Mechanisms of squat initiation and propagation on rail surfaces

Ali Ahmed Ali Al-Juboori

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Mechanisms of squat initiation and propagation on rail surfaces

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Abstract

Squats are recognized as an important rail track issue and occur in many railway networks across the world. Consequently, squats exacerbate the deterioration of track and some vehicle components. Associated excessive noise and vibration have increased the complaints against the rail operators. The area of squat initiation is still a matter of controversy. Research efforts have included possible relationships between the narrow and brittle white etching layers (WELs) observed on rail surfaces and squat formation in rails. However, based on current literature, the nature of WELs and their formation mechanisms are not fully understood, as is their relationship to the squat formation. In order to implement a more effective preventive and early corrective management strategy, it is critical to determine the root causes of squat defects and understand the mechanism of squat initiation and subsequent crack propagation on the rail surface.

In this thesis, the damaged ex-service rails containing various stages of squats and WELs have been comprehensively investigated by field examination; advanced experimental techniques including Gleeble Thermal-Mechanical experiment, High Pressure Torsion (HPT) test and Leaking current pulsed GTAW experiment; and high resolution metallurgical analysis such as Focused Ion Beam instrument (FIB)/Transmission electron microscopes (TEM), Scanning Transmission electron microscopes (STEM) and Synchrotron X-ray Diffraction etc.

Current work confirms that there is a correlation between squat formation and the occurrence of WELs on the rail surface. Synchrotron XRD results obtained from squat defects and WELs at the vicinity regions of squats are consistent, which provides a clear link between squats and WELs. Microstructure observations found that the cracks initiated from WELs, extending down into the rail to form the squats.

It has been found that there are two distinguishable types of WELs based on different operational conditions. At the heavy braking rail regions, WELs consists of martensite and retained austenite; and undeformed pearlite exists at the transition zone between the WELs layer and steel matrix. This type of WEL is likely induced by temperature and pressure changes; therefore, it was referred as TP-WEL. The existence of TP-WEL links with a new type of squat (stud). At the steady traffic regions (low braking utilisation), WELs contains nanocrystalline plate martensite, fine ferrite and finely fragmented
cementite, which is caused by severe plastic deformation. This type of WEL is termed as SD-WEL which can initiate squat defects.

Nature and formation mechanism of WELs were investigated by Gleeble Thermal-Mechanical experiment, High Pressure Torsion (HPT) test and Leaking current pulsed GTAW experiment. It was found that (i) arcing from leaking current between wheel and rail is considered as an alternative formation mechanism of WELs based on thermal phase transformation; (ii) WEL occurs under accumulated severe shear deformation by rolling contact fatigue in rail surface; (iii) WELs can be formed below critical pearlite to austenite transformation temperature under a combination of thermal input and high contact pressure.

A localized section of rail associated with water dropping from an air-conditioning system located in a tunnel roof represents an ideal controlled location to examine the formation of WELs under both dry and wet conditions and the hypothesis of the hydraulic entrapment of water on crack propagation. The presence of oxides associated with an absence of shear deformation on the cracks faces reveals that crack growth was driven by Mode I loading (tensile opening). Stress corrosion cracking was involved in both crack growth and the observed formation of secondary cracks. Current work provides strong evidence that the presence of water plays a significant role in crack growth and development under the service condition.

A case of squat-induced rail fracture failure reveals that the cracks can turn down and grow into the parent rail on a transverse plane. However, the abnormal microstructure in rail steel, such as brittle martensitic patches and inclusions due to the un-proper heating treatment during the rail manufacturing process, results in the rail fracture failure at the low tonnage.
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1. Introduction

1.1 Preamble

Squats are recognized as an important rail track issue caused by either mechanical and/or thermal actions inducted on rails by wheels. Squat defects can result in rail surface degradation, representing a significant challenge for railways track management around the world. Thus, squats need to be treated by rail grinding when they are at an early stage of development; otherwise, a replacement process has to be eventually applied to the damaged rails resulting in a significant cost in railway networks.

Squats are railhead defects containing cracks. The cracks are developed on the running band surface and gauge corner of rails, which usually propagate down in the longitudinal and lateral direction across the running band area within shallow angle to the rail surface. The subsurface cracking causes a lateral decline of the area overlapping crack planes, which results in local depression. Squat depression increases the vertical dynamic load of wheelsets on rails, which consequently exacerbates the deterioration of the track structure and other vehicle components. In some cases, squats can develop the most dangerous defects called transverse defect (TD) which potentially results in completely rail failure. This is the main risk correlated with squat defects concerned by the railway's management across the world, especially with increasing numbers of squats in the recent few years. Thus, an investigation into squats is important, particularly in relation to the initiation and root causes of the defects, propagation and fracture failure mechanism.

Squats formation is often associated with surface initiated rolling contact fatigue (RCF) cracks. However, research efforts have included the study of the relationships between the narrow and brittle white etching layers (WELs) observed on rail surfaces and squat formation in rails. This type of surface layer has been found at rail regions where squat occurs or in the vicinity of squat type defects. In some research, WELs are thought to be created by severe plastic deformation on the rail surface. However, WELs are also thought to be induced by the thermal process between wheel and rail. This finding is mainly correlated with the absence of plastic deformation underlying the pearlite microstructure. This means that the mechanism of RCF has not involved in WEL formation. This type of white etching layer has been considered the main cause of crack initiation and thermal damage called “Stud” which is quite similar to squat defects.
1.2 Research significance

1- Although squats have been investigated previously, the area of squat initiation is still a matter of controversy. Most research into squat cracks initiation was conducted under the vast majority classification of rolling contact fatigue damage. Recently, a relationship between the mechanics of squat cracks formation and the occurrence of white etching layers on the surface of rail has been reported and discussed. However, it is still lacking solid metallurgical evidence from microstructural examination and phase analysis to support a correlation between WELs and squats formation, which was conducted in detail on ex-service rails.

2- There is an argument about the intrinsic nature of the WELs. Although WELs in rail steels have been investigated for many years, the interpretation of their exact structure, nature, and formation mechanisms are still not fully obtained. For example, theoretical research hypothesised that: (i) WELs are the result of the thermal process caused by wheel slip; (ii) WELs are mechanically induced by severe plastic deformation; (iii) WELs are caused by a combination of thermo-mechanical process. Based on these perceived mechanisms of formation, the microstructural nature of WELs has been interpreted differently. Therefore, there is no general agreement about white etching layers formation in rails. Understanding of the nature and formation of WELs in rails is very essential because of its perceived link with rail damage and the high cost of track maintenance procedures such as rail grinding designed to remove the damaged surface on rail.

3- Numerous researchers have concentrated on the mechanics of squat cracks propagation and growth. The influences of dry and/or wet conditions on crack growth have been investigated and several hypotheses and theories such as “dry”, “wet lubricated”, “hydraulic pressurisation”, “liquid entrapment” and “squeeze fluid film” have been proposed. Although numerical modeling and laboratory tests support the hypothesis of hydraulic entrapment on squat growth, there is no report on the direct metallurgical examination of rails damaged by the presence of water under service conditions in a rail track network. The requirement for fluid
to cause small cracks to propagate provides a rational of why squats vanish as soon as the railway enters a tunnel. The questions are being raised concerning the physical evidence of water entrapped in cracks for the in-service track, and what evidence if any, is there that the surface water promotes crack growth? In addition to the mechanics of hydraulic entrapment on crack propagation, is there any other likely mechanisms operative?

It is reported that squats are possible to induce rail fracture failure in a rare case. If the secondary cracks branch from the ordinary cracks (leading and trailing) and grow down on the transverse plane of the railhead, a most dangerous transverse defect (TD) can be formed leading to rail breakage. However, there is no study providing a comprehensive understanding of why and how squat cracks development into the transverse defects and the role of pre-existed metallurgical defects in rail steels on the formation of TD cracking.

1.2 Aim and specific objectives

The aim and specific objectives of this study are to:

1- Investigate the possible linking between squat formation with the occurrence of white etching layers on the surface of rails.
2- Provide a fundamental understanding of the intrinsic nature of the WELs.
3- Examine the root causes of white etching layer formation in rails.
4- Investigate the formation mechanism of the white etching layer in rails.
5- Determine the role of water on the mechanics of squat cracks growth and development
6- Identify the root causes of squat-induced rail fracture failure.

1.2 Methodology

In order to achieve the main objectives of this research, the following procedure will be implemented.
- The ex-service damaged rails containing the defects including squats, white etching layers and wears at various damage stages were used for investigation. Based on the degree of damage, samples were selected and sectioned. Structural and microstructural evaluation of the surface and subsurface layer of squats and the adjacent region were characterised using high-resolution techniques, including advanced methods of optical and electron microscopies, and synchrotron diffraction. Detailed the intrinsic and precise nature of the WELs taken from different rails were examined and discussed.

- Laboratory experiments were carried out on the rail samples, which were sectioned from the railhead of the ex-service rails, and then machined based on equipment requirements and standardization. The experiments are:

  - Electrical leakage current experiment.
  - High pressure torsion test.
  - Thermo-Mechanical simulation using the Gleeble system.
  - Fatigue crack growth experiment.

- All facilities and equipment are available at the University of Wollongong (UOW) except High Pressure Torsion machine which is available at The University of Sydney.

- Results from the structural and microstructural investigation on the damaged rails and from lab experiments were analysed, discussed and evaluated.

1.5 Experimental and laboratory instrument

The following section briefly introduces the experimental instruments and techniques used in this study, and the experimental procedure and sample preparation methods are described in detail.

1.5.1 Gas Tungsten Arc Welding (GTAW) machine

Gas tungsten Arc welding (GTAW) machine was used in this experiment to generate leaking current. The experiment was used to investigate the formation mechanism of the
white etching layer under the leaking current (arcing). In railway operation, the phenomenon of arcing has been often occurring between the wheel and the rail. Therefore, the GTAW machine is considered a useful technique for processing arcing phenomena. Figure 1.1 shows a GTAW machine before and during the test. The experiment was conducted at the University of Wollongong.

![GTAW machine with setting up samples and during testing](image)

**Figure 1.1** GTAW machine with setting up samples and during testing.

### 1.5.2 High Pressure Torsion (HPT) machine

High Pressure Torsion (HPT) machine was used to validate the formation mechanism of WELs by severe plastic deformation. The experiment was proposed to examine whether white etching layers can be induced by severe plastic deformation. HPT is the most effective technique for processing severe plastic deformation because the material is subjected to compressive hydrostatic stress and simultaneous an extremely large torsional straining. HPT was used because the loading condition is similar to that of the rail-wheel contact, such as high contact pressure and severe plastic deformation. Therefore, a microstructure change induced by HPT is similar to that found in the plastically deformed surface layer of rails. Figure 1.2 shows an HPT machine that is located at The University of Sydney.

![HPT machine](image)
1.5.3 Gleeble simulation system

A Gleeble 3500 thermo-mechanical simulation system was used for fast heating and quenching treatment with a combination of high pressure contact experiment. The machine was used to simulate the white etching layer on the rail surface under the thermal process combined with the high contact pressure. The contact condition at the rail and wheel interface is very complicated. Large normal contact pressure is applied on the rail by the wheels during rolling contact. Along with the compressive load, heat is generated at the wheel/rail contact zone during train passage, resulting in a temperature rise of the material. The combination of heat and pressure results in surface structural evolution and phase change. Thus, the Gleeble thermo-mechanical simulation system can successfully simulate the realistic wheel/rail contact loadings conditions. Photos of the Gleeble 3500 simulation system located at the University of Wollongong is shown in Figure 1.3.
1.5.4 Fatigue testing machine

A high-performance Instron 8801 fatigue and the tensile tester were used for fatigue life experiments. The 8801 servo-hydraulic uniaxial testing equipment is used for a wide range of dynamic and static loading testing. The machine was used to determine fatigue crack propagation of the samples made from two different rails, and to examine how the pre-existed inclusions in the rail material affect fatigue crack growth. Figure 1.4 shows the front view of Fatigue and Tensile tester that is located at the University of Wollongong.

Figure 1.4 Instron 8801 fatigue and tensile tester.

1.5.5 Sample preparation for metallography

1.5.5.1 Cutting, sectioning and machining samples

In this work, cutting, sectioning and machining samples were performed by using different techniques based on the area of interest, section direction and the size of samples. The rails 2-3 meters in length, provided by railway operators, were first cut into shorter sections that were considered the area of interest for investigation. In this stage, the samples were only cut through the transverse direction and by using a semi-automatic Parkanson advanced cutting machine. Further sectioning in lateral and transverse directions and machining standard samples for experiments were performed by using Wire EDM machine and Waterjets cutting machine. For metallography examination, samples were prepared by cutting into small pieces using Struers Accutom precision cut-
off machine. The figure shows all the cutting machines mentioned above and used in this work. The machines are located at the University of Wollongong.

![cutting machines](image)

Figure 1.5 cutting machines used in this work.

1.5.5.2 Mounting machine

The small pieces interested for metallurgical investigation were mounted in Polyfast powder (conductive resin), by using Struers Citopress 20 hot mounting instrument, as shown in Figure 1.6.
1.5.5.3 Grinding and polishing machine

After the mounting process, the samples were grounded and polished by using a Struers Tegramin grinding/polishing machine shown in Figure 1.7. The grinding and polishing procedure used are given in Table 1.1. After polishing, the samples were etched by using a 2-2.5% Nital etchant solution to reveal the microstructure.
1.5.5.4 Optical and stereo microscope

Optical and stereo microscopes were used to examine the microstructure of the material under visible light with different magnification. The Stereomicroscope was mainly used for low magnification observation such as a full view of squat cracks development and growth underneath the surface layer. Figure 1.8 shows a Leica DM6000 optical microscope and Leica M205A stereo microscope that were used in this study.

<table>
<thead>
<tr>
<th>procedure</th>
<th>Abrasive &amp; cloth</th>
<th>Time, minute</th>
<th>lubricant</th>
</tr>
</thead>
<tbody>
<tr>
<td>grinding</td>
<td>220#, 500#, 1200# SiC paper</td>
<td>2 minute each</td>
<td>water</td>
</tr>
<tr>
<td>Polishing</td>
<td>9 μm Largo cloth</td>
<td>5</td>
<td>Green -base lubricant</td>
</tr>
<tr>
<td>Polishing</td>
<td>6 μm Dac cloth</td>
<td>5</td>
<td>Green -base lubricant</td>
</tr>
<tr>
<td>Polishing</td>
<td>3 μm Dac cloth</td>
<td>5</td>
<td>Green -base lubricant</td>
</tr>
<tr>
<td>Polishing</td>
<td>1 μm Dur cloth</td>
<td>8</td>
<td>Green -base lubricant</td>
</tr>
</tbody>
</table>

Figure 1.8 Leica DM6000 optical microscope and Leica M205A stereo microscope.

1.5.5.5 Scanning electron microscopes (SEM)

Two types of scanning electron microscopies SEM were used in this study: JEOL JSM-6490LV SEM and JEOL JSM-7001F FEG-SEM. The JEOL JSM-6490LV SEM is a 30 kV conventional tungsten filament, variable pressure scanning electron microscope, and was used to characterise the cracks and texture evolution that is induced in the surface and subsurface layers or rails caused by rolling/sliding contact loadings, as well as to
characterise the fracture surface of the broken rails. In addition, it was also used for Energy-dispersive (EDS) X-ray spectroscopy analysis.

The JEOL 7001F FEG-SEM is a 30 kV analytical thermal field emission gun scanning electron microscope capable of 3 nm spatial resolution and was used for Transmission Kikuchi Diffraction (TKD) analysis. TKD is an advanced technique for detailed crystallographic analysis. JEOL 7001F FEG-SEM was also used to obtain high-resolution imaging. Figure 1.9 shows JEOL 6490 SEM and 7001F FEG-SEM that are located at the University of Wollongong.

![JEOL 6490 SEM and 7001F FEG-SEM](image)

Figure 1.9 JEOL 6490 SEM and 7001F FEG-SEM.

### 1.5.5.6 Transmission electron microscopies (TEM)

The samples were further characterised by transmission electron microscopes (TEM) to provide detailed information about the microstructural features and crystal structure in high magnification, as well as to identify the phase and the composition of the sample in nano-scale.

Two types of TEM microscopes were used in this work: JEOL JEM-2011 and JEOL ARM-200F. The JEOL JEM2011 is a 200 kV conventional LaB$_6$ transmission electron microscope with a spatial resolution of 0.16 nm. The JEOL JEM-ARM200F is a 200kV probe corrected scanning transmission electron microscope (STEM) capable of atomic resolution imaging. The TEM instruments shown in Figure 1.10 are located at the University of Wollongong.
1.5.5.7 Focused Ion Beam instrument (FIB)

The FEI Helios NanoLab G3 CX Focused Ion Beam (FIB) milling instrument was used to prepare a transparent lamella for high-resolution TEM observation and for TKD SEM analysis. The FIB instrument is a state-of-the-art dual (electron and ion) beam analytical scanning electron microscope and highly targeted sectioning tool. Figure 1.11 shows the FIB instrument that is located at the University of Wollongong and used in this work.

![Figure 1.11 FEI Helios NanoLab G3 CX Focused Ion Beam (FIB) milling instrument.](image-url)
1.5.5.8 Hardness tester

Two kinds of hardness tester were used in this work: Macro identical Vickers hardness tester and Digital Micro Vickers hardness tester (TH715) as shown in Figure 1.12. The hardness measurement of the surface of the rail was performed by using a Macro hardness tester with an identical load of 1000 g. The hardness measurement across the etched section was performed using Micro-hardness tester with an identical load of 100g. The reported values were taken from an average of three indentations measurements.

![Macro identic Vickers hardness tester](image1.png) ![Digital Micro Vickers hardness tester (TH715)](image2.png)

Figure 1.12 Maco and microhardness tester used in this work.

1.5.5.9 X-ray diffraction techniques

In this study, the structure and phase analysis was performed using High-Resolution Synchrotron X-ray Diffraction and a conventional GBC MMA X-Ray diffraction unit. High-resolution synchrotron X-ray experiments were performed using the Powder Diffraction (PD) beamline at the Australian Synchrotron, Victoria, Australia. The beamline was used in glancing incidence geometry with a monochromatic X-ray beam energy 15 keV (λ=0.8265 Å). Phase analysis was performed in- range between 20° and 100°. Depending on the sample size and the thickness of WELs, different beam size of 0.1 x 0.5 mm, 0.2 x 0.5 mm (Vertical x Horizontal) and glancing angles (ω) of 0.2°, 0.5°, 1° and 2° were used for the acquisition of diffraction data (Bragg angle 20°≤ 2θ ≤ 100°) using the Mythen detector with exposure time in the range of 0.5-8 min.
A conventional X-ray diffractometer with Cu Ka (wavelength = 1.54056Å) radiation was used for thin film and phase analysis. Figure 1.13 shows both X-ray instruments used.

![Figure 1.13 High-Resolution Synchrotron X-ray Diffraction and conventional GBC MMA X-Ray diffraction unit.](image)

**1.6 Thesis outline**

This thesis is organised into ten chapters as described below;

Chapter 1 is the preliminary chapter, which presents the introduction of the research, research significance, the aim and specific objectives of the study, methodology, and instruments used in this work and finally provides a brief outline of the thesis.

Chapter 2 is the literature review, which presents a comprehensive overview of squat defects in rails. In this chapter, detailed information about the mechanics of squat initiation, propagation and rail fracture is provided. Factors that affect squats to form and
further developing are discussed. Methods and hypotheses focused on squat cracks growth were presented as well.

Chapters 3 to 9 are the experimental results and discussion. Chapter 3-7 considers the mechanism of squat initiation; Chapter 8 considers the mechanics of squat propagation; Chapter 9 considers the mechanics of squat fracture failure.

Chapter 3 presents the investigation on the squat’s formation and its correlation with the occurrence of the white etching layer on the surface of rail steel. In this chapter, the experimental work is directly conducted on the ex-service damaged rails affected by squats and white etching layer. All the test results were presented and discussed accordingly.

Chapter 4 describes the microstructure features, compositions and phase transformation of white etching layers obtained from two different re-railing sites. The results were then characterised, compared and discussed.

Chapter 5-7 are laboratory experiments. Chapter 5 introduces an alternative formation mechanics of the white etching layer on the railhead based on thermal phase transformation caused by leaking current (Arcing phenomenon). A comparison between acing induced WELs and Rail WELs were made and discussed. Chapter 6 presents the formation mechanism of the white etching layer due to severe plastic deformation. A comparison between deformation-induced WELs and Rail WELs was made and discussed. Chapter 7 presents the formation mechanism of the white etching layer under the application of thermal and mechanical loadings conditions. The results were characterised and discussed.

Chapter 8 presents the role of water entrapment mechanism on the growth and development of the squat crack. The experimental work was conducted on the ex-service damaged rails containing defects from water dropping. Detailed information was presented and discussed.

Chapter 9 presents the investigation of the causes of squat cracks induced fracture failure. The experimental work was conducted on a specific broken rail caused by squat. In addition, the laboratory experiment associated with the fracture failure and fatigue life
was also presented and included in this chapter. The test results were thoroughly presented and discussed.

Chapter 10 is the final chapter which discusses the overall conclusions of this research and provides recommendations for future research.
2. Literature review

In this chapter, a comprehensive literature review on squat defects is provided, including the mechanics of squat cracks initiation, propagation, and fracture failure. The historical background of squat defects is extensively reviewed, along the surface and subsurface cracking characterisation, hypotheses, and theories proposed for squats formation on rails.

2.1 Background

Squats are rail surface cracking defects and considered a significant problem that damage railway system around the world. For instance, in Dutch railways, there are about 22% of ProRail (1405 km length of rail) damaged by squats (Badcock 2008). In Australia, Sydney Trains (previously called RailCorp) network announced that a large proportion (about 18%) of rail was affected by squats in 2008 (Kerr et al. 2008). Currently, there is a considerable increase in squats number to be reported in 2014 that more than 20 km of track within the Sydney Trains (ST) system has classified as un-detected rail (squat affected rail) (Grassie 2014). In addition, the repair of rails damaged by squats is economically costly. For example, squat-associated expenses are more than 5000 Euro per kilometer annually in Europe (Molodova et al. 2014). They also cost about 15,000 euro/km of each replacement and welding process in another railway network (Li, Zhao, Dollevoet, et al. 2008).

The major risk with squats is that they may result in rail fracture. If the secondary cracks branch from the ordinary cracks (leading and trailing) and grow down on the transverse plane of the railhead, a dangerous defect, called transverse defect (TD) can be generated, which leads to a potential rail fracture failure. This defect can be prompted by the inclusions that existed inside the rail material due to steel manufacturing (Cannon et al. 2003). Therefore, it should be detected early for proper rail maintenance otherwise the replacement of the damaged section is essential. Although these transverse defects are relatively infrequent, five cases were recorded in the Sydney Trains system in 2012 (Wilson 2012). Moreover, squat cracks propagated horizontally at the rail sub-surface can shield the ultrasonic signals to detect dangerous TD deeper cracks during a normal examination. Furthermore, the track components can also be worsened by squats due to high frequencies of impact load caused by surface local depression. For example, Kaewunruen et al. (2015) carried out field investigation on the ex-service rail containing
Squats using axle box acceleration. Data analysis showed that high vibration and dynamic frequency were determined at squats region, confirming that a significant wheel dynamic loading was caused by squats with possible transmitted to track settlement. In addition, squats can cause environmental damage by raising the noise level. For instance, the noise level increases over 10 dBA when the wheels pass over 7-8 meters of rail section containing squats (Kerr et al. 2008). The increasing noise level might disturb people who are living close to the railway. For example, a collective group of people in Woolloomooloo, Sydney, noticed a particular noise exacerbated by squats damage (RailCorp Noise and Vibration Unit 2011).

Squat defects can usually find on all kinds of track system. They can damage all lines such as passenger, freight or mixed, heavy haul, tram and metro lines, tracks with concrete or wooden sleepers, slab and ballast (Li 2009; Zoeteman et al. 2014). Additionally, squats can occur in the tangent and curved track, low and up rails, transitions, signals, turnout locations, and in Standard Carbon (SC) and Head Hardened (HH) rail material.

Squats are railhead defects containing cracks that have initiated at the rail surface or subsurface and propagated down in the longitudinal and lateral directions of the railhead. These cracks usually follow the sheared layer of the rail surface material and they are strongly influenced by a high traction ratio and brake effort in addition to high contact, thermal, residual and bending stresses. Also, it has been found that water entrapment plays a crucial role in crack growth.

Based on research, several studies and investigations have been conducted to clarify squats and their root causes. Clayton and Allery (1982) presented the metallurgical characteristics and site investigations of squat defects. This study showed that squats are surface-initiated defects resulting from severe plastic deformation due to high frequent rolling contact fatigue. In addition, small indentations produced in the rail by foreign objects inserted between the wheels and the rails during normal track operation services have a considerable impact on the initiation of squat cracks.

In Japan, the historical background of Shinkansen shelling defect which is completely similar to squats was displayed and their causes, growth, and detection were reviewed by Kondo et al. (1996). They found that the presence of horizontal cracks under the rail surface leads to formation a surface shelling defects and in some cases, the horizontal
cracks have changed direction into transverse planes resulting in transverse defects (TDs) which potentially terminates with rail breakage. They also considered that the causes of these defects were due to rolling contact fatigue and the rate growth of shelling defect raised with passed tonnage.

In Australia, an extensive review of the squat and its related defect have been reported by Marich and his colleagues (2008; 2001). They classified the squat defect into two types; running surface squat and gauge corner squat. In addition, it was proposed that squat growth is probably driven by excessive sub-surface shear stresses. Furthermore, it was claimed that the white etching layer observed in their samples is more likely to be formed by an adiabatic shear mechanism caused by micro slip under traction.

Bogdański (2005; 2014) and Bogdański et al (2008; 2005; 1996; 1998) embedded a series of numerical and modeling analysis to explore the possible mechanism of crack propagation and growth in the squats. They focused on some factors that influence on the crack growth rate, particularly the effect of entrapped liquid. It was concluded that the mechanisms of crack growth are correlated with (1) reduced friction inside the crack faces (2) hydraulic pressure exerted on the crack faces (3) fluid entrapment in the crack interior or the squeeze fluid film.

In the Dutch railway system, Li and his colleagues (2009; 2011; 2008; 2008; 2013; 2014) concentrated on the mechanism of squats initiation and growth as well as their root causes. Data from field monitoring and observations of squats were analysed and compared with the FEM simulations. The results showed that squats are grown by high dynamic contact loads due to surface irregularities.

Mechanical and metallurgical investigation on rail type R260Mn containing squats was conducted by Steenbergen and Dollevoet (2013; 2013). The study was focusing on squats origination and growth. It was found that that the directions of tangential shear stress and plastic deformation flow located at the subsurface of the rail play a crucial role to originate the leading and trailing cracks. It addition, it was found that the formation of a leading crack is orthogonal to the resultant of tangential shear component and it is deeper and longer than trailing crack.

White Etching Layer (or WEL for abbreviation), usually observed on the rail surface, is often formed near the squats defects and it is expected to have a critical role in squats
formation on the rail surface (Clayton & Allery 1982; Kerr et al. 2008). Normally, WELs are characterised by a white appearance and sharp edges under light microscopy. Several studies have been conducted to investigate the WELs, however; formation mechanics, root causes and structure composition are still a matter of controversy (Baumann et al. 1996; Clayton & Allery 1982; Djahanbakhsh et al. 2001; Jirásková et al. 2005; Kerr et al. 2008; Lojkowski et al. 2001; Newcomb & Stobbs 1984; Österle et al. 2001; Pyzalla et al. 2001; Wang, L et al. 2003; Zhang et al. 2006). Based on these investigations, it was claimed that the WELs are formed due to thermal (rapid heating and cooling) and/ or mechanical processes (severe plastic deformation) induced on the rail surface.

Grassie (2012; 2011) and Grassie et al (2011) reported that Stud defects, which have certain similar characteristics to squat defects, were observed recently on the rail surface and considerably increased. Studs were classified as thermal damage linked with the formation of the white etching layer. Comparatively, studs are not related to rolling contact fatigue as initiation is not associated with severe plastic deformation.

A report on the site work in Sydney Trains (ST) network was published by Grassie (2014). The defects founded on rails were classified as “Studs”. The classification was based on the following phenomena, firstly, Studs are comparatively benign than classical squats. Secondly, Studs can develop quickly down to the surface of rail within a low tonnage of traffic. Thirdly, Studs cannot develop into a transverse defect. However, the metallurgical examination on the ex-service rail of Sydney Trains (Al-Juboori et al. 2017) found that severe plastic deformation was taken place at a region under the white etching layer which is controversy to the definition of a stud.

Finally, several researchers presented a comprehensive literature review associated with squats defects. For example, Cannon et al (2003), Ekberg et al (2014; 2005) and Zerbest et al (2009; 2005) introduced rolling contact fatigue, rail damage, and fracture. Li (2009) and Grassie (2012) reviewed update information about squat defects in the rail surface based on the mechanics of squat formation, field investigation, and monitoring. In spite of extensive studies on squats, the mechanisms of squat cracks initiation, propagation and fracture failure are still not fully understood.
2.2 Characteristics of squat defects

2.2.1 Superficial appearance

Squats normally contain two surface-breaking cracks: leading and trailing. Leading cracks develop in the rolling direction while trailing cracks grew in the opposite direction of traffic. When leading and trailing cracks develop on the rail surface, the U, V, Y or semi elliptical-shaped cracks can be produced, and the opening side of the shape is often toward the gauge corner, as illustrated in Figure 2.1.

![Figure 2.1 Typical squat defect formed on running band area (photon taken from site investigation at Sydney Trains).](image)

Squats defects can be visually recognised due to the formation of some particular features. The main characteristic of squats in addition to surface-breaking cracks is the presence of darkened area on the rail surface, at the running band area or near the gauge corner. The darkening area is due to an accumulation of corrosive products in the depression which
is caused by a lateral decline of the area overlapping crack planes. This area seems to have a certain shape like two lobes that resemble a permanent deformation.

Another fundamental feature of squats is the widening of the running area at the squat location which results from severe plastic deformation induced by a high dynamic load of wheels on the rails.

2.2.2 Surface cracking morphology and growth pattern

Squat development on the rail surface follows three different stages. The first stage of squat is characterised by the formation a single surface-breaking crack (leading crack) on the running band area, with an angle in a range between 45°-90° to the rail axis, in the parallel direction of traffic (Steenbergen, Michaël & Dollevoet, Rolf 2013). The crack length is associated with the growth stage and the crack propagation is moving toward the gauge corner, crossing the running band area. The second stage of squat development is considered when a second crack (tailing crack) branches from crack initiation point and develops toward the gauge corner but in the opposite direction of vehicle motion. The surface cracking pattern of squat, in this case, appears typical, with the formation of two cracks of the V-shaped pattern. The severe cases and most development stage of the squat are that when the two surface-breaking cracks have fully developed on the rail surface, forming another V-shaped crack on the field side of the rail surface.

2.2.3 The subsurface cracking morphology and growth pattern

Under the rail surface, the V-shaped surface-breaking cracks propagate down in the longitudinal and lateral direction of the railhead within a shallow angle of about 15°-30° to the rail surface. The cracks often develop horizontally within a plane depth of about 3-6 mm. (Kerr et al. 2008; Pal, Valente, et al. 2012). Figure 2 shows the schematic illustration of the squat stages of development.
2.2.4 Squats classification

Based on the stages of squat development, squats in Europe are classified into three categories; light, moderate and severe squats (Li et al. 2010). Squats are described as light or early-stage when small defects have initially formed on the rail surface while they are categorized as moderate and severe defects when light defects grow into a typical mature (both leading and trailing cracks develop completely on the rail surface). In Australia, squats stages are divided into mild, moderate, severe and very severe defects (Kerr 2013; Kerr & Wilson 2012; Kerr et al. 2008). The final stage of squat is classified as a very severe defect because a part of the rail surface is lost. Figure 2.3 shows typical examples of the squat at different stages.
Figure 2.3 Typical examples of squats at different stages of development (photos taken from rails supplied to the UOW for investigation).

2.2.5 Surface profile of squats

The surface profile of squat defect at the mature stage can exhibit a W-shape as illustrated in Figure 2.4. It is noticed that the total length of W-shape which corresponds to the two lobes is about 80 mm and the wavelength of squat is about 40 mm. However, in some severe cases, the wavelength could reach 60 mm (Li et al. 2011)
2.2.6 Squat types

On the railhead, there are two types of squats noticed in the Sydney Trains system, including running surface squats and gauge corner squats (Kerr et al. 2008). The first type is squats that form on the running band of the rail surface. The second type of squats is formed on the gauge corner regions and usually developed from rolling contact fatigue cracks. Figure 2.5 shows an example gauge corner squat.
2.3 Squat occurrences and track structure

2.3.1 Squats at a turnout

Turnout is a particular component of the railway track that allows a train to change its direction from the original track. It is an essential and vulnerable part of the railway because transferring from track to another causes high-frequency impact forces excited between wheel and rail. As a result, many defects can occur at turnout locations, particularly those attributed to rolling contact fatigue such as squats. Figure 2.6 shows squats defects at the turnout.

![Squats at turnout](image)

Figure 2.6 Squats at turnout (Daniel 2012).

2.3.2 Squats at sleepers

Sleeper is described as a foundation structure laid under the rails which are used to support track and transfer the traffic loads to the ballast. Sleeper is made from either concrete or timber. Therefore, the foundation stiffness is different due to material type. Squats were found on the rails with both types of the sleeper, but the rate of popularity is relatively various. For example, the data obtained from field investigation at the Sydney Trains networks found that the high density of squats formation was recorded on rails with timber sleepers compared to concrete sleepers (Kerr et al. 2008). This was confirmed with other research based on the numerical analysis presented by (Farjoo et al. 2013; Farjoo, Daniel, et al. 2012). In this study, a 3D finite element method was used to determine the
stress intensity factors at the crack tip when the contact patches move over the crack located on the soft (timber) and hard (concrete) foundations. As a result, it was concluded that squats defects are more likely to take place at locations with a lower stiffness foundation (timber). This is because the soft foundation causes high compressive bending stress which leads to enlarging the shear mode of stress intensity factor. As a consequence, squats occur with timber more than concrete sleepers.

In addition, based on-site investigation conducted in the Dutch rail network by ProRail Li, Zhao, Esveld, et al. (2008), it was found that the locations of squats also relatively vary to sleeper’s position. For instance, the analysis indicated that about 74% of squats were found on rails above the sleepers (part I) compared to that not on the sleepers (part II), as shown in Figure 2.7. This interprets the effect of stiffness and other track parameters, such as bending stresses, on squat formation.

![Figure 2.7 The identification of part I and part II of rail (Li, Zhao, Esveld, et al. 2008).](image)

### 2.3.3 Squats at switches, crossings and insulated rail joints

Squats can be occasionally noticed at particular locations such as switches, crossing and insulated rail joints. This is because the track stiffness at these locations is changed suddenly from the surrounding rail. In addition, the axle dynamic load amplifications are considerably high at joints due to rail discontinuity and relatively surface dipping. As a result, a mechanical failure such as rolling contact fatigue can be accelerated at these locations and result in squats defects. For example, a numerical analysis using a multibody finite element model was performed by Li, Zhao, Dollevoet, et al. (2008) to identify the influential factors that cause squats at insulated joints (fishplate end). The findings showed that when the stiffness is changed due to loose fishplate bolts, the rolling
contact force becomes considerably high and this can lead to differential wear and/or differential plastic deformation which eventually results in squat formation. An example of a squat location at switches, crossings, and insulated joint, as well as the vibration of vertical contact load at insulated rail joint, can be seen in Figures 2.8 and 2.9 respectively.

Figure 2.8 Squat at switches, crossings and insulated joint (Zhao 2012).

Figure 2.9 Dynamic contact load along rail including insulated rail joint (Zong et al. 2013).
2.3.4 Squats at different rail curvatures

Generally, the occurrence of squat defects can be observed in either low and high rails at both tangent and curved tracks. However, there is a considerable variation in squats proportion. Figure 2.10 displays the percentage of squats with various curvatures and rail types in the Sydney Trains network. The data shows that a large proportion of defects are predominantly formed on high rails and curved tracks. On the other hand, squats can be absent on the sharply curved tracks and low rails. This phenomenon is more likely caused by the wear process-induced between wheel and rail. Generally, at a tight curve, the excessive wear and friction can remove the small cracks and therefore they cannot further develop into squats. As a result, the concentration of squats is correlated with rail curvature.

Figure 2.10 Percentages of squats in various ranges of curves in Sydney, ‘Low all curves’ considers the rates of the squats defects on the all low rail (Daniel et al. 2013).

2.3.5 Squats at stations and signals

Squats defects were expected to occur easily near to stations and signals. The rail section at this track position is always experienced excessive tractions and braking efforts that are needed to accelerate or decelerate a train’s speed. Consequently, friction level and slipping processes between wheel and rail become significantly high resulting in squats formation. For instance, in a particular line of Dutch network, the correlation of squats positions with the relative speed of a train is shown in Figure 2.11. As can be seen, several
squats at moderate and severe stages (illustrated in different symbols) were concentrated around 26 km where the train’s speed is relatively low (Li et al. 2011). This could indicate that high deceleration, which is usually occurred near to stations or signals, can lead to the occurrence of high tangential force and also wheel slip which may result in squats. This characteristic was also observed on the particular lines of the London underground network (Grassie, S. et al. 2011). In Sydney networks, there was no relationship between squat occurrences and station locations (Kerr et al. 2008).

![Figure 2.11 Correlation between squat occurrence and train's speed, red squares refer to severe squat and yellow diamond denotes moderate squats (Li et al. 2011).](image)

### 2.3.6 Squats at the tunnel

Although squats rail defects can be present everywhere, they can be absent in the tunnel. Generally, squat absence in the tunnel has been claimed by many researchers (Ishida et al. 2003; Kerr et al. 2008; Kondo et al. 1996). This conclusion is often associated with the presence of fluid as well as environmental contaminations which play a major role in cracks development particularly, in the traffic directions. (The effect of water on squat formation is discussed later).
2.3.7 Squats correlated with different rail material grades.

In general, there are commonly two types of rail used in the railways; head hardened and standard carbon steel. Squats occur in both types of rail. However, although the head hardened rail type has been developed to reduce wear condition as well as to improve the total life of rolling contact fatigue; it has noticed that squats are a greater predominance in head hardened (HH) compared to standard carbon (SC) rails (Grassie 2014). This could be related to an unsuitable balance between wear rates and rolling contact plastic deformation. More investigation on this phenomenon is under progress in the Sydney Trains network.

2.3.8 Squats correlated with rail age

The consideration of rail age with squats occurrence was explained in Figure 2.12 (Kerr et al. 2008). As can be seen, there is no clear relationship between squat occurrence and rail age.

![Figure 2.12 Distribution of squats rail age (Kerr et al. 2008).](image)

2.4 Mechanisms of squat cracks formation, propagation, and fracture

Squat defects on rails have been noticed since the 1950’s; however, mechanics of squat cracks initiation, formation, and growth as well as fracture failure are still not fully understood. The most popular argument is that squat initiation is attributed to two different formation mechanics, such as mechanical and/or thermal origination. Similarly, several studies hypothesised various mechanics of squat growth and development down
the rail. However, metallurgical evidence from the ex-service rail is insufficient and controversial results were presented. This is particularly due to the complicated contact condition between wheel and rail, and multiple influence factors. In addition, the case of fracture failure of the rail associated with squat defects is rare and only occurs in certain cases.

The following literature review presents the most popular theories and hypotheses being proposed for the mechanisms of squat cracks initiation, propagation, and fracture failure.

2.5 Initiation: Mechanisms of squat formation

Squats formation is often associated with surface-initiated RCF cracks (Grassie 2012). The mechanism of RCF is a common phenomenon but not fully understood, particularly in relation to crack initiation (Carroll 2005; Daniel 2012). The RCF cracks initiated by either ductility exhausting of the surface material (ratcheting) or by plastic shakedown. Both of these failure mechanisms are caused by the accumulation of plastic deformation on the rail surface due to high contact stresses, which eventually can lead to squats formation. In addition, research efforts have indicated the relationships between the narrow and brittle white etching layers observed on rail surfaces and squat formation in rails. White etching layer has been observed in the vicinity of squat type defects (Clayton & Allery 1982; Fletcher & Sanusi 2016; Kerr et al. 2008; Simon et al. 2013). In some research, WELs are thought to be created by severe plastic deformation on the rail surface. However, WELs are also thought to be induced by the thermal process between wheel and rail (Grassie, S. L. et al. 2011; Scott et al. 2014). This finding is mainly correlated with the absence of plastic deformation underlying the pearlite microstructure. This means that the mechanism of rolling contact fatigue has not involved in WEL formation. This type of white etching layer has been considered the main cause of crack initiation and thermal damage called “Stud” (Grassie, S. L. et al. 2011). Further detailed about RCF and WELs are given bellow in different sections.

