UTS.

Two other main input properties such as density and elastic property Poisson’s ratio, was treated as bulk value and were maintained constant between each phase. These values were based from measured and published readings. A mass scaling factor of $10^3$ was applied to reduce computation time. This generalised technique at the current factor has a reported 2.2% influence on plastic strain from the base model [114].
5.2 Results

Cutting simulations generated by the FE model correlated well with experimental results. Figure 5.7(a) shows the formation of chip segments generated by the FE model. A serrated edge chip profile distinguishing to stainless steels was produced. The profile of chip segments for both 2205 and 2507 was similar compared to frozen chip root samples produced at comparable conditions, shown in Figure 5.7(b). Similar features between the FE model and experimental chip included chip segment size and shear angle. For example, the 2205 simulated cut displayed in Figure 5.7(a) predicted a °44 shear angle. The chip root produced under the same cutting velocity $V=74\text{m/min}$, generated a °46 shear angle, Figure 5.7(b), equating a negligible 4.3% error.

Figure 5.7 Duplex 2205 FE Model simulating chip formation at $V=74\text{m/min}$ (b) frozen chip root sample, interrupted at $V=74\text{m/min}$
Due to a lack of published literature regarding a duplex FE cutting model, there is less support for comparing computed data. Regardless, the effective plastic strain ‘PEEQ’ data shown in Figure 5.7(a) and Figure 5.8 came to good agreement with experimental strain value calculations from the previous chapter, see Figure 4.7. For example, calculated strain averages at the stagnation region based on grain size measurement, included $\varepsilon_\alpha = 3.09$ and $\varepsilon_\varphi = 2.09$ for duplex 2205 sample frozen at $V = 74 \text{ m/min}$. The stress and strain values were found to be of similar profile compared with a close match to duplex model, an Austenite 316L FE model by Nasr [95].

![Figure 5.8 Duplex 2507 FE Model simulating chip formation at $V = 48 \text{ m/min}$ (b) frozen chip root sample, interrupted at $V = 48 \text{ m/min}$](image)

Figure 5.8 Duplex 2507 FE Model simulating chip formation at $V = 48 \text{ m/min}$ (b) frozen chip root sample, interrupted at $V = 48 \text{ m/min}$
5.2.1 Stress distribution during shear plane formation

Figure 5.9 displays the stress distribution during the formation of the primary shear plane and subsequent chip segment. The contour plots revealed a concentrated high stress region originating at the tool nose from penetration. This concentrated stress region was shown to grow with tool advancement, developing a perimeter surrounding the increasing chip-tool interface region. The high stress region continued to grow until it extended toward the top surface layer, as shown in the 3rd sequence of Figure 5.9. With nowhere further for the stress to redistribute, the formation of the primary shear plane followed ‘abruptly’ once the concentrated region of high stress reached the top layer. There was also the development of a secondary shear which was seen in Figure 5.9 to be of smaller scale. Both shear planes were found to have originated from the tool nose, similar to the region of high-stress.

In the high-stress region, austenite elements would withstand the higher loading while ferrite succumbed to plastic strain. This would be due to a higher austenite UTS compared to the ferrite elements. Subsequently, the origin of shear was found to occur from a region of ferrite elements along the tool nose, shown in Figure 5.9. Upon shear, the expanded high stress region was found to dissipate as the fractured segment i.e. the chip, was formed. Then what is shown in the final sequence of Figure 5.9, is the process repeating itself with the re-development of an initial high stress region.

The plot shown in Figure 5.10 tracks the stress of austenite and ferrite elements during its transition into the chip region, at two different depth locations. It showed the elements located near the tool nose at a 0.178mm depth, experienced higher maximum stresses compared to the elements located near the top surface, at a 0.045mm depth, for a 0.200mm total depth of cut. The plot also highlights the transitioning into chip region, which shows the phase elements are subjected to minimal stress levels once the phases are a part of the chip.

The stress contour profile and maximum stress levels were found to be typically the same for both alloys 2205 and 2507, and similar at varying speeds between 48m/min to 74m/min. For further detail, the full stress contour plots of both alloys at these two cutting speeds can be viewed in the appendix section, Appendix A5.1.
Figure 5.9 Stress distribution sequence during cutting 2205 at $V = 48$ m/min
5.2.2 Strain magnitude during transitioning

The magnitude of strain was found to be higher in ferrite elements compared to austenite, which can be related to a lower ferrite yield. Of the observed layers from the workpiece surface highlighted in Figure 5.11, including a sub-layer (Layers 0-5), the largest measured strain paths occurred near the tool nose radius, layers 1 and 2. The maximum strain was found to occur after tool contact in the ‘transitioning region’. A region designated as past the point where the advancing tool-interface exceeded the probe element’s original position. This position is highlighted ‘x’ in Figure 5.11.
Figure 5.12 Strain profile to maximum at different depths from the workpiece surface during tool advance, at (a) layers 1-3 and (b) layers 4, 5 and 0, 2205 duplex cut at $V=48\text{m/min}$

Figure 5.12 shows a plot of the measured strain rates reaching their maximum. There was a noted difference with $\Delta \varepsilon$ between the two phase elements in layer 1 and layer 2, shown in Figure 5.12(a). The strain profile of layer 1 describes the path of phase...
elements which do not shear into the chip but remain embedded as a part of the machined surface. The layer 1 profile showed that both austenite and ferrite elements are subjected to a sharp, high and equal strain rate in the transitioning region caused by compression shear from the tool nose.

In contrast, the path of layer 2 elements do transition into the chip, after tool nose contact through a secondary shear zone. The strain plot revealed ferrite elements in this path were subjected to higher strain rates compared to austenite during transitioning,

![Figure 5.13 location of probe elements during shear plane formation](image)

The strain profile of layer 3 described the path of elements which merge into the primary shear plane. Its strain plot in Figure 5.12(a) shows a steady ramp increase occurring before the designated transitioning point. This confirms that most of the deformation occurs prior to the tool advancement for phase elements that cross into the chip region through the primary shear. This can be seen in contour plot Figure 5.13 showing the layer 3 probe element undergoing extensive strain ahead of the chip-tool interface. The steady ramp increase and reduced strain difference $\Delta\varepsilon$ between phase elements, were indicators that the phase elements transitioning into the chip through the primary shear was smoother compared to transitioning through the secondary shear. The Figure 5.12(b) plot revealed phase elements on layers 4 and 5 maintained a smooth transition into the chip under minimal strain, as well as the compression strain detected by the probe element at layer ‘0’ located 0.025mm from the machined surface, recording a maximum strain $\varepsilon \approx 0.02$. 

90
5.2.3 Cutting speed influence on strain

The cutting speed was varied between 48m/min and 150m/min to observe how it would influence the strain response at the tool nose where the highest strain was detected i.e. layer 2 in the previous section, see Figure 5.11. Based on maximum strain plots presented in Figure 5.14, it was found mainly austenite elements showed an increasing strain response with cutting speed. The data was taken from maximum averages at four locations along the tool nose, depth d=0.178mm from the original surface.

![Figure 5.14 Influence of cutting speed on the maximum strain of phases detected at the tool nose, at depth d=0.178 for duplex (a) 2205 (b) 2507](image)

Ferrite generally maintained a high strain rate $\varepsilon > 3$ at increasing speed while the trend showed the austenite strain was progressively increasing, effectively reducing the
strain difference found in both modelled alloys $\Delta \varepsilon$, as highlighted in Figure 5.14(a). This increasing austenite strain was found to be caused by reduced plastic strain recovery at higher cutting speeds, which can be seen in the stress and strain rate plots shown in Figure 5.15.

![Figure 5.15 Stress and strain rate plots for duplex 2507 at tool nose, depth $d=0.178\text{mm}$ at (a) $V_1=74\text{m/min}$ and (b) $V_2=150\text{m/min}$](image-url)
At cutting speeds $V > 110\text{m/min}$, the austenite strain was found to match the ferrite strain at some of the measured locations, as shown in Figure 5.15(b). The stress plot revealed there was no plastic strain recovery at these location. An increasing shear angle was also a product of increasing cutting speed. The shear angle increase combined with matching high phase strain was shown to generate unstable chip formation at high cutting speeds for $V > 140\text{m/min}$ shown in Figure 5.16.

### 5.2.4 Modelling a stagnation zone

The conditions of a stagnation zone could be simulated by increasing the current plastic strain failure criterion by $\varepsilon_f > 20\%$. This extended the damage degradation range in the ductile damage model, resulting to a delay in complete element failure after damage initiation. The resulting effect is the formation of stagnation zone which shows to develop in front of the tool nose as shown in Figure 5.17. A full contour plot sequence of the showing the growth of the stagnation zone can be viewed in the appendix chapter, section A5.5, page 145.
Figure 5.17 Contour plot sequence of modelled stagnation zone during cutting 2205 at \( V=74 \text{m/min} \) (a) initial development (b) mid-stage growth (c) chip segment shearing along stagnation zone
As shown in Figure 5.17, the stagnation zone developed into a full sized built-up layer in the form of a lower-stress region compared to the bulk region in shear. The shape formed was similar to a BUE, lying on the tool surface. The modelled built-up layer also showed similar behaviour by acting as the cutting edge, as shown in Figure 5.17(c) where a shear segment shows to originate from the nose of the built-up layer.

![Figure 5.17](image1)

**Figure 5.18** Velocity (XY) contour plots of modelled built-up layer in the (a) horizontal and (b) vertical direction, 2205 model at $V = 74\text{m/min}$, legend unit in $V(\text{mm/s})$

Velocity plots displayed in Figure 5.18 verified the modelled built-up layer was stationary relative to the cut tool, low vertical magnitude. Meanwhile the material
above and ahead of the built-up layer displayed greater velocity projection into the chip, i.e. the vertical direction, as shown in Figure 5.18(b). Similar studies [65, 80] validated the development of a stagnation zone in their FE models based on velocity plots which displayed a low velocity region, Arrazola [65] defined this velocity as $V < 0.05\text{m/s}$.

### 5.2.5 Strain of phases in the stagnation region

Figure 5.19 displays the strain profile of phase elements entering the stagnation zone region compared to normal conditions. The time where the phase element is seen entering the stagnation zone is marked by the solid line black. The plot reveals the maximum strain is higher, almost double for austenite compared without a modelled stagnation zone. Also, the time to reach maximum strain is also found to be longer under ramp increase, as shown in the plot.

![Figure 5.19 Strain profile of phase elements in the path of the stagnation zone, located along the tool nose, d=0.178mm](image-url)
5.3 Discussion

The stress contour plots shown in Figure 5.17, characterise the built-up region to be of lower stress compared the bulk region in shear. This agrees with experimental findings reported by Bao and Stevenson [115] who found a substantial drop in measured stress from the rake face leading into the chip, under simulated machining conditions where BUE was known to form during the turning of aluminium alloys. Bao and Stevenson hypothesized a negative stress gradient from the rake face into the chip as being the main cause of built up edge formation. Based on the current model, it can be stated that low stress is a product of the stagnation by which the material is no longer a part of the flowing material i.e. material in shear. It is therefore at a lower stress state as depicted in the contour stress plots.

The strain values presented in this chapter correlated well with similar FE models tied with austenite stainless steel 316L [95, 116]. The predicted values from the current model were similar to those reported such as Maranhão [116] who predicted the plastic strain in his 316L model to range between 3.5-4.0.

The output model data does provide admissible description of strain behaviour of phase material during workpiece to chip transitioning. The strain profile shown in Figure 5.12 of elements transitioning into the chip through different paths, does correlate with experimental data reported by Stevenson [117] who displayed in his strain field map, the secondary shear zone and surrounding chip-tool interface to be of greater strain than the primary shear zone. Such was the data generated by the FE model. Furthermore the higher strain values found to occur in phase elements which develop in the built-up layer region, Figure 5.19, are a reasonable assumption since it is well-known for BUE to consist of highly strained material. With strain hardening being a product of plastic strain, Trent [60] reported a 600HV increase in a BUE structure compared to 200-250HV measured in the bulk material.
5.4 Summary

The finite element (FE) model presented in this chapter has provided significant data detailing the plastic behaviour of individual phases of a duplex microstructure during chip formation. The FE model showed ferrite obtained a higher plastic strain compared to austenite. Moreover, strain values were higher in the modelled stagnation zone, $\varepsilon_a^{(max)} = 4.56$, compared to the austenite strain $\varepsilon_y^{(max)} = 3.00$. It should be noted, the strain values were significantly influenced by the input material properties, particularly the yield point. Ferrite elements initiate plastic strain earlier than austenite due to a lower ferrite yield point, thus having a higher degree of strain.