2.5.1 Surface cracks initiated by rolling contact fatigue

Due to the transport operation, the topmost rail surface is experienced severe work hardening, which results in plastic ratcheting. Even though the amount of plastic deformation in the rail after each cycle may be very small, it becomes large after many cycles of loading (Kapoor 1997). When the ratcheting strain reaches the limiting ductility
of the rail, the rail will fail at the local material point, which corresponds to the initiation of wear or rolling contact fatigue (Kapoor 1994; Su & Clayton 1997; Tyfour et al. 1996), for example, in the form of squat or head checks in the railhead. Hence, the squat is recognized as a Rolling Contact Fatigue (RCF) problem. RCF analysis involves the solution of wheel/rail contact, crack initiation, growth of crack and fracture of the components; and it is still a challenge in practice due to the high complexity of the problem.

The initiation of fatigue crack in rails is caused by the accumulation of shear deformation due to repeat rolling - sliding contact loading. At high traction, the plastic flow is induced by the stresses at the rail surface. The material near the surface experiences a non-proportional cycle of tension, followed by shear, and then by compression, which makes it difficult to predict the position and orientation of crack initiation.

Ringsberg et al (2000; 2001) defined fatigue parameter $FP$ as a weighted of the sum of stress and strain components. In this model, it is assumed the RCF occurs due to the ratcheting and loading conditions above the shakedown limit. It combines the multiaxial fatigue model, the critical plane concept and the rule of fatigue damage summation to predict the RCF crack initiation. The ratcheting and multiaxial fatigue damage model was also used by Jiang et al (1999) to calculate the rolling contact crack initiation of manganese steel. Tyfour et al (1996) combined both ratcheting and crack pressurization parameters to estimate the fatigue life of two rolling bodies. An empirical model, which has gained acceptance in the prediction of fatigue crack initiation owing to ratcheting only, was proposed by Kapoor (1994).

**2.5.1.1 Material response and the shakedown map.**

Rails are normally exposed to a high wheel load through a relatively small contact zone. During their operational life, rails typically experience millions of load cycles from the passage of train wheels. The response of material subjected to repeated loading is governed by the stress, strain, wear, and temperature change (Dieter & Bacon 1986), and can be any 4 types shown in Figure 2.13, including elastic, elastic shakedown, plastic shakedown, and ratcheting.
As maximum Von-Mises stress within the contact zone can be larger than the yield point of the rail materials (Vo et al. 2014), two material responses including plastic shakedown and ratcheting normally occur in the contact zone of railhead, cause the fatigue initiation of surface cracks, such as “squat” or “head checks”.

The diagram named “shakedown map” (Brunel et al. 2010; Farley et al. 2010; Ponter et al. 1985) has been shown very efficient in estimating the material response and anticipating the position of the greatest fatigue damage. The shakedown map identifying the response of material for the railhead and rail gauge of low rail and high rail in the curved track is plotted in Figure 2.14 (Vo et al. 2015). As shown in the map, for those contact zones at the railhead, the responses of material mainly fell in the ratcheting region which was due to high traction force and high contact load exerted on the rails. The damage at this location was predicted to occur on the surface rather than the sub-surface, whereas the response of material fell in the elastic shakedown region for the contacts at the rail gauge. However, on the high rail (points circled in red), the pressure at the rail gauge corner was quite high compared to that at the railhead. In this case, the response of the material at the rail gauge corner was close to the ratcheting region, whereas that at the railhead was in the elastic shakedown region. This prediction is confirmed by the field data of the head check or gauge corner squat (Kerr et al. 2008).
2.5.1.2 Life prediction of RCF crack initiation

The fatigue life prediction strategy in Figure 2.15 was employed by Ringsberg et al (2001) to estimate the position, the orientation, and crack initiation life. The FE tool was used to simulate the wheel-rail rolling contact phenomenon. The crack initiation analysis was carried out as a post-process using the results from the FE analysis. In the post-process, the most critical material point and plane for crack initiation (the crack plane) were identified with respect to fatigue damage. It is defined as the plane subjected to the largest value of the multiaxial fatigue parameter, $FP_{max}$, during one load cycle in Equation (1). The details of the equation can be found in (Ringsberg 2001).

$$FP_{max} = \left(\frac{\sigma_{max}}{2}\right) \Delta\varepsilon + J \cdot \Delta\tau \cdot \Delta\gamma \quad \ldots \ldots (1)$$

Figure 2.15 A strategy for fatigue life prediction of crack initiation (Ringsberg 2001).
When the crack plane has been identified, the fatigue life of crack initiation is computed by using a damage summation rule of Palmgren – Miner (Ringsberg et al. 2000). To determine the number of cycles for crack initiation by ratcheting, the pertinent damage per load cycle, \( n \), is thereafter calculated as:

\[
\left( \frac{dD_{LCF}}{dN} \right)_n = \left( \frac{1}{N_f} \right)_n \quad \text{(2)}
\]

The FE simulation was carried out on the high rail of a curved track with rail profile UIC60 and pearlitic UIC 1100 rail steel. The axle load of 12 tons was used for a suburban train. The material behavior in the rail gauge corner showed a ratcheting material response with an almost constant rate for the last five simulated wheel passages, while the railhead showed a plastic shakedown material response. The fatigue calculations showed that the number of wheel passages to crack initiation was \( N_f = 79,600 \) and \( 66 \times 10^5 \) for rail gauge corner and railhead respectively.

### 2.5.1.3 Microstructural evolution of the rail surface layer by RCF

The plastic deformation caused by RCF can be clearly shown in the transverse and lateral sections of squats when they are revealed under microscopy. Figure 2.16a illustrates the strips of plastic deformation for the squat sample sectioned longitudinally. As can be seen, the deformed grains of material are oriented along the rail axis, usually adverse of traffic direction. While in the transverse section of squats the shared grains (plastic flow) are often directed toward the gauge corner of the rail, Figure 2.16b.

![Figure 2.16 Plastic deformation on the rail (x 120); (a) longitudinal section; (b) transverse section (Clayton & Allery 1982).](image-url)
The cracks initiate under the rail surface following the deformed microstructure passing through the boundaries of sheared pearlite grains that result from high traction ratio (shear stresses). This can be clearly shown in Figure 2.17.

Figure 2.17 Crack follow shared grains (note: rolling direction referred by arrow) (Ringsberg 2000).

Microstructural changes caused by accumulated plastic deformations result in the inhomogeneity characteristic of rail mechanical properties. Such the hardness measurement at deformed regions is relatively higher than the undeformed area of the material. This can be proved by Figure 2.18 which illustrates the micro-hardness values measured cross the rail samples presented in Figure 2.16.

Figure 2.18 Microhardness test across rail samples (Clayton & Allery 1982).
2.5.1.4 Relationship between rolling contact fatigue crack initiation and wear

It has been found that there is a relation between rolling contact fatigue crack initiation and wear. Wear can eliminate squat cracks formation; however, too much wear results in rail life reduction. Therefore, balancing a wear rate for example by a well scheduled grinding process is important to maintain a railhead profile and prevent cracks formation.

2.5.2 Squat formation and the occurrence of white etching layer on the rail surface

Recently, research efforts have indicated a possible relationship between the narrow and brittle white etching layers (WELs) observed on rail surfaces and squat formation in rails. WEL plays a critical role in crack initiation on the rail surface and correlations have been observed between WELs and the formation of squat defects (Bernsteiner et al. 2016; Horstead 2013; Sanusi & Fletcher 2017; Wilson 2011, 2012). In addition, it was found that the formation of WEL on the running band surface can significantly reduce the RCF life of the rail and promote squat defects (Steenbergen 2016). WEL is usually characterised as a hard, brittle structure that is susceptible to breaking, cracking and in some cases surface spalling. Consequently, due to wheel/rail contact, small cracks can initiate within this layer and/or at its edges and propagate into the parent rail material resulting in the generation of squat defects. The rail surface quality can be adversely degraded and significantly decrease of the total life of the rail.

The formation of white etching layers on the surface of rails has been comprehensively investigated. However, a clear understanding of the premise that cracks in squats originate from an extension of cracks in WELs, are still in progress.

A number of studies have been carried out on rails containing surface defects to examine the root causes and origin of damages. Clayton and Allery (1982) conducted a metallurgical analysis on ex-serviced rails containing squats. The results of the metallurgical investigation showed that small cracks were observed to initiate from WELs that was believed to be thermally induced. The existing of WELs was linked with wear damage and corrugation formation on the rail surface. Similarly, Dikshit et al. (1991) examined head-hardened rails containing surface spalling defects and it was found that cracks were initiated from WEL and developed down into the rail surface within depth about 3 mm, resulting in surface spalling defects. Carroll and Beynon (‘Rolling contact fatigue of white etching layer: Part 1 - Crack morphology’ 2007) investigated the
behaviour of WELs and crack initiation mechanism under the condition of rolling contact fatigue, by using a twin discs experiment. Based on the results obtained, it was suggested that the occurrence of WELs plays a significant role in crack initiation and formation of defects such as spalling which in turn leads to increase wear rate. In addition, it was suggested that the cracks initiated from WEL may have a tendency to grow deeper into the rail resulting in larger defects such as squats. Carroll and Beynon (‘Rolling contact fatigue of white etching layer - Part 2. Numerical results’ 2007) also investigated the behaviour of WELs by using modelling analysis. They found that cracks were initiated from the interface layer between WEL and bulk material due to anisotropy variation. In addition, crack growth corresponded to the deformation layer of pearlite underneath WEL. Kerr et al. (2008) conducted an extensive investigation of squats defects. It was reported that squats on the running band area were developed from white etching layers while squats on the gauge corner were developed from gauge corner cracking. This conclusion was drawn from the observation that WELs with cracks extending from their surface were found in the vicinity of running band squat while only plastic deformation was present in the gauge corner squat. In addition, it was recommended to use a high resolution technique for further understanding. A metallurgical investigation on the rails containing squats carried out by Simson and Mohan (2010), Valente et al. (2010) was found that the formation of squat cracks was driven by WEL and RCF. However, Grassie, S. L. et al. (2011) and Grassie (2014) suggested that white etching layers rather than rolling contact fatigue are the main cause of rail surface defect called “Stud” which is very similar to a squat. Linking between rolling contact fatigue and the presence of white etching layer on the rail steel material were examined by Seo, Kwon, Jun, et al. (2011), using twin disc testing. After certain cycles of loadings, the cracks formed at the interface region between WEL edge and bulk material, and at the middle of WEL. Pal, Valente, et al. (2012) examined the damaged rails containing squat defects. Metallurgical analysis showed that WELs were present near squats region. Observation showed that the cracks initiated at the WELs edge. It was suggested that WELs may cause fatigue crack initiation and squat formation. On the other hand, Pal et al. (2013) examined a different patch of damaged rails containing squats, to determine the role of white etching layer on squats formation. It was found that WELs was not present near squats and therefore did not play a role in squat cracks formation. Daniel et al. (2013) suggested, based on coherent investigations on the Australian rail network, that white etching layers assist squats to
originate. Simon et al. (2013) performed tribological and metallurgical investigations on squat and the adjacent regions. WELs initiated cracks were observed in the vicinity of a squat. Based on the microstructural observation, it was concluded that the microstructural evolution associated with WELs formation was a key role for squat initiation. Steenbergen, Michaël and Dollevoet, Rolf (2013) investigated the formation mechanism of squats in a tangent track. The microstructural analysis found that some patches of white etching layers near squats were present, and the cracks were initiated at the edges of WELs and propagated down into the rail following the plastic follow direction underneath WEL. Based on this observation it was suggested that WEL played a crucial role in cracks formation and origination of squat defects in tangent rails. In contrast, Steenbergen in (2015) stated that white etching layers do not play role in cracks initiation and squat formation. This conclusion was drawn after conducting microstructural investigation on squats formed in curved track rails, and even though after the observation of cracks formation within WELs, as shown in Figure 2.19. Zhou et al. (2016) studied the characteristic of WELs by conducting two twin discs experiment. Due to rolling/sliding contact loading, WELs have induced on the surface layer and the cracks have initiated at WELs’ edge and propagated along the interface between WELs and matrix leading to surface spallation. Steenbergen (2016) examined the ex-serviced rails after grinding process and the following train operation. It was found that the combination of residue large portions of WELs after grinding and plastic deformation induced by passing wheels results in severe cracks initiation and spalling formation. It is interesting to note that improper rail grinding could accelerate the rail degradation instead of extending rail life expected. Rasmussen et al. (2017) performed a metallurgical analysis on the ex-serviced rail that being newly grinded. The results found that WELs were caused by grinding and a high density of cracks were initiated from WELs. These cracks propagated down into matrix and then returned up to rail surface casing shallow spalling defects. Cracks within WELs were throughout to be a direct consequence of grinding. Sanusi and Fletcher (2017) modelled the effect of WEL on formation of large defects in rails. The results found that the presence of WEL results in an increase of crack growth rate inside the railhead.

Based on the literature review, a fundamental understanding of the correlation between squats formation and the occurrences of white etching layers on the surface of rail is high of importance. That is to implement more effective, preventive and early corrective
actions related to squat initiation or the early stages of squat growth, as well as reduce their detrimental consequences, the detailed investigation leading towards a comprehensive understanding of the structure and microstructure nature of the WEL is required.

Figure 2.19 surface characteristic of damaged rail containing squats, showing cracks initiated at interface layer between WEL and matrix and propagated down into the rail, (a) longitudinal section (b) transverse section, (Steenbergen 2015).

2.5.2.1 Characteristics of White Etching Layer (WEL)

The white etching layer is a microstructural phenomenon, characterised by brittle, hard layer, formed on the rail surface due to the actions between wheels and rails. Figure 2.20 shows typical white etching layer formation on the running band area, taken from field investigation.

Figure 2.20 White Etching Layer on the rolling band area after etching by 2% Nital (photo taken from field investigation at Sydney Trains.)
The WEL reveals a white appearance and sharp edges under light microscopy due to high resistance to etching by Nitric acid (Nital solution). WEL has been recognized as one of the main causes of rail surface damage because WEL exhibits exceptional fragility and is demonstrated to result in the initiation of cracks in the railhead (Clayton & Allery 1982; Kerr et al. 2008). Figure 2.21 shows typical examples of surface-initiated cracks from the white etching layer.

![Fig 2.21 Typical example of surface-initiated cracks from the white etching layer](Rasmussen et al. 2017)

2.5.2.2 Mechanical properties and microstructure characterization of WEL

A distinct feature is that a WEL has a high hardness (700 - 1200HV) (Baumann et al. 1996; Clayton & Allery 1982; Zhang et al. 2006), and this can cause brittle crack initiation and subsequent fatigue crack propagation. In addition, due to the high hardness property of WEL compared with bulk materials, the wear rate is reduced where the WEL is located. This phenomenon results in a localised differential wear on the rail surface. The harness variation measured from the rail surface is displayed in Figure 2.22. It can be seen that the presence of the WEL contributes a high hardness value of about 1000HV, and sharply drops to the 400 HV of the parent material. In addition, the hardness distribution across WEL was not uniform, varying between 900-1000 HV, which indicates that WEL is not homogenous (Pan et al. 2017; Zhang et al. 2006; Zhou et al. 2016)
The measurement of residual stresses in a WEL was conducted by Österle et al. (2001) using X-ray diffraction analysis. It was found that the compressive stress is significantly present both along the rolling direction and transverse direction. As shown in Figure 2.23, within the regions where the WEL is present these compressive residual stresses are approximately 200 MPa higher than in the other parts of the rail. Wang et al. (Wang, L et al. 2003) also measured the residual stress in a WEL and found the results similar to those of Österle (2001). That is, a WEL has high compressive residual stresses of around 600 MPa.

Figure 2.23 Residual stresses at WEL location and rail material (Österle et al. 2001).
2.5.2.3 Structure nature of white etching layers

Even advanced experimental techniques were used to characterise the structure nature of the white etching layer on the rail surface, the composition and structures of WEL are still arguable.

Österle et al (2001) found that the WELs have a martensitic microstructure with a high dislocation density. This result is widely accepted by lots of researchers and engineers from the fields (Clayton & Allery 1982; Grassie 2012; Kerr et al. 2008). As shown in Figure 2.24, the asymmetric profile shape of synchrotron X-ray diffraction is especially pronounced for the 200-peak, therefore, a tetragonal distortion of the lattice in the WELs is present which indicates the martensite formation. Phase analyses in the WELs did not reveal the presence of retained austenite in any of the layers. The c/a ratio of the martensite in the WEL-band varies between 1.002 and 1.0095. The dislocation density also varies across the band, and the average dislocation density is $7 \times 10^{11}$ cm$^{-2}$. In addition, the elemental distribution of the white etching layer taken from the rail track was determined by using atom probe tomography (ATP) analysis (Takahashi et al. 2010). It was found that the concentration of manganese and silicon elements was varied across WEL, which confirms that WEL is a martensite structure. R.Pan et al (Pan et al. 2017) characterised the structure of WEL taken from the surface of treads of U71Mn rail by SEM and TEM. It was found that the microstructure of WEL has comprised predominantly martensite, with few amounts of undissolved cementite. Furthermore, a nanocrystalline $\alpha$-phase structure was observed at the topmost layer of WEL. L. Beneš (Beneš 2012) examined the white etching layer by using the TEM and X-ray diffraction techniques to identify the microstructure of WEL. It was found that WEL contained nano-structured martensite morphologies, very small ferrite grains as well as very fine carbide precipitations from lath martensite. The microstructure of WEL was also found to have high dislocation density.

In contrast, Lojkowski et al (2001), Djahanbakhsh et al (2001) and Baumann et al (1996) proposed that the WELs are nano-crystalline Fe-C alloy with the grain size ranging from 15 to 500 nm. The microstructure transformation of rail surface may take place at the rail - wheel contact temperatures less than 230°C. Zhang et al. (2006) found that WELs are composed of severely deformed pearlite, nano-crystalline martensite, austenite, and cementite, etc. The 3DAP analysis confirmed that carbon with an average concentration
of 4 at. % (atomic percentage) is distributed nearly uniformly in the topmost surface of the WELs. Zhang et al. (2006) found that WELs are composed of severely deformed pearlite, nano-crystalline martensite, austenite, and cementite, etc. The 3DAP analysis confirmed that carbon with an average concentration of 4 at. % (atomic percentage) is distributed nearly uniformly in the topmost surface of the WELs. Based on conventional X-ray and synchrotron X-ray analyses, Wang and Pyzalla et al (Pyzalla et al. 2001; Wang, L et al. 2003) found that the microstructure of WELs on the rail surface changes distinctly due to loading. The strong loading leads to the formation of martensite, which contains up to 13 volume percentage of retained austenite. The presence of retained austenite within the WEL indicates that its’ formation is accompanied by a considerable rise in temperature. The microstructure of the white etching layer formed on the surface of rail grad R260Mn was characterised by electron backscatter diffraction (EBSD) and transmission Kikuchi diffraction (TKD) in SEM (Petrov & Sietsma 2017). The EBSD mapping showed that WEL is characterised by ultrafine grain structure of ferrite, martensite like morphology and retained austenite.

Figure 2.24 Synchrotron x-ray pattern of 200 peak in the WEL and the adjoining area(Österle et al. 2001).
2.5.2.4 The formation mechanisms of white etching layers

In addition to the argument on the nature of WELs, it is important to note that the formation mechanics of WELs is also a matter of controversy (Baumann et al. 1996; Clayton & Allery 1982; Djahanbakhsh et al. 2001; Jirásková et al. 2005; Kerr et al. 2008; Lojkowski et al. 2001; Newcomb & Stobbs 1984; Österle et al. 2001; Pyzalla et al. 2001; Wang, L et al. 2003; Zhang et al. 2006). Until recently, the following hypotheses consider the formation mechanics of white etching layer: (i) WELs are the result of the thermal process caused by wheel slip; (ii) WELs are mechanically induced by severe plastic deformation; (iii) WELs are caused by a combination of thermo-mechanical process. Based on these perceived mechanisms of formation, the microstructural nature of WELs has been interpreted differently in section 2.5.2.2.

- **White etching layer thermally induced**

WELs thermally induced are composed predominately of martensite structure and some retained austenite and some undissolved carbide (Al-Juboori et al. 2017; Beneš 2012; Hosseini et al. 2015; Pan et al. 2017; Takahashi et al. 2010). WELs containing martensite along with the presence of retained austenite provide direct evidence of phase transformation from pearlite into austenite and a subsequent partial transformation into martensite during rapid quenching. In a general concept, reaching the austenization temperature in steel means that the material needs to be heated up to a flash temperature of 727 °C and then rapidly cooled to form martensite. However, the actual required temperature for phase transformation at the rail surface is below 727 °C. High contact pressure between wheel and rail assists the transformation from pearlite into austenite at a relative lower temperature (Ahlström & Karlsson 1999; Baumann et al. 1996; Daniel 2012; Gautier et al. 1995). Experimental studies also confirmed that the martensitic patches were formed on the surface of rail steel material after heating under hydrostatic pressure to temperature below the critical pearlite-austenite transformation point (Wu, Petrov, et al. 2016a). It was found the transformation temperature required for austenization decreased by 30 °C under a hydrostatic pressure of 1.8 GPa. In addition, the microstructural evolution of the top surface layer such as grain size refinement of the pearlite structure due to severe plastic deformation structure caused by RCF could also reduce the heat required for austenization (Carroll & Beynon ‘Rolling contact fatigue of
white etching layer - Part 2. Numerical results’ 2007). Support of WELs being caused by thermal phase transformation theory also comes from the sharp transition between WEL and parent rail material. Grassie et al (Grassie, S. L. et al. 2011) found that WELs can exist without severe plastic flow at underlying pearlite microstructure. Observations by atomic probe tomography (ATP) confirmed the absence of severe plastic deformation in the subsurface layer at the transition zone of WELs; and identification of manganese atoms distributions in the cementite lamella of WELs. ATP results support the theory of the martensitic transformation after austenization (Takahashi et al. 2010).

- **White etching layer mechanically induced**
On the other hand, WELs is mechanically induced by severe plastic deformation rather than thermally induced by phase transformation process (Baumann et al. 1996; Djahanbakhsh et al. 2001; Lojkowski et al. 2001; Simon et al. 2013; Zhang et al. 2006). This finding is associated with the Nano-crystal structure nature of WELs, which cannot be associated with the structure nature resulted from thermal phase transformation. When the top surface layer of pearlite structure is subjected to high contact shear stresses, the ferrite lattice is destroyed resulting in high dislocation density. In this stage, the carbon atoms in cementite structure are accommodated causing cementite dissolution which leads to the formation of a high destroyed lattice. The characteristic features of this layer, including nano-sized grains, high hardness, cementite dissolution, and the observation of the deformed pearlite structure at the transition zone indicates mechanical alloying.

- **White etching layer thermally-mechanically induced**
However, if the severe plastic deformation is the only mechanism that transforms the pearlite into WELs, it would be expected to produce a less sharped transition (Steenbergen, Michaël & Dollevoet, Rolf 2013). In addition, if the mechanism of WELs formation is totally driven by thermal phase transformation caused by Wheel slip, WELs would be generated just after one cycle of loading, which is controversy to the fact that WELs are loading history-dependent. Hence, WEL was suggested the result of a combination of mechanical and thermal processes (Beneš 2012; Kondo et al. 1996; Österle et al. 2001; Petrov & Sietsma 2017; Zhang et al. 2006). High contact stresses at the wheel and rail interface results in the cracking of cementite lamella inside the ferrite
matrix. Due to repeated cyclic loadings, the fractured cementite lamellas are extremely refined into non-sized fragmentations. Simultaneously, the thermal energy cycle on the rail surface caused by wheel slip results in the formation of austenite which subsequently transforms into fine martensite after subsequent cooling. As a result, WEL is formed with a composition of fine martensite, nanocrystalline cementite, ferrite and retained austenite.

- **Brown etching Layer**

Recently, it was reported that there is a distinguished layer, identified in parallel to WEL called brown etching layer (BEL). Investigation on this layer was suggested that BEL is the decomposition product of martensite (Li et al. 2016; Wu, Petrov, Li, et al. 2016). Detection of retained austenite in the microstructure of BEL could also interpret the contribution of thermal activity. So far, the formation mechanism of BEL is still not well-known. Figure 2.25 shows WEL and BEL formed on the rail surface.

![Image of WEL and BEL formation on the rail surface](Li et al. 2016)

**2.5.3 Factors and parameters enhanced squats crack formation**

It is important to investigate the causes of squat defects and determine the factors that promote cracks and generate squats on the railhead. Site monitoring and field
observations have been conducted to provide the possible root causes of squat defects. Li et al (Li et al. 2011; Li, Zhao, Esveld, et al. 2008; Zoeteman et al. 2014) classified the cause of squat defects into two categories: internal and external. Internal causes are associated with track properties, including corrugation (short pitch), material and surface irregularities due to weld, wear and local plastic deformation. External causes are correlated with external factors and unusual interaction between wheel and rail, for example, indentation, slipping, skidding, and all mechanical and thermal activities. A statistical data collected from railway networks in the United Kingdom showed that about 75% of squats are correlated with the corrugations, welds as well as indentations which may be caused by hard particles inserted between wheel and rail (Seo, Kwon & Lee 2011). In the Netherlands, about 74% of squats were observed with corrugations (Li et al. 2011). All external and internal causes result in settlement and irregularities in the rail surface which eventually leads to high contact dynamic load and stress concentration at the specific area. As a result, a contact stress level can exceed yield point of rail material allowing for a plastic deformation to be taken place; and accumulate continually due to RCF; eventually, result in squat formation if wear is insufficient. This mechanism of squat formation has been considered by numerous studies and investigations which consider squat defects generated due to rolling contact fatigue.

2.5.3.1 Squats induced by indentation.

Indentations can lead to the formation of defects on the rail surface. When hard objects or particles such as balls from ball bearings and ballast stones are trapped between wheel and rail, the rail surface may be depressed by these objects during the passing wheel and result in small indentations at particle size. Consequently, with a period of time the repeated contact load and wear process can result in flattening these circular indentations and followed by the formation of a lip at its edges. As a result, due to plastic accumulation and the consequential exhaustion of ductility of the rail material, the cracks can form at indentation’s edges where the maximum shear is located and develop down into rail material.

Squat defects correlated with indentations have been reported in the previous studies (Clayton & Allery 1982; Kondo et al. 1996; Li et al. 2011; Li, Zhao, Esveld, et al. 2008; Seo, Kwon & Lee 2011). Most of these were obtained from field investigations and monitoring. Figure 2.28 illustrates that a squat was induced from a small indentation on
the rail surface. As can be seen, a wave pattern, which was visible in 2007 was completely disappeared in 2008. However, those waves have returned again and become visible in 2009. This phenomenon might conclude that the indentations may be removed by the wheel/rail contact. On the other hand, the defect was still growing and broadened in rolling and transverse directions as shown in photos obtained in October 2008 and May 2009.

Figure 2.26 Squat formed at indentation (Li et al. 2011).
In addition, the surface profiles of the above defect at the early stage of initiation in Figure 2.26 were measured and shown in Figure 2.27. It is clearly noticed that the depth of the indentation is considerably reduced over time due to wear and plastic deformation. However, the pattern of the vertical-longitudinal profile has gradually become a W-shape which corresponds to the shape of the squat at development stages.

![Figure 2.27 Surface profile of squats induced from indentation (Li et al. 2011)](image)

Experimental and numerical studies were carried out to determine the effect of dent size on squat development. The result showed that the cracks can initiate and propagate due to rolling cycle fatigue when the size of defects is exceeded than a certain size because the wear process removed small indentations after a few cycles (Seo, Kwon & Lee 2011).

### 2.5.3.2 Squats induced by corrugations.

Corrugations are considered as a short and long pitch pattern that is located on the rail surface and can be described as one of the serious issues on the rail system. This is because they can reduce rail life and eventually require expensive rail replacement. Squats are strongly correlated with the occurrence of short pitch corrugations. For example, it has found that about 72% of squats observed with corrugations and expected that about 33% of them were caused by corrugations. However, there are still lots of unknowns about the relationship between squats and corrugations. They may have developed independently...
due to same track conditions; or the corrugations result in rail surface irregularities, which can raise a high dynamic force at wheel/rail interface and lead to the squats; or it results from the combination of two mechanisms above (Deng et al. 2018; Li, Zhao, Esveld, et al. 2008). A typical example of squats occurrence within corrugation can be seen in Figure 2.28.

Figure 2.28 Typical example of squats formation within corrugations(Deng et al. 2018).

Squats on the rail surface can also induce corrugations-like wave patterns as illustrated in Figure 2.29. This wave pattern has a wavelength range of about 20 to 40 mm and it could reach to 60 mm in some severe cases of squats. This pattern correlated with squats may disappear or reappear. Further information will be discussed in the next section.

In addition, another feature observed was the presence of a white etching layer at the crests (peaks) of the corrugations while there was no WEL at trough regions. This was confirmed by the microhardness test examined longitudinally between two corrugations (2 peaks and troughs) as shown in Figure 2.29. The maximum value of about 1000 HV
was measured at peak region while there was a slight difference between the plastic deformation region (430 HV) and un-deformed area at the trough region (250 HV) (Clayton & Allery 1982). In conclusion, the rail surface irregularities due to corrugations and the presence of the white etching layer at those peaks can cause squats or accelerate their development.

![Micro-hardness of two corrugations, 0.02 mm below the surface](image)

Figure 2.29 Micro-hardness of two corrugations, 0.02 mm below the surface (Clayton & Allery 1982).

### 2.5.3.3 Squats at weld section

Welding is one of the most common processes used to connect rails and however, it can induce surface defects. During the welding process, the rail material is exposed to high temperatures that lead to inhomogeneity of material. Additionally, there are the geometric imperfections occurred at the cross-section of the weld. Another effect is that the surface grinding after the weld is not implemented correctly. Consequently, the rail surface at the weld zone can become vulnerable to squat initiation. For instance, the statistics showed that about 10 to 15 percent of the total 72 percent of squats were located at welds (Li et al. 2011; Li, Zhao, Esveld, et al. 2008; Zoeteman et al. 2014).

There are two common types of weld used in rails including flash butt and alumina-thermic. Squats can occur at both types as shown in Figure 2.30. Although squats can be visually distinguished from their superficial appearance, the squats at welds do not
display a typical two lobes and V-shaped surface-breaking crack patterns. However, it is characterized by just one or two isolated lobes without surface cracking.

2.5.3.4 Squats induced by head check defects.

Head checks are caused by rolling contact fatigue and described as fine cracks located on the railhead, particularly at the gauge corner of the rail section. Head check defects are usually occurred and concentrated on high rails in the curved track because of excessive tangential shear stresses induced by wheel and rail interaction. In many cases, some of these checks develop under the rail surface and become severe resulting in one side of V-shaped crack (leading crack). Consequently, due to subsequence brittle fracture failure, the trailing crack is formed resulting in gauge corner squats. Figure 2.31 shows a gauge corner squats developed from head checks defects (Kerr & Wilson 2012). In fact, there is
no doubt that head chicks are considered as one of the rail surface problems and might be considered the root origin of squats if not worn off.

Figure 2.31 Squat developed from head checks cracks (Grassie 2014).

2.5.3.5 Squats correlated with the grinding process

Rail grinding is one of the effective maintenance methods to remove or eliminate squats and other types of rail surface defects. When about 4-6 mm of metal has been ground from the railhead, most squats are more likely to be completely removed because the depth of cracks related squats is normally between 3 to 6 mm (Kerr et al. 2008). However, insufficient or deep rail grinding could lead to a negative result. If the metal removal is less than the required limit, the substantial residual squats still exist which may become the source of squat defect during the subsequent service. For example, Figure 2.32 presents photos taken for squat after grinding conducted on a specific part of the Sydney Trains network. It can be seen that many squats initiated and developed after 12 months of rail grinding which reflects the effect of inadequate grinding. On the other hand, deep grinding can result in rail surface irregularities which cause high contact load and consequently squats formation. Therefore, the grinding process should be well controlled.
An interesting study was conducted by Steenbergen (2016) to examine the microstructural evolution of the surface layer of two rail grades, standard carbon (SC) and head hardened (HH), immediately after grinding process and after few days of train operation. Metallurgical investigation results found that, on SC rail steel, a non-uniform WEL was induced from grinding process due to frictional heating and after few days of train services pre-induced WEL has almost completely removed due to traction and the surface layer has plastically deformed.

On HH rail steel, the surface characteristic appears completely different. After grinding, large grinding grooves with significant frictional heating induced WEL patches at grooves edges, as well as localised but not uniform plastic deformation were evolved in the surface layer of the rail. After a few days of train application, the spalling wear from rolling contact resulted in flattening the surface layer by removing grinding grooves with
some associated WELs. However, some individual WELs were pressed into the surface and cracks were initiated from WELs and extended down within the localised pre-existed shear strain. Figure 2.33 shows the surface characteristic of SC and HH rails steel in the longitudinal direction, immediately after grinding and after few days of train operation.

Grinding induced white etching layer was also confirmed by another study (Rasmussen et al. 2017), conducted on the rail section removed immediately after the grinding cycle from the Danish rail network.

2.5.3.7 Squats induced by wheel burn

Wheel burn is considered as sliding and skidding marks that are located on the rail surface. These defects are caused by recurring slippage of the driving axle during traction
and braking processes. Sliding and skidding defects can be one of the important factors for squat initiation and growth in rails.

2.6 Propagation: Mechanics of squat cracks growth and development

After the initiation at the rail surface, the cracks propagate down into the subsurface layer. The cracks growth and squat development are driven by multi mechanisms. Numerous researches have concentrated on the mechanics of crack propagation and squat growth in order to provide an explanation of how a mild squat grows into the mature squat. Several hypotheses and theories have proposed depending on numerical analysis and laboratory investigations. The following literature summarized the mechanics of crack propagation and squat growth.

2.6.1 Cracks propagation driven by dynamic force

It was suggested that cracks propagation are driven by a dynamic force (Li 2009; 2011; 2008; Li et al. 2006; 2008; 2013; 2014). A small rail top geometry defect such as light squat results in surface irregularity, which in turn, leads to exciting dynamic contact force of certain wavelengths with a series of peaks. Such peaks are repeated at every wheel passage at the same location, causing localised ratcheting. The deformation caused by the first two force peaks, $F_1$ and $F_2$, as illustrated in Figure 2.34 will eventually cause $A_1$ and $B_1$ (as shown at the top chart of Figure 2.34 to become the two lung-like parts $A_2$ and $B_2$ of squat $B$ (as shown at the bottom chart of Figure 2.34. Squat $B$ will further grow until the rail failure if no remedial action is taken.

It was confirmed that the squat is formed from an existing surface irregularity or other discontinuity by high-frequency (950-1900 Hz) interaction between wheel and rail, leading to cyclic ratcheting. Hence, the local eigen characteristics of the wheel and system play a key role in squat development.
Figure 2.34 An illustration of the postulated squat growth process by the dynamic contact force (Li et al. 2011).

Steenbergen and Dollevoet (2013) also suggested that rapid internal crack growth is caused by dynamic wheel-rail interaction with high-frequency impact. However, the role of cracks, either surface-breaking or inside the ‘body’ of the railhead, are ignored in Li’s study (Li 2009; 2011; 2008; Li et al. 2006; 2008; 2013; 2014).

The life cycle of fatigue cracks consists of three phases; crack initiation, crack propagation and fracture failure, as illustrated in Figure 2.35 (Kapoor et al. 2002). The transition from initiation mode to propagation mode is driven by high contact stress. The crack growth rate is enhanced by the presence of liquid (rain, moisture or lubricant) on the rail surface. The subsequent shear and tensile stresses, which are induced by bulk
stresses, accelerate the crack growth rate and result in rail fracture. Therefore, the overall responses result in a W-shaped curve as shown in Figure 2.4. The role of liquid on crack growth is discussed later.

Figure 2.35 Stages of crack type squat initiation and propagation (Kapoor et al. 2002).

Squats growth and development on the rail were measured by using an ultrasound technique (Kaewunruen & Ishida 2014; Kaewunruen & Ishida 2016). Based on the collected data, a linear relationship pattern was the characteristic of cracks growth with respect to accumulated passing tonnage, as shown in Figure 2.36.

Figure 2.36 Squat growth behaviour via accumulated of passing tonnage (Kaewunruen & Ishida 2014).
2.6.2 Cracks propagation enhanced by liquid.

The influences of wet (lubricated) conditions on crack growth have been investigated by several hypotheses and theories (Bogdański 2002; Bogdański 2005; Bogdański 2014; Bogdański, S. & Lewicki, P. 2008). Possible mechanisms to explain the role of lubrication have been put forward.

- **Mechanics of “reducing friction inside the crack”**

  The friction at crack face will be reduced when the fluid is inside the crack, which promotes Mode II (shearing) crack growth, as shown in Figure 2.37(a).

- **Mechanics of “hydraulic pressure” exerted on the crack face**

  The fluid may be forced to the crack interface by the pressure of the rotating wheel and transmitting the contact stress to the crack faces by hydraulic action, which produces tensile stress at the crack tip and increases the crack growth rate in Mode I (opening). This effect will prevent crack closure, as illustrated in Figure 2.37(b).

- **Mechanics of “fluid entrapment” in crack interior**

  Fluid can be trapped in the crack when the wheel moves over the crack. The entrapped fluid will cause a significant increase in the Mode I stress, as shown in Figure 2.37(c). Please note that the crack mouth is seal which is different from the mechanism of ‘hydraulic pressure’.

![Figure 2.37 Schematic illustration of crack propagation by (a) the shear mechanism, (b) the hydraulic pressure mechanism and (c) the fluid entrapment mechanism(Wang et al. 2017).](image)
Many models about fatigue crack growth including 2D and 3D configuration have been developed to examine the mechanics of pressurisation of fluid on crack propagation. Olzak et al (1991) used a 2D FEM model to calculate stress intensity factors (SIFs) at the front of surface- and subsurface- cracks and it was concluded that the liquid pressure inside the crack faces has a significant impact on the stress state at the crack tip. Datsyshyn et al (Datsyshyn 2005; Datsyshyn & Kadyra 2006) found that SIFs at the crack tip are related to the crack length, crack angle with respect to the running surface, size of the contact patch, the friction coefficient between crack faces and pressure of an entrapped liquid, for example, rainwater. Bogdański et al (Bogdański 2002; Bogdański et al. 2005) investigated the hydraulic pressure profile of the entrapped liquid and found that the entrapped water increased $K_I$ (stress intensity factor) up to 1000%. Bogdanski et al (Bogdański 2014; Bogdański, S. & Lewicki, P. 2008) also developed a 3D crack model and found that the opening mode can be significantly increased by crack pressurization leading to a higher crack growth rate. The effect of water entrapment on the crack growth direction was studied by Farjoo et al (Farjoo, Pal, et al. 2012) using a 3D FEM model of squat crack and it was found that the pressurized water inside an oblique crack changes crack direction parallel to the surface resulting in squat formation. S. Ancellotti et al (Ancellotti et al. 2016) used a 2D FEM model to examine the effects of fluid pressurization and entrapment mechanisms on the stress intensity factors SIFs of cracks initiated under lubricated rolling / sliding contact fatigue. The results showed that the SIFs in mode II (shearing) is relatively high in comparison with SIFs in mode I (tensile) when using only fluid entrapment mechanism and the crack propagation is influenced by SIFs in mode II (shearing) in this condition but both have almost same influences when composing the pressurization mechanism.