Figure 5.20. Displaying individual phases (a) austenite and (b) ferrite, during formation of a stagnation zone in 2205 duplex model at $V=74$ m/min
As a consequence of the modelled softer ferrite, the ferrite elements displayed a higher degree of deformation, which was evident in the simulated stagnation zone shown in Figure 5.20. The output strain difference triggered the response of the softer phase being squeezed, while the harder phase ‘austenite’ remains to dominate the region. Most of the deformation occurred inside the simulated stagnation zone, as indicated in the previous strain plot, Figure 5.19.

The display of austenite dominating the stagnation zone apparently did not correlate with the chip formation or characterisation study which confirmed the region being more concentrated with ferrite. The results show there is a possible inaccuracy to the material property description. As previously mentioned, yield values were based on nano-indentation results which determined ferrite as the softer material. The indentation data was also used to approximate the compression yield. Material data was also supported by literature where Johansson [85] reported austenite in lean duplex as having a higher yield and hardness compared to ferrite.

This chapter concludes the experimental work presented in this thesis. The data acquired from this chapter will be discussed in collaboration with the chip formation study and phase characterisation study from the previous two chapters in the following discussion chapter.
CHAPTER SIX

6.0 Discussion

The objective of this thesis was to acquire a better understanding of the long-established issue of frequent BUE found when machining duplex stainless steels. The course of experimental studies was primarily focused towards the key research question which was “what are the mechanisms triggering frequent built-up edge (BUE) during the machining of duplex stainless steel alloys?” This question itself evolved during the progression of each study as more became known about the subject matter.

In summary, an observation was made in the chip formation study in Chapter 3, where a ferrite build-up was found at the stagnation region in frozen chip root samples. This ferrite build-up was identified as a probable triggering mechanism to the formation of BUE. Micro-cracks were found developing along ferrite phase that was part of an initial built-up layer in the stagnation region. These micro-cracks were similar to those which lead to BUE formation. The characterisation study in Chapter 4 sort to “quantify the ferrite build-up”. Electron backscatter diffraction (EBSD) mapping of the stagnation region confirmed ferritic bands were collecting at the stagnation region, with up to 85% of ferrite makeup was detected in the region. The purpose of the concluding finite element study was to explain “how the ferrite build-up was occurring”, by modelling the plastic behaviour of the individual phases during chip formation. Yet the plastic behaviour of phase elements in a simulated stagnation zone, did not match the ferrite build-up as observed in the experimental chip formation and characterisation studies.

This chapter reviews the results from the experimental chapters with the main focus directed at addressing how the ferrite build-up is collecting at the stagnation region. The subsequent mechanisms are identified and presented with corroborating evidence and supporting literature.
6.1 Mechanism for ferrite build up in the stagnation region

The finite element model portrayed ferrite elements at a higher degree of strain than austenite during both chip formation and while collecting at the simulated stagnation zone. This also agreed with strain approximations in Chapter 4.3 which was based on the change in the average measured grain size. Despite these findings, there is evidence to suggest that under the influence of high strain, the austenite phase loses its strength and becomes softer, subsequently developing greater strain than ferrite.

A number of studies [85, 92, 96, 118, 119] have reported that in a duplex material, austenite plastically deforms at a higher degree than ferrite. Johansson [85] observed this through X-ray diffraction (XRD) and transmission electron microscopy (TEM) in a 2304 duplex alloy during cyclic loading. Johansson mentioned that despite austenite obtaining a higher yield and hardness, it still underwent higher plastic deformation than ferrite. Johansson suggested was due to residual stresses present in the material. Furthermore, Wroński [92] reported austenite has a higher dislocation density than ferrite, indicating the rate at which low angle grain boundaries (LAGB) appear is greater. The grain boundary mapping in Chapter 4.4 of the characterisation study, detected a large population of LAGB’s in both austenite and ferrite which signified the occurrence of high level dislocation, refer to Figure 4.10. If Wroński’s observation was correct, then the rate of dislocation and related strain that occurs would be increased in the austenite phase. This would explain how ferrite bands are collecting, since these bands are deforming to a lesser degree and at a lesser rate than austenite.

Figure 6.1 Indicated austenite flow directions in the stagnation region of a 2205 chip root sample, produced at \( V = 94 \text{m/min} \)
Therefore, based on the proposed behaviour, the austenite flow paths highlighted in Figure 6.1 of a stagnation region would indicate the austenite grains are effectively straining more and at an increased rate than ferrite. These austenite grains would tend to flow quicker into the chip during tool advancement through the primary or secondary shear zone, or separate at lower region and remain compressed as part of the machined surface. Ferritic grains appear to flow in the same directions as austenite but this would be at a relatively slower rate. Given these differences the ferrite would tend to show less deformation being able withstand more loading, which is highlighted in Figure 6.1 by the display of larger ferrite grains compared to austenite.

### 6.2 Plastic behaviour of austenite under high strain

To state the austenite phase is plastically deforming to a greater degree than ferrite would imply the phase is essentially becoming softer despite austenite being well-known for its high work-hardening ability [44, 51]. Studies involving cyclic loading [120, 121] reported the austenite phase does become softer after work-hardening under increasing strain, and continues to soften until fracture. Mateo [120] indicated the increasing plastic deformation in austenite was due to the activation of new slip systems. This activation of new slip systems, moreover multiple slip systems would be a suitable candidate to explain the plastic softening behaviour in austenite.