In addition to numerical research, laboratory investigations have been carried out to measure the role of water on squat initiation and growth. Kaneta et al (1998) have induced a squat defect on a circular disk specimen after applying regular repetitions of dry and wet running on twin disc machines. They found that the water accelerated crack propagation which in turn accelerated squat formation, confirming the presence of water inside squat-crack faces and tips by metallurgical examination. Fletcher and Beynon (1999) examined the effect of lubrication method on rolling contact fatigue and found that rolling contact fatigue life of rails was relatively reduced under water-lubricated contact compared to dry and solid lubricant dispersed in liquid media. They (Fletcher &
Beynon (2000) also performed three methods of testing: dry, dry-wet (using molybdenum disulphide suspended in oil, dry-wet (water lubrication) and found that pressurization effects play a more significant role on crack growth through reduction of crack faces, friction and the cracks developed faster under water lubricant. This finding was recently confirmed by an experimental laboratory twin disc testing conducted by Maya-Johnson et al. (2017). Carol (2005) confirmed that water acts as a lubricant to reduce friction and minimize crack initiation at low contact cycles; however, water promotes crack growth with the increased contact cycles. W. Wang et al (2017) examined by lab testing the different types of fluids on the crack growth rate. It was found that depending on the viscosity of the lubricant the cracks growth rate is highly enhanced by fluid entrapment mechanism when water is used, while the crack growth rate is driven by hydraulic crack growth mechanism when oil and grease are used.

Furthermore, a full-scale test carried out by Fletcher et al (2008) was used to determine the effect of water penetration on the growth of surface-breaking cracks in rails. It was found that surface-breaking cracks were penetrated by fluid after many cycles of loading, which confirms the mechanism of water entrapment.

Although numerical modelling and laboratory tests support the hypothesis of hydraulic entrapment on squat growth, there is no reported direct metallurgical evidence of rails damaged by the presence of water under service conditions in a rail track network. The requirement for fluid to cause small cracks to propagate provides a rational of why squats vanish as soon as the railway enters a tunnel. The questions can arise concerning the physical evidence of water entrapped in cracks for an in-service track, and what evidence if any, is there that the surface water promotes crack growth? In addition to the mechanics of hydraulic entrapment on crack propagation, is there any other likely mechanisms operative?

**2.6.3 Squat cracks development associated with track type**

Squats have been found in almost all types of rail geometry such as tangent and curved tracks and turnouts, different rail types including mild carbon and hard hardened rails, and traffic conditions such as passenger and mixed traffic (Kerr et al. 2008; Li 2009; Wilson et al. 2012; Zoeteman et al. 2014). However, the absence of squats inside tunnels has been reported by numerous researchers for different railway networks. In Australia, a set of statistical data obtained from Sydney Trains and conducted on a wide proportion
of the network confirmed that there are no squats in the tunnels (Kerr et al. 2008). In the UK, Grassie et al. reported that the defects on the London Underground network occurred almost exclusively in the open track rather than in tunnels (Grassie, S. L. et al. 2011). Observation confirmed by Dutch experience is that squats do not occur in tunnels (Steenbergen, M & Dollevoet, R 2013). Squats, or rail shelling defects as called in Japan, usually occur outside tunnels unless contamination is present (Ishida et al. 2003; Kondo et al. 1996). The absence of water on the rail surface in tunnels can be considered as a critical exception factor for squat formation (Grassie 2012; Kerr & Wilson 2012; Kerr et al. 2008; Wilson et al. 2012). This conclusion is often associated with a converse argument, where the presence of fluid and contamination is found to play a major role in crack development, particularly in the traffic direction.

It has also been found that the existence of water or other fluid contamination on the wheel and rail contact interface inside tunnels has a great influence on rail surface degradation, as it does in an open track. In addition, the environmental conditions in contact regions can affect both adhesion and wheel slip between the wheel and the rail (Lyu et al. 2015; Zhu 2011; Zhu, Olofsson & Chen 2013; Zhu, Olofsson & Söderberg 2013). For example, under dry conditions, friction is significantly higher resulting in the enhanced adhesion between the wheel and the rail. In contrast, wheel/rail adhesion under wet conditions declines considerably resulting in increasing wheel slip ratios (Arias-Cuevas et al. 2011; Arias-Cuevas et al. 2010; Chen et al. 2002; Tyfour et al. 1996; Wang et al. 2011; Zhu et al. 2016). Thus, the wheel/rail friction is relatively higher in tunnels than in open track indicating that the condition of wheel slip is less likely to occur in tunnels, which can be interpreted as evidence for the lack of rail surface defects inside tunnels (Grassie, S. L. et al. 2011). In fact, the contact condition between wheel and rail is neither continually dry nor permanently wet. This means that the same portion of rail could experience alternative conditions due to external circumstances.

2.6.4 Squat growth during service

Squats initiation and development are correlated with the passed tonnage. It was found that a few amounts of MGT of traffic are adequate to initiate squats. Figure 2.38 explains the relationship between squat formations and passed tonnage that was monitored on the Japanese rail network. It can be clearly seen that squats originate with about 40 MGT and develop down when the passing tonnage reaches to 100 MGT. Consequently, squats
become severe and more dangerous at about 400 MGT services because the transverse defects are more likely to initiate and cause track failure. Therefore, the grinding process is essential to undertake at that time (40 MGT) to arrest squat development (Kondo et al. 1996).

![Diagram of squat initiation and growth based on passed tonnage](image)

Figure 2.38 squat initiation and growth based on passed tonnage (Kondo et al. 1996).

### 2.7 Fracture: Squats cracks induced fracture failure

As mentioned before, the cracks initiated at the surface of rail and propagated down in lateral and longitudinal direction across and along the running band. The cracks often develop horizontally within a plane depth of about 3-6 mm. (Kerr et al. 2008; Pal, Valente, et al. 2012). The cracks then grow up toward the field side and then result in spalling when they have terminated the top surface layer. Squats induced surface spalling can be seen in Figure 2.3, under the consideration of a very severe class defect.

In a few cases, the cracks turned down changing their direction from the sallow angle, typically between 15-30° to an acute angle of about 70°. When the cracks reach the edge of the residual compressive stress layer, typically at length between 5-8 mm (Grassie 2014), they sustainably develop into the transverse plane which eventually forms a transverse defect and terminates with rail breakage. Figure 2.39 shows a typical example of a broken rail caused by squat. Generally, the fracture surface of the broken section is
very characteristic. It contains two different zones can and can be visually distinguished: fatigue zone and brittle fracture zone.

![Figure 2.39 One example of a broken rail caused by squat (Photo taken from Rails supplied to the UOW).](image)

The growth of cracks into the transverse plane is mostly driven under the action of longitudinal stresses, including bending stress; thermal stress and residual stress (Grassie 2014; Kerr & Wilson 2012; Kondo et al. 1996). Based on the stages of squats development as shown in the schematic illustration Figure 2.38, the transverse crack would develop down into the rail after about 350 MGT with a growth rate of 0.014 mm per MGT, associated with cracks depth about 5 mm from the rail surface. However, it is still unknown why there are only a few cases of squats that their cracks developed into transverse plane, as well as why does the fracture failure in some broken cases occur suddenly under low tonnage and without warning. Because of the few cases of rail breakage caused by squats, there is also a lack of associated investigations on the causes of fracture and formation mechanics of failure, which can provide a clear understanding of how cracks develop down into the transverse plane and result in transverse defects (TDs). In addition, there was no direct report providing physical evidence that shows connecting between squats and transverse defects (TD’s) occurrence.
Sugino et al. (1996) conducted a metallurgical investigation and FEM modeling on the transverse defect extended from surface-initiated gauge corner RCF cracks in heavy haul railway, to clarify the formation mechanics TDs defects at an early stage. The results found that the fracture failure was driven by tensile fatigue and no metallic inclusions were observed at the crack front. Cannon, D. F. et al (2003) reviewed that there are three groups of fracture failure; (i) those developed from pre-existed defects such as tache oval or kidney defect (ii) those developed from damage of defects such as wheel burn from spinning wheels which is due to poor installation and handling process, (iii) those developed from defects such as squats due to fatigue fracture. Steenbergen (2015) presented a broken rail section and suggested that the fracture failure was originated from head checks defect at the gauge corner and the deep growth and fracture failure mechanism was driven by bending fatigue stresses from large axle loadings of passing vehicles. Kerr et al. (2008) provided extensive detailed information of squats their related defects from site investigation and depending on collected data, it was concluded that there was no report confirming that the fracture failure from transverse defect was associated with squats. Grassie (2014) published a report providing an intensive investigation on squats. He confirmed that the cracks turned down the rail, passing through the layer of compressive residual stress into the transverse plane under fatigue banding stress. He also confirmed that squat cracks can progress into fracture failure and formation a TD’s defect. However, there is no evidence that TD’s can develop from stud defects.

2.7.1 Bending stress

Bending stresses arise principally from the axle load which is usually between 8 and 22.5 tons (for heavy haul in Australia up to 40 tons) and its dynamic magnification by a moving train. While the lateral bending stress contributes to rail failure, the vertical bending stress dominates fracture.

2.7.2 Thermal stress

Rail axial thermal stresses are produced due to seasonal differences in the temperature because built-in rails cannot elongate and shrink with increasing and decreasing temperatures. These are tensile stresses at lower temperatures and compressive stresses
at higher temperatures. In some countries at winter, high tensile thermal stresses at low temperatures are combined with relatively low toughness of rail material to contribute to crack growth and rail fracture.

2.7.3 Residual stress

Residual stresses in the rail are introduced by different mechanisms. Primarily they come from the manufacturing process, such as heat treatment and roller straightening. These stresses are subsequently modified by plastic deformation occurring in the wheel and rail contact zone. By this mechanism, the residual stresses are redistributed as illustrated in Figure 2.40. Traditional destructive methods, for example, strain gauge and sectioning, are relatively coarse and unsuitable when the steep stress gradients and multiaxial stresses are present. Non-destructive methods, such as ultrasonic (Hirao & Ogi 1997; Hirao et al. 1994) and neutron beam diffraction (Finstermann et al. 1998; Kaiser et al. 2014; Sasaki et al. 2008; Tawfik et al. 2006) have been introduced to measure the residual stresses in the rails.

Figure 2.40 Redistribution of the residual stresses during service (Schajer 2010).
2.8 Squat/Stud: differences and similarities

Squats are generally described as rolling contact fatigue defects that initiated on the rail surface due to the ratcheting phenomenon. This phenomenon is defined as cyclic plastic shear deformation of the surface layer. In recent research, Grassie (2012; 2016) and his collages (2011; 2011) argued that a certain type of railhead defect named “stud” has become considerably emerged particularly, with installing modern rolling stocks. Studs at the beginning were classified as a squat defect for the reason that this defect has completely similar characteristics features. However, there are significant differences in the mechanical and metallurgical aspects between studs and squats. In addition, it was suggested that studs are thermal damage and severe plastic deformation has not a contribution. Therefore, studs cannot be classified as a rolling contact fatigue defect. Due to this, a different name called “stud” has been suggested by Grassie (2012) to distinguish with the historical squats defect.

Based on the metallurgical analysis conducted by Grassie et al (2011), there are some observations associated with stud defects. Firstly, there is almost no accumulated plastic deformation at stud except at the crack mouth. If the plastic deformation is present, it could be very localised and no significantly distributed along the rail. Thus, it cannot be considered into the level of ductility exhaustion that is typically seen in RCF crack. Secondly, there are patches of the white etching layer located closely to the defect which suggested that the mechanism of its formation is due to thermal phase transformation. Indeed, the presence of WEL indicated that rail steel was heated up at high temperatures and followed by rapid cooling. In addition, the pro-eutectoid ferrite boundaries were observed within the white etching layer’s structure and this confirms how little amount of plastic deformation taking place. These observations are shown in Figure 2.41 which reveals a transverse section of the white etching layer taken from the vicinity of the stud defect.
To compare with historical squat, Figure 2.16 (b) shows a transverse section taken at squat defect by Clayton et al (1982). It can be seen that the grains are severely flattening in the central region due to rolling contact fatigue, which often changed to an outward shearing of the grains at the edges of the running band.

The hypothesis has been proposed to explain the mechanism of stud formation. Only some key points are summarized as follows:

- Studs are initiated by thermal damage of the rail, which results from limited wheel slip, possibly associated with localised areas of poor adhesion.

- Studs are associated with vehicles having AC traction or thyristor-controlled DC traction because wheel slip is better controlled to permit operation at higher traction ratios.

- The mechanism by which studs propagate is unclear but maybe a low cycle fatigue mechanism associated with contact stresses.

The similarities and differences between squats and studs showing below have been summarized by Grassie et al (2011).
2.9 Summery

In this chapter, a comprehensive review of squat defects in rails was presented. Squats are recognized as an important rail track issue and represent a serious problem affecting significant parts of railway networks across the world. The mechanism of squat initiation; propagation and fracture were presented and discussed. Numerous field investigations, metallurgical analysis, modelling and laboratory studies have been carried out on the formation mechanism of squat defects; however, it is still a matter of controversy.

Squats are considered as a consequence of rolling contact fatigue (RCF) damages. But there is still an argument on how to promote crack initiation due to RCF. Research efforts have included possible relationships between the narrow and brittle white etching layers (WELs) observed on rail surfaces and squat initiation in rails. Studies were considered to
be caused by rail thermal damage linked with the formation of the white etching layer. However, based on current literature, the relationship between WELs and squat initiation is not fully understood. The metallurgical evidence from ex-service rails is not examined in details and a direct correlation between squat and WELs is still missing.

The nature of WELs and their formation mechanisms are still a matter of controversy. WELs are believed to be created by (i) a result of the thermal process caused by wheel slip; (ii) mechanically induced by severe plastic deformation; (iii) a combination of the thermo-mechanical process; squats can develop below them. Controversy continues partly because depending on the particular study performed, the microstructures of WELs have been both identified and described differently as; martensite, nanocrystalline Fe-C alloy; deformed pearlite, nano-crystalline martensite, austenite and cementite; and martensite and retained austenite.

White etching layers are expected to occur at the running band region of wheel and rail contact because only contact patch on rail surface is exposed to highest shear stresses and heat caused by wheel slip during excessive traction or braking effort. However, the question can be raised: if the WEL is found away from the running band region, it could consider another precise formation mechanics. Accordingly, there is no systemically research on the electrical leakage between the wheel and rail (arching) as a formation mechanics of WELs on rails. In the railway system, particularly the electrified railway, arcing phenomenon (electrical leakage) has been occurring more often and regularly on the rails. In the electrified railway, the current is delivered to the train traction motor via contact wires and then returned to the station via the contact between the wheel and the rail. If the contact between the wheel and the rail is temporarily loosing due to dynamic vibration, entrapped dirt, water, and lubricant, or random existence of oxidation/rust on rail/wheel surface, the high current can jump through the gap and results in electrical arcing. During the arcing, extensive heat is generated and speared over a large area of the top surface of the rail. Thus, arcing can be considered as another heat source in the railhead (rather than wheel slip) those results in a change of microstructure and the formation of the white etching layer.

Squats growth and development were found to have driven by rolling contact fatigue and the presence of fluid such as water significantly enhances the crack growth rate. Although numerical modelling and laboratory tests support the hypothesis of hydraulic entrapment
on squat growth, there is no reported direct metallurgical evidence of rails damaged by the presence of water under service conditions in a rail track network. The requirement for fluid to cause small cracks to propagate provides a rational of why squats vanish as soon as the railway enters a tunnel. The questions can arise concerning the physical evidence of water entrapped in cracks for an in-service track, and what evidence if any, is there that the surface water promotes crack growth? In addition to the mechanics of hydraulic entrapment on crack propagation, is there any other likely mechanisms operative?

Based on the literature review, the development of the cracks can be decomposed into three phases, including crack initiation, crack propagation and fracture failure. The cases of squat induced fracture failure have been very few reported, and there is also a lack of associated investigations on the causes of fracture and formation mechanics of failure, which can provide a clear understanding how cracks develop down into the transverse plane and result in transverse defects (TDs). In addition, the physical evidence that shows connecting between squats and transverse defects (TD’s) occurrence is insufficient.
3. Mechanism of squat initiation and correlation with the occurrence of white etching layer on the rail surface

3.1 Introduction

Squats are recognized as an important rail track issue related to rolling contact fatigue and represent a serious problem affecting significant parts of railway networks across the world. However, the mechanics of squat initiation is a still matter of controversy. Recent researches have indicated a possible relationship between the narrow and brittle white etching layers (WELs) observed on rail surface and squat initiation. Grassie (2012; 2016) and Grassie et al. (2011) argued whether the rail surface defects were actually rolling contact fatigue induced squats. He proposed the concept of “stud”, which develops without severe plastic flow and associated with WELs at the top layer of rail under a lower borne tonnage. In this chapter, the structural and microstructural features of ex-service damaged rails containing cracks (squats) and WELs were investigated in order to determine whether there is a correlation between squat defects and the presence of white etching layers.

3.2 Ex-service damaged rails

Thirty-four ex-service damaged rail sections, cut from different re-railing sites in New South Wels (NSW), Australia was provided by Sydney Trains and delivered to the University of Wollongong for investigation. Figure 3.1 shows the rail samples as received.

The samples consisted of two types of steel grades; head hardened and standard carbon pearlitic steels. The chemical composition and mechanical properties of rail samples are given in Tables 3.1 and 3.2, respectively. The rails are manufactured to meet Australian standards AS 1085.1 – 2002 (2002).
Figure 3.1 Ex-service damaged rails (As received).
The damaged rails were visually inspected after the rust removal from the rail surface using WD-40 lubricant cleaning product. Upon inspection, the rail steel sections were found to have contained defects, including surface cracks, squats at mild, moderate, severe and very severe stages as classified by Kerr et al. (2008), as well as spallation, grinding marks and other forms of wear associated with various damage stages. Squats were observed on both running band surface and gauge corner position, as well as squats, were also found at the weld section. Figure 3.2 shows typical examples of rail surface defects including squats.

Table 3.1 Chemical composition of rail samples.

<table>
<thead>
<tr>
<th></th>
<th>C,%</th>
<th>Mn,%</th>
<th>Si,%</th>
</tr>
</thead>
<tbody>
<tr>
<td>Standard carbon (SC)</td>
<td>0.78</td>
<td>0.9</td>
<td>0.2</td>
</tr>
<tr>
<td>Head hardened (HH)</td>
<td>0.78</td>
<td>1.0</td>
<td>0.2</td>
</tr>
</tbody>
</table>

Table 3.2 Mechanical properties of rail samples.

<table>
<thead>
<tr>
<th></th>
<th>Typical tensile strength, MPa</th>
<th>Typical elongation, %</th>
<th>Hardness, HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>Standard Carbon (SC)</td>
<td>920</td>
<td>11</td>
<td>280</td>
</tr>
<tr>
<td>Head hardened (HH)</td>
<td>1200</td>
<td>10</td>
<td>380</td>
</tr>
</tbody>
</table>
Mild and moderate squats

Mild and moderate squats

Very severe squat at weld
Figure 3.2 Typical examples of rail surface squat defects.
3.3 Methodology

Based on the severity of squat defects, four damaged rails were selected for investigation. All these rails are head hardened types. Each rail consists of squats at different stages of developments associated with the presence of WELs on the rail surface. The degree of damage on each rail is varied between the samples, showing some cracks and material surface damage development along the running band area. High-resolution synchrotron X-ray diffraction, optical and electron microscopy, as well as hardness indentation, were introduced to (i) investigate the surface evolution associated with the crack development and WEL formation along the running band region; (ii) understand the microstructural changes of the rolling contact regions, structural and microstructural characteristics of the top surface and subsurface regions (cross-sectional area). Samples were obtained from WELs regions, surface cracking area of squats at mild-moderate-severe stages, and from the running band regions where no cracks appear. Samples from bulk material (undamaged region) were also obtained and considered as a reference sample.

3.4 Experimental Details

3.4.1 Rail samples examined

Four sections of worn rails were chosen based on the degree of damage. Each rail section was about 1-2 m in length and taken from two different lines (passenger and mixed). Rails No. 1 and 2 were replaced from a section where the vehicle's speed is steady (low braking utilisation) while Rails No. 3 and 4 were cut from a track where the train approaches the station (high braking utilisation). It means that each group of the rails experienced different load history and speed conditions. The information about the load history, speed and geometric location of the section examined are summarized in Table 3.3. Each section was thoroughly cleaned and etched using a 2% Nital solution in order to identify the WELs on the rail surface.
<table>
<thead>
<tr>
<th>History and geometric location</th>
<th>Rail No.1-2</th>
<th>Rail No.3-4</th>
</tr>
</thead>
<tbody>
<tr>
<td>Rail type</td>
<td>High rail</td>
<td>Low rail</td>
</tr>
<tr>
<td>Track type</td>
<td>Mixed (passenger and freight)</td>
<td>Passenger</td>
</tr>
<tr>
<td>Axle load (ton)</td>
<td>22</td>
<td>19</td>
</tr>
<tr>
<td>Curvature radius (m)</td>
<td>1000</td>
<td>1598</td>
</tr>
<tr>
<td>Track speed (km/hr), maximum</td>
<td>80/115/125</td>
<td>80</td>
</tr>
<tr>
<td>location</td>
<td>Steady speed through the track section (low braking utilities)</td>
<td>Approach the station (high braking utilities)</td>
</tr>
</tbody>
</table>

Twenty locations including WELs associated with mild and moderate squats on the rail surface have been chosen for characterization. The state of rail surface at each location chosen from Rails No.1, 2, 3 and 4 is shown in Figure (3.3), (3.4), (3.5) and (3.6), respectively. For the crystallographic structure of surface analysis, samples from (i) WELs; (ii) surface cracking area (Squats); (iii) uncracked region, were all located on the running band area along the rolling direction. Samples from the gauge corner and field-side area were also cut from some particular locations and examined in order to investigate the microstructural changes of the surface in both longitudinal and transverse directions.
Figure 3.3 Rail No.1, showing the locations examined, including WEL, squat and area on the rolling band that is no visible cracks and WEL formed on the surface.
Figure 3.4 Rail No.2, showing the locations examined, including WEL, squat and area on the rolling band that is no visible cracks and WEL formed on the surface.

Figure 3.5 Rail No.3, showing the locations examined, including WEL, squat and area on the rolling band that is no visible cracks and WEL formed on the surface.
3.4.2 Phase and microstructure analysis

Samples obtained from the top surface of the rails were examined by high-resolution synchrotron X-ray diffraction in order to determine the changes in the phase structure of the rail surface at each location toward the rolling direction. High-resolution synchrotron X-ray experiments were performed using the Powder Diffraction (PD) beamline at the Australian Synchrotron, Victoria, Australia. Detailed information of beam energy and radiation used has been described in chapter 1 in section 1.5.

The quantitative analysis was performed of each diffraction pattern in order to identify the phases in each sample. The content of retained austenite (RA) was quantitatively determined by the calculating the integrated intensity of (110), (200) and (211) peaks of ferrite/martensite and the (111), (200) and (220) peaks of austenite. The volume fraction of austenite was calculated using the formulas provided by ASTM standard (2008).

The lattice parameters, $c$, and, $a$, were determined based on Bragg’s law in order to identify the crystal structure of the top surface layer. The $c/a$ ratio was calculated to determine whether the lattice structure has a tetragonal crystal associated with the formation of supersaturated of solid carbon (martensite structure). When the martensite
is present, the diffraction pattern of the (200) reflection will be splitting into different crystals planes. Therefore, the (200) peak was chosen for close analysis as it can indicate either Ferritic or Ferritic/ Martensitic phases. A double pseudo-Voigt function was used for peak analysis and c/a determination. The percentage of carbon content was also calculated by applying a linear relationship function with the c/a tetragonality of martensite that was found in the equation supplied by Bhadeshia and Honeycombe (2017).

For the metallographic analysis and observations, samples were cut from squats and from WELs that was located in the vicinity of squats in order to investigate the microstructural changes on the top surface layer. Then, they were prepared by using a standard metallographic procedure described in chapter 1 in section 1.5.5.

Hardness indentations were performed on both the top surface layer and cross-sectional area. For the top surface samples, the hardness test was measured at twenty locations shown in Figures (3.3), (3.4), (3.5) and (3.6), using macro Vickers hardness tester with a 1 kg load. For cross-sectional samples, the microhardness test was performed on the etched surface, from the top surface down to a certain depth using a micro Vickers hardness tester with 100 g load.

3.5 Results and Discussion

3.5.1 Reference Samples

The reference samples are the specimens obtained from the bulk material of the tested rails. They were cut from the 15 mm depth bellow rail surface, where the material is considered to be unaffected by any circumstance that occurred due to train passage. The samples were examined to determine the predominated phases and microstructural characterisation of the bulk material and then to compare with the results of the samples obtained from the top surface layer. This can provide the details regarding the structural and microstructural changes of the surface material which can lead to squats formation on the railhead.

Figure 3.7 shows a full XRD diffraction pattern of the reference samples measured at 15 keV. After indexing, the reflections indicate that the material of tested rails consists of a predominately alpha-iron phase (α), and has a body center cubic structure (bcc). The
phase of austenite structure (fcc), as well as the characterisation of asymmetric profile shape, were not detected on both profiles. The only slight variation was found on the diffraction pattern of Rails No. 1&2, showing the presence of a very weak reflection of (211) peak of cementite at a diffraction angle of 25.9°. It appears that the strongest reflections of the samples measured at 15 keV are located between the diffraction angle of 20°<2θ<45° which corresponds to (110), (200) and (211) of ferrite peaks comparing with other planes. Therefore, the only peaks with the mentioned range of the diffraction angle will be just considered in the analysis.

![Figure 3.7 Full XRD diffraction pattern of the reference samples at 15 keV.](image)

3.5.2 Synchrotron diffraction investigation

In order to identify the crystal structure of the rail surface at defects and adjacent layers, a high-resolution synchrotron X-ray diffraction experiment was performed on the twenty locations on four rails. Radiation was applied to the samples obtained from (i) squat defects, (ii) WEL regions in the vicinity of squats and (iii) running band regions that may
contain some WEL patches. The X-ray results on the samples are presented in the following.

The enlarge section of the x-ray results within a diffraction angle between 20° and 45° has been plotted for Rail No. 1, 2, 3 and 4. Figure 3.8 shows the diffraction profile of the surface of Rail No. 1, through point 1: WEL followed toward rolling direction by 2: Squat (Mild), 3: Rolling Band (RB) sample, 4: Squat (Moderate), 5: WEL, 6: Squat (Mild) and 7: Squat (Moderate). The result shows a significant variation in the crystal structure. The full-width half-maximum (FWHM) increased gradually in the sequence: reference sample, running band, squats, and WELs (near squats). In addition to this trend, the peak intensities decreased considerably and the peak locations shifted toward lower diffraction angles. Regarding phase structure, significant variations were observed on the diffractions patterns between the samples obtained from the rail surface and the bulk material sample (reference). The first critical observation is that the two strongest peaks (212) and (115) of carbide, which was indexed based on the standard diffraction pattern of carbide Fe₃C at 15keV, have appeared in the diffraction spectra of the WELs and squats region. The less strong peak (200) of carbide was also observed in the diffraction pattern of 1:WEL and its adjacent region of point 2:squat. The strongest peak of carbide, (121) is thought to be completely overlapping with the strongest peak of (110) reflection. It is also shown that the peaks of Iron Oxide (Fe₂O₃) were found within the diffraction angle between θ=20-22° in the diffraction spectra of points 1, 2 and 7 correspondings to WEL, mild and severe squats.

It can be seen that the single peak reflection of (200) α in the diffraction pattern of the reference has split into two overlapping peaks of (002) α’ and (200) α’ crystal planes, indicating to the distortion of ferrite into tetragonal martensite. In addition, splitting peak characterisation was also observed in the diffraction peak of (211) near the diffraction angle of θ~41°, particularly at the diffraction pattern of point 1, 5 for WELs and point 2, 4 for squats.

Regarding to Austenite peak intensities, based on the standard diffraction pattern of Austenite structure (γ) shown below the figure, the austenite peaks were not found in any diffraction profiles measured along the surface of Rail No.1. The absence of austenite peaks reflection in the diffraction patterns could indicate the absence of thermal activity induced on the rail surface.
The X-ray results indicate that there were significant changes in the crystallographic structure of the rail surface along the running band area, compared to those obtained from the bulk material. The microstructural changes were found to be associated with the state of the rail surface. Samples from squats and their adjacent regions containing the white etching layer reveal the similar corresponding results. Samples from the rolling band area, where cracks and WEL patches were not detected show almost similar characterization of phase structure to the bulk material. This finding concludes that there may be a correlation between WEL formation and squats occurrences.

Figure 3.8 Comparison of Synchrotron XRD results of the different regions on the surface of Rail No.1.
Figure 3.9 shows the enlarged diffraction patterns of Rail No.2 obtained from different six locations. Radiation was applied on point 8: WEL accompanied with Squat (Mild), 9: WEL only, 10: Squat (Mild), 11: Rolling Band (RB) region (crack and WEL free), as well as two samples obtained from location 12, from rolling band and from field-side region (away from rolling contact region). Generally, the diffraction patterns of the rail surface whether on the rolling contact region or field-side region are significantly different than reference material. The diffraction spectra of points 8, 9, 10 and 12 correspondings to WELs and squats have shown the presence of the strongest carbide peaks of (212) and (115) of crystal plans. The strongest peak intensity of (121) is completely overlapping with the peak intensity of (110) reflection. The carbide peaks observed on both WELs and squats reflection was totally absent on the diffraction spectra of point 11 (RB) in spite of a very close location. The (200) peak of Fe$_3$C was found in the diffraction pattern of WEL (point 9) and squat (point 10). It was also found the existence of iron oxide (Fe$_2$O$_3$) peaks within rang angle of $\theta=20$-$22^\circ$.

To compare with the reference reflection, the peak intensities have shifted toward the lower diffraction angle (clearly seen at the intersection of (220) peaks of points 8, 9, 10 with reference peak). In addition, peak splitting phenomena were observed in the diffraction patterns of WEL and squats while the single peak reflection was only the characterisation of the RB spectrum. The intensity of cementite peak that was observed in the bulk material (displayed in Figure 3.7), has totally disappeared in the diffraction patterns of all considered samples. In relation to peak broadening, the (110), (200) and (220) peaks in the profiles of WELs and squats are shown to be substantially broadened, indicating to the refinement of crystallite size. As well as the reference samples, the austenite peaks were not detected in the diffraction pattern of WEL and squats.

However, the austenite peaks are clearly visible in the diffraction pattern of point 12 and point 12a. The (200)$_\gamma$ and (220)$_\gamma$ peaks of austenite, which are indexed based on the standard reflection of Austenite at 15 keV were found to be taken apart in the diffraction pattern of the both running band and field-side samples. The (111) $\gamma$ peak of austenite is clearly visible in the diffraction pattern of field-side specimen. For running band sample (point 12), the carbide peaks that were found in the diffraction pattern of the WEL and squats sample (points 8, 9, 10) samples were also detected but become barely visible.
The X-ray results indicate that there are two different phase structures of WEL: one associated with the carbide (Fe₃C) phase structure and another is associated with the presence of retained austenite peaks in the diffraction patterns. Correlation between WEL and Squats was also confirmed by the above X-ray analysis.

Figure 3.10 shows the calculated diffraction pattern of Rail No.3, measured at different five locations. The measurements were performed on the samples, cut from running band area toward the rolling direction in the sequence: 13: Rolling Band (RB) area, 14: WEL, 15: Squat, 16: WEL, and 17: Squat (severe) (as displayed in Figure 3.5). The x-rays result
shows a totally different crystal structure than those observed in the previous samples of Rails No. 1 and 2. The (200), (212) and (115) peaks of carbide (Fe₃C) were not detected on the diffraction spectra of WELs and Squats sample (point 14,15,16 and 17). Instead, the (111) γ, (200) γ and (220) peaks of austenite reflection were in the diffraction pattern of points 14, 15, 16 and 17(WELs and squats sample). The (311) peak of austenite was only observed in WELs spectra. These austenite peaks were not found in the diffraction pattern of the RB sample (point 13), as well as in the reference samples. The presence of austenite peaks in both diffraction pattern of WELs and their adjacent regions (squats), and the absence of these peaks in the rolling band area where the cracks and WELs were not observed, indicates both WELs and squats are related.

In comparison with the diffraction pattern of the undamaged sample (reference), The peaks shifts very slightly toward the lower diffraction angle and the asymmetric profile phenomena were found in the (200) and (220) reflections of WELs and Squats.

Figure 3.10 Comparison of Synchrotron XRD results of the different regions on the surface of Rail No.3.
For Rail No.4, the synchrotron radiation was used to determine the crystal structure not only on samples from the running band area but also from the field side and gauge corner regions. Three locations (18, 19 and 20) containing squats and WELs were chosen for investigation. The XRD results for all samples are shown in Figure 3.11.

Samples from location 18 show a different crystal structure. In the field side region, the diffraction pattern reveals the (111) γ, (200) γ, (220) γ and (113) γ crystal planes of austenite reflection while these peaks were not found in the diffraction pattern of running band and gauge corner. This confirms that WEL was not always located on the rolling contact region. In addition, the (200) peak in the profile of WEL was found to have an asymmetrical shape, compared to the symmetrical peak shape of RB and GC reflections. Comparing between RB and CG reflections, the peaks broadening were seen to be slightly increased in the diffraction pattern of RB, indicating microstrain produced in the lattice structure, possibly due to plastic deformation.

In the same way, the crystal structure of the rail surface at location 19 was determined. The diffraction pattern at squat (19A) and GC (19C) looks almost identical, showing no additional phase structure other than ferrite structure with some peaks broadening due to grain refinement. While the x-ray result obtained from the field-side area (19B) confirms an additional austenite phase in the crystal structure.

Similarly, the XRD profiles of location 20, obtained from the surface of the squat (20A), field-side or WEL (20B) and gauge corner (20C) show the almost similar crystal structure of those in the locations 18 and 19.

Generally, the diffraction pattern shows that the crystal structure along the gauge corner and running band area containing squats is almost identical; however, it appears significantly different along field side regions due to WEL existence, showing the formation of austenite phase structure. The XRD results also show that the peak intensities of austenite along the field side region decreased considerably from location 18 to 20, indicating that the field side region was not continuously covered by WEL.
The peak profile 200 of WEL, Squat and rolling band area

Figure 3.12 shows the diffraction pattern near the (200) reflection of the ferrite phase (body-centered cubic, BCC), obtained from regions of, WEL, Squat and rolling band area of Rail No.2. The peak was chosen for close analysis because it can indicate either ferritic or ferritic/martensitic phases. It can be clearly observed that the peak profile of WEL has an asymmetric shape which indicates that a separation of the (002) α to the different crystal planes associated with the distortion of the ferrite into tetragonal martensite. At squat defect, the peak profile (200) has almost corresponded to the peak profile of WEL, in which showing an asymmetric shape, reflection related to two overlapping peaks and...
belonging to 002/220 reflections of martensite. In contrast, the peak reflection at the rolling band area where cracks and WEL patches were not observed on the surface is distinctly symmetric and has a very narrow full width at half maximum (FWHM). It is clearly distinguished that the asymmetry of the 200 reflection gradually decreased in the sequence of WEL, Squat followed by undamaged area, indicating that the crystal structure of the rail surface has significantly changed within a very short distance, showing a clear correlation between WEL and squat.

For Rail No.3, the peak profile (200) of the rolling band area, WEL, and Squat is shown the Figure 3.13. It appears that the characterisation of the peak profile of these regions is consistent with those from Rail No.2. The symmetric profile shape in the region just before WEL was found, but become obviously asymmetric shape in the adjoining regions: WEL and squat. This indicates that the phase structure has changed from ferrite at rolling band area to martensite at both WEL and squat. It means that martensite formed locally on the rolling band area. As squat contains martensite structure, it is likely for squat to be related to WEL.
Figure 3.12 Characterisation of the 200 peaks of WEL, squat and rolling band area (undamaged region), on the surface of Rail No. 2.
Figure 3.13 Characterisation of the 200 peaks of rolling band area (undamaged region), WEL and squat, on the surface of Rail No. 3.
Crystal structure, Lattice parameter, and carbon concentration

The lattice parameter, *c*, *a* of the tetragonal crystal structure of the examined rail surface were quantitatively calculated based on Bragg’s law. Figure 3.14 shows the distribution of the *c/a* ratio of martensite along the running band area, from point 1 (WEL) to point 7 (Squat). The result indicates that the maximum value of the tetragonality of martensite (*c/a* ratio) was found at WELs peaks, reached to 1.011 at S1 and 1.008 at point 5. Slightly lower values were measured at Squats defects. The *c/a* ratio equals 1 at the rolling band area, indicating that the lattice is fully cubic (bcc) crystal structure. The *c/a* ratio of martensite is a function of carbon content in the structure. Based on the linear correlation between *c/a* ratio and the percentage of carbon content (wt%), the distribution of carbon content was calculated. Figure 3.15 shows the distribution of carbon content along the rolling contact area containing WELs patches, squats and rolling band regions (squats and WEL free). The maximum amount of carbon content reaches to 0.26% wt and 0.18% wt on WELs regions (points 1 and 5). It means that about one-third of the carbon concentration content of the steel material was dissolved in the WEL.

![Figure 3.14 Lattice parameter a, c and c/a ration along the surface of Rail No.1.](image-url)
Figure 3.16 shows the distribution of the c/a ratio along the rolling band area of Rail No.2. The results indicate that both squats at mild stage (point 10) and the adjacent area containing WEL patches (point 8, 9) have a tetragonal lattice structure compared with the region just followed by mild squat (S11). A relatively small c/a ratio was found in the region adjoin obtained the moderate squats. At the field side corner (point 12a), 20 mm distance (in the y-direction) from point 12, away from the rolling band area, the tetragonal structure of martensite was determined. The result shows that the higher c/a ratio was found in the field side region, reached of 1.011, which indicates that WEL was also present in the field side region, not only the rolling band region. The corresponding carbon content percentage along the rail surface of Rail No.2 is shown in Figure 3-17. It appears that the solute carbon content decreased gradually from 0.17% wt in WEL to 0.12% wt in the squat and become 0.0% wt in the rolling band area where neither squat nor WEL is present. In addition, there is a significant variation in carbon content between field-side region (point 12a) and rolling band or squatty region (point 12) although they are located in the same vertical direction. In field-side region, the carbon content reached up to 0.22% wt while it is just 0.05% wt of carbon content in the adjoining squatty region.
Figure 3.16 Lattice parameter $a$, $c$ and $c/a$ ration along the surface of Rail No.2.

Figure 3.17 Distribution of carbon content along the surface of Rail No.2.
Figure 3.18 and 3.19 shows the distribution of c/a ratio and the carbon concentration along the surface of Rail No.3. A relatively higher c/a ratio was measured in WEL patches. Although the c/a ratio was by comparison small in squats region, the structure of the surface from squats still contains a tetragonal martensite structure. The corresponding carbon concentration in the WELs patches reaches 0.2% and 0.16% wt, while in squats reaches 0.08% and 0.04% wt.

Figure 3.18 Lattice parameter a, c and c/a ration along the surface of Rail No.3.
Figure 3.19 Distribution of carbon content along the surface of Rail No.3.

For Rail No.4, the tetragonal structure of martensite was determined along the rail surface in both transverse and longitudinal directions. Figure 3.20 and 3.21 reveals a variation of c/a ratio and the corresponding carbon concentration in the transverse direction through the locations of 18, 19 and 20. A large variation of a tetragonal distortion (c/a ratio) was measured across location 18. The maximum c/a ratio was found in the field side region, reached up to 1.011, while c/a equals 1.002 and 1 in the rolling band area and at the gauge corner. Similarity, the distribution of c/a ratio through location 19 was almost consistent to location 18, with some variation in the c/a ratio measured at field side region. However, a very slight variation in the tetragonal distortion (c/a ratio) was found between the gauge corner, rolling band area, and field-side region through the location 20 from. The distribution of carbon content shows compatibility with the variation of the lattice structure, in which maximum carbon content concentration was found in the field side region where the WEL is present.
In the longitudinal section, the lattice structure parameter and the corresponding solute carbon content were determined along the rolling band area, gauge corner and field-side corner. Figure 3.22 and 3.23 presents the distribution of c/a ratio and carbon content as a
function of distance. It appears that no much change in the lattice structure was found along the rolling band area and gauges corner even though the measurement was conducted on uncracked area, mild squat and moderate squat. In contrast, a large change in the lattice structure was found along the field side region where the WEL exists. The c/a ration significantly decreased toward the rolling direction, from 1.011 at location 18 to 1.002 at location 20, indicating that WEL was not homogenous over the field side region, possibly due to squat formation. Correlation with the distribution of carbon concentration along the field side region was confirmed, showing maximum carbon amount dissolved in the WEL region while this amount significantly decreased when squats took place on the rail surface.

Figure 3.22 Lattice parameter a, c and c/a along the surface of Rail No.4.
• **Determination of Retained Austenite**

The volume fraction of retained austenite was determined by calculating the integrated intensities of the austenite peaks to the ferrite/martensite peaks in the diffraction pattern. The quantity of retained austenite was 0% on the rolling band surface of Rail No.1, Figure 3.24, even though martensite was detected in the WELs and squats region. For Rail No.2, the retained austenite was not found on the part of the rolling band area containing WELs and mild squats, Figure 3.25. However, the retained austenite was detected on location 12 where two moderate squats have taken place on the rail surface. A significant quantity of retained austenite, about 17%, was found on the field side corner, compared to just 2% retained austenite on the adjacent running band area.
Figure 3.24 Retained austenite content along the rolling surface of Rail No.1.

Figure 3.25 Retained austenite content along the rolling surface of Rail No.2.
On the other hand, the measurement confirms that the retained austenite content presents along the rolling band area of Rail No.3. Figure 3.26 shows the distribution of retained austenite along the surface containing squats and WELs. The maximum amount of the retained austenite appears at points 14 and 17 where the white patches present on the rail surface. While, the content of retained austenite relatively decreased on the regions just next to WEL, at points 15 and 17 where the squats have formed on the rail surface. In addition, the retained austenite was not discovered on some part of the rolling band area (Point 13) which is free of surface damage such as squats and WELs.

The retained austenite was measured along the rolling band area, gauge corner and field-side corner of Rail No.4, Figure 3.27. The result shows that neither rolling band area nor gauge corner contained retained austenite in the structure. However, a frequent amount of retained austenite was found on the field side region. At point 18 on the field side region where WEL formed, the retained austenite content reached up to 13% and decreased less than one quarter on the point 19 and become 0.72% on the point 20.
• **Hardness measurement**

Figure 3.28 shows the distribution of hardness along the rolling band surface of Rail No.1, through point 1: WEL to point 7: Squat. It appears that the hardness values were not consistent along the rolling surface, indicating a highly inhomogeneous microstructure of the top surface layer of the rail. The maximum hardness values, about (650-710HV) were measured in the WELs region (points 1 and 5). On the squats region, the hardness reaches 600 HV. The lowest hardness value, about 470 was found on the rolling band area where the area is clear of squats and WELs. In comparison with the hardness of the bulk material (380HV), the hardness on the rail surface became nearly double.

In the same way, the distribution of harness along the rolling band area of Rail No.2 including points 8 to 12 is almost comparable, in which the hardness value decreased in the sequence of WEL, squat and uncracked area (RB), Figure 3.29. At point 12, the hardness near field-side region (12a) is significantly high, making up 800HV, compared with the adjoining region on the rolling band region.
The distribution of hardness along the contact surface layer of Rail No.3 shows a similar attitude of those in Rail No.1 and 2. The maximum hardness values of more than 730HV, are reached on the WEL (point 14).

For Rail No.4, hardness distribution of the rail surface is shown in both transverse and longitudinal directions, Figure 3.31 and 3.32. In the transverse direction, a large variation was found at point 18. The hardness value, measured on the rolling band area was almost close to this on the gauge corner, while the hardness on the field side region was significantly high, reaching up to 800HV. Similar behaviour of the hardness distribution was found across points 19 and 20, but the variation along the three regions (rolling band, gauge corner, and field side) was less intense. In the longitudinal direction, the measurement shows that hardness was almost homogenous along with the rolling band and gauge corner although the measurement includes the surface of mild and moderate squats region. Whereas, the hardness values along the field side region were significantly varied, showing higher value at point 18 and lower value at point 20. This is due to squats formation on railhead.

![Figure 3.28 Hardness distribution along the rolling surface of Rail No.1.](image-url)
Figure 3.29 Hardness distribution along the rolling surface of Rail No.2.

Figure 3.30 Hardness distribution along the rolling surface of Rail No.3.
Figure 3.31 Hardness distribution across surface of Rail No.4.

Figure 3.32 Hardness distribution along the rolling surface field and gauge corner of Rail No.4.
3.5.3 Microstructure analysis by microscopes and microhardness measurement

The microstructural characterisation of the top surface layer on the rail cross-section was investigated. The observation was performed using optical and scanning electron microscopy. The investigation was focused on the WELs region and the adjacent squat at the early stage of development.

(1). Rail No.1

Figure 3.33 shows part of the surface of the Rail No.1, containing mild squat and WEL. The WEL band appears on the slightly bright shining rolling surface area with a width of about 10 mm. Squat with cracks length of 18 mm and 6 mm of leading and tailing cracks were found to have formed on the rolling band area within the WEL band. Samples A and B were obtained from WEL and Squat and then prepared for observation in both longitudinal direction (LD) and transverse direction (TD).

![Figure 3.33 Part of the surface of Rail No.1, showing samples A and B taken from WEL and squat for metallographic examination.](image)
Figures 3.34 and 3.35 show results of the microscopic analysis for longitudinal and transverse cross-sections respectively of sample A, revealing the white etching layer (WEL) on the surface. The longitudinal section at low magnification shows White Etching Layer WELs with a bright featureless contrast, characterized by homogenous structure, sitting on the top surface layer of the rails. It appears that the white etching layer (WEL) was present on a part rather than the entire top layer surface of the rail and that the WEL has formed in the different thicknesses. Small WEL patch with thickness about 15 μm and shape of half an ellipse was found to have individually formed, as seen in the middle. For further details, part of the rail surface, as referred by the red box has been enlarged. Observation from the SEM image shows two important microstructural characteristics of the WEL. The first is the formation of fine vertical cracks through the WEL and the second is the presence of plastic deformation underneath WEL. The crack was found to have initiated at the edge of the WEL where the stress state at this point (between WEL and deformed rail surface) is varied. The crack initiated from WEL edge and transmitted the whole WEL with a direction angle of 38° and then propagated down into the rail bulk material. Inspection of the crack reveals propagation into the rail steel along the shear deformation direction of the deformed pearlite (underneath WEL). At a certain depth where the plastic deformation becomes less severe, the crack has branched into two different cracks. These cracks developed down within the direction angle of 66° and 98° to the rail surface. The growth behaviour of these cracks may indicate that the cracks propagated through the grain boundaries. In addition, the grain boundaries of ferrite can be clearly seen, extending from WEL down into the rail material.

The micro-indentation hardness distribution of WEL with depth was measured and plotted, as shown in the corresponding figure. WEL exhibits high hardness values- with a range of up to 770 HV. However, the hardness across WEL depth was not uniform. The top and lower surface layers of WEL showed less hardness than the middle region of WEL. As the depth increased from the top surface layer, the hardness values suddenly decreased from 690 HV at the transition zone to about 450 HV at the plastic deformation region underneath WEL. The hardness then gradually decreased as the depth increased, until the value reached 380 HV at depth about 40 μm from the top surface which includes about 30 depth of WEL and 10 μm of the deformed region.
In the transverse direction (TD) (Figure 3.35), a white etching layer with a uniform thickness of about 22 μm was found to cover the entire surface of the rail contact layer. Not only at the WEL edge but vertical cracks were also observed in the middle region of WEL. The top surface layer of WEL appears to have spalling. The SEM enlarged image displays part of the rail surface in the spallation region. Two fine cracks were found to have initiated from WEL, transmitting the whole WEL thickness. One of the cracks just turned down to the material following the sheared grains and the other developed vertically, crossing the ferrite grain boundary existed in the WEL region. Underneath WEL, the pearlite lamella structure is seen to have heavily deformed indicating the presence of the work hardening process. The strained lamella microstructure is orientated in the gauge corner direction.

The corresponding microhardness measurement performed across the transverse section shows similar distribution of that measured on the longitudinal section, in which the maximum hardness value was measured at the middle layer of WEL compared to upper and lower regions, as well as the hardness values reached the material standard value at depth about 45 μm from the rail surface.

Figure 3.36 shows the surface characterisation of the rail sample B, obtained from the squat area, just close to cracked WEL (sample A). At low magnification optical image, a squat initiated crack was found to have formed on the rail surface and propagated down into the rail material in the rolling direction with an angle of about 35° to the rail surface. At the rolling surface layer, small patches of the white etching layer were present at the different locations along the surface. To obtain more information, parts of the rail surface have been characterised by SEM. The enlarged image shows the microstructural characterization of the squat surface region, characterised by the presence of fractured WEL, as appeared in dark grey colour. The observation has also confirmed the presence of plastic deformation just below the WEL. There was also a sharp transition layer between the deformed layer and WEL. In addition, the upper part of the surface layer appears to be irregular, indicating that WEL was partly removed due to a high impact load.

Figure 3.37 shows the SEM images with the corresponding microhardness as a function of depth from the top surface layer down to the rail bulk material. These images display the characterisation of the surface layer of the area before crack mouth in relation to the
rolling direction, in both longitudinal and transverse directions. It can be seen that non-homogenous patches of WEL were irregularly distributed on the top surface layer of the squat. The WEL thickness within the short distance is significantly varied, the thick layer reached to 20 μm while the thin layer was less than 5 μm. Spalling was also found in the top layer of WEL. Below the WEL, there is a transition zone characterised by a sharp layer, associated with the severe plastic deformation. The plastic deformation found in the longitudinal direction appears more severe than this observed in the transverse direction.

The microhardness distribution across the longitudinal and transverse cross-section is shown with the corresponding images. The measurement was performed through the maximum thickness of WEL, from the direction referred by the arrow. High hardness values were measured at the WEL, reached about 800 HV. Below the WEL, the microhardness values decreased from about 450-500 at the deformed region to about 380-390 at the bulk material.

In summary, the microstructural examination confirmed the existence of a white etching layer on the rolling band surface in the squat and its adjacent WEL region. The white etching layer was found to have varied thicknesses. Cracks were observed to have formed from WEL and extended into the rail material. Underneath WEL, the microstructure feature was characterised by heavily deformed of pearlite grains. Microhardness measurements showed that high hardness values were recorded at the WEL and the hardness decreased sharply as the depth increased toward the matrix region. Comparing with the synchrotron results, this type of WEL that associated with the ongoing severe plastic deformation is composed of a very fine martensite structure with some undissolved fine carbides particles. As the reflections from the squat region and its adjacent WEL area were compatible with the microstructural observations, the results from synchrotron and metallurgical analysis are consistent.
Figure 3.34 Characterization of the top surface of sample A in the longitudinal direction.
Figure 3.35 Characterization of the top surface of sample A in the transverse direction.
Figure 3.36 Characterization of the top surface of sample B at squat region.

Figure 3.37 Longitudinal and transverse section of the region just before the crack mouth in relation to rolling direction.
(2). Rail No.2

Five samples from Rail No.2 were cut for metallographic observation. Samples were obtained from rolling band area including WEL (A), mild squat (b), running band area (C), squatly region (D) and from the field-side region (E), as shown in Figure 3.38. The observed plane is illustrated by arrows, considering the longitudinal cross-section direction (LD) parallel to the rolling direction (RD).

Figure 3.38 Rail No.2, showing samples A, B, C, D and E taken from WEL, squat, undamaged area, squatty region on rolling band and field side region for metallographic examination.

Figure 3.39 displays an optical micrograph, showing the nature of the rail surface layer at region A which was considered the location of WEL. A white etching layer with a thickness of about 20 μm was found on the top surface of the rail sample. Cracks were also observed in the WEL and can be cleanly recognized their existence associated with the spallation of the top surface layer. The cracked WEL area has been enlarged in order to examine the microstructural feature of the surface layer. As the previous samples, WEL appears to have a homogenous structure, characterised by featureless appearance. Below WEL, there is an accumulation of severe plastic deformation, characterised by flattened ferrite/cementite lamella, oriented toward in opposite to the travel direction. The existence of plastic deformation indicates a high shear strain induced on the top surface layer due to action by the wheel on the rail. WEL developed fine vertical cracks, once the cracks reached the transition zone (deformed region), the growth direction is changed to be compatible with the flow direction of the plastic deformation. From the corresponding microhardness measurements, the distribution of hardness values across the WEL and the
deformed region shows a similar pattern of that measured through the samples containing WEL in Rail No.1. The maximum hardness values were measured in the middle region of WEL, reached about 740 HV and the lower hardness, about 380-390 HV was found at depth of about 30 μm from the top surface.

Figure 3.40 shows a low magnification SEM image of the longitudinal cross-section of the squat at region B. It is clearly visible that the crack initiated from the top surface and propagated down into the rail material with the angle about 45°. At about 2 mm depth the trailing crack branched from leading crack and developed almost in the horizontal plane, in the direction opposite to rolling direction (RD). It can be seen that the trailing crack developed inside the material and did not appear on the rail surface.

To examine the microstructure feature of the squat surface layer, three regions 1, 2 and 3, as illustrated by white box were chosen for characterisation. These regions include crack mouth and the area at the front and back of the crack mouth. An optical micrograph, Figure 3-41, shows the crack mouth associated with the presence of a very thin shiny white etching layer along the surface just close to it. Severe plastic deformation can also be observed at the bottom of WEL. The observed features of crack initiation and propagation associated with the presence of the WEL in a squat region strongly correlate with the formation of cracks observed in the adjacent WEL sample (region A).

Figures 3.42 and 3.43 show the surface feature at the front (region 2) and the back (region 3) of the squat surface layer respectively with the corresponding microhardness distribution across the depth. It can be seen that WEL here has a very thick layer, reached up to 40 μm of depth compared with these adjacent to crack mouth. At the front region, although WEL seems to have a less homogenous structure than those observed in the WEL region, microstructural characterisation of the surface layer appears identical. The microhardness measurement shows the random distribution of the hardness values across WEL depth, ranging between 820 to 700 HV, and confirming the irregularity structure of WEL. From the transition zone (deformed region) to a depth of 80 μm below the surface, the microhardness ranges slightly between 460 HV to 380 HV.

On the other hand, the microstructural feature of the surface layer at the back area (region 3) looks unusual. The plastic deformation flow can be clearly recognized inside the WEL region, having two different opposite directions, as referred by arrows. At about 22 μm
depth below the WEL, the pearlite structure was severely deformed, and the shear direction was completely toward the rolling direction. In contrast, the plastic flow at the depth of about 32 μm was normally opposite to the rolling direction. Changing in the plastic flow direction within a short distance could be related to high braking effort induced on the rail surface by the wheel, which results in reversed plastic straining. In addition, an irregular part of WEL was found at the subsurface layer, taken a place inside the pearlite structure. Interpretation of this layer indicates that this WEL was present at the top surface but due to a particular action caused by the wheel on the rail, the top surface layer containing WEL has been severely rolled resulting in the destruction of the upper layer of the surface. It can be seen that the cap was induced inside the material, which can be harmful as it can easily develop into a crack.

The micro-indentation hardness measurement was performed vertically through lines 1 and 2, as shown in the figure. The distribution of hardness values shows consistently with the microstructural features observed in the inspected sample. Across line 1, the hardness varies between 760 HV to 690 HV within depth of about 20 μm and then significantly dropped to about 460 HV at depth of 22 μm, and then suddenly jumped to about 800 HV at the WEL that was found inside the material, and then becomes close to bulk material after more than 70 μm depth. Across line 2, the hardness measurement shows a similar distribution behaviour of the previous sample.

The microstructural examination results of the rolling band area (region C), where is no visible surface damage or effect on the material, is shown in Figure 3-44. It is clearly obvious that there is not WEL on the rolling band surface, instead, a high shear strain characterised by flattening grains can be seen directly below the surface of the rail. The microhardness was measured across the surface, showing a slight variation in the hardness values with respect to depth. The severely deformed region exhibits relatively high hardness values, reached up to 450 HV compared with the bulk material.

An optical micrograph on the longitudinal cross-section reveals two distinguished WELs on the surface, one observed on the rolling band area within the squatty region, Figure 3-45 and the other found on the field side region away from rolling contact area, Figure 3-46. WEL formed on the rolling band area was present in individual patches (not continuous), having almost half an elliptic shape and does not have a high thickness. The sublayer region as appeared in dark grey colour is characterised by heavily deformed
pearlite structure, and dark colour indicates high dislocation density induced by high shear strain. Hardness results were consistent with the characterisation observed in the examined sample, in which WEL exhibits the highest hardness value (620 HV).

In contrast, WEL formed at the field side of the rail appears significantly different. WEL here is characterised by a homogenous structure, uniform thickness reached between 40-50 μm, and no visible spallation and/or cracking occurrence. In addition, the pearlite structure at the transition zone just below the WEL does not show any apparent deformation. The distribution of the hardness with depth shows almost uniform variation across WEL depth with the relatively higher value (about 80 HV) remaining in the middle section of the WEL. At the interface layer between WEL and pearlite structure, the hardness reached 550 HV and then decreased to nearly the same level of bulk material at a depth greater than 50 μm.

In summary, the white etching layer associated with different thicknesses was found in the squat area and its adjacent region. Vertical tiny cracks were formed from WELs and transmitted down into the rail material in the parallel direction of the deformed texture. Below the WEL, the microstructure was characterised by flattening grains confirming the presence of severe plastic deformation at the transition zone of WEL. WEL was also observed on the field side region, characterised by an undeformed underlying structure. Revered plastic straining was seen on the squat surface region.
Figure 3.39 Characterization of the top surface of sample A in the longitudinal direction.
Figure 3.40 Longitudinal section of squat (region B), showing subsurface cracks propagation inside the material, as well as showing the location of three regions (indicated by boxes with numbers) that have been enlarged for characterisation.

Figure 3.41 Optical micrograph image of crack mouth (region 1), showing the presence of thin layer of WEL just close to crack mouth.
Figure 3.42 SEM image showing subsurface feature of the front region (region 2), with corresponding hardness measurement.

Figure 3.43 SEM image showing subsurface feature of the front region (region 3), with corresponding hardness measurement.

Figure 3.44 SEM image showing subsurface feature across rolling band area (region C) where squat and WEL was not observed on the surface with corresponding hardness measurement.
Figure 3.45 Optical micrograph image showing a very thin WEL with heavily deformed structure underneath WEL, obtained from rolling band area of squatty region (sample D), with corresponding hardness measurement.

Figure 3.46 Optical micrograph image showing a very thin WEL with undeformed structure underneath WEL, obtained from field side region of squatty area (sample E), with corresponding hardness measurement.
(3). Rail No.3

Figure 3.47 displays a part of the running band surface of Rail No.3, showing visible WEL patches occurred randomly and associated mild squat formation. Regions A, B, and C obtained from WEL and squat were investigated. The microstructural observation was examined across the longitudinal and transverse cross-section in the direction illustrated by arrows.

![Figure 3.47](image)

Figure 3.47 Part of the surface of Rail No.3, showing samples A, B and C taken from WEL and squat for metallographic examination.

Figure 3.48 shows a typical optical micrograph of the transverse cross-section area of the region A. WEL was present on the entire surface of the rail, having a uniform thickness reached up to 50 μm. Within WEL, there is a visible tiny short crack. Beneath WEL, a typical microstructure of the rail carbon steel material, consisting of ferrite and cementite can be clearly recognized. There is also a transition zone, located between WEL and bulk material, which is barely visible in low magnification. Using SEM, the enlarged image shows the undeformed structure and the plastic deformation does not take place in the structure below the WEL. The crack initiated from WEL and transmitted vertically through the whole thickness but has not developed further into the bulk material. Some ferrite boundaries can be clearly visible, extending from bulk material through the white
etching layer. Spalling was also found on the upper layer of WEL. The corresponding microhardness distribution within the WEL through the rail surface is attached. In a depth of about 50 μm from the top surface, the highest hardness values were measured, varying between 700 HV to 770 HV, confirming WEL formation. In a depth exceed than 50 μm where the transition zone is located, the hardness values decreased sharply to nearly the level of bulk material, confirming the absence of work hardening layer that usually shows a slight increase in hardness values compared to the base material.

The microstructural examination results for the longitudinal cross-section of region B is shown in Figure 3.49. WEL here was seen to have been destroyed. This could be caused by a high impact load induced by the wheel on the rail. The breaking up of the WEL caused several cracks formation and spalling off to the top surface layer. The area at the large piece breaking off has been enlarged. It shows that cracks initiated in the location where the part of WEL has broken up away from the layer. The cracks then extended down into the rail bulk material. It appears that the cracks growth direction was associated with the structure of the transition zone underneath WEL. As there was no plastic flow underneath WEL, the cracks developed vertically in the rail material.

The microhardness measurement was performed through WEL down to base material. The results across the transverse section show similar hardness distribution behaviour of those measured through the longitudinal section (region A).

Figure 3.50 is the SEM micrograph, showing crack mouth in the longitudinal direction of the squat at region C. A patch of WEL with thickness about 25 μm was observed just at the crack mouth. Inside WEL, some regions having dark grey colour seem to have been oxidised. Underneath WEL, the microstructure feature was not clearly visible to identify. In addition, part of the material at both crack moth sides has also been transformed into oxidation. Regarding crack initiation direction, the crack first developed down to the material with direction almost vertically with a slight deviation toward the opposite direction of rolling (RD). At a few microns depth from the rolling surface, the crack then changed the growth direction to be corresponding with the rolling direction. It can also be seen that the back area of the crack mouth has been depressed down compared to the front area. This is due to the formation of subsurface cracks, resulting in surface depression.
The front and the back-surface area of the crack mouth have been characterised, Figure 3.50. As can be seen, patches of WEL were detected on the rolling surface area at both sides of crack's mouths. WEL was not present in homogenous thickness and some part of WEL has been fully oxidised, particularly those on the back-surface area. The microstructure feature underneath WEL looks generally undeformed, but a very slight deformation was locally observed on the part of the rolling surface layer.

In summary, WEL associated with the undeformed structure at the transition zone was found on both squat area and the adjacent WEL region. Cracks were seen to have initiated from WEL and propagated into the rail material. The cracks developed on the rail surface where WEL has broken up.
Figure 3.48 Characterization of the top surface of sample A in the transverse direction, showing a thick WEL associated with crack formation and undeformed pearlite structure underneath WEL, with corresponding hardness distribution.
Figure 3.49 Characterization of the top surface of sample B in the longitudinal direction, showing a fractured WEL associated with crack formation and propagation into bulk material of undeformed pearlite structure, with corresponding hardness distribution.
Figure 3.50 Characterization of the top surface of sample C in the longitudinal direction, showing the presence of fractured WEL at crack mouth and at both sides of crack mouth.
(4). Rail No.4

Samples A, B, and C on the field side region and samples D, E and F on the running band region, as shown in Figure 3.51, were cut for microstructural characterisation. Samples from running band were taken from surface damage area (squats) whereas field-side samples were obtained from the area where no visible cracks were present on the surface as it is located out of the rolling band range. All samples were observed through the longitudinal cross-section in the direction illustrated by arrows. Microhardness was performed through the cross-section area of all samples.

Figure 3.51 Rail No.4, showing samples A, B, C obtained from field side region containing WEL, and samples D, E and F taken from rolling band region containing squat, for metallographic examination.

At field-side region, Figure 3-52 shows the microstructural examination results of samples A, B, and C respectively. In sample A, a large WEL patch with thickness reached up to 40 μm was observed on the surface layer. It has a very fine homogenous structure compared with the bulk material. Some ferrite boundaries appeared upon WEL, extending up to the rail surface from bulk martial through WEL. Below WEL, the undeformed microstructure can be clearly observed, showing the absence of plastic deformation. In comparison, a very thin WEL, only with a thickness of fewer than 5 μm was observed on the surface layer of sample B. The underneath WEL structure does not show any apparent deformation but the pearlite lamella just blow WEL may have been subjected to compression by high impact dynamic load. So far, no WEL existed on the surface layer of sample C, and the top surface structure is characterised by very fine pearlite lamella,
compared with the other pearlite patches inside the bulk material, indicating that the local stress resulted in elongation of the pearlite lamella.

Results from micro-hardness measurement confirmed the presence of WEL on the surface of samples A and B with the absence of it on the surface of sample C.

For sample A, high hardness values, varying between 750HV and 690 HV, was measured across a 40 μm depth of WEL, and the hardness becomes nearly compatible to the hardness level of the bulk material from depth 40 μm down to the base material. No transition microhardness value was found in the transition zone.

For sample B, higher hardness value was measured only within a very little depth form the top surface. At depth between 5 μm to 10 μm, which is considered the transition zone area, the hardness values showed a slight increase in hardness values compared to those measured in the same region of the previous sample (sample A).

For sample C, the upper layer of the surface, characterised by the absence of WEL, was found to have a relatively high hardness value (about 430 HV) compared with the bulk material that was found just over a depth of 5 μm from the surface.

At the running band region, Figure 3.53 shows the surface characterisation with the corresponding microhardness results of the samples D, E and F. Each sample considers different surface damage levels. Sample D obtained from the surface that is no visible cracks and other material effects on the surface. Sample E and F consider mild and moderate squat. The results showed that there is no white etching layers formed on the rolling surface area. Instated, severe plastic deformation was observed directly under the surface, indicating that the surface layer has been rolled and shear deformation is along the longitudinal section of the rail, opposite to the rolling direction. It is clearly visible that the deformation of the surface is different on the samples. Based on the shear angle of pearlite bands from the worn surface, the angle of deformation was reached about 25° to the rolling surface of sample D, and became more severe on the surface of sample E, making up 12° angle with respect to the rolling contact. On the surface of sample F, the plastic deformation is very severe, and the structure is characterised by almost flattened grains. In addition, due to plastically deformed surface material, small surface cracks were initiated.
The corresponding miro-indentation hardness measurement shows varying distribution through depth. The maximum hardness, ranging between 450 HV to 400 HV was measured through a depth of about 15 μm, 20 μm and 25 μm across the deformed layer of samples D, E and F respectively. The hardness varies between 400 HV to 380 HV on the elongated pearlite grains area at depth from 20 μm, 25 μm and 30 μm to rail surface of samples D, E and F respectively. When the plastic deformation has ceased at depth over of 25 μm (sample D), 30 μm (sample E) and 40 (sample F), the hardness is reached the material level.
Figure 3.52 Characterization of the top surface of Rail No.4 along field side region showing the gradually removed WEL, (a) (b) and (c) longitudinal section of sample A, B and C respectively, with corresponding hardness distribution.
Figure 3.53 Characterization of the top surface of Rail No.4 along rolling band region showing different level of plastic deformation on the surface layer, (d) (e) and (f) longitudinal section of sample D, E and F respectively, with corresponding hardness distribution.
In summary, the white etching layer was found on the field side region but not found on the rolling band area. WEL was removed from some region on the field side due to squat formation. Plastic deformation was the predominant characterisation of the rolling band surface. The deformation level was associated with the damage level of the rolling surface. Microhardness measurements confirmed the microstructural features observed through the longitudinal section of the samples from the field and running band area.

3.6 Discussion

3.6.1 Structure changes on Rails

The structure on the surface of rails having different load history, geometrical location and working conditions evolved due to actions induced between the wheel and the rail. Rails from the area where a high braking effort reveals completely different crystal and phase structure than those removed from the section where the trains are steady speed through (less braking implementation). Nevertheless, squats and white etching layers were formed on the rail surface where a high distortion in the lattice structure was produced. Based on synchrotron X-ray radiation results, the phase structure along the rolling band area was not identical. The phase structure at squats and WELs was different than the bulk material. In contrast, the area that no squats and WEL has taken place shows a similar phase structure of the bulk material. This indicates that structure evolutions on the material along the rolling band area have locally occurred. Evidence was obtained from the distribution of the lattice parameter with respect to distance showing different degrees of distortion induced on the rolling surface layer. Results from peak profile analysis confirmed that the 200 peak profile has changed from asymmetry shape on the WEL and Squat to symmetry after a few distances from the defect location. Accordingly, large quantities of carbon dissolution were found in the WEL and squats region while the amount of carbon content is still constant in the same section of rail but different locations. Interpretation of this may be related to unstable contact condition between the wheel and the rail.

In addition, the different loading and different work conditions induced significant different phase structures on the surface of rails. In track region with steady speed traffic condition, Rail 1 and 2, the rail surface layer was found to have martensitic structure associated with the formation of white etching layers and squats. In track region with high
braking application, Rail 3 and 4, the rail surface layer was found to have austenite with martensite structure, detected at WELs and squats. Observation from microstructural examination confirmed evolution on the surface texture has also induced across the depth.

### 3.6.2 White etching layer formation.

Synchrotron XRD analysis of the rail surface, combined with microstructural and microhardness investigation, revealed that there are two distinguishable types of WEL; one containing fine martensite and the other comprising both of retained austenite and martensite. WELs containing ultra-fine martensite with some undissolved carbide particles was found associated with severe plastic deformation at the transition zone. In contrast, WELs containing a significant amount of austenite and martensite were associated with the undeformed pearlite structure at the interface layer. Each type of WEL was found to be correlated with the load and working conditions induced by the wheel on the rail. WEL comprising martensite with a significant amount of retained austenite was observed in the heavy breaking regions. It means that these Rails experienced a high level of wheel slip process caused by high rolling contact stress and high contact temperature WEL contained austenite suggests temperature-induced phase transformation process. In contrast, WEL contained nanocrystalline martensite was found in low braking utilisation regions where the traffic speed is steady.

Based on the working condition, along with the structural characterisation of WEL and the transition zone (underneath WEL), it strongly suggests that WELs characterised by fine martensite and undergone deformed structure at the interface layer is mechanically induced via severe plastic deformation. This type of WEL is termed as SD-WEL. In comparison, the white etching layer having martensite and austenite particles, alongside undeformed pearlite at the interface layer between the WEL and the base railhead is thermally or thermomechanically induced by a pure temperature or a combination with pressure changes. Thus, this type of WEL is termed as TP-WEL.

In addition to temperature arise caused by wheel slip, it also possibly be caused by the arcing process. Support for WEL type being caused by the arcing process-induced on the rail surface is that the existence of WEL on the field side region rather than rolling contact area on the top surface of Rail No.4.
To compare between the two types of WEL, SD-WEL type was only found on the rolling band while TP-WEL was found on the rolling band and field-side region. Peak profile analysis revealed that the 200 peak of both WELs type was characterised by an asymmetric shape, confirmed that a body center tetragonal (bct) is present in the lattice structure of WELs. The 200 peak in the profile of SD-WEL was found to relatively broadened and shifted toward the low diffraction angle, compared to TP-WEL, indicating grain size refinement and microstrain induced in the lattice structure due to high shear strain. Results from the quantitative analysis showed that the distortion of the lattice structure is highly induced in both types of WEL resulting in the formation of oversaturated Fe, C alloy. Determination of retained austenite showed a significant amount of retained austenite was found in the TP-WEL structure in comparison with 0% amount in the majority SD-WEL type. However, retained austenite with a volume fraction less than 1.5% is present in the only SD-WEL type that was obtained from the rolling band area of location 12 of Rail No.2. As the formation of TP-WEL was found in the same area but at the field side region, it is expected that the pre-existed SD-WEL was affected by temperature extended from the adjacent area.

The microstructural examination results confirmed that TP-WEL is always thicker than SD-WEL. Results from micro-hardness indentations showed that SD-WEL is relatively harder than TP-WEL. Increasing the hardness values is due to increasing the distortion of the lattice structure associated with the carbon content resulting from the dissolved cementite lamella. Distribution of hardness measurements across the depth of WELs showed that the maximum hardness values were measured in the middle section of WELs compared to the upper and lower layer, confirming the inhomogeneity structure of WELs. In addition, the observation confirmed cracks formation in both WELs types, which is due to the high hardness characteristic. Spalling associated with the formation of oxides particles was found on the upper layer of both WELs types, with a large proportion in SD-WEL type.

On the other hand, the surface morphologies of WELs from different locations were examined and compared revealing significant location-dependent differences, depending on the closeness of the running band regions to squat defects. The sample shown in Figure 3.54a was taken from a region of running band located some distance from any visible squat defect, whereas the sample of Figure 3.54b was obtained from a region very close
to a squat at the early and the moderate stages of development. In Figure 3.54a, the top view indicates that the surface layer is mottled, comprising bright regions of WEL and associated darker surface dimples, which could be associated with embedded oxide particles. In contrast, an examination of the top and plan views shown in figure 3.54b revealed that WEL’s in the vicinity of the squats typically comprised smoother, thinner lenticular strips, with few additional surface particles. Because the later type of WEL is much thinner and shows little evidence of accumulation of oxide particles, it can be inferred that WEL in the vicinity of squats is associated with relatively new surface damage.
Figure 3.54 Top view of rail samples, showing surface morphologies of WELs from different locations.
3.6.3 Correlation between squat initiation and white etching layer

Microstructural characteristics of the WELs regions confirmed the formation of vertical fine cracks in WELs. The characteristics of WEL in which the brittleness and high hardness nature as well as stress variations with the adjacent region (bulk material) are considered the cause of cracks formation under the cyclic wheel/rail contact loading. These cracks have a tendency to propagate into the rail material, as clearly shown in the enlarged SEM images. In addition, the cracks are observed which have initiated from the edge and middle WELs and propagated down into the rail material. Inspection of the cracks reveals that propagation into the rail material was associated with microstructure underneath WELs. Cracks developed down the rail surface along the shear deformation direction when the plastic deformation is present at the subsurface layer. The cracks developed vertically in the rail material passing undeformed pearlite structure located just below WELs. Such observed features of crack initiation and propagation strongly correlate with the formation of squat defects on the rail surface.

Microstructural and phase structure analysis also supports the conclusion that there is a correlation between squats formation and white etching layer. Synchrotron XRD results obtained from the region of the squat defect and WEL shows similar phase structure induced on both regions compared with the area on the same rail surface but does not contain squat and WEL. In addition, high distortion in the lattice structure corresponding with the dissolution of the high amount of carbon content was locally induced on the rolling surface containing WELs and squats. Microstructural examination obtained from squat defects shows the existence of fractured WELs just close to the crack mouth. An explanation of the absence of a continuous WEL along the surface layer of squat defect may simply be that this brittle surface is preferentially removed by a combination of wheel abrasion and battering over the surface of the squat. Evidence of this is obtained from the microstructural observation over a visible WEL that existed on the rail surface of Rail No.4. It can be clearly seen that the existence of WEL on the surface was associated with the size of the squat formation. WEL was partly removed when a mild squat takes place on the surface layer and completely removed at moderate squat formation.
3.6.4 Crack Initiation and growth.

The crack initiation and growth on the rail surface is highly dependent on the structure nature of the rail material. To prove this, the sample containing WEL and vertical cracks, initiated at different positions along WEL and propagated down into the rail material, was metallography examined. The type of WEL examined is associated with the subsurface plastic deformation. Cracks were observed to have initiated on the WEL at the front and back sharp edge in relation to the rolling direction and at the middle section of WEL. Figure 3.55a shows crack initiated at the front edge of WEL and propagated down into the rail, following the sheared grains direction. The crack depth is about 35 μm from the rail surface. Within the same layer, a crack was observed in the middle region of WEL, Figure 3.55b, crossing the interface region toward the direction of plastic deformation with depth about 30 μm. Figure 3.55c shows a tiny crack formed at the back edge of WEL and propagated in the direction opposite to rolling direction, crossing the sheared pearlite lamella. As these three cracks were located within the same layer of the WEL, they experienced the same loading cycles. However, each crack showed different crack growth rate and development. It indicates that the stress status at mentioned three points (front-middle-back) of the WEL was varied. In addition, the growth rate of cracks at these three points is also promoted by the sheared pearlite structure underneath the WEL. In the front edge, the crack propagates along the interfacial boundary between WEL and the rail matrix, and to further grow with the sheared grains. In the middle edge, the crack grows vertically before reaching the deformed pearlite below the WEL. In the back edge, the crack propagates in two tendencies, first following the interface boundary between the WEL and the matrix and then crossing through the bais material and gradually aligning with the deformation direction. This leads to the conclusion that fatigue life at the front edge of the WEL is lower than other positions such as middle and back edges(Lian et al. 2019).
Figure 3.55 SEM images showing crack formation at a different locations within WEL.
3.7 Conclusions

The structural and microstructural evolution of the top surface layer of rails were investigated to determine whether there is a correlation between squat initiation and the occurrence of the white etching layer on the top surface. Based on load history, geometric location and working conditions, two different groups of rails containing surface defects including squats were used for investigation. Based on the results obtained, the following conclusions are drawn.

1. White etching layers play a significant role in crack initiation on the rail surface. Support for cracks being formed by the white etching layer is confirmed by microstructure and phase analysis. Synchrotron XRD spectrum measured at squat defects themselves is consistent with those obtained from the adjacent WEL region. Occurrences of cracks extending from WELs down into the rail in regions near the squat was confirmed by microstructural observations, consistent with the premise that cracks in squats originate from an extension of cracks in WELs.

2. Cracks were observed to have formed at different locations within WEL, for example, initiated at the front and back sharp edge of WEL in relation to the rolling direction, and at the middle section of WEL. Crack propagation on the subsurface layer of rail follows the shear direction when the plastic deformation is present.

3. Synchrotron XRD analysis revealed two distinguished structures of WEL; one associated with the formation of fine martensite and the other with the presence of a mixture of retained austenite and martensite. The metallographic analysis showed that the microstructure of the transition zone beneath the WEL comprised either deformed or undeformed pearlite.

4. Load and working condition is associated with the type of white etching layer formation on the rail surface. WELs containing martensite were induced by severe plastic deformation at the wheel-rail interface, which was associated with the deformed subsurface pearlite structure. WELs containing both of retained
austenite and martensite were thermally or thermomechanically induced, which were correlated with the undeformed pearlite underlying structure.

5. White etching layers were not only found on the rolling band surface, but they can be formed on the field side of the rail. The white etching layer existed on the rolling band region is not similar to those on the field side region, in which they exhibit different structure nature.

6. Retained austenite was only detected on the WELs that is associated with the undeformed structure of the transition zone.

7. The microhardness measured on the WEL is significantly high compared with that of the transition zone and base material, which confirms the brittle nature of this type of WEL. The microhardness values at the topmost surface layer of squats themselves are considerably low than that measured on the WEL in regions near the squat. An explanation is that the brittle surface layer can be preferentially removed by a combination of wheel abrasion and battering over the deformed surface of the squat.

8. Different degree of distortion of the lattice structure was induced on the rail surface material due to unstable contact condition between wheel and rail. High distortion in the lattice structure was induced for both squats and WELs on the rail surface, showing a high carbon concentration amount. Increasing the hardness values at squats and WELs regions is due to the distortion of the lattice structure associated with the carbon content resulting from the dissolved cementite lamella.
4. Advanced characterization of two distinct types of white etching layer formed on the rail surface

4.1 Introduction

Based on the literature, the structural nature of WELs, formation mechanisms and a relationship with squat formation are not fully understood. WELs are believed to be created by either thermal martensitic phase transformation and/or severe plastic deformation on rail surface and squats can develop below them. Controversy continues partly because depending on the particular studies performed, the microstructures of WELs have been both identified and described differently as; martensite, nanocrystalline Fe-C alloy; deformed pearlite, nano-crystalline martensite, austenite and cementite; and martensite and retained austenite.

In the previous chapter, the correlation between WELs and squat initiation has been confirmed. WELs associated with distinct rail track regions were examined and found to be different, depending on whether the track was subjected to damage induced by severe train braking (wheel slip) or by repeated deformation via rolling contact fatigue. In order to obtain a fundamental understanding of the nature of two distinct types of white etching layers, advanced material characterisation technologies were introduced for the particularly interesting sites of two distinct types of white etching layers in the current chapter. The examination focuses on the detailed microstructure nature of these WELs, the composition comparison between them, morphology of the interface regions between the WELs and pearlite, and the correlation with operating conditions. The transformation mechanism of two types of white etching layer was discussed. In heavy braking rail regions, WEL containing martensite and retained austenite was observed, alongside undeformed pearlite at the interface layer between the WEL and the base railhead. This type of WEL, referred as TP-WEL, is found to be induced by temperature and pressure changes. In contrast, in low braking utilisation regions where the traffic speed is steady, WEL caused by rolling contact fatigue deformation contained nanocrystalline martensite and a region of severe plastic deformation immediately below the WEL. This type of WEL is termed as SD-WEL.
4.2 Methodology

Samples containing WELs obtained from the top surface layer of tracks with different loads and traffic conditions (Load history and geometric location for rail samples can be found in Table 3.3 in Chapter 3), were prepared by the metallographic procedure and then observed by optical and scanning electron microscopy. Figure 4.1 a, b shows the longitudinal section of two different types of WEL, including TP-WEL and SD-WEL. Thin WELs lamellas were prepared by FIB using FEI Helios NanoLab G3 CX instrument and then characterised by TEM using JEOL JEM-ARM200F with accelerating voltage of 200 kV, working with SAED, conventional brightfield (BF) and dark field (DF), as well as HRTEM. In addition, elemental analysis was performed using Energy Dispersive X-ray Spectroscopy (EDS) in the Transmission and Scanning Transmission electron microscopy mode (TEM) and (STEM).