![Figure 6.2 Schematic of slip activation in an austenite grain under loading](image)

An apparent model was derived to describe the hardening-softening transition of austenite, which is displayed in Figure 6.2. The model was based on the tensile study by Feagus [122]. The model ties in the activation of multiple slip with the evolving...
annealing twin structures that was observed in the characterisation study, Chapter 4.5. Annealing twins have been associated to maintaining the initial microstructure, as it had been reported [123] that their presence acts as a barrier for slip motion. In his tensile study on 316L austenite stainless steel, Feagus [122] observed back stresses in the form of single slip pileups collect along grain and twinning boundaries. These pileups create intergranular stress concentrations, which the boundaries was reported to withstand up to a maximum strain $\varepsilon_{\text{max}} = 1.5$. At which point, multiple slip systems, including cross slip would activate to relieve these stresses, as shown in Figure 6.2. A known product of cross and multiple slip systems are the formation of heterogeneous structures [98, 124], such as those detected in the stagnation region of chip root samples in Chapter 4.4, see Figures 4.8 and 4.9. Two studies [118, 124] have related Feagus’s multiple slip activation threshold to occur in duplex stainless steel alloys within an austenite phase. Hedström [124] observed the softening occur in single austenitic grains, in a 2304 duplex tensile sample using X-ray diffraction.

The thesis data to support austenite softening had occurred in the stagnation zone was firstly, the detection of heterogeneous structures from EBSD grain boundary mapping in Chapter 4.4 and secondly, two strain approximations, based on FE model in Chapter 5.2.2 and the measured grain size in Chapter 4.3, which estimates the occurring strain to be well over the reported $\varepsilon_{\text{max}} = 1.5$ threshold reported by Feagus [122]. The evidence is reasonable to support austenite softening is occurring under high strain at the stagnation region.

### 6.3 Modelling austenite softening at high strain

With the hypothesis that austenite softening is causing ferrite build-up at the stagnation region, a concluding study was conducted with the aims to validate the hypothesis and to improve on the FE model. The validating study would utilise the FE model of Chapter 5 and take into account austenite softening to observe whether it would trigger an increase in austenite strain, subsequently triggering ferrite build-up in the stagnation zone.

The material curve in the FE model for austenite elements was modified to account for softening under high strain. A softening threshold of $\varepsilon_{\text{max}} = 0.15$ was assigned, which matched the threshold value reported by Feagus [122]. The softening region of the
curve was modelled as a linear decline as shown in Figure 8.3. The decline represented the weakening of the phase, showing austenite being unable to support the exerted load.

Figure 6.3 Austenite strength curve incorporated with softening region after strain threshold for \( \varepsilon = 1.5 \)

A softening gradient ‘\( G \)’ was assigned to govern the rate of loss in strength. This linear degradation for austenite was merged with the Modified Ramberg Osgood equation, as shown in Equation 6.1.

\[
\sigma = \begin{cases} 
E\varepsilon - \alpha\varepsilon^\beta & \text{for } \sigma \leq \sigma_y \\
(\sigma_u - \sigma_{0.2})e^{\frac{\ln(\varepsilon)}{m}} + \sigma_{0.2} & \text{for } \sigma_y < \sigma \leq 1.5 \\
\sigma_{1.5} - G\varepsilon & \text{for } \sigma > 1.5 
\end{cases}
\]  

Equation 6.1

The softening gradient was varied between 1 – 10 GPa to observe the influence of different softening gradients on the model. Figure 6.4 displays the modified strength curves submitted to ABAQUS. The original ferrite curve based on Equation 5.2 in Chapter 5.1.2, was maintained constant for the ferrite elements in the model.
The results displayed in Figure 6.5 show austenite softening hypothesis does have an influence to which phase governs the stagnation zone. This is substantiated since the ferrite strength curve was constant during these simulations. Results show as austenite became softer, it deformed to a greater degree than ferrite since it could no longer maintain loading, not even at the lower loading state of ferrite. The amount of relative deformation between the two phases could also be observed with the strain difference. Figure 6.6 plots the maximum strain of phases at a fixed location in the stagnation region.

What Figure 6.5 and the strain plot in Figure 6.6 indicates is that the greater the strain difference $\Delta \varepsilon$, the greater the imbalance of phases in the stagnation zone. A softening gradient of -1 GPa was not sufficient to overcome the lower ferrite yield influence. Therefore, ferrite maintained high strain and with a significant strain difference between the phases $\Delta \varepsilon = 2.35$, saw the lower strained austenite dominating the stagnation zone.

A transition region was noticed, which occurred between the softening gradient of -2 to -4 GPa. Both phases appeared equally deformed in the stagnation zone and
Figure 6.5 Duplex cutting model 2205 at V=74m/min, with austenite softening at various softening gradients ‘G’, Ferrite (red) and Austenite (blue)
furthermore, the strain difference was minimised, $\Delta \varepsilon = 0.01$ at -2 GPa and $\Delta \varepsilon = 0.17$ at -4 GPa. Beyond the transition region i.e. $G > 4$ GPa, the austenite phase begins to deform more and the ferrite phase begins to dominate the stagnation zone, mirrored by an increasing strain difference, shown in the Figure 6.6 plot. The ferrite phase dominance, particularly for the simulation at $G = -10$ GPa, Figure 6.6, appears to match the SEM and EBSD phase map images of the stagnation zone in the previous chip formation and phase characterisation studies, in Chapters 3 and 4.

What the data from the modified model concludes is that it is highly credible the ferrite collecting at the stagnation zone may have occurred through austenite softening. Therefore, if ferritic bands are responsible for the formation of BUE, as was concluded from the discussion in Chapter 4 then the relationship in Equation 6.1 which states the strain difference between phases at the stagnation zone correlates to the formation of BUE.

$$\Delta \varepsilon \propto BUE$$  \hspace{1cm} \text{Equation 6.1}

### 6.4 Effect of cutting temperature

The effect of cutting temperature on the duplex microstructure was not investigated in this thesis. There has been little research effort focused on examining the thermal response to duplex machining, outlining a potential gap in literature. Temperature is
generally an important aspect to consider. It has been linked to phase transformations such as martensite formation and re-austenising along the secondary shear zone \[125\] and machined sub-surface \[39\] in machining hardened steels (\(\%C > 0.45\)).

Despite this, an early temperature investigation found that the heat generated from duplex cutting would have minimal an effect on the microstructure. This focused the thesis direction towards investigating the effect of strain mechanisms. Figure 6.7 displays the raw data from the temperature experiment that was conducted.

![Figure 6.7 Raw temperature readings in turning operation, duplex 2205, cutting at \(V_1 = 88\text{m/min}\) and \(V_2 = 48\text{m/min}\), feed = 0.2mm/rev, DoC = 2mm, K-type thermocouple (measurement range 200-1350°C), sample rate 0.2s](image)

The aim of the experiment was to observe the steady state temperature \(T_{ss}\) by conducting one continuous cut, in a turning operation. The material workpiece was 2205 duplex, machined in As-supplied condition, in \(\Phi20\text{mm}\) round bar form. An \(\Phi1.5\text{mm}\) diameter tip K-type thermocouple of 60mm length, was inserted into a solid carbide tool insert. These were the same WNMG inserts used in the chip formation study in Chapter 3. A wire-cut (EDM) slot was produced along the underside of the carbide insert, providing an insertion for the thermocouple. This located the thermocouple measurement tip point \(d = 3.0\text{mm}\) from the tool nose.
Tests were performed at two cutting speeds, $V_1 = 88\text{m/min}$ and $V_2 = 48\text{m/min}$. Due to the nature of the produced chips, the steady state temperature at $V_1 = 88\text{m/min}$ could not be achieved due to heavy blockage from continuous chip generation during cutting. Its temperature profile was projected in Figure 6.7 based on its initial curve and the $T_{ss}$ profile of $V_2 = 48\text{m/min}$. The steady state temperature range at these cutting speeds was approximated around $175^\circ\text{C} < T_{ss} < 250^\circ\text{C}$ based on Figure 6.7.

There would be no occurrence of phase transformations at this range. Secondary phases such as ‘χ’ ‘σ’ and ‘ε-Cu’ precipitate in the range between 300-1050°C, and embrittlement occurs at 475°C [126]. Bordinassi [88] reported no microstructural changes occurred in turning super duplex 2507 at cutting speeds up to 150m/min. This would correlate to the present temperature data if Bordinassi’s comment on microstructural changes implied phase transformation. In regards to thermal softening, linear strain-temperature graphs published by Siegmund [42] revealed thermo-loading at $T = 250^\circ\text{C}$, influenced strain in the range between $0.0025 < \varepsilon < 0.0045$ in the longitudinal direction and $0.002 < \varepsilon < 0.003$ in transverse. Siegmund’s graphs was based on dilatometer tests, which measures the strain influence due to thermal loading. According to these figures, the affected strain due to thermal loading in duplex cutting would have minor influence in terms of the total strain due to cutting.

### 6.5 Contribution towards machining literature and industry

The findings of this thesis and its studies contribute significantly toward the field of machining. Its main contribution is a greater understanding of how BUE can occur in duplex stainless steel alloys. What the literature survey in Chapter 3 identified was that there has been no sufficient explanation to explain what triggers BUE in duplex stainless steel alloys. Carlborg [56] suggested the percentage of ferrite in duplex stainless steel alloys triggered increasing BUE formation. It is only until now, is it realised ferritic bands have the tendency of collecting at the stagnation zone, subsequently triggering the formation of a BUE.

The collective data from this thesis could aid the tooling and research industry towards developing more efficient tools or controlling machining environments that would be capable of minimising or even eliminating BUE in machining duplex. An area of particular interest would be the strain difference $\Delta \varepsilon$ highlighted recently in Figure 6.6.
and its relation to BUE in Equation 6.1. If the increasing strain difference $\Delta \varepsilon$ between phases which occur during cutting correlates to the formation of BUE, then the focus would be to minimise the strain difference $\Delta \varepsilon$. The FE study in Chapter 5 highlighted an interesting strain response where it was shown that increasing the cutting speed ($V$) correlated to reducing the strain difference $\Delta \varepsilon$, refer to Figure 5.14 in Chapter 5.2.3. This indicates a relationship exists between these two variables, highlighted in Equation 6.2.

$$V \propto \Delta \varepsilon \quad \text{Equation 6.2}$$

The relation between the BUE and cutting speed $V$ indicated by the connection between Equations 6.1 and 6.2, have been well-known in literature. BUE with duplex stainless steel alloys is generally reported to form at lower cutting speeds [3, 4, 49, 56]. Carlborg [56] indicated this in his tool life charts, stating the lower cutting limit velocity for 4 min tool life was limited by BUE. The knowledge that BUE is reduced at higher cutting velocities not only supports the relationship in Equation 6.2, but verifies through experimental literature, the feasible link between the strain difference $\Delta \varepsilon$ and BUE.

In conclusion, there is significant potential in the FE cutting model’s concepts, to be utilised by the finite element community. As the first known cutting model based on duplex stainless steel alloys, the model can be further developed and employed to obtain machining data based on various machining parameters and including different alloyed microstructures, which would not be limited to the duplex microstructure alone. Finally, the application of a dynamically simulated stagnation zone would be of significant interest to machining studies, particularly those focused with studying BUE formation.
CHAPTER SEVEN

7.0 Conclusion

This thesis investigated the mechanisms which trigger the formation of BUE while machining duplex stainless steel alloys. The presented studies identified two main candidates to be primary mechanisms for BUE. The first mechanism, is the austenite phase softening under high strain. This occurs in the stagnation zone during chip formation, and is responsible for austenite plastically deforming more than the ferrite phase in this region. The strain difference $\Delta\varepsilon$ between austenite and ferrite in this region is responsible for the second mechanism, which is the build-up of ferritic bands in the stagnation zone. Ferrite is highly prone micro-cracking under high cyclic loading. The collection of ferrite bands in a highly strained region such as the stagnation zone would trigger the occurrence of micro-cracks, and the greater the strain difference $\Delta\varepsilon$ the greater the ferrite build-up. Micro-cracks are the well-known triggers to BUE. These cracks develop in size, activating the separation between the built-up layer and the chip, forming a BUE.

This chapter provides a summary and main conclusion of the experimental studies presented in this thesis. This chapter also addresses limitations to the current thesis work and highlights areas for continuing future work in the current field of study.