In order to investigate the microstructure of the white etching layer in different positions, lamella samples were taken from both vertical and horizontal planes of WELs. The lamellas obtained from the vertical plane were considered to examine the microstructure evolution through WEL depth including the underneath structure with respect to rolling direction (RD). For the horizontal plane, the lamellas were considered to investigate the microstructure characterization of the WEL in the transverse direction (TD) with respect to the rolling direction. The horizontal lamellas were taken from the middle section of WELs. Figure 4.2 shows cutting sections from vertical and horizontal planes of WELs before being lifted out and welded on the holder and final thinning to electron transparency.

Figure 4.1 Two different types of WELs, (a) TP-WEL (b) SD-WEL.
4.3 Results

4.3.1 Vertical Plane Lamellas: TEM and STEM observation

Figure 4.3 a, b and c shows the whole thin lamella sections of L1, L2, and L3, respectively, after welding, thinning and cleaning processes. It is clearly visible that the structure of lamella (L1 at TP-WELs) contained distinguishable coarser crystalline grains than the structure of lamella (L3 at SD-WELs). At the transition zone, the microstructure reveals a significant variation, in which the state of pearlite structure underneath WEL was found to be either heavily deformed (lamella L3 at SD-WELs) or undeformed...
(lamella L1 at TP-WELs). The phenomena, including the variation in the crystallite grain size at two types of WELs and different microstructural characterisation of the transition zone, strongly confirm that there are two distinguishable types of WELs induced by different formation mechanisms excited on the rail surface.

Figure 4.3 a, b and c shows the whole thin lamella section L1, L2 and L3, respectively.
To obtain further detailed microstructural evolution with respect to depth, the characterisation techniques including BF, DF, SAED, HRTEM and EDS provided by TEM were introduced to examine the regions from the topmost, middle and lower layer of WELs itself, and from the transition zone between WEL and matrix rail material. A comparison then was made between the two types of WEL at the different depth positions.

- **The topmost position of White Etching Layers.**

  The topmost region of both types of WEL within a depth of 4 μm from the surface was examined. TEM micrographs in Figure 4.4 a-d characterise the microstructure in the topmost position (close to the rail surface) of WELs of lamella samples (topmost position of L1 from TP-WEL and L3 from SD-WEL, respectively). Significant variations regarding the morphology structure and grain size refinement can be clearly recognised between them. The WEL grain as shown in Figure 4.4a was found with size in the range from 30 nm to 200 nm, compared with the fine grain size, ranged between 5-20 nm in Figure 4.4c. The topmost region of TP-WEL is composed of martensite and retained austenite as shown in Figure 4.4a. Plate-like crystals of martensite (PM) with thickness in the range of about 10-30 nm can be recognized at the middle and bottom of the image. Particles of retained austenite (RA) were found in parallel to the plate martensite, showing a correlation of the orientation relationship between martensite and parent retained austenite. The SAED pattern in Figure 4.4b, taken over plate martensite and adjacent region (indicated by a red circle in the TEM image in Figure 4.4a) shows martensite orientation with [001] bcc zone axis parallel to the [111] fcc zone axis of austenite. In addition, the diffraction spots of twins were identified in conjugate with the bcc matrix, as indicated by the circles. Splitting was found in the diffraction pattern, indicating two overlapping peaks of the (002) α’ and the (200) α due to transformation from ferrite to ferrite/martensite phase structure. Barely visible spots were found inner than (110) ring, which is unknown whether they are related to the diffraction pattern of cementite or iron oxide.

  The topmost layer of SD-WEL as shown in Figure 4.4c consists of a very fine polycrystalline structure, containing ultra-fine plate-like martensite with thickness in the range about 5-10 nm, and very few remanet thinner ferrite crystals as well as very fine
fragmented carbide particles. High dislocation density was observed in the microstructure. The SAED in Figure 4.4d shows concentric diffraction ring pattern, indicating the occurrences of high-angle of misorientation between grains. The diffraction spots of cementite, (200) Fe₃C and (211) Fe₃C were present inside and outside the diffraction ring of (110). Austenite with fcc structure was not found in the diffraction pattern. Splitting was also observed into (002) α’ and (200) α in the diffraction pattern, confirming the tetragonal distortion of the lattice structure induced at the topmost region of WELs. Diffraction spots expected to be related to iron oxide (Fe₂O₃) were found inner than the (110) ring.

Figure 4.4 Microstructure feature on the top surface layer of WEL taken from (a) Lamella L1 (from TP-WEL) and (b) corresponding diffraction pattern, (c) Lamella L3 (from SD-WEL) and its corresponding diffraction pattern in (d).
Further detailed substructures inside the plate martensite were examined. Figure 4.5 shows the BF and DF TEM images with the corresponding SAED pattern of the selected area (indicated by a red box in Figure 4.4c) taken over the plate martensite. The results showed that twins were found inside the plate martensite, induced by severe plastic deformation. From SAED, cementite particles with (101) Fe₃C and (202) Fe₃C were found in the diffraction pattern, confirming that nano-fragmentations of carbide embedded within plate martensite.

![TEM images showing twins martensite structure](image)

Figure 4.5 TEM brightfield (BF) and dark field (DF) with corresponding SAED image of the selected area indicated in the Figure 4-5 c, showing twins martensite structure with very localised fine cementite particles.

- The middle position of White Etching Layer

TEM and STEM micrograph with different magnifications, corresponding SAED and HRTEM in Figure 4.6 a-f shows typical examples of the composition and crystal structure of the middle region of the lamella sample from 5 to 12 μm down from the top surface (middle position at L1 from TP-WELs). From the BF and DF STEM images in the Figure 4.6a and b, it can be clearly observed that there are regions of different composition and microstructural morphologies associated with local grain size variations, resulting in inhomogeneous microstructure characterisation. The microstructure is composed of plate martensite, retained austenite, remanent ferrite, and some carbide precipitations. Plates like martensite morphologies were found with different thickness and distributed in
different ordination direction. The thicker plate martensite is considered to be a combination of many thin plates having the same variant. Thin elongated retained austenite grains were observed at the boundary of a large grain of martensite, having an ordination relationship with the adjacent thin plate martensite. Fine carbide precipitations particles were found elongated in parallel with the plate martensite. Some remanent ferrite particles were also recognised inside the microstructure.

It contained a mixture of both fine and coarse grain-sized microstructure. Large individual grain martensite with crystallite size in microns can be recognised inside the microstructure. This grain consisted of very fine plate martensite, characterised by a relatively dark grey colour in BF image and bright contrast in the DF image compared with the surrounding microstructure. The distinguished appearance of the martensite phase structure in this grain is due to the carbide precipitation which is possibly related to reheating or tempering process at the rail surface. High magnified BF STEM image in Figure 4.6c shows a detailed microstructure of the selected area taken from the martensite grain structure, confirming the composition of a very fine plate-like morphology. The dark films accommodated at plate boundaries (appeared in bright contrast strips) are related to cementite particles due to the decomposition of retained austenite. Figure 4.6 e shows the corresponding SAED pattern, which confirms the presence of internal nano-sized transformation twins as substructure features of plates. The diffraction pattern of the [011] zone axis of martensite, in which the white and yellow solid lines indicate the diffraction spots from martensite matrix (α’) and the associated spots from twins (t), respectively, shows parallel orientation relationship. Extra diffraction spots, as referred by arrows, were found at 1/3 (211) and 2/3 (211) to be added in the diffraction pattern of matrix and twins. The diffraction spots of cementite were totally overlapped.

TEM micrograph in Figure 4.6e characterises the microstructure of WEL, taken from another part in the middle region of TP-WEL. The microstructure here shows different morphology structures than those observed in the previous area, considered by a mixture of lath martensite and retained austenite. Lath-shaped crystals martensite with an average width of about 200 nm was found having an almost parallel arrangement. The presence of lath martensite features together with the plate-like martensite microstructures in the WEL could be related to the inhomogeneous distribution of carbon content in the material. In addition, high dislocation density in the lath martensite can be clearly
recognised considering the substructure of the lath. Retained austenite films were found to be accommodated between lath martensites. The presence of retained austenite films at lath boundary was confirmed by HRTEM observation in Figure 4.6e. The corresponding FFT diffractogram of selected area in Figure 4.6f, revealed fcc crystal structure with the streaking in the [111] orientation.
Figure 4.6 Microstructural characterisations of the middle section of WEL sample L1, (a), (b), BF and DF STEM image (c) and (d) High magnified BF STEM image of the selected area and corresponding SAED pattern showing plate’s martensite with twined structure. (e) and (f) TEM with HRTEM image characterised different part on the middle region of WEL, showing lath like martensite with the presence of retained austenite at the lath boundaries.
For the middle region of lamella sample, L3 from SD-WELs, the BF and DF in STEM image, given in Figure 4.7a and b, reveals ultra-fine microstructure containing a mixture of plate-like martensite with different thickness and random orientation, fractured ferrite layers as well as heavily fragmented nano-sized cementite particles that appeared bright in DF image due to higher atomic number. Areas with relatively high mass thickness contrast in the DF image could be related to scattering electron of the grains having different crystal orientations.

As shown in Figure 4.7c, BF in the STEM image of the selected area with high magnification reveals heavily deformed microstructure, characterised by severely residue ferrite layers associated with a high density of dislocation. The corresponding SAED in Figure 4.7d shows that microstructure is composed of polycrystalline ferrite and cementite rings with a non-uniform intensity of the ring pattern. Splitting spots were observed in the diffraction pattern which indicates the tetragonal structure martensite. Spots associated to the austenite reflection were not observed in the diffraction pattern.

Figure 4.7 e shows another example of the microstructure of WEL in the middle region of SD-WELs. The residue of the lamella structure was found in the microstructure and associated with high dislocation density. HRTEM micrograph in Figure 4.7f reveals the lattice image of the selected area, showing structural defects in the lattice, as marked by a red ellipse. The defects considered as the dislocations and both low and high angle grain boundaries. A detailed view at dislocation indicated at the top right corner of the image. Due to high dislocation, large strain fields were induced in the lattice structure. The corresponding FFT shown at the bottom of the image reveals the diffraction pattern of ferrite (bcc) crystal structure in the [111] orientation.

The microstructure at the middle regions between L1 from TP WELs and L3 from SD-WELs shows significant variations. TP-WEL (L1) is comprised predominantly of plate martensite, few lath martensites, some retained austenite and few cementite precipitations particles emended between plate boundaries. Whereas the microstructure of SD-WELs (L3) was characterised by ultra-fine martensite, residue deformed ferrite and fragmented undissolved cementite particles. TP-WEL (L1) was composed of a mixture of coarse and fine grains while SD-WELs (L3) in comparison contained predominately ultra-fine grains, forming a more homogenous microstructure.
Figure 4.7 microstructural characterisations of the middle section of WEL lamella sample L3, (a), (b), BF and DF STEM image (c) and (d) High magnified BF STEM image of the selected area with the corresponding SAED pattern showing nanocrystalline structure with the presence of fine cementite particles. (e) and (f) TEM with HRTEM image showing a high concentration of defects induced in the lattice structure due to high deformation.
The lower position of white Etching Layer

The lower position of the WELs is the region near to the transition zone, but still in WELs. For the thick TP-WEL, the lamella sample (L2) will be used for characterisation. The BF and DF STEM micrograph in Figure 4.8a, b characterise the microstructure of TP-WEL about 40-50 μm depth from the top surface of the rail. The microstructure at the lower region of TP-WEL appears much finer than that at the top and middle regions of TP-WEL. Figure 4.8 shows a martensite structure, mostly composed of typical morphology of plates without any particular orientation. Plate martensite can be distinguished by the twined structure that reveals a different contrast. Some twins are not clearly visible, and some reveal bright contrast, as seen in the DF image. Variation in the contrast is due to different crystal orientation direction of the twined martensite structure. In addition, an individual thin film of cementite lamella (C) was found in the microstructure, confirming that the cementite lamella was not fully transformed.

To obtain a more detailed elemental composition, EDS mapping was used for elemental analysis. Figure 4.8 c and d shows the distribution mapping of carbon (C-k) and iron (Fe-K) for the selected area. The other chemical elements such as silicon (Si) and manganese (Mn) were ignored due to low concentration. The distribution of carbon in this specified area containing nanostructure plate martensite does not show any particular orientation, indicating that the amount of carbon content decreased sharply, certainly due to the dissolution of cementite. A segregated high carbon content layer can be clearly observed in the C-K mapping, confirming the presence of thin cementite film. From Fe-K mapping, the distribution of the component of iron was not uniform, showing some local concentration of iron characterised by high contrast appearance. The area of high iron concentration indicated by arrow was found to have an orientation relationship with the thin cementite film shown in the carbon map. This configuration consists of the typical layer of pearlite lamella.

The BF and DF TEM image in Figure 4.8 e-f characterise another part in the lower region of TP-WEL. The microstructure here is also comprised predominately by a mixture of fine and coarse plate martensite with random orientation. The twined structure can be recognised as a substructure of plates. Undeformed ferrite matrix was found in the microstructure, lined up in parallel direction with grain of plate martensite. Dislocation defect indicated by dark spots can be observed in the ferrite layer. An individual spherical-
shaped of unresolved cementite particle was found inside the ferrite matrix, indicated by an arrow in the DF image. In addition, there are a very few precipitations of cementite particles, characterised by bright contrast spots in the DF image, embedded between plate martensite boundaries. Retained austenite particles were not really recognised between martensite grains.

The lower region of SD-WEL (from L3 sample) was 6-8 μm from the top surface as shown in Figure 4.9a-f. The microstructure in the lower region seems relatively coarser than the middle region of SD-WELs due to containing many large scales of fractured pearlite lamella. Regions with locally bright contrast shown in the BF STEM image (Figure 4.8a) indicate lots of broken and deformed ferrite layers. Fragmentations of cementite lamella with relatively dark contrast appearance can be seen within the ferrite layer. The matrix region is mostly composed of ultra-fine plate martensite structure. The dark area shown in the DF STEM image is a deformed ferrite particle under a certain direction of the incident diffraction beam.

The high magnification image of the selected area is shown in Figure 4.9 c-d. Fractured cementite particles with a thickness of about 20 nm were found inside the ferrite matrix, Figure 4-9c. The dark contrast appearance of the cementite was not uniform due to the partial decomposition of cementite in the ferrite matrix. In Figure 4.9d, cementite particles can be seen in different forms. Broken cementite lamella within the ferrite layer can be clearly observed. Fragmented cementite particles with spherical and rod-shaped crystals were also found in the adjacent region. The cementite particles on the region at the bottom of the image were not clearly recognised, indicating that the cementite particles were partially or fully dissolved due to severe plastic deformation. The heavy deformation also induced dislocation defect in the ferrite lamella, as indicated by dark spots.

EDS mapping was used for the chemical characterisation of the selected area indicated by the dashed line in Figure 4.9 d. From C-K mapping in Figure 4.9e, the distribution of carbon content on this area was not uniform, showing a much-localised concentration of carbon with random orientations. The localised high carbon content corresponds with the cementite particles. The random distribution of carbon content determined by EDS mapping strongly confirms with the observation that the cementite layers have heavily fractured into differently shaped particles with completely random crystallographic
orientation, due to the high shear stress of rolling contact. From Fe-K mapping in Figure 4.9f, the distribution of iron content was relatively homogenous.

Both TP-WELs and SD-WELs at lower regions contained the plate martensite and some remnant ferrite lamella. A slight variation was found associated with the presence of cementite particles inside the microstructure of each WEL type. The cementite lamella in the TP-WEL (L2) was almost completely dissolved, while the fragmented cementite particles with different size refinement and crystal-shaped morphology were found inside the microstructure of the SD-WELs (L3). This observation was consistent with the distribution of carbon content characterised by EDS mapping. Iron mapping showed that the distribution of iron content was relatively less homogenous on the TP-WEL type due to the existence of some localised area of high iron content. Retained austenite particles were not much recognized on the microstructure at the lower region of TP WEL (L2) and were not present on the SD-WEL (L3).
Figure 4.8 Microstructural characterisations of the lowest section of WEL lamella sample L1. (a), (b), BF and DF STEM image (c) and (d) EDS mapping of the selected showing the distribution of carbon and iron elements on the microstructure. (e) and (f) BF and DF TEM image showing another example in the lowest region of the WEL.
Figure 4.9 Microstructural characterisations of the lowest section of WEL lamella sample L3, (a), (b), BF and DF STEM image (c) and (d) High magnified BF STEM image showing cementite particles associated with different shaped crystals. (e) and (f) EDS mapping of the selected showing the distribution of carbon and iron elements on the microstructure.
The transition zone at the interface boundary of the lamella sample (L2 of TP-WEL) and (L3 of SD-WEL) exhibits a completely different microstructure as shown in Figures 4.3b and c. The pearlite structure close to the interface boundary of L2 from TP-WEL does not exhibit any plastic deformation showing a sharpened transition region; while the heavily deformed pearlite structure (severe plastic deformation) can be clearly observed at the transition zone of L3 from SD-WEL. This phenomenon confirms that two types of WELs are associated with different formation mechanics.

Further detailed microstructure feature at the transition zone is shown in Figure 4.10 a and b, for the TP-WEL (L2) and SD-WEL (L3) respectively. For TP-WEL type, the pearlite layers uniformly oriented at the transition zone is partially disappeared at the interface layer and completely decomposed into different phase structure at the WEL. The structure of WEL has been identified as the martensitic structure. As the plastic deformation was not found in the transition zone, the transformation from pearlite to martensite structure was induced by the thermal process. The selected area electron diffraction SAED pattern was taken over the undeformed pearlite and the adjacent WEL region (indicating by a red circle in the TEM image). From SAED, the orientation relationship between ferrite and cementite was identified in the diffraction pattern at the undeformed region, showing a relationship between the [-111] zone axis of the bcc crystal of ferrite to the [100] zone axis of cementite (Fe₃C). At the adjacent region (the lower region of TP-WEL), the diffraction shows polycrystalline rings. From SAED, the diffraction spots of (100) Fe₃C was found in the diffraction pattern inner than the (110) α′ ring, while no cementite lamella was observed. Further, the SAED shows that the diffraction peak of ferrite {121} α is split into two rings very close to each other. Peak splitting indicates the tetragonal structure, which means the formation of α′-phase martensite. The transformed of α′-phase martensite can also be confirmed by the presence of twin spots in the diffraction pattern as indicated by solid circles.

On the other hand, the interface region of the SD-WEL samples (L3), Figure 4.10b, reveals a different microstructural characterization than that of the TP-WEL sample (L2) in Figure 4.10a. High fragmentations of cementite plates, with size in the range from 10-30 nm can be seen at the deformed region near the interface boundary (underneath WEL). The cementite particles are randomly oriented, and no lamella structure was observed in
this area. The SAED taken over the deformed region shows spotty rings from the ferrite phase with the [-111] zone axis and some single spots and blurred rings from cementite were revealed. Moving up the deformed region, there is a nanocrystalline layer, characterized by ultra-fine structure, due to grain refinement, fractured and dissolved of cementite particles. Only some large individual cementite and ferrite particles can be observed. From SAED image in this area, the full diffraction rings were identified with nonuniform brightness distribution of diffraction rings, due to random misorientation nanostructure of cementite grains which was found overlapping the (011) diffraction ring of martensite.

Figure 4-11 displays the DF STEM image and its corresponding Carbon (C-K), Oxygen (O-K), Silicon (Si-K), Manganese (Mn-k) and Iron (Fe-K) mapping for the TP-WEL at transition region of lamella sample (L2). From C-K mapping, significant changings in the microstructural arrangement of cementite lamellar from the matrix pearlite towered the nanocrystalline layer over the interface border can be clearly recognized due to the high carbon content of cementite. This confirmed that the originally lamellar shaped of cementite with uniform bright contrast lines at the undeformed region gradually dissolved at the interface boundary and almost disappeared at the WEL region resulting in the formation of a nanocrystalline supersaturated Fe-C solid solution. From (O-K) mapping, a very little oxide content was seen with no much variation between the matrix and the WEL. The distribution of Silicon and Manganese components over the transition zone was almost uniform with a very slight concentration of Si found at the interface border. From Iron mapping, the distribution of Fe-K was slightly varied, particularly the area near the interface boundary. The dark region appeared along the right half of the mapping area was related to the thickness variation of the lamella due to partly diffusion of the protective Pt layer during the final thinning process.

The component distribution of SD-WEL, lamella sample (L3) at the transition zone were shown in Figure 4.12. From C-K mapping, there was no clear uniform arrangement of the cementite layer at the matrix region; instead, there were heavily fractured films of cementite observed in the area. The cementite particles were much finer in the adjacent nanocrystalline region (WEL), due to partly or completely decomposition. The distribution of oxygen, silicon, manganese, and iron were relatively homogenous and no distinguished variation in the content between the deformed matrix and WEL was
detected. The only very localised area having much Mn concentration was found in the WEL region near the interface boundary.

In comparison, the configuration of cementite lamella at the transition zone for TP-WEL and SD-WEL was significantly varied. The cementite layer from the undeformed matrix to TP-WEL was suddenly disappeared while the cementite lamella from the deformed matrix to the SD-WEL was gradually fractured and then dissolved. This indicates that each WEL was driven by different formation mechanics. WEL associated with the undeformed matrix (TP-WEL) was thermally or thermomechanically induced. Whereas, WEL associated with the deformed matrix (SD-WEL) was more likely to be mechanically induced.
Figure 4.10 characterisation of the transition zone for (a) Lamella sample L2 (b) lamella sample L3.
Figure 4.11 EDS mapping showing elemental distribution at the interface of lamella L2.

Figure 4.12 EDS mapping showing elemental distribution at the interface of lamella L3.
• **The matrix region just below the transition zone**

A high magnification bright and dark field STEM image in Figure 4.13 a and b shows pearlite structure characterization at WELs just below the transition zone between WELs and matrix. As shown in the bright field image, a typical pearlite structure with an average interlamellar spacing about 75-150 nm depending on the grain orientation was clearly recognized underneath TP-WEL (lamella sample L2). It confirms the absence of plastic deformation underneath TP-WEL. A very few individual dislocation densities within the ferrite lamella was observed with very few fragmentations of cementite. In the dark field image, the cementite layer appeared in bright strips shows a uniform orientation relationship with the ferrite matrix, indicating the absence of any work hardening process and confirming a single crystal structure.

In constant, the pearlite structure of the SD-WEL type (L3), as shown in the bright field image (Figure 4.13 b) is not discernible, characterised by flattened and almost elongated parallel to the rolling direction, due to shear stress caused by rolling and sliding contact loading. The interlinear spacing between ferrite and cementite lamellae becomes much smaller than those observed in the TP-WEL type (L2), with average spacing about 25-50 nm. The dislocation density is much higher than that in the undeformed perlite. The cementite layer as shown in the dark field image appears much thinner and randomly oriented due to the heavy fragmentation of cementite strips into nano-sized particle grains.
Figure 4.13 DF and BF STEM images characterised the subsurface layer underneath WEL, (a) and (b) Lamella sample L2 showing undeformed pearlite structure, (B) Lamella sample L3 showing heavily deformed pearlite structure.

4.3.2 Horizontal Plane Lamellas: TEM and STEM observation

Figure 4.14 a, b shows the horizontal lamella samples L4 (TP-WEL) and L5 (SD-WEL) after final thinning and cleaning processes. Each sample is about 8 μm in length, 8 μm in width and about 5-10 nm in thickness. The viewed plane considers the transverse direction with respect to the rolling direction.

The low magnification STEM image in Figure 4.14c and d reveal the microstructure of whole lamella samples of L4 and L5, respectively. Although both WELs have very small
grain-sized martensite structure, a slight variation in grain size refinement can be recognised. The microstructural features in the L5 of SD-WEL appear a slightly finer but both of the samples were composed of a mixture of fine and coarse grains. In addition, the crystallographic direction in the L5 was mostly elongated to the rolling direction, compared with the random orientation of the crystals in the L4 of TP-WEL. As seen in the Figure 4.14 e and f, taken from the middle area of WELs, a collocation of martensite features can be clearly recognised, confirming the martensitic structure nature of both of TP- and SD-WELs.

Figure 4.14 Horizontal-plane WEL lamella taken from the middle section of the two different kinds of WELs samples. (a) Lamella sample L4 taken from the undeformed WEL sample (b) Lamella sample L5 taken from the deformed WEL sample. (c) (d)
microstructure characterisation of the whole lamella samples, (e) and (f) microstructure of the selected area showing the characterisation of the middle region.

- **TP-WEL type**

BF and DF TEM images in Figure 4.15a and b show the detailed microstructure characterisation of TP-WEL, taken from the middle region of the sample. It is a typical martensite transformation steel structure, characterised by a composition of fine and coarse features with random crystal origination. Lens-shaped martensite structure was formed with a thickness ranged between 50-100 nm. Needle shaped crystals of plate martensite with a thickness between 2-5 nm, was also found. The twined structure was found as a substructure of plate martensite. The retained austenite particles and precipitations crystals of cementite were not observed in the diffraction pattern from the SAED result. This indicates that the cementite lamella in this area has fully dissolved and the phase structure has fully transformed into a martensite phase structure. In addition, the formation of a fine grain structure with different sizes and refinements could be associated with various cooling rates during the quenching process.

Another part in the WEL lamella sample was characterised by BF and DF with corresponding high-resolution STEM (HRSTEM) imaging, shown in Figure 4.15 c-f. The microstructure here also reveals fine plate martensite morphologies with twins as a substructure of the plate. Part of the area appears to have a structureless feature, expected to be supersaturated Fe-C with the ultra-fine size of crystals. DF image in Figure 4.15d shows the non-uniform distribution of bright contrast regions between the plates. The bright contrast here is less likely to be correlated to the crystallographic orientation. To obtain more detailed microstructural features, HRSTEM imaging was taken at the twin boundary, indicated by a small box in the BF image. From HRTEM, the fringes of the matrix lattice (α’) are symmetrical with the lattice fingers of the twins. The twin appears to have a sharp boundary with the matrix, formed with an angle of about 60°. The corresponding FFT image in Figure 4.15 f shows the lattice orientation relationship between the twins (indicated by a dashed line) and matrix (indicated by a solid line) about the crystallographic axis of [011]. From FFT, it was confirmed that the most common (112) twinning system type was formed in the martensite. Extra spots were found at the position of 1/3 and 2/3 {112} (as indicated by circles in the diffraction pattern), which was thought to be related to double diffraction of the twins. In addition, the spot
diffraction of cementite was not clearly visible in the diffraction pattern, indicating that they were either fully overlapped with the spot diffraction of the matrix or fully dissolved. While the diffraction spots of the (001) iron oxide (Fe₂O₃) (indicated by arrow) was found in the diffraction pattern. Further, the high magnified inverse FFT image (shown at the right up corner of the HRTEM image) shows the interface lattice structure, between the matrix and twins. It can be seen that the lattice near to the interface seems to have not fully straight fingers which indicates that the crystal distortions were induced in the lattice structure. From HRTEM and FFT, the retained austenite was detected at the interface boundary.

- **SD-WEL type**

BF and DF TEM with corresponding SAED images in Figure 4.16a and b reveals the microstructural feature of the SD-WEL at the horizontal plane. From the BF image, the martensite with the deformation twines was elongated toward the transverse direction of the rail. From the DF image, the twin boundary planes seem flat at the atomic scale. The SAED image shows diffraction spots of the matrix and twined structure as indicated by an arrow, having an orientation relationship about the zone axis [012]. Spots related to cementite and even retained austenite was not found in the diffraction pattern, indicating fully decomposition of cementite lamella in the microstructure of this specified region. The absence of cementite particles and the existence of the deformation twins in the microstructure confirm the martensitic structure nature of the deformed WEL type. However, it was hard to identify whether the transformation structure into martensite was caused by only severe plastic deformation or with a combination of the thermal process.

Figure 4.16 c and d reveals the DF and BF HRSTEM images taken from the middle section of the WEL sample. Variations in the microstructure feature can be clearly recognised. The microstructure in this region was comprised of remanent pearlite lamella, the region of high dislocation density as well as very fine plate martensite. Part of a typical pearlite lamella structure was found in the microstructure, characterised by shear band existence and high defect of density. Fractured cementite layer can be also observed which is possibly due to shear and plastic deformation induced on the rail surface. It was also noticed that the region with high dislocation density, appeared brighter in the DF image, was found at the boundary of the broken pearlite lamella. This indicates that the original ferrite and cementite grain was fully transformed into polycrystalline ferrite.
However, due to the heavily deformed region, it was hard to identify the dark field image whether the cementite layer has fully decomposed or there were still some very fine fractured cementite particles. To obtain a detailed microstructure feature, HRSTEM was used to characterise the region of a high density of defects. Figure 4.16 e shows the fringes lattice structure of the selected area indicated by the box in the DF image. The HRSTEM confirmed the high dislocation density induced in the lattice structure. The high magnified inverse FFT image of the elected area shows that large microstrain was induced in the fringes lattice causing high distortion in the lattice structure. The corresponding FFT image in Figure 4-16 f reveals the diffraction pattern along the bcc [0-11] crystallographic structure. It can be seen that the diffraction spot of the (011) plane shows high contrast appearance which indicates overlapping with the cementite diffraction spots. Despite the overlapped [011] ring of cementite, the diffraction spots of the (001) and (002) plans of cementite (indicated by arrow) are clearly visible in the diffraction pattern. The presence of cementite diffraction spots in the diffraction pattern indicates that the cementite layer has not fully decomposed but fractured to nanocrystalline sized particles. In addition, retained austenite reflection spots were not seen in the diffraction pattern.

For comparison, two types of WELs at a horizontal plane were predominately composed of plate martensite associated with twinned structure as a substructure of plates. A slight variation in crystallographic shape was recognised with the plate martensite morphologies. For TP-WEL the lens-shaped crystals of plate martensite were found in the microstructure, showing curved characterisation of the twins boundaries in the dark field image. For SD- WEL, the fine, and needle-shaped plates martensite were found in the composition structure. The existence of cementite particles was confirmed on the microstructure of the SD- WEL type but was not confirmed on the microstructure of TP-WEL due to the presence of double diffraction of twins. Cementite in the undeformed WEL was either overlapped twining boundaries or fully dissolved. In comparison, cementite particles found in the microstructure of the deformed WEL sample was the residue of the fragmented cementite lamella. The retained austenite was not found in both WEL types.
Figure 4.15 Examples of the microstructure characterisation, taken from the middle region of the TD-WEL sample. (a) (b) BF and DF TEM image with corresponding SAED image showing lens-shaped crystals of plate martensite. (c) (d) (e) and (f) BF, DF, HRSTEM, and corresponding FFT image, showing the crystal structure at the twin boundary.
Figure 4.16 Examples of the microstructure characterisation, taken from the middle region of the SD-WEL sample. (a) (b) BF and DF TEM image with corresponding SAED image showing very fine deformed twined martensite. (c) (d) (e) and (f) BF, DF, HRSTEM and corresponding FFT image, showing the high dislocation density and fragmentation of cementite particles embedded in the microstructure.
4.4 Discussion

4.4.1 The microstructure of White Etching Layer

Based on the operational traffic conditions, it was confirmed there are two distinct types of white etching layers formed on the rail track surface. Different WEL types experienced various loading and operational conditions. There were significant variations in the microstructural feature and grain size between the two different types of WELs. At transition zone, TP-WEL was associated with the undeformed pearlite structure comprised of relative largely grained structure; while SD-WEL was associated with plastic deformation and comprised the ultra-fine nanocrystalline structure. A mixture of fine and coarse plate martensite with a few amounts of lath like martensite retained austenite and very few precipitations of cementite were found in the structure of TP-WEL type. In comparison, the microstructure of the SD-WEL was composed of nanocrystalline martensite, fractured cementite and some deformed and defected ferrite films, with the full absence of retained austenite particles.

The microstructural evolution along the depth for both types of WELs was examined, showing the non-homogenous distribution of the microstructural features from the topmost surface down to the matrix region. The topmost surface (close to the rail surface) of the TP-WEL was confirmed to consist of plate martensite and retained austenite. The twined structure was found as a substructure of plate martensite. Retained austenite with shaped-like films was found between the plate martensite grains, showing approximately parallel to a close-packed direction in the martensite. The diffraction pattern confirmed that cementite particles were fully dissolved in this nanocrystalline region, instated diffraction spots related to iron oxide was detected. In the middle region of TP-WEL, plate-like martensite was slightly thicker than those on the top surface, associated with the existence of lath-like martensite. The presence of lath martensite was possibly due to the different distribution of carbon concentration in this region. Retained austenite particles were still found in the middle region of WEL and confirmed by HRTEM imaging. Fine carbide precipitations were found between the plates' boundaries (as confirmed by SAED), which is more likely to be precipitated from retained austenite due to the tempering process. At further depth near to the transition zone, the microstructure contained some individual cementite particles as well as segregated ferrite layer, confirmed by EDS analysis. The presence of cementite particles in the microstructure
indicates that the cementite lamellas have not fully dissolved. No retained austenite was detected in this region which indicates that austenite was fully transformed into martensite. The presence of retained austenite at the upper section of TP-WEL with the full absence at the bottom region is possibly due to the influence of cooling rate, as the top surface region is cooled slower than the bottom position of WEL which results in a high proportion of retained austenite during the transformation. At the transition zone between the TP-WEL and matrix region, it is clearly visible that the pearlite structure close to the interface boundary of WEL (matrix region), does not exhibit any plastic deformation which confirms that no work hardening was induced on the pre-existing pearlite structure. In addition, the uniform pearlite lamellae (strips of ferrite and cementite layers) have almost completely transformed into martensite phase structure at WEL. However, a very slight amount of cementite particles was still observed on the WEL area close to interface boundary, as confirmed by SAED imaging, which indicates that the cementite lamella has not fully dissolved.

In comparison, the topmost region of the SD-WEL exhibited a severely nanocrystalline microstructure, characterised by the very fine plates martensite with very few fragmented cementite and individual ferrite with high dislocation density. The SAED confirmed the tetragonal structure characterisation, by showing a separation of the (002)α to different (002)α’ and (200)α’ crystal planes associated with a distortion of the ferrite into tetragonal martensite. In addition, twinned martensite was also observed on this microstructure. Retained austenite particles were not detected in the topmost region of SD-WEL. Down within a few micrometer depths from the top, some individual fragmented pearlite lamella was seen on the microstructure, giving a slightly coarse-grained structure. The HRTEM image confirmed high structural defects induced in the lattice, considered by a high density of dislocation associated with high and low angle grain boundaries caused by large shear strain. In the lower region, the high local concentration of carbon particles was detected, which is associated with the occurrence of fragmented cementite. It can be recognised that the size of the cementite particle gradually increased with respect to depth. At the transition zone, there was a severely deformed pearlite structure underneath WEL, confirming the presence of work hardening caused by shear plastic deformation.

The horizontal plane of both WELs, taken from the middle section only was also characterised. The microstructural investigation confirmed a variation in the grain size
refinement between the two types of WELs. The crystallographic direction in the SD-WEL was elongated to the rolling direction, compared with the random distribution direction of the crystals in the TP-WEL type. The microstructure of both WELs types comprised a different grained size of plate martensite associated with twinned structure as a substructure of the plate. A slight variation was found with the crystallographic shaped of crystals, in which lens-shaped crystals were observed in the microstructure of the TP-WEL type. Regarding cementite particles, none was confirmed on the microstructure of the TP-WEL type. In contrast, the finely fragmented cementite was found in the area of high shear bands and dislocation density for SD-WEL. Retained austenite particles were not observed on the microstructure of both types of WELs.

For comparison, there was a significant variation in the grain size refinement between two types of WELs and across the depth of the WELs. SD-WEL type appeared to have a very fine nanocrystalline grained structure in the range between 5-20 nm compared with the relatively coarse-grained structure of the TP-WEL type, ranged between 30-200nm. This is because the microstructure of SD-WEL was severely refined due to repeated cycles of rolling/sliding contact fatigue. Across the depth, the grain size refinement was not uniformly distributed. The large-grained microstructure was seen within the upper section of the TP-WEL, and the grain size gradually decreased with respect to depth. In contrast, the nanocrystalline dimension of the grains across the small depth of the SD-WEL increased gradually due to the refinement of the cementite particles. The distinguished differences of grain size refinement with respect to depth could be related to the type of action induced by the wheel on the top surface of the rail to form WEL. The large-grained microstructure is most likely to be associated with the diffusion rate. The high grain growth results from the diffusion at high temperatures, which leads to an increase in the size of grains. In a different way, the long-time mechanical milling induces finer grain size near the surface region of the rail.

There is a strong agreement between a detailed microstructural examination by TEM and STEM characterisation in the current chapter and crystallographic structure analysis by synchrotron XRD in the previous chapter. Two distinguishable types of white etching layers on the rail surface were confirmed.
4.4.2 The mechanism of White Etching Layer formation

The formation mechanism was suggested based on the structure of WELs. TP-WEL containing martensite and austenite phase structure appears more likely to be the result of the martensitic phase transformation of the top rail surface material caused by thermal and/or thermomechanical process. Undeformed structure characterisation at the interface layer between TP-WEL and bulk material provides a solid support for the WEL formation being caused by phase thermal transformation. In contrast, the SD-WEL structure characterised by very fine nanocrystalline martensite and fragmented cementite, associated with the presence of severe plastic deformation at the transition zone is strongly indicative of WEL being induced by severe plastic deformation (mechanically process). Thus, it was concluded that the rail surface layer contains two different types of WEL and each type is caused by different actions on the contact zone between wheel and rail.

The nature of the WEL structure, as well as its formation mechanism, were reported by several previous studies. The laboratory simulated WEL by the mechanical process was found to have nanocrystalline martensite, fragmented cementite and ferrite with a total absence of austenite (Zhou et al. 2016). Similarity, martensite and retained austenite were the compositions of WEL structure being formed in the lab by a thermomechanical process (Wu, Petrov, et al. 2016b). However, only ferrite and cementite were observed in the WEL structure being induced in the lab by severe plastic deformation, while thermally induced WEL contained martensite, austenite and cementite(Hosseini et al. 2015). WEL obtained from the trade of U71Mn rails was found to be composed of martensite, austenite and ultra-fine undissolved cementite, and it is mainly caused by thermal action (Pan et al. 2017). The tetragonal crystal lattice distortion of martensite (bct) accompanied by the ferrite crystal structure (bcc) was identified in the WEL by synchrotron X-ray diffraction experiment, suggested thermally induced by phase transformation (Österle et al. 2001; Wang, L et al. 2003). However, observations from the deformed structure at the transition zone and from the WEL has shown almost similar characteristic feature such as cementite dislocation, confirming that WEL was the result of plastic deformation (Baumann et al. 1996; Lojkowski et al. 2001).
4.5 Conclusion

To understand the nature and formation mechanics of two distinct types of white etching layers formed on rail surface at different track/operational conditions, advanced characterisation technologies, such as Focus Ion Beam (FIB) milling, high resolution transmission electron microscopy (HRTEM) and selected area electron diffraction (SAED), were introduced to compare the detailed microstructure, the composition and morphology between two WEL types. The microstructure evolution along the depth from the topmost region down to the matrix were examined and discussed for the first time. The following conclusions are drawn:

1- There are two distinguishable types of white etching layers formed on the surface of the rail track; TP-WEL is composed of martensite and retained austenite, and the SD-WEL contained ultra-fine martensite crystals structure and finely fragmented cementite. TP-WEL is much thicker than SD-WEL.

2- The microstructure of the transition zone beneath the WEL comprised either deformed or undeformed pearlite. TP-WEL was associated with the undeformed pearlite while the SD-WEL type was correlated with the severely deformed pearlite.

3- Based on the microstructure characterisation, it was concluded that each WEL type was formed by different mechanisms of formation. TP-WEL associated with undeformed pearlite is thermally or thermomechanically induced by phase transformation. SD-WEL associated with the deformed subsurface pearlite is mechanically induced caused by severe plastic deformation.

4- The significant variation in grain size refinement and nanocrystalline dimension was observed between two types of WELs. TP-WEL contained a relative coarse-grained structure. In contrast, SD-WEL consisted of a very fine-grained structure.

5- TP-WEL type appears less homogenous than the SD-WEL type, due to containing a mixture of fine and coarse grain, particularly at the topmost region near to the rail surface. The coarse grain at the topmost region is most likely to be formed due to the diffusion at the high temperature.
6- The microstructural evolution varies with depth of WEL. For the TP-WEL type, austenite particles were found on the top section of the WEL and gradually disappeared with increasing depth, instead, cementite particles were fully dissolved at the surface while they were slightly present at the middle and near the transition zone. For the SD-WEL type, cementite particles were present along the depth with different crystallographic size and concentration as they were gradually fractured and dissolved.

In order to investigate different formation mechanics of WEL (thermal, thermomechanical, and mechanical condition), a series of advanced laboratory experiments, such as high-pressure torsion, Gleeble thermal-mechanical simulation, and electrical leaking current, etc, were carried out and discussed in following chapters.
5. Mechanism of WEL formation: Electrical Leakage between Wheel and Rail – Arcing phenomena

5.1 Introduction

A number of studies have focused on the formation mechanism of the white etching layer and its origin in the railway operation. Until recently, the following hypotheses consider the precise mechanics of WELs formation: (i) WELs are the result of the thermal process caused by wheel slip; (ii) WELs are mechanically induced by severe plastic deformation; (iii) WELs are caused by a combination of thermo-mechanical process. The mechanisms discussed above lead to the occurrence of white etching layers on the area of wheel and rail contact. This is because the contact patch which is the active point of the wheel on the rail is exposed to highest shear stresses which result in localised severe plastic deformation; and the highest rate of heat caused by wheel slip during excessive traction or braking effort. However, if the WELs are not on the running band area, it would suggest that there is another cause of WELs formation.

In the railway system, particularly the electrified railway, arcing phenomenon (electrical leakage) has been occurring more often and regularly on the rails. In the electrified railway, the current is delivered to the train traction motor via contact wires and then returned to the station via the contact between the wheel and the rail. If the contact between the wheel and the rail is temporarily losing, due to dynamic vibration, entrapped dirt or water, lubricant effect or oxidation occurrences, the high current can jump through the gap and results in arcing. The other resources of arcing also include the wheel passage the insulated joint and lightning on a train during bad weather. During the arcing, extensive heat is generated and speared over a large area of the top surface of the rail. Thus, arcing can be considered as another heat source for the railhead which, in addition to wheel-slip, might result in a change of microstructure and the formation of a WEL.

So far, there is no systematic research on electrical leakage between the wheel and rail, loosely defined as arcing, as a possible generator of WELs on rails. The present work investigates possible formation mechanisms of WELs on the railhead based on thermally induced transformations associated with electrical arcing. Samples of rail material obtained from the ex-service rail were examined under different laboratory leakage
current conditions. Structural and microstructural characterisation of the WEL formed via an arcing mechanism of current leakage was investigated in detail. The variations and similarities in both morphology and structure of WELs generated by arcing in the laboratory and after service in a metropolitan rail track network were compared and discussed.

5.2 Material and experimental procedure

A head hardened (HH) rail steel, cut from a curved rail track was used for the investigation. The microstructure of this steel is a predominately pearlitic structure, with the average grain size of 15 μm. The chemical composition and mechanical properties are given in Table 5.1.

<table>
<thead>
<tr>
<th>Grade</th>
<th>C, %</th>
<th>Mn, %</th>
<th>Si, %</th>
<th>Tensile strength, MPa</th>
<th>Hardness, HV</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>HHR</td>
<td>0.78</td>
<td>1.0</td>
<td>0.2</td>
<td>1200</td>
<td>350-380</td>
<td>10</td>
</tr>
</tbody>
</table>

Rectangular prism-shaped samples (15*10*8) mm were sectioned from a depth of 10 mm below the rail surface, where the material is considered to be still within the hardened layer but away from any microstructural changes on the top surface layer caused by train passage. As shown in Figure 5.1, samples were subjected to electrical discharges by using Gas Tungsten Arc Welding (GTAW) machine with an arc length between 3.5 to 4 mm. Thermocouples were placed just close to the sample surface to measure the temperature changes during the testing. The arc current was controlled and moved along the sample’s surface in the direction of travel, as indicated by an arrow. Five repeated runs under the same arcing condition were applied on the surface of the sample. Each run was separated by a 2 mm interval after 2 minutes dwell time from the previous run to ensure the arcing over the virgin rail surface.
There is little literature on the nature of the sporadic arcing observed between wheels and rails. In practice, it appears that this would be difficult to measure. In order to model a range of arcing applications four different sets of energy inputs were used, each condition involving a different arcing current (Figure 5.1b) and travel speed. Four individual samples were tested under the sets condition listed in Table 5.2, with parameters set so that no visible arcing damage was produced on the test specimens. Five repeated runs under the same set condition were applied in each sample. Figure 5.2 shows samples before and after testing using Condition S3. Based on the thermal energy input, each sample was subjected to a specific localized heating effect and then followed by subsequent self-quenching to ambient temperature, which is similar to that expected during arcing at the wheel/rail interface.
Table 5.2 details of the parameters used.

<table>
<thead>
<tr>
<th>Sets</th>
<th>Travel speed, mm/min</th>
<th>Current I, amps (A)</th>
<th>Voltage, Volts (V)</th>
<th>Total arc energy input, J/mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>#1</td>
<td>285</td>
<td>23</td>
<td>11</td>
<td>53</td>
</tr>
<tr>
<td>#2</td>
<td>196</td>
<td>23</td>
<td>11</td>
<td>77.5</td>
</tr>
<tr>
<td>#3</td>
<td>98</td>
<td>23</td>
<td>11</td>
<td>155</td>
</tr>
<tr>
<td>#4</td>
<td>98</td>
<td>32.2</td>
<td>11</td>
<td>217</td>
</tr>
</tbody>
</table>

Figure 5.2 Samples before and after testing.

Samples containing WELs, taken from an ex-service rail, as shown in Figure 5.3, were also considered in this study for comparison. It has been found that in this sample, WEL occurred uniformly at the region between the running band and the rail field side, away from the rolling contact patch. The existence of WEL located away from the running contact band is inconsistent with a deformation-induced phenomenon and is considered highly likely to have been induced by a leaking current rather than wheel slip or rolling contact, as these phenomena only occur in the vicinity of two contact surfaces.
For valid comparative purposes in this study, it is important to note that the laboratory arcing test pieces were cut from the same HH rail section near the pre-existing WELs. In order to distinguish between these samples, the name “Sx” is used for laboratory samples, with the letter S is used to describe a set condition and x used to describe the set number whereas the term “RW” is used for the ex-service rail WEL samples.

A small slice piece was cut thoroughly the transverse section of the samples for metallurgical investigation. Microstructures were characterised by optical and scanning electron microscopy. The samples were mounted in resin, ground to 1200 grid, polished to 1-micron diamond and etched with 2.5% Nital. Hardness investigations were performed using a micro-Vickers hardness tester with a 100 g load.

Thin WELs lamellas, of length about 20 μm were lifted from simulated WEL and from Rail WEL samples using the Focused Ion Beam (FIB) technique (FEI Helios NanoLab G3 CX focus ion beam). These lamellas were lifted out from the interface zone (the area between WEL and matrix). Samples were characterised by transmission electron microscopy (TEM) using a JEOL JEM-ARM200F TEM instrument, operating at 200 kV. Figure 5.4 shows the focused ion beam process for lamella sample preparation. Transmission Kikuchi Diffraction (TKD) was performed on the WELs lamella for microstructural and nanocrystalline analysis, using JOEL JSM-7001F SEM, with accelerating voltage of 30 kV and step size of 15 by 15 nm.
X-ray diffraction was performed on both the Simulated and Rail WELs, to determine the variations and similarities in morphology, the crystallographic structure, and phase present between WELs induced by electrical discharge and those formed on the rail surface.

Figure 5.4 Focused ion beam process, (a) selected area of interest, (b) Pt layer deposited onto area (c) milling both sides of the samples to produce a trench, (d) left out the sample, (e) and (f) bright and dark field STEM image after final thinning and cleaning process.

5.3 Results

5.3.1 White etching layer formation by arcing

Sections of WEL into the base rail are shown in SEM images, Figure 5.5 a, b, c and d for samples S₁, S₂, S₃ and S₄, respectively. White Etching Layers or WEL patches were found to form along the top surface layer of samples where the current leakages were applied.
The formation size and thicknesses of the WELs on the surfaces of each rail steel sample were found to depend on the conditions applied. Small WEL patches were formed under the lowest arcing current and the size of the WEL patches significantly increased when the total arc energy input increased, due to either decreasing the travel speed or increasing current. In addition, observations from optical microscopy revealed that there was a distinguishable contrast difference in WELs associated with different energy inputs. WELs appeared much brighter under low energy input conditions, as shown in Figure 5.5 a, b and c, but darker under higher energy inputs used, as shown in Figure 5.5d. The amount of energy input is an inverse function of travel speed. For slow movement both the heat source remains above the surface for longer periods and there is more eddy current heating beneath the surface. The darker appearance of the layer (Figure 5.5d, which appears brown under white light) is consistent with the formation of the so-called brown etching layer (BEL) commonly found on the rail surface in conjunction with the white etching layer. Formation of both WEL and BEL on the rail steel due to deliberate arcing current represents strong evidence that arcing associated with electrical leakage can result in WEL formation on the surface of the rail.

![Figure 5.5 Cross-sectional area showing the formation of WEL and BEL on the rail steel due to current leakage. (a) (b) (c) and (d) shows the surface layer of sample S1, S2, S3, and S4 respectively.](image)

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5.3.2 Brown etching layer formation by arcing

Figure 5.6a shows the formation of two distinguished overlapping patches; white etching layer (WEL) and brown etching layer BEL. The dark region is BEL while the light region is WEL, as it appears under optical microscopy. This characteristic was only observed on the samples that were subjected to high energy inputs (Set #4 conditions). From the schematic illustration shown in Figure 5.6b, BEL was found at Patches #1, #2, #3 and #4 while WEL was only induced at Patch #5. The overlapping characterisation view in Figure 5a considers the cross-sectional area between Patch #4 (BEL) and Patch #5 (WEL), as indicated by the dashed box in Figure 5.6b. Practically, each patch was generated after 2 minutes of holding time and a 2 mm distance from the previous patch. It can be concluded that Patch #1 was affected by a tempering process associated with the arcing process following Patch #2, as is for Patches #2, #3 and #4. Whereas, after producing Patch #5 the sample was not subjected to a temper reheating process as there is no following neighbour arcing. The interpretation of this is that WEL was formed originally, and later transformed to BEL due to tempering. During reheating, the pre-formed martensite layer was tempered at a temperature below the critical point, resulting in changing of the microstructure and mechanical properties.

Changes in the mechanical properties can be clearly seen in the variations of hardness measurement between WEL and BEL. The maximum hardness values, ranging between 700-760 HV were found on the WEL. The hardness gradually degreased to about 600-650 HV at the transition layer. The hardness then significantly dropped to about 400-485 HV in the BEL. In addition, the hardness distribution across the BEL was not uniform indicating that the structure of BEL is not homogenous.

The SEM micrographs in Figure 5.6 c and d show representative microstructures of the white etching layer and brown etching layer, respectively. It can be clearly seen that there is a significant difference in the morphology between the WEL and BEL. The microstructure of WEL contained much finer features compared with BEL. The microstructure of WEL comprised predominately very fine needle-shaped crystals, revealed by TEM to comprise carbide in the martensite. Whereas the BEL microstructure consists of a mixture of tempered martensite and coarser carbides. In the BEL, the martensite has partially decomposed into carbide and iron, and the distortion of crystal
structure produced is consistent with the drop in the hardness from ∼700HV in the WEL to ∼400HV in the BEL.

The microstructures of white etching layers that were simulated in Samples S1, S2 and S3 appear identical, as shown in Figure 5.5 a, b and c, respectively, and also are consistent with the WEL in Patch #5 of Sample S4 (Figure 5.6 c). The WELs comprise very fine particles and there is the complete disappearance of the lamellar pearlite structure, indicating that partial or complete decomposition has occurred. Therefore, due to the structural similarities between samples S1, S2, and S3, only one sample was tested under Set #1 condition (S1) and considered for advanced structural and microstructural analysis. An intensive investigation was also conducted on the Rail WEL (RW) sample, for comparison.

On the other hand, the evolving microstructure of the simulated brown etching layer appears much different than those of WELs, either from the lab simulation or from the ex-service rail. Further investigation of the BEL and comparison with Rail BEL will be the subject of a different study.

![Figure 5.6 Overlapping characteristic of WEL and BEL, (a) WEL and BEL with microhardness distribution, (b) Schematic illustration shows the formation of BEL and](image-url)
WEL on the surface of the sample ($S_1$), (c) and (d) SEM micrograph showing the texture structure of BEL and WEL.

5.3.3 Microstructures of simulated WEL and that from ex-service rail

The microstructural features of the Simulated WELs under Set #1 condition ($S_1$) were characterised and compared with WELs (RW) formed on the ex-service rails, using optical microscopy, microhardness measurement, high resolution imaging by SEM, TEM and STEM characterization with electron selected area diffraction SEAD imaging and EDS analysis, as well as TKD in SEM analysis.

5.3.3.1 Optical micrograph and microhardness studies

Surface characterisation of ($S_1$) and (RW) samples, along the cross-section, perpendicular to the travel direction is shown in Figure 5.7 a, b. Similar characteristics were observed in both samples, with the existence of WELs with a bright featureless contrast on the top surface layer of the samples, and a sharp change beneath to a distinct pearlite microstructure. For the RW sample, WEL occurred uniformly on the rail top, in the region between the running band and the field side, away from rolling contact patch. The presence of WEL here strongly confirms that neither severe plastic deformation nor frictional heating by wheel slip was the cause of this WEL. The WELs on the surface of ($S_1$) and (RW) were about 75-80 μm in depth. The top surface layer of (RW) sample appears to be pitted, possibly due to high impact loads of wheels on the rail. Underneath both WELs, a sharp interface layer was observed between the WELs and the matrix. The region at the interface the WEL is characterized by an undeformed layer of pearlite, indicating an absence of severe plastic deformation. At further depths, no grain elongation was recognized and a microstructure of pearlite colonies (the matrix structure).

Figure 5.7c shows the distribution of microhardness values as a function of depth, measured from the surface of the WELs down to the matrix region of both the Simulated ($S_1$) and Rail WELs (RW). The hardness profiles for both WELs appear almost identical. The hardness values of the two WELs were very close to the topmost region. At depth about 30 μm from the surface, a slight variation in the hardness between $S_1$ and RW samples was found, showing a slight increase in the RW sample and a slight decrease in the $S_1$ sample. The hardness then sharply dropped down to about 400-380 HV at a depth
85 μm, near the interfaces layer. Underneath both WELs from depths more than 100 μm the hardness was consistent with that of bulk material of 380Hv.

From both the optical micrographs and hardness distribution profiles, there is high consistency between the Simulated WEL due to electrical current arcing, and the WEL observed in ex-service rails.

![Figure 5.7](image)

Figure 5.7 (a) Rail WEL (RW), (b) Laboratory Simulated WEL, sample (S1) and (c) Hardness profile for both RW and S1 Samples.

### 5.3.3.2 SEM investigations

Figure 5.8a and b show SEM micrographs for both the lab simulated WEL(S1) and Rail WEL (RW), obtained from an area close to the interface between the pearlite and martensite. For the RW sample, a region of pearlite lamellar structure can be observed in the lower right part of Figure 5.8a. A much finer crystalline structure can be recognized in the area of WEL which was revealed by TEM (Section 5.3.3.3) to comprise martensite, and regions, described as interface layer (Figure 5.8a) is located between the pearlite
colonies and the WELs. The result for the WEL is consistent with a decomposition of cementite lamellae caused by the temperature rise of the rail surface, resulting in formation austenite and its subsequent transformation to martensite.

In comparison, similar characteristic features were observed in SEM images of the microstructure of the lab sample ($S_1$), as shown in Figure 8b. The dashed line indicates the boundaries between the pearlite (lamellae of ferrite and cementite), and WEL or interfaces layer. An apparent partial fragmentation of the cementite lamella appears in the interface region, characterized by a combination of rod and spherical like-particles inside a ferrite matrix. The WEL region has similar features to the Rail WEL.

In fact, the microstructure of WELs being induced leaking current experiment and those taken from the rail surface were severely refined, which make it difficult to identify the microstructural features by SEM. Thus, further detailed microstructure features of both WELs based on TEM and STEM observation results were analysed and discussed in the next section.

![Figure 5.8 SEM micrograph obtained at the Interface layer showing WEL, transition, and matrix for (a) Rail WEL (RW) sample, (b) lab simulated WEL in the sample (S1).](image)

### 5.3.3.3 TEM and STEM studies

Figure 5.9 shows STEM micrographs, obtained from the transition zone located between the pearlite matrix and WEL region, for both Rail WEL and lab, generated WEL samples. The high angle annular bright-field (HAABF) and dark-field (HAADF) images in Figure
5.9 a, b are from the transition zone. In the full region, at the right of the image, the structure comprises undeformed colonies, with an absence of evidence of plastic deformation. For the orientation of the region examined, the average interlamellar spacing is about 50-100 nm, with a thickness ratio of around 1 to 3 between ferrite and cementite strips. A microstructural transition zone is indicated by a dashed line on the left side nearly all of the pearlite lamellae is almost disappeared, possibly due to thermal-induced decomposition, which indicates the beginning of phase transition (martensite transformation). The dislocation was observed in this region to be generated in the ferrite layer, while the cementite films that appear brighter in the pearlite matrix in the dark field image, Figure 5.9 b, are partially dissolved (as referred by arrow). On the left side of the image, the perlite structure has completely replaced by a martensitic structure that corresponds to the WEL region. WEL exhibits a complex structure, containing a mixture of a very small percentage of lath martensite, plate martensite, some non-transformed retained austenite, and remanent of ferrite particles. Lath martensite morphology can be seen at the bottom region of the brightfield image where carbide is partially dissolved. The presence of different martensite morphologies in the microstructure could be related to the variation in carbon concentration, which means that plate martensite is associated with the region of higher carbon content while lath martensite is formed at a region of lower carbon content. Indeed, steel with carbon content ranging between 0.5 to 0.8 wt.% is generally complicated, and all martensite type morphologies in addition to retained austenite can exist in the microstructure (Tkalc 2004). In addition, some carbide precipitations were found in the WEL region, having a much bright contrast as shown in the dark field image, Figure 5.9b, properly due to a high concentration of carbon. A very thin film of retained austenite was found to be located along martensite plate boundaries, as indicated by arrows in bright and dark field images, which is strongly support the thermal phase transformation of austenite to martensite. In comparison, the crystalline structure at the transition zone of WEL that is induced by arcing in the lab has almost similar features of Rail WEL, as shown in the bright and dark field image, Figure 5.9c, d. Pearlite structure at the right side of the image exhibits a relatively coarse spacing; with an average lamellar spacing of 100 -150 nm and ferrite to cementite thickness ratio are about 1 to 2. Manganese sulphide (MnS) inclusions were found in the matrix region crossing pearlite colonies. Some cementite particles, with rod and spherical like shape morphologies, were found in a ferrite matrix at the termination
of the lamella, indicating that the transformation was mainly from fragmented cementite lamella. As Rail WEL structure, both lath and plate martensite, as well as retained austenite can be recognized in the martensitic matrix phase (WEL). Retained austenite was observed at the pearlite grain boundary, confirming its location in the austenite phase map, as presented in Figure 5.11a. In addition, some ferrite grains were recognised in the microstructure as seen at the bottom of the image, which may be related to high tempered temperature or long period tempering. In the brightfield images, some darker spots that appear brighter in the darkfield images were recognized within the WEL area, which is related to different diffraction contrast associated with particles that have a different crystallographic orientations.
Figure 5.9 TEM micrograph obtained at the transition zone of WEL, (a) and (b) bright field and dark field image for Rail WEL, (c) and (d) bright field and dark field image for lab simulated WEL.

Figure 5.10 and 5.11 shows the distribution of chemical elements taken by EDS analysis over the interface border of Rail WEL sample and simulated WEL sample, respectively. The trace of Carbon (C-K), Oxygen (O-K), Silicon (Si-K), Manganese (Mn-k) and Iron (Fe-K) were mapped. It can be seen that both WELs exhibits almost similar behaviour, in which the distribution arrangement of elements at the transformed and untransformed region. Form C-K mapping, both WELs reveals significant changing in the microstructural arrangement of cementite lamellar from the matrix pearlite towered the nanocrystalline layer over the interface border. This confirmed that the originally lamellar shaped of cementite with uniform bright contrast lines at the undeformed region gradually dissolved at the interface boundary and almost fully decomposed at the WEL region resulting in the formation of a nanocrystalline supersaturated Fe-C solid solution. From (O-K) mapping, a very little oxide content was seen with no much variation between the matrix and the WELs. The distribution of Silicon and Manganese components over the transition zone was almost uniform for both WELs, with a very slight concentration of Si found at the interface border. From Iron mapping, no much variation was seen on the matrix and nanocrystalline region of both WELs. The dark region appeared along the half of the mapping area was related to the thickness variation of the lamella due to partly diffusion of the protective Pt layer during the final thinning process.
Figure 5.10 EDS mapping of Rail WEL, taken over the interface layer between matrix (pearlite region) and WEL, showing the distribution of chemical elements.

Figure 5.11 EDS mapping of simulated WEL, taken over the interface layer between matrix (pearlite region) and WEL, showing the distribution of chemical elements.
Figure 5.12 shows a typical brightfield TEM image with corresponding selected area electron diffraction (SAED) pattern of both investigated WEL samples, obtained from regions located in the middle of the WELs (microns above the transition zone, toward the top surface). For both WELs types, the structure is characterized by plate martensites with internal nano-sized transformation twins as substructural features within the plates. For the Rail WEL, the SAED pattern in Figure 5.12 b shows an [0̅11] bcc zone axis, in which the solid and dashed lines indicate diffraction spots from the bcc matrix and the associated spots from twins, respectively. Extra diffraction spots, indicated by the arrows, were found at 1/3 (2̅11̅1) and 2/3 (2̅1̅̅1) are also associated with the matrix and twins.

In the lab simulated WEL, Figure 5.12c, distinguishing features included both twined and untwined grains. The untwined region at the upright side of the image is believed to comprise ferrite or martensite particles, with contrast due to high dislocation densities. Formation of martensite and/or some individual ferrite grains with a very high density of dislocations is possibly due to grain size refinement. It is possible that large grains of austenite transform to martensite while small grains of austenite transforms to ferrite. Martensitic transformations are mainly governed by free energy associated with shear mechanism induced in the lattice. The free energy induced in the small size grains can be insufficient to support a martensitic transformation and thus a transformation is to ferrite more favorable. Figure 5.12d shows a SAED pattern taken over the twined region. From SAED, two martensite orientations with {111} type bcc zone axes were identified in the diffraction pattern. The solid circle in Figure 5.12d indicates a twin spot. It can be seen that diffraction spots consistent with the carbide phase are present, in addition to the matrix and twins in the diffraction pattern. The result indicates that nano-precipitates of carbide are embedded within the plates, a result attributable to a tempering process. The fine twins and high dislocation densities are consistent with the deduction that the structure is predominately martensitic.
Figure 5.12 TEM micrograph showing the structure of WEL corresponded with the diffraction pattern, (a) and (b) for Rail WEL, (c) and (d) for lab simulated WEL.

5.3.3.4 TKD characterisation analysis

Figure 5.13 shows the analysis results of the microstructures of the WELs, determined by TKD scans on the thin lamella prepared by focus ion beam (FIB) milling. The results are obtained from interface regions in both the lab simulated sample ($S_1$) and Rail WEL sample (RW). Greyscale image quality (IQ) maps, imposed by austenite phase maps for both samples are presented in Figure 5.13a, b. Variations in the grey colour indicate the pearlite strips (matrix region) which are light grey colour and regions of martensite (WEL) which appear much darker in comparison. The different grey levels indicate
different dislocation densities associated with the formation of martensite. In the WEL regions, the structure is not completely martensitic but contains retained austenite phase, indicated by red colour. Detection of retained austenite in the microstructure of WEL indicates that incomplete phase transformation has occurred during the quenching process. A comparison between the phase maps for both samples shows significant variations in the area of a fraction of retained austenite. WEL formed on the rail surface contains much higher amounts of retained austenite, as shown in Figure 5.13 b, while only a small amount of retained austenite is detected in the lab formed WEL located close to pearlite boundaries, as shown in Figure 5.13 a. The occurrence of retained austenite at the boundaries with the pearlitic regions is likely attributed to the more irregular arrangement atoms and an inability of martensite to no longer form (Callister). The difference in the proportion of retained austenite phases may be related to differences in thermal history and increment of peak temperature. The presence and amount of retained austenite can affect the mechanical properties of the material (Rafi et al. 2014). The relative decrease in the hardness values measured in the WEL region for the Rail sample compared to lab simulated samples could be associated with the higher amounts of retained austenite in the microstructure in addition to grain refinement effects of hardness.

In order to identify the changes in the microstructural parameters including the crystallographic orientation, EBSD scans were performed on the FIB lamellae. Figure 5.13 c, d shows inverse pole figure (IPF) maps with respect to Z- direction, imposed with grain boundary (BC) map for both the laboratory simulated and Rail WEL samples, respectively. Crystallographic development can be clearly recognized in the WEL texture. The grains in the WELs region appear to have different orientations, associated with different colours, indicating grain fragmentation that causes distortion of the crystals. For the lab formed WEL, the grains are slightly elongated at flattened, possibly due to recrystallization and growth processes, resulting in a more plate-like morphology compared to that in the Rail sample.
Figure 5.13 IQ ferrite, martensite, and austenite map for (a) simulated WEL (b) Rail WEL; IPF map with respect to Z-direction for (c) simulated WEL (d) Rail WEL (c); enlarged map showing the relationship between austenite and ferrite for (e) simulated WEL (f) Rail WEL.
Figure 5.13e and d show enlarged views of the selected area, indicated by the dashed square shown in Figure 5.13 a and b. The red areas represent the austenite phase (fcc) inside the ferrite matrix, which is characterised the grey pixels. It can be seen that the austenite particles are located at the grain boundaries. The white line that is present at the austenite boundary represents an observed orientation relationship between austenite phase structure as; (011) fcc is parallel to (111) bcc.

Figure 5.14 displays the relative frequency (%) as a function of grain size distributions of the laboratory simulated WEL in S1 and Rail WEL. This quantitative calculation was only performed for the WELs regions and completely excluded the pearlite areas. It can be seen that the highest diffraction ratio was associated with the finer grain sizes range between 60-80 nm for the laboratory simulated WEL sample and 100-140 nm for Rail WEL sample. For comparison, the grain size distribution shown in Figure 5.14a is almost comparable for both WEL regions. The maximum grain diameter, about 5.993 μm was found in the Simulated WEL, which is larger than that in the Rail WEL, 1.6 μm. The presence of coarse grain in the microstructure of WEL is attributed to the presence of some pearlite islands, which remaining undissolved. For overall grains and particles, the mean size measured in the laboratory WEL is 212 nm in compared with 274 nm of Rail WEL.

The distribution of the local misorientation angle was determined in order to estimate the strain distribution and deformation localisation in the WELs. Figure 5.14b shows the misorientation angle distribution, measured for the low angle sub-grain boundaries (θ < 5°), and calculated with respect to third nearest neighbours in regions of the Rail WEL and lab simulation WEL. The analysis of both WELs generally shows a high fraction of low angle boundaries, concentrated in the range of angles 0°- 1°. Both samples appear to have almost similar distributions, indicating that a lack of plastic deformation in the vicinity of the WEL on the rail surface. In addition, it should be noted that areas with a high fraction of low angle boundaries are characteristic of regions of high dislocation densities (Wright et al. 2011).
Figure 5.14 Relative frequencies (%) as a function of grain size distributions in (a) and local misorientation in (b) for simulated and Rail WELs.

5.3.4 Structural analysis of simulated WEL and Rail WEL.

Figure 5.15 shows the XRD profiles of the lab simulated WEL sample ($S_1$) and Rail WEL (RW). The (200), (220) and (311) peaks of austenite are present in the WELs profiles, showing a relatively high amount of retained austenite in the Rail WEL structure compared with that generated by experiment, and confirming the results obtained by TKD. The peaks of cementite were barely visible but are still present in the diffraction pattern, at angles between 47° and 50°, indicating that the cementite is either not fully dissolved and/or has formed due to decomposition of retained austenite.

Figure 5.15b and c show the diffraction pattern at angles between 63° and 67° which includes the 200 reflection of ferrite for both WELs reflections. It can be seen that the profiles show a distinct asymmetry, indicating a distortion of the ferrite lattices into
tetragonal martensite. The (200) α has separated to (002) α’ and (200) α’, consistent with martensite formation and confirmed by TEM and STEM observations. The lattice parameters a and c were determined based on Bragg’s law using a double pseudo-Voigt function for peaks analysis and the results showed that the c/a ratio for both WELs peaks are approximately 1.09-1.011, which indicate the formation of oversaturated Fe-C alloy.

Figure 5.15 XRD profiles of the Rail WEL, lab simulated WEL sample S1 (a) full pattern showing the presence of austenite peaks in the both WELs and weak peaks of cementite, (b) & (c) 2 00 plan showing the asymmetric reflection of WEL profile indicating to martensite structure.

5.4 Discussion

Microstructural evolution associated with WEL formation on the surface of the rail was found in regions where arcing is deliberately applied. Leaking current or arcing is a common phenomenon that is more often occurred between the wheel and the rail. For the electrified railway system, the power is generated and used to operate the traction motor, and then transmitted to the rail via contact point between the wheel and the rail. When
there is a problem in contact condition due to certain circumstances such as dynamic vibration or presence of unconducive media between the wheels and the rails as well as different levels of wear conditions, the traction current leaks through the small gap, resulting in arcing phenomenon. During arcing, intense heat is generated and dissipated on the surface of the rail, causing microstructural changes of the material associated with the formation of the white etching layer.

Previous studies found that thermally induced white etching layers were principally developed in the rolling contact area from the frictional heating process by wheel slip. However, if the WELs are not only present on the rolling band area but also on the different locations of the rail surface, it would suggest that there is another source of heat generation. In the current chapter, the laboratory experiment provides evidence that leaking current results in an intensive heat production on the rail surface material, which leads to WELs formation by phase transformation. In railway, arcing can take place on the entire area of the rail surface. Therefore, WELs can be found everywhere on the rail surface including rolling contact area and outside of running band.

The experiment was conducted on the rail steel material, taken from the subsurface of ex-service rails, under different sets condition. The conditions and parameters used in the experiment can be achieved by a real application. It is difficult to measure the exact amount of leakage current induced on the rails during service, current research provides a clear understanding of the arcing mechanism as a formation mechanics of WELs on the rail surface.

Figure 5.16 shows that the depth and size of white etching layers were a function of the conditions used. It can be seen that WEL depth dropped by traveling faster at the same current applied while WELs depth significantly increased when the current increased under the same travel speed condition. Lower travel speed results in higher energy input while higher travel speed results in lower energy input.
Brown etching layer (BEL) was formed on the rail material by arcing mechanism. Until recently, the formation mechanism of BEL is still unclear. Recent research was suggested that BEL is the decomposition product of martensite, as well as the detection of retained austenite in the microstructure of BEL, could also interpret the contribution of thermal activity (Wu, Petrov, Li, et al. 2016). In current work, BEL was simulated on the samples by arcing after applying higher energy input. BEL was thought to be the result of the tempering process. When a higher energy input was passed along the surface of the sample, a large patch of WEL was generated by rapid subsequent cooling at room temperature. Then, after 2 minutes freely cooling, the current was applied again at 2 mm distance from the center of the pre-generated WEL patch at the adjacent region. As a result, the pre-generated WEL was re-heated and tempered and transformed into BEL. It has been found that BEL contains a coarser-grained structure and shows a lower hardness value compared with those of WEL.

The microstructure of the simulated WELs by leaking current in the lab is very similar to that of WELs induced on the rail during service. For both of Rail WEL and simulated WEL, the interface between WEL and transition zone exhibited a sharped layer, confirming that plastic deformation does not take place. From the microhardness measurement, both WELs exhibits a high value of about 780-800 HV. SEM observation revealed that the simulated and Rail WELs were composed of fine-crystalline features.
characterized by a mixture of needle and blocks shaped particles in small-sized range, indicating to a heavy decomposition of cementite lamellae in the solid solution. Both WELs contained the martensite and retained austenite with a very small amount of undissolved cementite. The existence of austenite in the microstructure of WELs indicates insufficient phase transformation occurred during the cooling process. From TKD analysis, the volume fraction of retained austenite between simulated and Rail WELs was varied. WEL formed on the rail surface contained a much higher amount of retained austenite while only a few amounts of retained austenite is detected at the interface boundaries of the lab formed WEL. The occurrence of retained austenite at the boundaries is properly due to more irregular arrangement atoms, and thus the martensite can no longer form (Callister). The difference in the proportion of retained austenite phases may be associated with thermal history and increment of peak temperature. Nevertheless, the presence of retained austenite phase in both WELs structure, having a parallel orientation relationship with the adjacent ferrite phase, provides evidence of phase transformation from pearlite into austenite and a subsequent partial transformation into martensite during rapid cooling.

5.5 Conclusion

A formation mechanics of WELs on the rails under the effect of the leaking current (arching phenomenon) has been proposed and investigated. Microstructural evolution associated with WELs formation on the surface of the rail was found where arcing is applied. Structural and microstructural characterisation of arcing induced WEL and pre-existed WEL obtained from the ex-service damaged rail were measured by using optical and scanning electron microscopy (SEM), microhardness measurements and high-resolution technique such as transmission Kikuchi diffraction analysis and FIB/TEM. The main results can be summarised as follows:

1. White etching layers can be thermally induced by arcing mechanism. The dimension of WELs was a function of thermal energy input associated with the travel speed and current amount
2. Microstructural feature of the arcing induced WEL and pre-existing WEL is similar, which are composed by fine martensitic units (fine plates and very few lathes), undissolved cementite lamella, remnant pearlite lamella and some grains of retained austenite.

3. Undeformed pearlite structure occurs in the transition region between WELs and matrix for both WELs, confirming the absence of shear deformation. The lack of plastic deformation in the vicinity of both WELs is supported by the crystallographic analysis from TKD characterization, which shows low misorientation occurrences between the grain boundaries.

4. The detection of austenite phase in conjugate to ferrite phase, having a parallel orientation relationship supports the thermal phase transformation mechanism of both arcing induced WELs and field pre-existing rail WELs. The twinned structure observed by TEM supports the shear mechanism caused by atom movements during martensite transformation.

5. Brown etching layers (BELs) were formed in parallel to arcing induced WEL for the case of high energy input. It was found that BEL was formed as a result of the tempering process of the pre-generated WEL.

6. The correspondence between simulated WELs and field rail WELs leads to the conclusion that electrical leaking current arcing is an alternative formation mechanics of white etching layer on the railhead based on phase thermal transformation.

7. The rail surface damages such as studs (types of squats defects) are often linked with the thermal damage induced on the rail surface. Therefore, the formation mechanism of WELs by arcing could be a root cause and origin of a stud.

8. The understanding of formation mechanics of WELs is of high importance as its perceived link with rail damage and the high cost of track maintenance procedures such as rail grinding designed to remove surface damage. A different track maintenance strategy is required to reflect the various types of formation mechanics of WELs.
6. Mechanism of WEL formation: Severe plastic deformation induced WEL

6.1 Introduction

As discussed in the previous chapter, there are two common hypotheses of white etching layer formation; phase transformation induced martensite (thermally induced) and severe plastic deformation (mechanically induced). WELs thermally induced by phase transformation due to electrical leakage was confirmed. In this chapter, WELs mechanically induced by severe plastic deformation are discussed.

On the surfaces of both rails and wheels, severe plastic deformation is accumulated in the surface layer due to cyclic rolling/sliding contact loading. The accumulated plastic deformation results in working hardening lead to a decrease in the fatigue resistance which, in turn, promotes rolling contact-induced defects. In pearlitic rail steel materials, severe plastic deformation results in microstructural evolution, which leads to change the mechanical properties of the material such as yield stress and fracture toughness. The change in the material characteristic influences the fatigue performance and consequently leads to cracks formation.

The most common microstructural change observed near the rail–wheel contact surface of the rail is called the white etching layer (WEL). WEL is characterised by a high level of hardness, therefore, by notable fragility, this layer represents a critical location for the occurrence of cracks. The cracks initiate at the edge of WEL and within the WEL and then propagate into the parent material. The detachment of the WEL and cracks within the WEL are proposed as the most likely mechanism for the initiation of squat type defects. In general, it is accepted that squats can develop from the WELs which are believed to be generated by thermal martensitic phase transformation and / or severe plastic deformation on the rail surface.

In order to investigate whether severe plastic deformation results in the formation of a white etching layer, High Pressure Torsion (HPT) experiment was performed on the pearlitic rail steel. HPT is considered the most powerful technique for producing severe plastic deformation because the material is subjected to large compressive hydrostatic stress and simultaneous extremely large torsional straining (Park et al. 2016; Zhilyaev &
Langdon 2008). This process results in high plastic strain and grain refinement. HPT was used because the loading condition is similar to that of the rail-wheel contact. Therefore, a microstructure induced by deformation of rail steel material samples being tested by HPT is similar to that found in the plastically deformed surface layer of rails.

6.2 Experiment

A head hardened (HH) rail steel, containing 0.78% C - 1.0% Mn- 0.2% Si with the hardness of about 370 HV and tensile strength of 1200 MPa, was used for investigation. Samples were taken from 10 mm below the surface of the railhead where the material structure is considered to be undeformed. Disks of 10 mm diameter and 1.5 mm thickness were machined and subjected to high-pressure torsion (HPT) tests using Walter Klement high-pressure torsion HPT system. Figure 6.1 shows the schematic illustration of the sample position in relation to the rail section, sample dimension and the testing set up. Samples were placed between the lower anvil (stationary) and the upper anvil (rotating). High pressure was applied concurrently with the torsional straining by turning the upper anvil clockwise relatively to the non-rotating lower anvil, with no-slip between the anvils and the surface of the samples. The test was conducted at room temperature, under hydrostatic pressures of 2, 4 and 6 GPa and torsional staining corresponding to 1, 2 and 5 turns at speed of \( \omega = 1 \) revolution/ minute.

![Figure 6.1 Schematic illustration of sample position in relation to the rail section, sample dimension and the testing set up.](image)

After testing, the HPT processed samples were prepared by standard metallographic procedures and then characterized by optical and scanning electron microscopy using JEOL JSM-7001F. The circular cross-section of the sample at radius R=5 mm (the edge)
was considered an observed plane for all OM and SEM images. The microhardness was measured using the Vickers diamond with 100g load. X-ray diffraction technique was performed to identify the crystallographic structure of samples using the GBC MMA XRD instrument. Cu k-alpha radiation (wavelength $\lambda=1.54056$ Å) was used and phase analysis was performed in a range between 40° and 120°.

Further detailed microstructure features of the HPT processed samples were characterised by transmission and scanning transmission electron microscopy TEM and STEM. Sample subjected to the condition pressure of 6 GPa and 5 turns was selected for characterisation. Thin foil samples were prepared for TEM by Focused Ion Beam (FIB) technique using FEI Helios NanoLab G3 CX focus ion beam. The thin foils were lifted out from the region where the high shear strain was induced (the circular cross-section at radius $R=5$ mm, the edge), Figure 6.2 shows the steps of TEM sample preparation. The samples were characterised by bright field (BF) and dark field (DF) imaging, the selected area electron diffraction (SEAD), Energy-dispersive X-ray spectroscopy (EDS) analysis and high-resolution imaging (HRTEM) provided by TEM and STEM using JEOL JEM-ARM200F TEM instrument with accelerating voltage of 200 kV.

The microstructural features of the severely deformed region induced by the HPT experiment were discussed and compared with the mechanically induced white etching layer sampled from the ex-serviced rail. The lamella sample (L3), given and discussed in chapter four, was used here for comparison.
Figure 6.2 Focused ion beam process, (a) selected area of interest, (b) Pt layer deposited onto area (c) milling both sides of the samples to produce a trench, (d) left out the sample, (e) and (f) bright and dark field STEM image after final thinning and cleaning process.