7.1 Thesis summary

This section provides an executive summary of the main findings of the experimental work conducted in this thesis. A conclusion of findings are detailed below, from the machinability and chip formation study in Chapter 3, the EBSD characterisation study in Chapter 4, to the two-phase modelling work presented in Chapters 5 and 6.

7.1.1 Machinability and Chip formation study

Chapter 3 presented two observational studies. The first, a machinability study that observed the machining behaviour of duplex stainless steel alloys during a drilling operation. The second study focused on observing the plastic flow of the duplex microstructure through the implementation of the quick-stop test in a turning operation. The main conclusions from this chapter are as follows.
Machinability study

- Adhesion and abrasion were the most common wear mechanisms found in the flank and rake face region in drill tools. Adhesion wear was the most dominant wear mechanism along the flank. Both duplex stainless steel alloys 2507 and 2205 experienced a higher tendency to BUE compared to austenite 316L. BUE is responsible for adhesion wear.

- Among the three materials considered in the machinability study, the highest cutting force, machined surface roughness and generated flank wear was found with drilling duplex 2507, followed by duplex 2205 and concluding with austenite 316L. The trend of these rankings were influenced by tool wear and primarily the presence of BUE.

Chip formation study

- 32-44% hardness increase was evident in the chip region compared to workpiece in chip root samples produced at 94-65m/min. On average, duplex 2205 displayed a slightly greater sensitivity to work-hardening than 2507.

- Scanning electron microscope (SEM) images of chemically etched ‘chip root’ samples revealed a build-up of ferrite phase collecting at the stagnation zone. The austenite presence was less dominant in this region, indicated by smaller grain size comparison.

- Micro-cracks detected at the chip-tool interface in an initial built-up layer was identified as similar to those which initiate the formation of BUE. The crack had developed transgranular along collected ferrite grains in the stagnation zone.
7.1.2 Phase characterisation at the stagnation zone study

The Chapter 4 study employed the use of Electron Backscatter Diffraction (EBD) method on chip root samples to characterise the phases present in the stagnation zone of ‘chip root’ samples. Grain boundary mapping was also conducted to analyse the plastic strain behaviour of phases in the region. The findings from the study included the following.

- EBSD phase mapping detected a greater percentage of $\alpha$-ferrite phase in the stagnation zone. 65-85% more ferrite was detected compared to austenite validating a build-up of ferrite is occurring in the region. EBSD phase mapping indicated ferrite collecting in the form of ferritic bands.

- Grain boundary mapping of the stagnation zone region, revealed both austenite and ferrite grains evolved into heterogeneous structures. These structure typically form as an adapting mechanism towards handling high strain, typically $\varepsilon > 1$.

- Annealing twinning structures were found dissipating ahead of the stagnation zone. These twinning structures are misorientating beyond their ideal °60 orientation along the $<111>$ plane. As a result the twinning structures are being detected at lower misorientation angles.

- Planar character slip in austenite grains was detected in close proximity to the stagnation zone. The detection of planar slip are a visual indication of work-hardening. Despite their appearance near the stagnation zone, special-grain boundary mapping of the stagnation zone revealed no planar slip structures or annealing twinning structures appear as part of the stagnation zone makeup.

7.1.3 Modelling two-phase material metal cutting

The finite element (FE) study in Chapter 5 focused on modelling the plastic flow of duplex microstructure during chip formation. An FE mesh was created based on the physical duplex microstructure, allowing for individual austenite and ferrite elements to be modelled under shear. The model was further developed in Chapter 6 with the incorporation of an austenite softening behaviour under high strain. The main conclusions from the FE study progressed through Chapters 5-6 are detailed below.
- Material transitioning into the chip through the secondary shear plane exhibited the greatest amount of strain. Phase material flowing through this region experienced an abrupt sharp strain increase that occurs when approaching tool nose is in close proximity. In contrast, material phases flowing into the chip through the primary shear plane experienced a more gradual strain increase.

- Increasing the cutting speed reduces the strain difference, therefore \( V \propto \Delta \varepsilon \). At increased speeds, there is less strain recovery for the stronger phase, therefore both phases will undergo more equal degree of strain.

Discussion Chapter 6 conclusions

- A single phase build-up in the stagnation zone during cutting duplex can be related to the strain difference \( \Delta \varepsilon \) between austenite \( \varepsilon_\gamma \) and ferrite \( \varepsilon_\alpha \) phases in the stagnation zone. The greater the strain difference the greater the build-up of the individual phase that exhibits the lesser strain. Therefore in reference to the chip formation and characterisation studies, if the build-up of ferrite is triggering the formation of BUE, then the strain difference correlates to built-up edge i.e. \( \Delta \varepsilon \propto BUE \)

- Ferrite build-up in the stagnation zone is explained by austenite softening under high strain. Austenite softens under high strain due to loss of twinning structures, therefore will exhibit a greater degree of strain than ferrite, (i.e. \( \varepsilon_\gamma > \varepsilon_\alpha \)) despite ferrite having lower yield strength. Under these conditions, the FE model indicated austenite plastically deforms more significantly, triggering a dominant ferrite build-up in stagnation zone.
7.2 Limitations in the present thesis work

Limitations are unavoidable and exist in the present thesis work. Despite the thesis output contributing a better understanding of the mechanisms which underlie built-up edge (BUE) in duplex stainless steel alloys, the issue still largely remains unresolved. This section outlines main limitations in the current thesis work, which is listed below under the following headings.

- **“The cutting model (problem definition) is as much a 3D problem as it is a 2D problem”** – Chip formation is a 3-dimensional process, and a turning operation, which the present FE cutting model is related to, is not a purely orthogonal cutting process. However, the main objective of the model was to study the behaviour of individual phases under chip forming conditions. It was anticipated a simplified 2D finite element cutting model could readily meet objectives while maintaining sensible accuracy.

- **“The austenite hardening-softening mechanism still remains a hypothesis, it is yet to be fully validated”** – The hardening-softening transition of the austenite phase that is occurring in the stagnation zone, remains to be proven, since the dislocation mechanisms are yet to be fully characterised. However, the presented evidence based on the FE model and SEM and EBSD observation and characterisation work, strongly supports the hypothesis that the austenite hardening softening behaviour does occur in the stagnation zone.

- **“There is a lack of physical data of a fully developed built-up edge (BUE) structure”** – There was no physical experimental data presented on the formation of a fully developed BUE structure. Consequently, it implies there could still remain unknown mechanisms triggering BUE formation. Regardless, the captured physical data of ferrite build-up in the stagnation zone, combined with detection micro-cracking, which ferrite is known highly prone to, fully supports ferrite build-up as the most highly potential candidate triggering BUE in duplex stainless steel alloys under machining. This should remain accurate, until such a time when more conclusive data may be obtained.
7.3 Future recommendations

The thesis outcomes has raised some essential questions. This concluding section highlights some of these, in terms of addressing them as areas needing further investigation. A list of continuing investigations in the current field of study is listed below.

- Despite steady state cutting temperature $T_{ss}$ readings indicating heat generation will not play a significant role, effects of cutting temperature on microstructure should still be investigated for validation. Particularly for the heat generated in the secondary shear zone. A thermal model should be developed and incorporated into an FE cutting model.

- Austenite softening mechanisms in duplex stainless steel alloys should be investigated further. An investigation characterising the mechanisms triggering austenite to soften and determining experimental strain threshold value $\varepsilon_{max}$ would benefit the current model in gaining further accuracy and credibility. A potential candidate for the investigation method for the characterisation study would be the sourcing of a Transmission Electron Microscope (TEM) detector, which would be able to probe the microstructure under higher resolution and identify these mechanisms. These chip root samples should be with compared with highly strained microstructures developed under rolling or other simple mechanical loading applications.

- An investigation on tool coatings and its influence on tool-life in machining duplex, particularly the tool coatings effect on adhesion wear, would be of significant interest. Machinability drilling trials revealed, current tool coatings from recommended tools to be ineffective towards limiting the appearance of BUE in drilling duplex, despite the coating containing an amount of Titanium aluminium nitride (TiAlN) which is known to produce a protective oxide $\text{Al}_2\text{O}_3$, film under high temperatures.

- The generation of frozen chip root samples containing fully developed built-up layer region would hold significant data towards understanding the full life-
cycle of the BUE, from the generation in the stagnation zone to the parting from the chip region.

- A frictional model should be developed for the duplex microstructure and validated for modelling friction between the chip-tool and work-tool interfaces, for FE cutting analysis.
Bibliography


APPENDIX

A.1 Calculating strain elements

If we take element ‘e’ highlighted in Figure 4.1 it is defined by the position of its 3 nodes \((x_i,y_i)\) for \(i\) and similar for nodes \(j\) and \(k\). Therefore if we consider calculating the strain in element ‘e’ caused by applied external loads \(F_1, F_2\) and \(F_3\), firstly we would need to consolidate affected variables. The external loads would trigger a displacement ‘\(u\)’ in each node e.g. \((u_{x,i}u_{y,i})\) for \(i\). Similarly the surrounding elements would transmute the applied loading into equivalent forces in the \(x\) and \(y\) directions \((F_{x,i}F_{y,i})\).

![Figure A.1 Basic concept of a finite element representation of a thin-wall plate](image)

The displacement is first determined by interpolation according to the following

\[
\begin{align*}
\mathbf{u}_x &= a_1 + a_2 x + a_3 y ; & \mathbf{u}_y &= a_4 + a_5 x + a_6 y \\
\end{align*}
\]

Equation A1.1

The coefficients \(a_1\) \(a_2\) and strain is defined as the rate of change of displacement, therefore taking the partial derivate

\[
\varepsilon_{xx} = \frac{\partial u_x}{\partial x} = a_2 = \frac{(y_j-y_k)u_{x,i}+(y_k-y_i)u_{x,j}+(y_i-y_j)u_{x,k}}{2\Delta}
\]

Equation A1.2
Similarly for other strains $\varepsilon_{yy}$ and $\gamma_{xy}$ the solutions would be as follows, in corresponding to matrix algebra which provides a more structured layout for writing.

$$\begin{pmatrix} \varepsilon_{xx} \\ \varepsilon_{yy} \\ \gamma_{xy} \end{pmatrix} = \frac{1}{2\Delta} \begin{bmatrix} y_j - y_k & 0 & y_k - y_i & 0 & y_i - y_j & 0 \\ 0 & x_k - x_j & 0 & x_i - x_k & 0 & x_j - x_i \\ x_k - x_j & y_j - y_k & x_i - x_k & y_k - y_i & x_j - x_i & y_i - y_j \end{bmatrix} \begin{pmatrix} u_{x,i} \\ u_{y,i} \\ u_{x,j} \\ u_{y,j} \\ u_{x,k} \\ u_{y,k} \end{pmatrix}$$

Equation A1.3

Or Equation A1.3 is described in a simpler expression as

$$\{\varepsilon\}_{element} = [B]_{element} \{u\}_{element}$$

Equation A1.3b

The $B$-matrix is an important term, often referred to as the ‘strain displacement matrix’ and is affected by the position of the nodes and thus the shape of the element, as shown in Equation A.3.
A.2 Machine turning chips

Different speeds and feeds had an impact on the types of chips produced, as shown in Table 3. All chips however, displayed a similar segmentation profile, which is typical in stainless steel chips. The chip types produced for SAF 2205 was to be more stable. At the 0.15mm/rev feed, the chips produced short to long snarled at 94m/min, then long tubular chips was found at 65m/min. These modes showed the chips were produced from mostly up-curling off the tool rake face. At higher feed 0.2mm/rev, discontinuous arc chips were produced. These were considered more desirable and are produced from combination of up and side-curling. The arc chips were observed at both 74m/min and 48m/min.