6.3 Results

The pearlitic steel material was deformed by severe plastic deformation caused by torsional shear stress applied parallel to the flat surface of the sample disc. The shear plane is considered as a tangential plane with respect to the sample surface, centered at the axis of the rotation of anvil. As the result of a combination of large shear and compressive loadings, the thickness of the samples decreased from the original value of 1.5 mm to a thickness between 1.1-0.7 mm, depending on the loading conditions applied.

6.3.1 Optical metallography

Figure 6.3 a-i shows the circular cross-sectional area of the samples processed by HPT, taken from the edge of the disc where that maximum shear strain occurred. Observation showed that there is a distinguished white layer formed at a radius of R=5 mm of the sample. The white etching layer formed at the top layer of heavily deformed pearlite region. The cracks were also observed within the white etching layer. The white appearance revealed under a light optical microscope was due to resistance to etching, similar to the white etching layer generated from the thermal process – thermally induced
martensite. The results of HPT tests confirm that the accumulation of plastic deformation via cyclic wheel/rail impact can result in WELs on the rail surface.

Figure 6.3 also indicates that the thicknesses of white layers are highly dependent on the load/strain applied to the samples. Microstructure evolution of the lamellar pearlitic structure underneath WELs was also clearly observed, showing a different level of fragmentations and refinements of perlite grains, corresponding to the changes of torsional straining condition. It appears that the torsional strain (number of revolution) has a much stronger influence on the microstructure evolution than the compression load (pressure). Deformation characterized by fragmentation of cementite and ferrite was seen on samples subjected to only one revolution at 2, 4 and 6 GPa hydrostatic pressures, as shown in Figures 6.3a, b, and c. The non-uniformly distributed microstructure refining was observed after applying the second cycle of torsional loading, as shown in Figures 6.3 d, e and f. 6 revolution deformation has resulted in the complete disappearance of pearlite structure, as seen in Figures 6.3 g, h and i. The microstructure is really fine and uniform. The white etching layers are clearly seen at the surface of the samples, the nature of these white etching layers is considered to be oversaturated F-C solid solution, which is the consequence of cementite decomposition resulted from severe plastic deformation. This oversaturated Fe-C solid solution has a similar feature with thermally induced martensite.
Figure 6.3 Optical micrograph of HPT samples, taken at the edge of the circular section of the samples, showing formation of WELs at the top layer caused by severe plastic deformation at different loading condition. (a) n=1, P=2 GPa, (b) n=1, P=4 GPa, (c) n=1, P=6 GPa, (d) n=2, P=2 GPa, (e) n=2, P=4 GPa, (f) n=2, P=6 GPa, (g) n=5, P=2 GPa, (h) n=5, P=4 GPa, (i) n=5, P=6 GPa.

Figure 6.4 shows the typical structure gradient of HPT tested sample at radius=5 mm (test condition of n=2 turns and P=2 GPa). Four zones were observed in sequence from the center of the sample: undeformed, elongated, severely deformed and transformed region (cementite decomposed region - WEL). The undeformed region reveals pearlite structure (matrix region) and these become elongated parallel to the shear plane with increasing the radius from the center. With increasing the radius toward the maximum value (R=5 mm), there was a heavy sheared pearlite zone, followed by a fully transformed region (WEL). The increase of distance from the center corresponds to an increase of shear strain and therefore the white etching regions correspond to higher degrees of a strain than the dark etching zones (deformed region).
Figure 6.4 Optical micrographs of the surface of the samples observed at the edge (radius=5 mm) of the circular section, under test condition of n=2 turns and P=2 GPa, showing the transformation levels of pearlite structure due to shear strain.

Figure 6.5 shows an example of micro-hardness measurement over the deformed and WEL regions. The white region revealed the highest hardness value, reached about 790 HV, which is consistent with the hardness level of WELs formed on rail track by the deformation (Zhou et al. 2016). At the heavy deformed region, the micro-hardness values were varied between 403-560 HV, showing an increase of hardness at the region with high-density white strips, which is related to WEL.

Figure 6.5 shows the distribution of hardness over the white and deformed region for the sample under test condition of n=5 turn and P=6 GPa.
Furthermore, microstructural observation revealed that cracks were formation at the WEL edges, extending into the deformed region, as shown in Figure 6.6. The formation of cracks in this region was certainly caused by high tangential force applied parallel to the flat surface of the cylindrical sample, and most likely due to the differences in the stress distribution and mechanical properties between WELs and their adjacent regions.

Figure 6.6 Optical micrograph showing cracks formation at the edge of WELs (the observed plane is the edge of the circular section of the sample with the test condition of n=5 turn and P=6 GPa).

6.3.2 SEM observation

SEM observation provides further detailed information on the evolution of the pearlitic lamella structure. In the initial deformation stage, Figure 6.7 a, the cementite lamella structure at the deformed region were fractured, and partly formed a long ribbon and rod-shaped particles. However, part of the pearlitic lamella was still present in the structure. With increasing the amount of deformation, the fragmented cementite lamellas were severely refined into spherical shaped particles inside the ferrite matrix, with an average diameter of 100 nm, Figure 6.7b. The light regions are cementite particles whereas the dark area is the ferrite matrix. With the highest shear strain, the cementite particles were almost fully decomposed forming a supersaturated Fe-C alloy, Figure 6.7c. The microstructure here becomes very fine with the highly distorted crystal structure, which in turn, results in high hardness and low ductility.
The microstructure evolution with increasing shear strain was depending on the orientation of pearlite lamella in relation to the shear direction. In addition, the lamella spacing was also affected by the shear plane. Figure 6.8 a, b shows the deformed pearlite lamella at large and low angle with shear direction, respectively. It is evident that under the large angle of strain, pearlite lamellas were bent, and part of cementite lamella was fractured and some of them re-oriented along the direction of shear strain. For the lamellar structure having a small angle with the shear direction, the pearlite lamella was stretched (from the deformation of ferrite) and became thinner, leading to a reduction in the lamellar spacing compared with those oriented perpendiculars to the shear direction.

Figure 6.7 Characterisation lamella structure based on the deformation level (test condition of n=5 turn and P=6 GPa), (a) Pearlite lamella with some broken cementite, (b) the cementite spheroidisation due to an increase deformation level, (c) the cementite was almost fully dissolved at the region of high shear strain (WEL).
Figure 6.8 Microstructure evolution based on a lamellar structure with respect to the shear direction (test condition of \(n=5\) turn and \(P=6\) GPa), at (a) large angle (perpendicular) (b) small angle (parallel).

### 6.3.3 X-ray analysis

Figure 6.9 shows the X-ray diffraction pattern of all the samples examined. It can be seen that the intensities of cementite diffraction peaks gradually decreased with increasing shear and compressive load. At 5 turns under all pressure conditions, the cementite peaks were barely visible in the diffraction pattern indicating that the majority of cementite was decomposed. The X-ray results are consistent with the microstructure observation of the samples subjected to different shear strain. The results confirmed that the cementite lamella has decomposed under heavy shear strain. In addition, it was observed that the reflection spectrum near (200) \(\alpha\), particularly at maximum pressure and shear deformation revolution, displays additional (200) \(\alpha'\) and (002) \(\alpha'\) martensite peak profiles matching the asymmetric WEL peak.

Figure 6.9 X-ray patterns showing cementite peaks decreased with increasing loads.
6.3.3 TEM characterisation

As mentioned previously, the sample subjected to the hydrostatic pressure of 6 GPa and 5 revolutions was selected for detailed TEM characterisation. A comparison was made between WEL generated by HPT and WEL induced by rolling contact on the rail track, aiming to reveal whether they have similar microstructure characteristics. Based on the degree of deformation, an observation was conducted at different regions in the sequences of the elongated region, deformed region, transition zone, and WELs region.

- **The elongated region**

Figure 6.10 shows a bright-field TEM micrograph with corresponding selected area electron diffraction (SAED) pattern, taken from the elongated pearlite region. In this area, the pearlite lamellas were bent and rolled with the direction of shear strain. As a result, part of cementite lamella was broken and fractured into small layers. In addition, dislocations in the ferrite lamellas were observed, showing various densities in the different thin ferrite films, indicating that the deformation was not homogenous. The region containing a high density of dislocations was found in the middle of lamella where the ferrite layer was deformed in the shear direction. The corresponding SAED pattern shows a single ferrite crystal orientation with a zone axis near to \{001\} ferrite (bcc) plane.

![Figure 6.10](image)

Figure 6.10 TEM micrograph with its corresponding SEAD taken at the elongated area, showing pearlite lamellas have rolled with shear direction resulting in broken perlite lamella and dislocation density in ferrite layers.
• **The deformed region**

Figure 6.11 a, b and c, d shows the bright field (BF), darkfield (DF) and corresponding SAED imaging in TEM, taken at the deformed region of HPT sample and ex-service rail sample, respectively. From (BF) image, both HPT and rail samples exhibit similar microstructure characteristics and deformation levels, in which the pearlite lamellas were severely deformed, resulting in a massive reduction in the lamellar spacing. The cementite layers, originally lamellar-shaped, were severely fractured into very fine particles in nano-sized grains ranging between 10-30 nm. Some cementite particles were decomposed and the other remained in the microstructure as fine fragmentations having random crystal ordinations, as shown brighter in the DF micrograph, obtained using matrix reflection. The ferrite layers in the microstructure were found with high dislocation density. The corresponding SAED image, taken over the deformed area shows diffraction ring pattern with a concentric ring of spots, indicating to small-grained structure and the occurrences of high-angle of misorientation induced between grains. It can be seen that the intensity of ring patterns is not regular, most likely related to the random crystal orientation of cementite. The diffraction spots of cementite were weak but still present in the diffraction pattern.
Figure 6.11 Bright and dark field with corresponding (SAED) TEM image showing the characterisation of the deformed region of (a) BF (b) DF of HPT sample (using cementite reflection; (c) BF and (d) DF of Rail sample using cementite reflection).

- **The transition zone**

Figure 6.12 shows bright-field (BF) TEM micrographs with SAED pattern, obtained from the transition zone between WEL and severely deformed region of HPT processed sample and rail sample, respectively. The microstructure evolution with increasing the deformation level toward the large radius of the HPT sample revealed similar characteristics of that process on the rail surface due to rolling contact. It can be seen that the lamellar structure characterised by heavily deformed pearlite gradually disappeared. The pearlite lamellas near the interface boundary (indicated by the dashed line) were aligned parallel to the shear direction. Small fragments of cementite lamella were
observed in the deformed region but almost disappeared in the WEL region. The area of WEL does not exhibit a uniform structure; the only structureless features were observed. The SAED taken at deformed region shows a slightly different diffraction pattern between the two adjacent deformed regions and WELs regions, in which the ring pattern in the diffraction of the deformed area appeared discontinuous containing discrete reflections, compared with much continuous angular distribution of spots in the diffraction pattern of WELs. This is due to the fine-grained polycrystalline structure of WELs.

Figure 6.12 TEM micrograph and corresponding SAED image showing the characterisation of the transition zone obtained from (a) HPT sample, (b) Rail sample.

- **White Etching layer**

The BF, DF and corresponding SAED TEM images in Figure 6.13 a,b characterise WEL that is induced by severe plastic deformation using HPT. From the BF image, it can be seen that WEL consisted of two different regions, the dark gray region at the upper right corner of the image and the light gray at the lower-left corner of the image. These regions represent different levels of plastic deformation. The upper dark region comprised an extremely fine-grained structure while the light region consisted of a relatively much
coarse-grained structure. From the DF image in Figure 6.13b (excited from Fe₃C and α’) the variation in the grain size refinement between these regions can be clearly seen, with no evidence of lamellar structure. The diffraction pattern of the selected area was collected from these regions. The SAED from the dark region shows a much continuous intensity of rings pattern, confirming a characteristic of polycrystalline structure with a heavy refinement of grains. The SAED from light is composed of concentric rings with discontinuous angular distribution of spots in the diffraction pattern, indicating that the grains are randomly oriented. In addition, the experimental profile was simulated from the SAED pattern and plotted. It can be seen that the peak profile of (200) has asymmetrically shaped reflection due to overlapping the peaks of (002)α’ and (200)α, associated with a distortion of the ferrite into tetragonal martensite (bct) phase structure. Furthermore, the SAED images of the two considered regions do not show any diffraction spots of cementite in the diffraction patterns, which indicates that the cementite lamella was fully decomposed, resulting in the formation of a supersaturated Fe-C alloy.

Figure 6.14a shows the STEM micrograph taken from the layer at the maximum radius (R=5 mm) which the shear strain is considered to be at the highest level. The microstructure here does not exhibit any clear crystalline structure because of an extremely grain size refinement. The further detailed crystal structure can be explained by high-resolution scanning TEM imaging (HRTEM), shown by the inverse FFT image in Figure 6.14 b. It can be seen that the lattice structure was extremely destroyed. The high magnified inverse FFT image of the selected area (A) shows that the fringes lattice are bent and severely deformed. In region B, the lattice structure contains edge dislocation defects which are caused by large microstrain induced in the lattice due to severe plastic deformation.
Figure 6.13 Bright and dark field with corresponding (SAED) TEM image showing the characterisation of the WEL processed by HPT (a) BF (b) DF (using cementite and α’reflections).

Figure 6.14 Bright and dark field (SAED) TEM with corresponding inverse FFT HRTEM image showing the characterisation of the WEL at the topmost layer (R=5mm).
Figure 6.15 shows the characterisation of the WEL that was taken from the rail surface. From the TEM micrograph, the microstructure is composed of nano-sized grains with full non-existence of lamellar structure. The much darker area is considered the region of high dislocation density. The corresponding selected area electron diffraction (SAED) shows a ring pattern confirming that the microstructure consists of a polycrystalline grained structure. It can be seen that the pattern of the ring contains discrete reflections which indicate that the grains are randomly oriented. Diffraction spots of cementite were completely disappeared showing that the cementite particles were fully decomposed. In addition, the reflection profiles simulated from SAED pattern shows that a deviation was induced in the (200) ferrite peak, as indicated by the arrow, which indicates that a tetragonal distortion was induced in the lattice structure.

The characteristics shown in this the WEL taken from the rail surface and caused by rolling contact is much consistent with that WEL being processed by severe plastic deformation using HPT. Therefore, it strongly suggests that mechanics formation of WEL is due to severe plastic deformation caused by rolling/sliding contact loading combined with compressive axial loading induced by the wheels on the rails.

Figure 6.15 TEM micrograph with corresponding SAED and simulated reflection profile, showing the characterisation of the WEL that is taken from rail surface.
6.4 Discussion

6.4.1 White etching layer induced by plastic deformation

An experimental investigation was carried out on the pearlitic rail steel material to determine whether white etching layers can be induced by severe plastic deformation. Large deformation was generated by applying a combination of a considerable shear strain along with large hydrostatic compressive pressure using high-pressure torsion (HPT). This technique was considered because the loading conditions are entirely similar to that induced on the rolling contact zone due to wheel/rail interaction. During the experiment, different range of shear strain corresponding to 1, 2 and 5 revolutions, and uniaxial compressive loads of 2, 4 and 5 Gpa was used at room temperature (t=20 ± 3 °C), in order to examine the microstructural evolution of the material with increasing strain and stress.

The microstructural investigation results confirmed that a distinguished white layer, as reflected under light microscopy due to the resistance to etching, was formed at the edge of the samples (the highest distance from the center of the sample, R=5 mm), where the shear strain is considered to be at the maximum level. Formation of a white layer on the pearlitic rail steel material subjected to similar wheel/rail loading conditions provides strong evidence that the white etching layer formed on the rail surface was induced by severe plastic deformation. Detailed information that supports the formation mechanism of WELs is discussed by the following findings.

The results showed that WELs processed by HPT revealed the highest hardness values, which is consistent with the hardness level of WELs that is thermally induced by phase transformation mechanism. The increase in the hardness level is mainly due to the nanocrystalline structure caused by severe plastic deformation. From microhardness measurement, significant variation was observed between WEL structure and the plastic deformation layer beneath. Along with the large difference in the grain size refinement, the yield stress in such regions became significantly varied. Large variations in the yield stress lead to crack formation. The characteristic of crack formation was shown in Figure 6.6.

Under the shearing strain and large axial load, the pearlitic steel is deformed, and the deformation level increased with increasing the distance from the center of the samples,
certainly associated with increasing the shear straining. Based on the deformation degree, three regions associated with different deformation levels were recognised at the surrounding area of the samples. These regions were found as a function of distance in the sequence: elongated, deformed and WEL. In the elongated region, the pearlite lamella was slightly deformed and in this region, dislocation density was observed in the ferrite lamella, as well as the cementite lamella was initially broken. Toward the large radius, the deformation level increased by increasing the shearing strain, which leads to that the pearlite lamella was strongly forced to bend and roll with the shear plane. In this situation, the cementite lamella was severely fractured resulting in fine particles distributed randomly in the texture. With further distance, the lamella structure was flattened parallel with the shear direction resulting in a significant reduction in the lamella spacing which was the important factor of the increased hardness. At the maximum distance (R=5) where the largest shear stain was located, the lamella structure was fully disappeared and replaced by a structure containing nanocrystalline grains.

The microstructural observation also showed that the microstructure evolution level was associated with the crystal orientation of pearlite lamella in relation to the shear direction. Large angle of strain resulted in severely broken of praline lamella while the small angle of strain leads to much thinning, stretching and elongating of lamella structure with the shear direction. This results in inhomogeneous deformation, and white etching layers form first in the lamellas that are oriented in parallel with shear direction (Ivanisenko et al. 2003). This characterisation can be seen in Figure 6.5, as white layers having high hardness values were embedded between the sheared grains and likely related to pre-formed WEL.

The configuration of the pearlitic steel material after being severely deformed by applying loads that exceeded the certain limit of yield stress was much similar to that of the rail surface layer containing WELs, in which the microstructure evolution plastic flow throughout the surface layer.

The theory of white etching layer formation by plastic deformation has been the subject of extensive researches. However, arguments on the mechanism formation of WELs do exist. The finding in this work provides evidence that WELs are the result of ongoing severe plastic deformation.
6.4.2 Structural and microstructural evolution by plastic deformation

Based on the structural and microstructural observation results, the white etching layer simulated by severe plastic deformation was found to have a nanocrystalline martensite structure. X-ray results combined with TEM observation confirmed a decomposition of 200 reflections into two sub-peaks which indicates that a deviation from bcc ferrite lattice was induced due to a tetragonal distortion. The lattice parameters a and c of the tetragonal crystal structure were quantitatively calculated based on Bragg’s law. Figure 6.16 shows the evolution of lattice parameters as a function of shear strain at different compressive loads. For all conditions applied, splitting between the lattice parameters a and c was clearly observed corresponding to a tetragonal martensitic structure with two different axes a and c. In addition, differences between the lattice parameters values significantly increased with an increasing number of turns as well as increasing the compressive load. For example, c/a ratio jumped from 1.002 at n=1 to 1.007 at n=5 under P=2 Gpa. Similarly, c/a increased from 1.007 at P=2 Gpa to 1.014 at P=6 Gpa under the same shear straining of n=5 turns. This indicates that both shear strain and compressive pressure contribute to the formation of a tetragonal distortion structure.

Figure 6.16 lattice parameters a and c as a function of shear strain at different load conditions, showing structural evolution of the deformed pearlite structure corresponding to a tetragonal distortion induced in the ferrite structure.
In order to provide a clear understanding the origin of the tetragonal distortion of the WEL structure, the carbon concentration was quantitatively calculated based on the relationship with the distorted lattice parameters, by using the formula of \( (c/a=1+0.045C) \), where \( c \) is the percentage of carbon content (wt\%) (Roberts 1953; Roitburd & Kurdjumov 1979). Figure 6.17 shows the calculated carbon content as a function of shear strain for different load conditions. The results showed that the percentage of carbon content increased significantly with increasing shear straining. Similarly, increasing compressive stress resulted in increasing the percentage of carbon concentration. This indicates that much carbon content was released into the ferrite matrix at \( n=5 \) and \( P=6 \) Gpa. Nevertheless, the parentage of carbon concentration dissolved in the WEL, processed under different conditions, was exceeded the limit of carbon C in the original material. It means that the WEL structure contained a supersaturated Fe-C alloy. As a result, the evolution of carbon content in the ferrite during the plastic deformation process was completely attributed to tetragonal distortion of the lattice structure, which eventually leads to the formation of a supersaturated Fe-C alloy. This observation on the material caused by severe plastic deformation is generally found from quenching martensite by phase transformation process (Djaziri et al. 2016). Therefore, the structural and microstructural analysis confirmed the martensitic structure nature of WEL being driven by severe plastic deformation.

![Figure 6.17 carbon concentrations as a function of shear strain.](image)

Figure 6.17 carbon concentrations as a function of shear strain.
Detailed information microstructural composition of WEL is shown by EDS mapping in Figure 6.18. The scanning was obtained from WEL, at the area containing a mixture of fine- and coarse-grained structures corresponding to different deformation levels. Variation in the composition distribution of such regions was clearly visible. From the C-K map, the carbon content appeared much uniform over the whole WEL region, indicating that the cementite lamella of the matrix pearlite was fully decomposed. For components distribution of oxygen (O-K), silicon (Si-K), manganese (Mn-K) and iron (Fe-K) across fine- and coarse-grained zones were distinct. The fine WEL region contained much more oxygen, manganese content than the coarse region while it had a less concentration of iron. Form Si map, a slight difference in silicon distribution were observed among such zones, although some segregated small regions having burble colour contrast was observed in the fine-grained zone, likely related to pre-existed defects. Variations in the elements between the two regions having different grain size refinement confirmed that severe plastic deformation could influence the weight composition of the material.

![Figure 6.18 EDS mapping showing the distribution of elements on WEL having two different grain size refinement corresponding to different deformation level.](image)
In order to further confirm that some type of WELs on the rail surface is caused by severe plastic deformation, a comparison was made between the WEL induced by severe plastic deformation from the HPT experiment and WEL obtained from the ex-service damaged rail. The observation was focused throughout WEL, transition zone and underneath subsurface layer. Results from TEM characterisation confirmed that the microstructural evolution in the mentioned regions of both WELs was highly consistent. Different deformation level was observed as a function of depth and WEL considered the area of the largest deformation level. From the deformed area up to WEL, the cementite lamella was gradually fractured and dissolved in the ferrite lamella, which leads to the formation of a nanocrystalline structure.

6.5 Conclusion

Formation mechanics of white etching layer from severe plastic deformation processed by high-pressure torsion technique on the pearlitic rail steel material was investigated. Detailed information structural and microstructural evolution of the material by severe plastic deformation were discussed and characterised by optical, scanning and transmission electron microscopy, and by conventional X-ray diffraction. A comparison with the white etching layer taken from the ex-service damaged rail was made. The results

1. White etching layer was induced by severe plastic deformation.
2. White etching layer was formed on the area of the highest degree deformation level.
3. WEL processed by severe plastic deformation is characterised by the high level of hardness, which is consistent with those induced by the phase transformation process.
4. The degree of deformation was associated with the crystallographic orientation of the pearlite lamella which means that WEL is formed faster when the shear direction is parallel to crystal ordination.
5. The structure of WEL was confirmed to contain a nanocrystalline martensite structure.
6. Martensite nature of WEL was confirmed by the formation of supersaturated Fe-C alloy, which was caused by the dissolution of carbon content, corresponding to
cementite lamella, in the ferrite grains linked observation of tetragonal distortion in the ferrite lattice structure.

7. The microstructural evolution of the WEL taken from ex-service damaged rail is consistent with the WEL induced by severe plastic deformation in the lab, which provides clear evidence that some type of WEL formed on the rail surface is caused by severe plastic deformation caused by rolling/sliding contact loading.

8. WELs formation resulted in a significant amount of anisotropy characteristic of the material evolved, which in turn, leads to crack initiation.
7. Mechanism of WEL Formation: Thermomechanically Induced

7.1 Introduction

Mechanics formation of the white etching layer on the rail steel material has been confirmed via two different routes. The first one was thermal-driven WEL by phase transformation and the second one was mechanical-driven WEL by severe plastic deformation. In this chapter, an experimental study was performed to examine the potential formation of the white etching layer (WEL) on rails by applying a combination of thermal and mechanical processes. The purpose was to explore whether a martensitic WEL can be formed on the surface of the steel after heating under pressure rail to temperatures below the critical pearlite-austenite transformation points.

In rail and wheel, the contact condition is very complicated. Large normal contact pressure, exceeded the limit of the yield stress of the material, is applied on the rail by the wheels during rolling contact. Along with the compressive load, heat is generated at the wheel/rail contact zone during train passage, resulting in a temperature rise of the material. The combination of heat and pressure results in characteristic surface structural evolution. White etching layer (WEL) is considered as one of the most particular structural phenomena that occurred in the rail surface layer. The occurrence of WELs on the rail surface layer is linked with the formation of the cracks and squats origin (as discussed in chapter 3). Therefore, to provide a clear understanding of the microstructural evolution after applying heat and pressure, an experimental study was carried on rail steel material.

7.2 Experimental Details

In order to determine the effect of the heating process combined with the high contact pressure generated by locomotive vehicles on the rail surface, an experimental study has been carried out using a Gleeble 3500 thermo-mechanical simulation system. The specimens used in this testing work were cut from an ex-service rail steel material with the chemical composition and mechanical properties shown in Table 7.1. The specimens were machined to dimensions of 20 mm in length, 15 mm width and 8 mm
thickness, and then set up in the Gleeble machine as shown in Figure 7.1(a). A schematic illustration of the test method is presented in Figure 7.1(b).

Table 7.1: Chemical composition and mechanical properties of the tested specimens.

<table>
<thead>
<tr>
<th>C, %</th>
<th>Mn, %</th>
<th>Si, %</th>
<th>Tensile strength, Mpa</th>
<th>Elongation, %</th>
<th>Hardness, HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.78%</td>
<td>0.9%</td>
<td>0.2%</td>
<td>920</td>
<td>11%</td>
<td>280-300</td>
</tr>
</tbody>
</table>

Figure 7.1: (a) Setting up the specimen in the Gleeble machine during the test, (b) schematic of the test method.

Initially, a sample was heated to a specified temperature with a heating rate of 200°C/sec. The temperatures used were 450, 500, 550, 600, 675, 730 and 850°C. When the temperature reached its specified limit, a 1 GPa pressure was applied and then the sample was cooled rapidly (by a quenching process). After the testing was complete, each sample was metallographically prepared by mounting in resin, grinding with 1200 grit, polishing to a 1 μm diamond finish and etching with 2.5% Nital. The etched surface of the samples was characterized by optical microscopy, followed by a micro-indentation hardness measurement performed on the etched surfaces using a Vickers diamond with a 500g load.
High-resolution synchrotron X-ray experiments were performed using the Powder Diffraction (PD) beamline at Australian Synchrotron, Victoria, Australia. The beamline was used in glancing incidence geometry with a monochromatic X-ray beam energy 15 keV (\(\lambda=0.8265\) Å). Phase analysis was performed in-range between 20° and 100°. Depending on the sample size and the thickness of WELs, different beam size of 0.1 x 0.5 mm, 0.2 x 0.5 mm (Vertical x Horizontal) and glancing angles (\(\omega\)) of 0.2°, 0.5°, 1° and 2° were used for the acquisition of diffraction data (Bragg angle \(20° \leq 2\theta \leq 100°\)) using the Mythen detector with exposure time in the range of 0.5-8 min. The quantitative analysis was performed of each diffraction pattern in order to identify the phases in each sample. The content of retained austenite (RA) was quantitatively determined by the calculating the integrated intensity of (110), (200) and (211) peaks of ferrite/martensite and the (111), (200) and (220) peaks of austenite. The volume fraction of austenite was calculated using the formulas provided by the ASTM standard.

Thin WELs lamellas were prepared by Focus Ion Beam (FIB) milling using FEI Helios NanoLab G3 CX instrument, and then characterised by Transmission Electron Microscopy (TEM) using JEOL JEM-ARM200F with accelerating voltage of 200 kV, working with selected area electron diffraction (SAED), conventional bright-field (BF) and dark field (DF), as well as high resolution transmission electron microscopy (HRTEM). In addition, the elemental analysis was performed using Energy Dispersive X-ray Spectroscopy (EDS) in the TEM and STEM mode. Figure 7.2 shows FIB process for sample preparation.
Figure 7.2 FIB process for TEM sample preparation. (a) The selected area of interest contacting martensite and perlite colony, (b) Pt layer deposited onto area (c) milling both sides of the samples to produce a trench, (d) left out the sample, (e) and (f) bright and dark field STEM image after final thinning and cleaning process.

7.3 Results

7.3.1 Microstructural evolution

The polished samples were observed under optical and scanning electron microscopes. Figure 7.3a shows a typical example of the microstructure evolution in the contact area. It can be seen that a small proportion of the surface material (contact area) where the 1 Gpa pressure was applied with heating at a temperature below the transformation point and rapidly cooled displays different microstructure characteristics than the bulk material. However, the non-contact area of this sample that was heated to the same temperature did not transform and remained a pearlitic microstructure. This is a strong confirmation that the pressure applied with heating plays a significant role in the phase transformation of rail steel. In other words, the critical point of hypoeutectoid steel (the critical point for phase transformation) may reduce when pressure is combined with heating. As a result, on rapid cooling, a martensitic white etching layer (WEL) can be formed below the normal transformation temperature.

The microstructure of rail steel obtained from the contact area, close to the surface of the samples that were heated to 450, 550, 600, 675, 730 and 850 °C, with application of 1 GPa pressure at the same time (for approximately 2 seconds) before they were rapidly cooled, are shown in Figure 7.3 b-g. As can be seen for quenching at 450 and 500 °C, a
pearlite microstructure, which is a major constituent of this near eutectoid steel, is predominant and comprises a fine combination of ferrite and iron carbide (cementite). As a result, it is apparent that the applied pressure did not change the rail steel structure within these specified temperatures.

On the other hand, when the heating temperature increased to 550 °C, the pearlite phase microstructure of the tested rail steel under the contact area changed to one consistent with a transformed martensite microstructure, as illustrated in Figure 7.3 d. With increasing the temperature to 600 and 675 °C under the same amount of pressure, the volume fraction of the transformed area increased, as shown in Figure 7.3 e, f. Quenching at high temperature, 730 and 850 °C, the pearlite phase structure has fully transformed into a martensite phase structure, as shown in Figure 7.3 g, h. In addition, it appears that quenching at 850 °C results in a much finer structure than those quenched at 730 °C under the same pressure.
Contact area, 450 °C

Contact area, 500 °C

Contact area, 550 °C

Contact area, 600 °C

Contact area, 675 °C

Contact area, 730 °C

Contact area, 850 °C
Figure 7.3 Microstructural features of the tested samples obtained from contact area: (a) low magnified image showing the characteristic of contact area where the pressure (1 GPa) was applied on the samples being heated at temperature below the critical transformation point, (b), (c) microstructure of the samples being heated to 450 and 500 °C, showing pearlite and ferrite microstructure with the absence of martensite structure indicating that WEL was not formed under these conditions, and (d), (e), (f), (g) and (h) microstructure of the contact area of the samples tested at 550, 600, 730 and 850 °C, showing formation of martensite structure under the contact area where a pressure of 1 GPa was applied.

In addition, it was noticed that the depth of the transformed region underneath the contact load had increased considerably when the heating temperature was raised, but with the same pressure applied. This feature is illustrated in Figure 7.4 for specimens which were heated to 550, 600, 675 and 730°C. The characteristic shown may suggest that at different temperatures different depths of austenite formation occur, reflected in variations in thicknesses of the resulting martensitic layers. This result is comparable with other studies (Al-Juboori et al. 2017; Pal, Daniel, et al. 2012) that found different thicknesses of WEL within the same section of rail sample. It can be assumed that either inhomogeneous distributions of temperature during the heating process on the rail surface, or loss of the contact condition between wheel and rail during slipping, might generate these observed irregular thicknesses of WEL. This reasoning is consistent with previous studies (Pal, Daniel, et al. 2012; Pau et al. 2001) which concluded that surface roughness of rails plays a significant role in decreasing the contact area and minimizing the heat exchange between the wheel and rail surfaces.
Figure 7.4 Development in the martensitic thickness at the contact area associated with increasing heating temperature: (a) T=550°C, (b) T=675°C, and (c) T=730°C.

Another feature that was observed is that of a transition region, which can be clearly distinguished beneath the martensitic layer, as shown in Figure 7.5. This region formed in the specimens that were heated to 550, 600, 675 and 730°C. It comprises fine grains compared with those of the unaffected layer (bulk material). This interface layer is most likely to be a heat-affected zone where the combined effects of maximum temperature achieved, and local pressure was insufficient to result in a martensitic structure on cooling. However, the temperature was still adequate to produce a microstructural transformation in the material (Rebelo et al. 1998). The observed grain refinement at the heat affected zone can be attributed to a phase transformation from pearlite to austenite. However, the peak temperature is apparently, is likely significantly below the austenite grain coarsening temperature (~1200°C), and the resultant parent microstructure has a very fine grain size.
Generally, under equilibrium conditions, the eutectoid point for steel that contains 0.77% carbon is 727°C (Krauss 2015). However, if the composition of the steel includes alloying elements and impurities, the critical point for phase transformation from pearlite to austenite will be modified. Additionally, during the heating process, the transformation between phases does not entirely take place under constant temperature. Because of this, a certain temperature range of transformation temperature has been calculated theoretically using empirical formulae. These equations are presented by Trzaska and Dobrzański (Trzaska & Dobrzański 2007) and displayed below in Equations 1 and 2 for the lower and upper limits, respectively.

\[ A_{C1} = 739 - 22.8C - 6.8Mn + 18.2Si + 11.7Cr - 15Ni - 6.4Mo - 5V - 28Cu \] …(1)

\[ A_{C3} = 937.3 - 224.5 \sqrt{C} - 17Mn + 34Si + 11.7Cr - 14Ni - 21.6Mo - 41.8V - 20Cu \] …(2)
Although these equations might not provide accurate values for critical transformation temperatures, they are a convenient and effective approach for prediction calculations (Pawłowski 2011). Therefore, they have been utilised to determine quantitative estimations for the rail steel used in this experiment. The findings have revealed the minimum and maximum transformation limits of approximately 719°C and 731°C, respectively. On the other hand, as previously mentioned, the experiment has been conducted at 450, 500, 550, 675, 730 and 850°C, and the transformation has started at 550°C when a pressure has been applied. Consequently, if the results obtained from these tests are compared with those calculated theoretically, a significant difference is clearly exhibited. This implies that the applied stress on the rail samples (which usually exceeds a critical limit for the material) during rapid heating can significantly lower the pearlite to austenite transformation temperatures. Therefore, it is suggested that the white etching layer observed at the rail surface can form at relatively lower temperatures under conditions involving high surface pressures.

### 7.3.2 X-Ray Diffraction Analysis

In order to identify the primary phase present in the transformed region, High-resolution Synchrotron X-ray diffraction pattern was generated from the samples that revealed a different structure. Figure 7.6 shows the X-ray diffraction profiles between the diffraction angle $2\theta < 50$, of a sample in two different regions. The first pattern was obtained in the contact region where the phase structure is completely transformed, and the second pattern was produced from bulk material (unaffected region). As may be seen, the peaks of the diffraction profiles are broad and asymmetric, which indicates that two phases are present in the crystal structure. This is caused by differences in grain size and refinement. It is apparent that the peaks of the contact region have moved toward a low angle direction, which is most likely to represent a martensitic structure. The peaks intensity of cementite (Fe3C) appeared clearly in the diffraction pattern of bulk material that has significantly decreased in the diffraction pattern of the transformed region, indicating that the cementite lamellae have partially or completely dissolved. A barely visible peak of austenite $(200)\gamma$ was present in the diffraction pattern of the contact zone, which indicates that some retained austenite particles remained in the structure and have not fully transformed into martensite.
In addition, the peak profile of 200 plan was enlarged for more detailed information. It can be observed that the peak profile of the contact zone appears to have an asymmetric shape, which indicates that the 200 peak of ferrite has split into 002/200 peaks associated with a distortion of ferrite into tetragonal martensite. This result is consistent with other studies (Österle et al. 2001; Wang, L. et al. 2003) which analysed the WEL using X-ray and synchrotron X-ray diffraction.

Figure 7.6 Synchrotron X-ray diffraction profiles obtained from the bulk material (unaffected region) and from the contact area. The Ferrite pattern was produced from the
region where the pressure was not applied (away from contact area), and the Martensite peak profile was obtained from the contact area where the pressure was performed on the sample. The enlarged image of the 200 reflection shows overlapping of the 002/200 peaks in the diffraction pattern obtained from the contact region, confirming a transformation from bcc ferrite into a tetragonal distortion of martensite structure.

7.3.2 TEM characterisation

The microstructure of the transformed area at the contact and around regions was characterised by TEM and STEM observation. Figure 7.8 shows typical examples of the microstructure characteristic taken from the transition zone, at the interface layer between the transformed region and the matrix. It can be seen that the cementite layers that appeared darker in the texture have partially or fully dissolved. Figure 7.7a shows that there are three different regions in the microstructure. Region I contained a uniform oriented pearlite lamella, region II shows partial dissolution of cementite with different shaped crystals and random orientation, region III that appeared darker is characterised by a high density of dislocation. The occurrence of these regions in the microstructure reflects the mechanism of transformation and formation process. The existing dislocation density at the interface boundary is likely due to the local strain induced in the lattice structure during the transformation process, and the applied pressure played a significant role. From Figure 7.7 d, a sharped transition between the pearlite lamella (matrix) and fully transformed martensitic region can be clearly recognized. Partial dissolution area containing cementite particles with darkened small spherical regions was observed at the boundary layer. The extended zone (fully transformed area) does not exhibit darker regions related to cementite which indicates that the cementite particles have fully dissolved.

Figure 7.8 shows the bright field (BF) and dark field (DF) TEM images and corresponding selected area of electron diffraction (SAED) pattern of the area obtained from the contact zone. From the BF micrograph, fine plate martensite with twinned structure can be clearly recognized. The SAED pattern confirmed the formation of the (112) type twins which is considered the most popular twin structure in the bcc. The diffraction spots from the bcc matrix are indicated by yellow lines in the SAED pattern while the diffraction spots from the matrix are indicated by white lines. The SAED pattern also shows that additional diffraction spots were observed at 1/3 and 2/3 (211) plan, as indicated by arrows. By using
the diffraction spot 1 (df1), taken at 1/3 (211) plan, the (DF) image in Figure 7.8 b shows the existence of nano-sized particles with a bright contest appearance, located between twins. These particles were indexed as carbide perceptions (Fe3C). In addition, the orientation relationship between twins and matrix structure can be seen in the DF micrographs imaged, Figure 7.8 c and d, after using the diffraction spots 2 and 3, as indicated in the SAED pattern.

Figure 7.7 TEM micrographs showing the characteristic feature at the transition zone between matrix (ferrite) and the transformed region (martensite). (a) (b) and (c) shows partial dissolution of cementite lamella, forming spherical and rod-shaped small crystals (darkened region) inside the ferrite matrix (light region), and (d) shows the sharped transition zone between ferrite and martensite layer with confirming the full dissolution of cementite particles in the martensite region.
Figure 7.8 (a) bright-field (BF) TEM micrograph with corresponding SAED pattern showing a typical example of the microstructure of the contact area after quenching under pressure (b), (c) and (d) dark field (DF) TEM micrograph of the corresponding diffraction spots 1, 2 and 3 shown in SAED pattern.

Figure 7.9 shows STEM micrographs with high-resolution HRSTEM imaging and corresponding fast Fourier transform (FFT) pattern, which characterises the microstructure of transmitted area (contact zone). It can be seen that an isolated pearlite island, having a non-uniform arrangement was observed inside the martensite region, Figure 7.9 a. Indeed, the formation of pearlite from austenite is associated with a very slow cooling rate. However, the cooling rate used to produce martensite under large pressure was very fast. Therefore, the existence of pearlite indicates that initial pearlite
has partially transformed into austenite and martensite formed due to subsequent fast quenching at a cooling rate of about 200 °C/s. Figure 7.9 b and c show high magnification images of the selected area, taken from the pearlite grain boundary.