<table>
<thead>
<tr>
<th>Table A2.1 Types of chips produced at the indicated machine settings</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>f = 0.15mm/rev</strong></td>
</tr>
<tr>
<td>V = 94m/min</td>
</tr>
<tr>
<td><img src="image1" alt="Image of SAF 2205 chips at 94m/min" /></td>
</tr>
<tr>
<td><img src="image5" alt="Image of SAF 2507 chips at 94m/min" /></td>
</tr>
</tbody>
</table>

Turning the SAF 2507 alloy produced a different set of chip profiles. Cutting chips were found more unstable. Some parameters saw the generation of multiple chip modes. E.g. long washer, conical helical and snarled chip modes were found to be generated at one single setting, observed at 94, 65, and 74m/min, shown in Table A2.1. At 48 m/min SAF 2507 produced a long straight curling profile. It should be mentioned, all chips produced in turning SAF 2507 were undesirable.
A.3 EBSD phase mapping of stagnation zone region

This section displays phase mapping data of two collections, refer to Chapter 4.2.

- Duplex 2205 and 2507 chip root samples produced at $V=74\text{m/min}$ and $48\text{m/min}$.
- Duplex 2205 and 2507 phase map of original As-Supplied microstructure

**2205 Chip root sample frozen at $V= 74\text{m/min}$**

![EBSD phase mapping image](image-url)
2507 Chip root sample frozen at V=74m/min
2205 Chip root sample frozen at V= 48m/min
2507 Chip root sample frozen at V= 48m/min
Phase map of original 2205 microstructure in As-supplied condition
Phase map of original 2507 microstructure in As-supplied condition
A.4 Modified Ramberg Osgood model

This section lists the material phase property input data for the modelling of strength curves using the modified Ramberg Osgood model, displayed previously in Equation 5.2. Accounting for true stress, Equation 5.2 becomes the following.

\[
\sigma_{\text{true}} = \begin{cases} 
(\varepsilon \sigma_0)(1 + \varepsilon) & \text{for } \sigma \leq \sigma_y \\
\left(\sigma_u - \sigma_{0.2}\right)e^{\left(\frac{\ln(\varepsilon)}{m} + \sigma_{0.2}\right)}(1 + \varepsilon) & \text{for } \sigma > \sigma_y
\end{cases}
\]  

Equation A4.1

Table A4.1 displays a list of input parameters used to model the austenite and ferrite strength curves. All parameters apart from the \( m \) exponent was determined based on experimental tensile data and Nano-indentation data mentioned in Chapter 5.1.2. The generated material curves are displayed in Figure A4.1. The plastic region was submitted to ABAQUS for analysis.

<table>
<thead>
<tr>
<th>Properties</th>
<th>2205</th>
<th>2507</th>
</tr>
</thead>
<tbody>
<tr>
<td>( E_\text{(GPa)} )</td>
<td>133</td>
<td>132</td>
</tr>
<tr>
<td>( E_{02}(\text{GPa}) )</td>
<td>0.745</td>
<td>0.738</td>
</tr>
<tr>
<td>( \sigma_{0.2}(\text{MPa}) )</td>
<td>667</td>
<td>661</td>
</tr>
<tr>
<td>( \sigma_\text{u}(\text{MPa}) )</td>
<td>830</td>
<td>776</td>
</tr>
<tr>
<td>( \varepsilon_y )</td>
<td>0.0054</td>
<td>0.0054</td>
</tr>
<tr>
<td>( \varepsilon_\text{u} )</td>
<td>0.204</td>
<td>0.204</td>
</tr>
<tr>
<td>Constants</td>
<td></td>
<td></td>
</tr>
<tr>
<td>( m )</td>
<td>3</td>
<td>4</td>
</tr>
<tr>
<td>( \alpha )</td>
<td>1.22E+37</td>
<td>8.76E+16</td>
</tr>
<tr>
<td>( \beta )</td>
<td>12.95</td>
<td>3.95</td>
</tr>
</tbody>
</table>

Recapping where \( \alpha \) and \( \beta \) are equal to

\[
\alpha = \frac{E \varepsilon_0 - \sigma_0}{\varepsilon_0^\beta} \quad \text{Equation 5.3}
\]

\[
\beta = \frac{E \varepsilon_0 - \sigma_0}{\varepsilon_0^\beta} \quad \text{Equation 5.4}
\]
Figure A4.1 Modelled strength curves for material phase austenite and ferrite, in duplex stainless steel alloys (a) 2205 and (b) 2507.
A.5 Finite element model

This section displays Von Mises Stress distribution contour plots for duplex 2205 and 2507 at two cutting speeds 48/min and 74m/min. zone cut, discussed in Chapter 5.

A5.1 2205 model cutting at $V=48$m/min

All legend units in (MPa)

Figure A5.1 Contour plots of sequence 1 – 3 of 6, 2205 model, $V=48$m/min
Figure A5.2 Contour plots of sequence 4 – 6 of 6, 2205 model, $V=48$m/min
A5.2 2507 model cutting at $V=48\text{m/min}$

Figure A5.3 Contour plots of sequence 1 – 3 of 7, 2507 model, $V=48\text{m/min}$
Figure A5.4 Contour plots of sequence 4 – 6 of 7, 2507 model, V= 48m/min
Figure A5.5 Contour plots of sequence 7 of 7, 2507 model, $V = 48$ m/min
A5.3 2205 model cutting at $V=74\text{m/min}$

Figure A5.6 Contour plots of sequence 1 – 3 of 6, 2205 model, $V=74\text{m/min}$
Figure A5.7 Contour plots of sequence 4 – 6 of 6, 2205 model, $V=74\text{m/min}$
A5.4 2507 model cutting at $V=74\text{m/min}$

Figure A5.8 Contour plots of sequence 1 – 3 of 6, 2507 model, $V=74\text{m/min}$
Figure A5.9 Contour plots of sequence 4 – 6 of 6, 2507 model, $V=74\text{m/min}$
A5.5 Stagnation zone simulation

2205 Model cutting at $V=74\text{m/min}$

Figure A5.10 BUE Contour plots of sequence 1 – 3 of 12, 2205 model, $V=74\text{m/min}$
Figure A5.11 Contour plots of sequence 4 – 6 of 12, 2205 model, $V=74\text{m/min}$
Figure A5.12 Contour plots of sequence 7 – 9 of 12, 2205 model, $V=74$m/min
Figure A5.13 Contour plots of sequence 10 – 12 of 12, 2205 model, $V = 74\text{m/min}$