Figure 7.9d and e are HRSTEM images with its FFT and inverse FFT pattern of the selected area, outlined by the small square in Figure 7.9 c. It can be seen that the fringers of the lattice structure in some regions, as indicated by circles, were not straight but slightly deformed, as well as edge dislocation defects were observed in the lattice structure, as indicated by the arrow. The deformed fringes and dislocation density existence are possibly caused by microstrain, induced in the lattice structure by large pressure applied. In addition, the HRSTEM confirmed that there were two different phases in the lattice structure: ferrite and austenite. The FFT diffractogram in Figure 7.9 f reveals a parallel orientation relationship between the [111] zone axis of the bcc crystal of ferrite to the [110] fcc zone axis of austenite. Form the enlarged inverse FFT images and their corresponding diffraction pattern in figures 10.7 g and h, the lattice structure of ferrite and austenite can be clearly recognized.

The results from TEM observation are totally consistent with the results obtained from Synchrotron diffraction.
7.3.3 Hardness measurement

A determination of the micro-hardness was made simultaneously on each of the samples after their structure had been revealed by microscopy. The hardness test was performed to complete the identification of the WELs, Figure 7.10 displays the changes in hardness profile measured across the surface where the load is applied. As can be seen, the increase in hardness values has significantly begun with heating to a temperature of 550°C, while very slight changes were observed with samples heated to 450 and 500°C. In addition, it is readily apparent that the hardness profiles of specimens that were quenched at 600, 675
and 730°C, have plunged sharply from about 800 to 450 HV. These results show that there were three distinct layers formed: these being martensite, a transition region, and an unaffected region. However, only one region was observed for the sample tested at a higher temperature (850°C). This indicates that the microstructure of this sample has become fully martensitic. As a result, the measurement of the changes in hardness formed by quenching was of considerable significance, in that it confirmed and explained some of the features observed in the microstructure.

![Hardness Profiles](image)

Figure 7.10 Micro-hardness profiles for all samples being heated to 450, 500, 550, 600, 675, 730 and 830°C. The hardness was measured across the surface where the pressure was applied.

### 7.4 Discussion

An experimental study was conducted on the rail steel material in order to examine the mechanism of the formation of the so-called white etching layer. The thermomechanical process was performed in the lab by using a Gleeble 3500 thermo-mechanical simulation system. This technique was used to simulate a loading condition that is similar to that induced on the rolling contact zone due to wheel/rail interaction. Large pressure, exceeding the material limit was applied to the samples at the same time with rapid heating and cooling application. The metallurgical examination results confirmed that
there was no influence of applied pressure on the microstructure change when the samples were heated up to 450 And 500 °C. However, the microstructural evolution corresponding with the formation of a distinguished white layer has begun with heating to a temperature of 550°C. The formation of a white layer at the contact zone with the application of low temperature (below the critical transformation point) strongly supports the transformation occurrences at lower temperatures with the assistance of pressure.

The required temperature for phase transformation without pressure application was quantitatively calculated based on the supplied formula given in equations (1) and (2). The results showed that based on the chemical composition of the material the rail steel would need to be heated about 719-731 °C to transform from pearlite into austenite which then transforms into martensite by the very fast cooling rate. In addition, the effect of applied pressure on the transformation temperature of rail steel material was predicted by using ThermoCalc software. Figure 7.11 shows the start BCC and the end FCC phase transformation temperature as a function of applied pressure. It appears that increasing pressure leads to a linear decrease in the eutectoid temperature. At pressure =1 Gpa, the minimum temperature for austenite to form is about 697 °C. Practically, the transformation temperature seems different from the prediction temperature. Martensite formed at a temperature below 600 °C under 1 Gpa.

![Figure 7.11 Prediction of transformation temperature under the influence of applied pressure.](image-url)
The structural and microstructural analysis confirmed a formation of martensitic layer by quenching of the material at a temperature below 600 °C with the application of 1 GPa pressure. The synchrotron radiation at the contact area of the sample tested at 600 °C shows a reflection pattern of martensite with the existence of retained austenite. Comparing with the reflection obtained from the unaffected area (away from contact zone), the reflection peaks were broadened and shifted toward the low angle, which is due to grain size refinement and micro-strain induced in the lattice structure. In addition, the peak intensity of cementite appeared clearly in the bulk material was barely visible in the diffraction pattern of contact area, indicating that the cementite lamellas have partially or completely decomposed.

From profile analysis, the 200 peak of contact zone revealed an asymmetric shaped pattern, confirming that a tetragonal distortion in the lattice structure from bcc into bct was induced. The lattice parameters a, c of the tetragonal crystal structure were quantitatively calculated based on Bragg’s law. It was found that the c/a ratio of the martensitic contact area reaches 1.009, indicating the formation of oversaturated Fe-C alloy. Accordingly, the solute carbon concentration in the martensite was determined by using the formula of \( \frac{c}{a} = 1 + 0.045C \), where c is the percentage of carbon content (wt%). It was found that about 20% of carbon concentration content was dissolved in the martensite. It indicates that more than a quarter of the carbon concentration of the rail steel material was dissolved in the contact region. As a result, the process of carbon dissolution in the material would not occur at a relatively low temperature without the assistance of high pressure.

The results from TEM observation are completely consistent with those obtained from synchrotron diffraction. Microstructural observation confirmed the martensitic structure nature of the contact region. High-resolution STEM (HRSTEM) imaging confirmed that dislocation density, as well as micro-strain, was induced in the crystal lattice structure, possibly due to the effect of pressure applied. In addition, the crystal structure of BCC and FCC were found in the lattice, having a parallel crystallographic orientation relationship, confirming the transformation from austenite into martensite.

The micro-indentation hardness distribution showed that the highest hardness values were measured at the martensitic region (contact area) while the hardness dropped significantly at bulk material and/or heat-affected zone, showing a sharp transition zone. The hardness
of the martensitic layer was about 720-820 HV, which is consistent with the hardness of rail WEL determined in chapter 3. In addition, it was found that the hardness values increased slightly with increasing the quenching temperature, which is due to the high distortion of the crystal lattice structure.

Comparing with the rail WEL that is discussed in chapters 3 and 4, the structural and microstructural characterisation seems almost similar, in which the microstructure of both rail WEL and simulated WEL has comprised predominately martensite with some retained austenite. In addition, the hardness was significantly consistent. On these bases, the observation strongly supports the formation mechanics of WEL by the contribution of thermal and mechanical processes.

7.6 Conclusion

In this study, thermo-mechanical simulation testing was carried out to explore the formation of a martensitic structure (white etching layer) in the rail steel material. The main results are summarised in the following:

1. Mechanics formation of the white etching layer is confirmed by the application of both thermal and mechanical processes on the rail steel material.

2. The white etching layer is caused by phase transformation from rapid quenching at a temperature below the critical point under an application of large pressure.

3. The thickness of white etching layer formation, the volume fraction of the transformed region and the grain size refinement at the contact zone were a function of quenching temperature rise.

4. The microstructure of WEL was predominately composed of martensite, with some isolated ferrite, some retained austenite, and very few nanoscale cementite particles.

5. Distortion of the crystal lattice structure, as well as dislocation density, were observed in the structure of the simulated WEL (contact region), which is caused by the application of large pressure applied.
6. About 20% of carbon concentration is dissolved in the martensite (simulated WEL) after quenching at 600 °C, which would not occur without the assistance of applied pressure.

7. The high hardness values, along with the martensitic nature of the contact zone, the results show significant consistent between the thermos-mechanical simulated WEL and Rail WEL.

In summary, based on the structural and microstructural investigation, formation mechanics of the white etching layer on the rail steel material can be driven by (i) thermally (ii) mechanically (iii) thermomechanically. White etching layers considered as the main causes of crack initiation and squat origin. As the root causes and mechanics formation of squat crack initiation defects on the rail surface are identified, the mechanics of crack development and growth into the railhead also need to be determined. Thus, the next chapter will discuss the role causes of water entrapment mechanism on the squat crack propagation and growth.
8. Mechanism of squat crack growth: role of water on crack evolution

8.1 Introduction

Squats have been found in almost all types of rail geometry such as tangent and curved tracks and turnouts, different rail types including mild carbon and hard hardened rails, and traffic conditions such as passenger and mixed traffic (Kerr et al. 2008; Li 2009; Wilson et al. 2012; Zoeteman et al. 2014). However, the absence of squats inside tunnels has been reported by numerous researchers for different railway networks. In Australia, a set of statistical data obtained from Sydney Trains and conducted on a wide proportion of the network confirmed that there are no squats in the tunnels (Kerr et al. 2008). In the UK, Grassie et al. reported that the defects on the London Underground network occurred almost exclusively in the open track rather than in tunnels (Grassie, S. L. et al. 2011). Observation confirmed by Dutch experience is that squats do not occur in tunnels (Steenbergen, M & Dollevoet, R 2013). Squats, or rail shelling defects as called in Japan, usually occur outside tunnels unless contamination is present (Ishida et al. 2003; Kondo et al. 1996). The absence of water on the rail surface in tunnels can be considered as a critical exception factor for squat formation (Grassie 2012; Kerr & Wilson 2012; Kerr et al. 2008; Wilson et al. 2012). This conclusion is often associated with a converse argument, where the presence of fluid and contamination is found to play a major role in crack development, particularly in the traffic direction.

Numerous researchers have concentrated on the mechanics of crack propagation and squat growth. Several hypothesis and theories such as “dry”, “wet lubricated”, “hydraulic pressurisation”, “liquid entrapment” and “squeeze fluid film” have been proposed for driven crack growth and development (Bogdanski 2002; Bogdański 2005; Bogdański 2014; Bogdański, S & Lewicki, P 2008). Laboratory investigations have been carried out to measure the role of water on squat formation and growth. However, although numerical modeling and laboratory tests support the hypothesis of hydraulic entrapment on squat growth, there is no reported direct metallurgical evidence of rails damaged by the presence of water under service conditions in a rail track network. The requirement for fluid to cause small cracks to propagate provides a rational of why squats vanish as soon as
as the railway enters a tunnel. The questions can arise concerning the physical evidence of water entrapped in cracks for an in-service track, and what evidence if any, is there that the surface water promotes crack growth? In addition to the mechanics of hydraulic entrapment on crack propagation, is there any other likely mechanisms operative?

In this study, a localised section of rail was found to suffer severe surface damage associated with water dropping from an air-conditioning system located in the tunnel roof. A detailed metallurgical examination has been conducted to examine the formation of WELs under both dry and wet conditions and their relationship with squat initiation, and to measure the effects of water on crack propagation and squat growth of ex-service rail. The work on the ex-service rail under locally wet and dry conditions has provided a unique opportunity to answer the questions of squat growth applied to railway networks.

8.2 Experiment

8.2.1 Ex-service rail samples

Damaged rail samples, shown in Figure 8.1, were removed from the high rail of curved track inside a tunnel. The rails have two contact points in the railhead and gauge corner. These worn rails include five parts, each of length around 1.5-2.5 meters, labelled from 1 to 5 along the traffic direction on the same running line. Part 3 was further cut into two sections for easy transporting.

As shown in Figure 8.1, there is no severe surface damage on rail Parts 1, 2, 4 and 5. In comparison surface, breaking-cracks (surface damage) can be clearly seen in Part 3 which was affected by water dropping. The higher magnification images in Figure 8.2 reveal different stages of crack development on the running band and gauge corner. Different surface damage features, such as light, moderate and severe surface-breaking cracks, as well as spallation, were detected on Part 3 showing rail surface deterioration with respect to traffic direction.
Normally, there are no squats on the rail in tunnels under dry service conditions. However, associated with water dropping from an air-conditioning system at tunnel roof, wheel and rail contact was changed from dry to wet, light and moderate cracks started to appear at the early stage of squat damage at the right side of rail where the trains were coming as shown in Figure 8.2. An interesting phenomenon has been noted, that the path of surface-breaking cracking is perpendicular to the traffic direction. Under the impact of water, severe rail surface damage occurred and there are three spots of large squat-induced surface spallation. After that, cracks at moderate and severe stages of growth can be observed, and one significant rail surface spalling off at the gauge corner. The shape of the squats is still similar to the two lungs. When the water vanished, the rail surface damage disappeared.
8.2.2 Experimental procedure

Based on the crack development stage, surface damaging level and wheel/rail contact condition, six positions on the rail surface were selected for metallurgical investigations. As shown in Figure 8.3, Samples 1 and 6 were sectioned from Parts 1 and 5 respectively, under dry conditions, and there are no visual surface-breaking cracks. These two sectioned samples were chosen to investigate the state of the rail surface under the dry wheel and rail contact before and after the influence of water dropping (before and after the distinguishable damage on Part 3). Samples 2, 3, 4 and 5 were selected from Part 3, corresponding to surface damage at wet-starting, fully wet, wet-ending and nearly dry contact conditions, respectively.
It is noted that the damaged rail length from samples 2 to 5 is around 0.72 meters. The distance between Samples 1 to 2 is 2.3 meters and that between Samples 5 and 6 is 2 meters, which ensures that Samples 1 and 6 are under dry conditions.

The six samples were firstly sectioned by using a band saw cutting machine (Parkanson). Then they were cut into small slices at the different locations of the running band along the traffic direction. Each slice was mounted in Polyfast resin and then ground, polished and finally etched in Nital. The etched samples were observed under optical microscopy (OM) and scanning electron microscopy (SEM). Scanning electron microscopy was performed using a JEOL 7001 field emission instrument with Oxford X-MAX energy dispersive spectroscopy (EDS) detection system. EDS analysis was performed to identify the composition of some interesting spots found inside the cracks.

![Figure 8.3 Positions of six samples for metallurgical analysis.]

**8.3 Results**

**8.3.1 Sample 1 for dry contact of rail (before water-dropping)**

Figure 8.4a shows Sample 1, which is characterised by undamaged surface and dry contact of rail. Two regions, one at the field side of the running band and the other in the middle, were chosen for investigation. These regions were cut longitudinally into small
slices and crack growth features and subsurface microstructure was observed from the direction illustrated by the arrows shown in Figure 8.4b.

![Figure 8.4 Rail Sample 1 for dry contact condition](image)

Figure 8.4 Rail Sample 1 for dry contact condition, (a) rail sample showing two locations for investigation, b) schematic illustration showing observation side with respect to the rolling direction.

An important feature observed on the surface layer of the running band (Regions 1) is the presence of White Etching Layer (WEL) up to 50\(\mu\)m in thickness for ex-service rail under dry contact condition. Figure 8.5a shows WEL associated with two distinguished sub-layers. The top layer corresponds to WEL and the bottom layer is the so-called Brown Etching Layer (BEL) due to brown exhibition under light microscopy (Li et al. 2016) and it is thought to be at the early development stage of WEL. BEL is characterised by its relatively coarse grains compared with those of WEL. Sheared pearlite grains can be found under the BEL. As shown in Figure 8.5b, cracks were initiated within WEL and propagated into the rail matrix. It confirms the previous findings in open tracks that WELs contributed to crack initiation (Al-Juboori et al. 2017).
Figure 8.5  WEL on the surface layer of running band- Sample 1, (a) Cracks within WEL contributing to squat initiation, and presence of BEL underneath WEL, (b) WEL associated with BEL and sheared pearlite grains, (c) microhardness variations in the three regions.

The transformation in the crystalline microstructure between WEL and BEL could be related to different levels of fragmentation and dislocation density of the cementite lamellae, associated with various and repeated actions between wheel and rail. Indeed, the relative differences in grain refinement shown in the structure of these layers can be the result of different loading and/or thermal cycles. This means that the pre-formed layer is mechanically or thermally tempered resulting in a microstructural transformation of the layer or part of it (possibly the topmost layer). The differences in the microstructural features affect the mechanical properties of the material. This is consistent with the variation in the microhardness values measured at these three regions, as shown in Figure 8.5c. The hardness measurement of BEL is relatively lower than that of WEL but higher than the matrix region. The hardness in the pearlite steel structure has a strong correlation
with the interlamellar spacing (Krauss 2015). Thus, BEL exhibits different microstructural characteristics to WEL.

Figure 8.6a reveals that many shallow cracks grew in samples in the middle of the running band (Region 2). It is very interesting to note that these cracks are characterised by a vertical direction of propagation with an angle of about 90° to the traffic direction after initiation at 29° and transition at 66°. These cracks only grew to a very shallow depth, around 0.5 mm beneath the rail surface. It can be seen in Figure 8.6b, that the surface layer at the running band was severely deformed by the repeated wheel and rail rolling contact.

Hence, it is confirmed that WELs are formed for rail in the tunnel under dry contact conditions and that crack initiation for rail is correlated with the presence of WELs on the rail surface, and plastic deformation.

8.3.2 Sample 2 for wet-starting of rail (beginning of rail damage area)

Sample 2 was sectioned from Rail Part 3 and is characterised by relatively light/moderate surface-breaking cracks under wet-starting contact conditions due to the leaking water from the tunnel roof. This sample is located at a distance of about 2.3 m from Sample 1 and considered the beginning part of the rail damaged area, which contains light (Cracks A) and moderate (Cracks B) stages of crack development, as shown in Figure 8.7a.
Figure 8.7 Rail Sample 2 in wet-starting contact condition, (a) Rail sample showing the feature of surface-breaking cracks and the location for detailed investigation, b) schematic illustration showing the observation side with respect to the rolling direction.

To investigate the evolution of the cracks, a metallurgical analysis was conducted on 3 particular spots along the crack path (as shown in Figure 7b): (a) the crack mouth to determine the nature of crack initiation and its surrounding microstructure; (b) the middle of crack path to examine the branching, deformation and features along crack faces; and (c) the end of crack path to explore the behaviour of crack growth inside the rail material.

Figure 8.8a shows an overall cross-section view of light Crack A, which is at the early stage of crack development at the field side of the running band. The crack was initiated at 60° to the traffic direction and then changed to 32° for growth. The general crack growth pattern of this crack is similar to those observed in the open tracks (Steenbergen, Michaël & Dollevoet, Rolf 2013). However, this crack growth path is different from that observed in Sample -1 for dry contact condition, which was initiated at 29° and propagated along a nearly vertical direction at 90°. The crack penetrated 1.6mm beneath the rail surface, which is higher than that of the only 0.5mm for dry contact condition of Sample 1.
Figure 8.8 Crack A, (a) overall cross-section view of crack A, (b) crack open mouth, (c) crack middle, (d) crack end.

Figure 8.8b, c, and d are high-magnification images at the crack mouth, middle and end of the crack path of Crack A, respectively. It can be seen that the plastic deformation of the pearlitic matrix has accumulated at both sides of the crack mouth. Crack faces at the opening mouth, middle and end were entirely predominated by iron oxide (confirmed by EDS analysis) and there is some debris in the crack gap. The oxide along the crack faces is characterised by its undeformed microstructure, which means that no friction was present between crack faces. Thus, it would suggest that the growth inside the material is likely driven by tensile stress (Mode I loading) rather than shear stress (Mode II loading) (Bogdański et al. 1998; Bower 1988) as there is no shear deformation existing in the regions of crack faces. In addition, the significant amount of oxides inside crack would interpret the presence of water which possibly entrapped inside the crack by wheel/rail interactions. When iron oxides are produced in steel material, the volume of the original region is significantly increased (Ghodrat et al. 2013). The expansion in volume results in high tensile stresses which induced inside the crack faces. As a result, the presence of
oxides significantly promotes crack growth and development. A large gap between crack walls also provides evidence of the volumetric expansion being caused by oxides.

An interesting feature for crack growth at the field side of the running band (Region 1) is shown in Figure 8.9. The behaviour of another crack growth shows a Zig-Zag pattern, which is significantly different from those related to plastic deformation. The crack first formed normally and then suddenly changed its direction in vertical and horizontal orientations. The changes in the crack direction were found to be associated with the orientation of the pearlite lamella structure. There was an absence of sheared layer on the crack faces and the presence of iron oxides along the crack path.

![Figure 8.9 Different crack growth behaviour, (a) Zig Zag behaviour, (b) Y- behaviour.](image)

Figure 8.9b shows Y-shape crack growth behaviour. It appears that the first stage of crack initiation is almost similar, but that the cracks grew into two ways: one up to the rail surface and the other down into the rail material.
Figure 8.10 (a) full view of subsurface cracks at Region 2, (b), (c) and (d) Region C1, C2, C3 and C4 respectively, showing the directions of tiny cracks branched from primary cracks.

Figure 8.10a shows a full view of moderate cracks B1, B2, B3 and B4 under the rail surface layer in the middle of running band (Region 2) of Sample 2. The behaviour of the crack initiation and growth seems to be relatively complicated as the cracks have intersected with each other and a subsurface crack network is generated. The majority of cracks were initiated on the rail surface within an angle of about 60°-80° except for the major crack B2 which is initiated with an angle of about 30°. The mechanism of crack
initiation can be linked to rolling contact fatigue damage as severe plastic deformation was observed on the rail surface layer.

Under the surface, the moderate cracks were found to have turned down with varying depths, with a maximum of up to 7 mm below the rail surface. Each crack has developed a subsurface crack in the opposite direction of motion after a few millimeters of depth and they usually grew slightly horizontal and jointed with the nearby main cracks which can eventually result in possible surface spalling. Rail surface spalling induced on the running band can cause high impact load applied to the rails and consequently the deterioration of both track and some vehicle components can be exacerbated (Kerr 2013; Wilson 2012). It can be noticed that the rail surface area which is very close to this sample (Sample 2) has been damaged by spalling, as shown in Figure 2.

In addition, several small cracks were found to have developed from large cracks growing across the material in many directions. Figure 8.10b, c, d, and e show a typical example of fine cracks obtained at regions c1, c2, c3, and c4, respectively. It appears that the small cracks were fully filled by oxides with significant accumulation at branching points. Although the growth direction of both small and large cracks is related, the change in the crack path associated with the formation of secondary cracks may suggest that a different mechanism has driven the crack branching.

Another subsurface crack feature observed is that cracks B2 have developed two subsurface cracks, for which major and minor names are suggested. The major and minor cracks can correspond to leading and trailing squat cracks. Both of these initiated on the rail surface and propagated slightly down in the opposite direction with a depth of 2-5 mm until they turned up and merged with the adjacent cracks.

The considerable amount of iron oxides observed along the crack faces supports a hydraulic fluid mechanism of crack growth. This is evidenced by the complete absence of shear stress state at the crack faces. The presence of water inside the crack interior plays a significant role in reducing friction between crack faces (Datsyshyn 2005; Datsyshyn & Kadyra 2006).
8.3.3 Sample 3 for fully wet contact of rail

The rail surface is severely damaged under totally wet conditions. Hence, moderately damaged sections were chosen for metallurgical examination as crack growth features cannot be revealed within the severely damaged area.

Figure 8.11 Rail Sample 3 for the fully wet contact conditions, (a) Rail sample showing surface-breaking cracks, (b) schematic showing the observation side in relation to the rolling direction.

Figure 8.11a shows Sample 3, obtained 0.14 m from Sample 2 and considered as a moderate stage of crack development under fully wet contact condition. Spalling damage is also found within this sample. Three regions corresponding to the field side, middle and gauge side of running band were chosen for investigation.

Figure 8.12a displays the crack initiation and propagation behaviour of Region 1 at the field side of the running band. These crack growth patterns appear similar to those observed in Sample 2 for wet-starting contact (Figure 8.8, early-stage Crack A). The cracks grew up to 3.8mm beneath the rail surface. Figure 8.12b reveals a subsurface cracks network in Region 2, which is the middle of the running band. The cracks grew in
many directions under the surface and they linked with each other to cause surface spallation. Compared with 3.8mm crack depth at the field side of the running band, the cracks could grow up to 5mm in the middle of the running band. Figure 8.12c shows the crack growth pattern under the surface layer in Region 3, which is at the gauge side of the running band.

Figure 8.12 Full longitudinal section views of (a) Region 1, (b) Region 2 and (c) Region 3.

Figure 8.13 shows the transverse section through the running band area of region 4. It can be seen that the growth pattern and cracks development down into the rail steel material was very severe and complicated. It appears that the many sub-surface cracks were connected forming cracks networks. The propagation behaviour may suggest the involving of multi mechanisms of cracks formation.

Figure 8.13 Full transverse section view of Region 4, showing very complicated cracks growth behaviour.
Figure 8.14a, b, and c show the growth pattern and branching behaviour inside the material. Many secondary cracks were observed to generate from primary cracks. It is interesting to note that the behaviour of crack growth observed at crack ends reveals very complicated patterns, which is typical of, and more likely to be associated with stress corrosion cracking (SCC) phenomena. The intergranular and transgranular modes of cracks growth observed by metallographic analysis, together with the presence of corrosive products that were found at the crack ends and confirmed by EDS mapping (as discussed in Section 4) strongly support the contribution of SCC phenomena in crack growth.

Figure 8.14 (a) , (b), (c) and (d) growth pattern and branching behaviour of secondary cracks at crack tips.

8.3.4 Sample 4 for ending wet region of rail

Figure 8.15 displays the surface morphology of Sample 4, which was obtained 0.3 m distance from Sample 3 at the end of the region wet by the dropping water. It contains only mild surface damage. The surface damaged area is characterised by five light cracks at an early stage of development. These cracks aligned on the running band close to the field side and the cracks were orientated normal to the rolling direction.
Figure 8.15 Rail Sample 4 for wet-ending contact condition, surface-breaking cracks, (b) schematic showing the observation side in relation to the rolling direction.

Figure 8.16a shows two cracks propagated underneath the rail surface at the field side of the running band in Region1. These cracks seem to have almost similar patterns of growth but the initiation angle is slightly different. The surface and subsurface behaviour of these cracks is identical to that in Crack A observed in Sample 2, which is also at the field side of running band for wet-starting contact condition. The crack penetrated up to 3.4 mm beneath the rail surface. Figure 8.16 b and c show the micrograph images obtained at the surface area close to cracks mouth, revealing the presence of iron oxides inside cracks that just initiated. This result confirmed again that the presence of water plays a significant role in crack growth.
Figure 8.16 (a) Low magnification image showing crack behaviour under the surface, (b), (c) micrographs showing the existing of iron oxides at both side faces.

Figures 8.17 shows the features observed in the middle of the running band in Region 2. This area is characterised by two distinguishing features: deep cracks and WEL around 10\(\mu\)m. Micro-cracks are expected to be initiated as a result of WELs and plastic deformation. Although WELs were not observed in wet-starting contact conditions, it is believed that once formed, they are removed by severe dynamic contact forces between wheel and rail.

Figure 8.17 (a) Shallow crack associated with plastic deformation, (b) a slight thickness of WEL on the rail surface of Region 2.
8.3.5 Sample 5 for nearly dry contact of rail

Figure 8.18a shows the surface morphology of Sample 5, obtained 0.28 m from Sample 4 and is characterised by free surface cracking for nearly dry contact conditions. Two regions were selected based on the crack locations in the previous sample.

Figure 8.18 Rail Sample 5 for nearly-dry contact condition, (a) Rail sample without surface damage, (b) schematic showing the observation side in relation to the rolling direction.

Figures 8.19a and b show the highly magnified images obtained at the field side and in the middle of the running band respectively. These images reveal the microstructure of the subsurface layer of the rail for nearly dry contact conditions. In Region 1, Figure 8.19a, the surface layer was heavily sheared, and surface-initiated rolling contact fatigue crack was formed. The root of this crack, as can be seen, is completely filled with a patch of WEL and oxide. However, WEL is entirely absent at the surface layer just above the crack location. The presence of WEL inside the crack path rather than the top surface (usually found) could be related to fragmentation occurrence. Interpretation of this is that WEL has formed on the rail surface at the beginning but due to high impact dynamic load associated with partly wet (nearly dry) contact, the WEL has fragmented. As a result, part
of the WEL has been inserted into the closest crack. This process is likely to occur due to the fact that WEL is susceptible to breaking due to hardness and brittleness characterisation. The occurrence of fragments of WELs located inside the squat crack has also been reported in another study (Kerr & Daniel 2011). This can support the expectation that the surface layer was originally covered by WELs but was removed by severe dynamic contact. Evidence is obtained from the microstructural analysis that shows the existence of a relatively thick WEL on the rail surface at the adjacent region. Figure 8.19b shows a surface layer of running band that was covered by a mixture of WEL with oxide, making up to 30 μm in thickness. It is expected that the moist environment has converted the surface layer from stable WEL to locally and partly corroded WEL.

![Figure 8.19 Longitudinal section obtained from Sample 5: nearly dry of rail, (a) at the field side of running band in Region 1, (b) in the middle of the running band in Region 1.](image)

8.3.6 Sample 6 for dry contact of rail (after water-dropping)

Figure 8.20 shows the surface morphology of Sample 6, located about 2 m distance from Sample 5, and where there was no water dropping. This sample is characterised by the undamaged surface and was selected to investigate the characteristics of rail surface for dry contact after the influence of water dropping.
Figure 8.20 Rail Sample 6 for dry contact condition (after water-dropping), (a) undamaged surface and (b) schematic illustration showing the observation side in relation to the rolling direction.

Figure 8.21a and b reveal the microstructure of the rail surface layer obtained at the field side and middle of the running band for dry contact without the influence of water dropping. Both regions show the presence of WEL on the rail surface with a thickness of up to 50 μm which is nearly the same thickness for dry contact before water-dropping for Sample 1. At the field side of the running band, a heavily deformed layer was observed underneath the WEL, which indicates that the WEL was induced by plastic deformation.
In the middle of the running band, fine cracks were found within the WEL, in addition to the presence of BEL. The formation of WEL on the rail surface is possibly due to a micro slip induced by the wheel on the rail. The friction traction between the wheel and rail was relatively high compared with Sample 5. Totally dry contact occurs for Sample 6 and the rail surface was not contaminated by water.

8.4 Discussion

Based on the results obtained, the part of rail (Rail Part 3), where the water was concentrated, is severely damaged in comparison with others (see Figures 1 and 2). Microstructural analysis showed that deep cracks, reaching up to 7 mm depth, were found at the affected area, at the Samples 2 to 5; while the cracks of only 0.5mm in-depth and associated with WEL were observed in the area of dry rail (before and after the water dropping region, Samples 1 and 6). This indicates that crack propagation and growth development was significantly influenced by the existence of surface water. These findings provide evidence of the hypothesis of crack propagation being driven by hydraulic entrapment of water.

In order to provide the physical evidence of possible effects of water on crack propagation, EDS map analyses were performed to determine local compositions in the vicinity of crack faces and crack branching root. Figure 8.22 shows a typical example
obtained at the crack tip, at the field side of the running band of Sample 2 under wet-starting contact condition. A specific area inside the crack surrounded by the dashed line was chosen for analysis. The electron image displays an enlarged image of the selected map area. The elemental mapping shows that the percentage of Oxygen (elemental O) inside the crack faces is considerably higher compared with that in the matrix region (pearlite lamella structure), which provides the evidence for the existence of iron oxides inside the crack due to the presence of water. In addition, varying percentages of chlorine (Cl) and sulphur (S) were found to be concentrated at crack regions as shown in element mapping. Although the mass percentage of these elements is relatively low compared with other elements measured at the adjacent region, they are considered as corrosive products. The chemical compositions of the selected regions A and B (as shown in the electron image), at crack interior and matrix region (pearlite lamella) are given in Table 8.1.

![Electron image](image)

Figure 8.22 Typical example obtained at branching crack showing the selected region with EDS mapping.
As has been discussed in the literate review, several mechanisms of crack growth by pressurisation of water inside the crack have been hypothesized. These include (i) assistance of crack opening by fluid trapped inside the crack (increase tensile stress at crack tips and promote mode I crack growth) (ii) the hydraulic transmission of contact pressure to the crack faces (the growth is driven by tensile stress Mode I loading) and (3) fluid lubricated crack faces (lubricant reduces friction between crack faces and promote mode II shear crack growth).

Table 8.1 EDS results obtained from EDS analyses of selected regions inside cracks and bulk material.

<table>
<thead>
<tr>
<th>Elements</th>
<th>Chemical composition, Wt%</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Crack interior, region A</td>
</tr>
<tr>
<td>C</td>
<td>24.57</td>
</tr>
<tr>
<td>O</td>
<td>33.32</td>
</tr>
<tr>
<td>Si</td>
<td>0.21</td>
</tr>
<tr>
<td>S</td>
<td>0.24</td>
</tr>
<tr>
<td>Cl</td>
<td>0.11</td>
</tr>
<tr>
<td>Mn</td>
<td>1.63</td>
</tr>
<tr>
<td>Fe</td>
<td>39.92</td>
</tr>
</tbody>
</table>

If crack propagation is caused by Mode II loading, the shear deformation exists at the crack face as a result of shear stress (Olzak et al. 1991). However, the plastic deformation layer of the crack faces was completely absent in all the samples examined. This represents evidence that the crack growth mechanism was mostly driven by Mode I loading (opening) rather than by shear stress (Mode II loading) and confirms the principle role of pressurization and entrapment mechanism of water in crack propagation.

The presence of oxides at crack interior may promote crack propagation. When iron oxides are produced inside the crack, the volume of the original region is significantly
increased. Thus, the local oxidation results in volumetric expansion inside the cracks. This process leads to the generation of stresses inside the crack faces causing local crack opening behaviour. This form of oxidation has a significant effect on the growth of deeper crack, and associated branching processes (Steenbergen, M & Dollevoet, R 2013). Thus, it is believed that the existence of these iron oxides plays a role in promoting crack growth and development.

In addition, the presence of corrosive products such as chlorine (Cl) and sulphur (S) elements inside the crack faces and tips, as confirmed by EDS analysis, can be a source of crack growth. It means that the mechanism of stress corrosion cracking (SCC) may contribute in crack growth. The local fracture behaviour of cracks, in which they have developed in multi directions associated with the formation of many secondary cracks that were totally filled by oxides is consistent with characteristic of stress corrosion cracking mechanism.

To summarize, due to the presence of water on the rail surface, the cracks are propagated under mode I tensile opening caused by pressurization and entrapment of water as evidenced by the existence of oxide along crack face and absence of shear deformation. In addition, the volume expansion of iron oxide assists mode I opening at crack tips. Indications of stress corrosion cracking (SCC) were also observed for crack growth and propagation.

### 8.5 Conclusion

Under normal dry conditions, no squats are observed on the rail in tunnels. A section of rail in tunnel affected by water dropping from the tunnel roof was obtained for a controlled experiment on ex-service rail involving detailed metallurgical examination of the effects of water on the formation of WELs, crack growth and squat formation. Six samples, corresponding to various wheel and rail contact conditions, including dry-contact (before-water dropping), wet-starting, fully wet, wet-ending, nearly dry and totally dry (after water-dropping), were examined in detail. The important findings are summarized as follows:
1. Water is not an essential condition for the formation of WELs.

2. For dry rails, WELs up to 50 μm in thickness were found to be formed and overlapping a related sublayer called Brown Etching Layer (BEL). The formation of two distinct layers is believed to be the result of different mechanical and/or thermal cycles. Confirmation was obtained from the variation in observed microstructural features of WELs and BELs, and the associated dropping in the microhardness values.

3. For wet affected rail only thin WELs, up to 10 μm, were observed and it is concluded that the majority of WELs were removed by severe dynamic contact between wheel and wet affected rail. Evidence was obtained for fragmentation of WELs, from microstructural observations showing the presence part of WEL inside a crack and absence on the surface of Sample 5, Region 1 at the nearly dry rail.

4. Based on the results obtained, crack initiation is associated with the formation of plastic deformation and the presence of WELs on the rail surface; while crack propagation is mainly influenced by the presence of water.

5. Support for crack propagation being caused by hydraulic entrapment of water is derived from the fact that the areas affected by locally water-dropping were severely damaged compared to other sections of the rail.

6. The microstructural analysis confirmed the presence of significant amounts of iron oxide along the cracks. Observations of the mode of crack propagation, combined with the detection of oxides at the crack faces and the existence of corrosive elements represent conclusive proof that water was present inside cracks under the wet rail surface, and plays a significant role in crack propagation.

7. The presence of large amounts of oxide along the crack faces as well as the absence of plastic deformation suggests that the growth mechanism is driven by
tensile stress (Mode I loading) rather than shear stress (Mode II loading) caused by pressurization and entrapment of water.

8. The volume expansion of iron oxide assists mode I opening at crack tips.

9. The distinctive complex fracture pattern and growth behaviour observed in cracks at wet rail contact, including the formation and growth of secondary cracks in random directions inside the material, are a likely consequence of a degradation mechanism involving stress corrosion cracking (SCC). Results obtained from EDS analysis confirm the presence of corrosive elements that are typically responsible for corrosion of medium carbon steels inside cracks.
9. Squat Induced Rail Fracture Failure

9.1 Introduction

In the previous chapters, the mechanics of squat initiation and propagation on the railhead were discussed. Typical squats contain two surface-breaking cracks on the running band, leading (principle) and trailing cracks. These surface-breaking cracks frequently develop down into the rail matrix in the horizontal plane with depth in a range between 3-6 mm; then join together and cause rail surface material to detach (spalling). Squats exacerbate the deterioration of track and some vehicle components. Associated excessive noise and vibration have increased the complaints against the rail operators. Under certain testing conditions, squats can compromise ultrasonic rail testing and hide potential turning down defects as the longitudinal propagated cracks may shield the signals.

Previous understanding viewed squat development as benign. They form much quickly under relative low tonnages compared to the typical RCF defects, for example, a depth of millimeters within 10MGT. But the main risk with squat defects is that the cracks may turn down and grow into the parent rail at a transverse plane, resulting in Transverse Defect (TD), with the possibility of a complete rail fracture failure if not detected in time. The formation of transverse defects can be significantly promoted by the impurity of rail material such as the inclusions formed during steel manufacturing. This is relatively rare and until recently was unknown in the Sydney Trains system. In this chapter the fracture failure of the rail caused by squat defects was present. The fracture failure by squat considers an advanced stage of squat cracks development and growth down the surface of the rail. This work proves the fracture failure mechanism being caused by squat cracks evolution inside the railhead containing pre-existed inclusions.

9.2 Experiment

9.2.1 Visual inspection on the ex-service broken rail

Broken rail samples as shown in Figure 9.1, were removed from the high rail of transition section (transition from tangent to curved track with 800m radius). The broken rail section was confirmed to be caused by a transverse defect (TD) developed from the squat.
A visual inspection was conducted on the broken rails after cleaning and rust removal to evaluate the level of surface damage. As shown in Figure 9.2, the broken rail has significant patches of surface damages, for examples, moderate squat defects were visibly found on the running band. There are two adjacent squats (Squats 2 and 3) within the distance of 38 mm and 50 mm from the crack initiation point at which the particularly Broken Squat was located. Other squats were located within a distance of about 200-300 mm to the Broken Squat. The intensity of squat formation within a short distance of rail (six squats including the Broken Squat within about 0.5 m length) demonstrates the severe status of rail surface degradation.

From the side view, the holes were found at the rail web on both sides of the broken section to install the fish plate after the rail broke and before the repair. The position of a broken section in relation to the sleepers can also be recognized.

After the cleaning process, the rail surface was etched by using a 2% Nital etchant solution to identify whether there is a microstructural change associated with the formation of the white etching layer. Figure 9.3 shows the existence of sliding marks on...
the sunning band area and a distinguished bright area was observed, believed to be associated with WEL and worn surface; but further metallurgical examination is needed.

Figure 9.2 Surface observation of the broken rail.
Figure 9.3 Surface statues after etching with 2% Nital, showing an existing of the sliding marks and a distinguished bright area.

Figure 9.4 Lateral views of the broken rail sections, showing the ductile and brittle fracture zones.

The lateral morphology (side view) of the broken rail sections is shown in Figure 9.4. There are two distinguishing zones, including one ductile crack growth zone, and another brittle fracture zone. It indicates the rail was firstly under the ductile crack development;
then followed by a rapid brittle fracture failure. The ductile crack growth zone could be further divided into two regions: shallow longitudinal cracking region and turning down cracking region.

The evolution of crack growth and critical turning down locations were shown in Figure 5. There are four stages of crack growth in the ductile crack growth zone. The first stage is corresponding to longitudinal crack growth and stages 2 and 4 are relevant to the TD crack propagation. The cracks initiated from the rail surface and propagated at an acute angle to the surface about 23° in the direction of traffic. When the cracks reached a depth of 8 mm, they started to turn down at an angle of 34°, followed by the increased angles of 53° and 60° until brittle fracture failure. During the shallow angle growth, the spallation of surface material occurred from the head of the rail, but the broken rail was caused by the turning down the crack.

Figure 9.5 Path of crack growth.
Three-dimensional morphology of the fracture face has also been measured and shown in Figure 9.6.

![3D profile of fracture face of the broken rail.](image)

Figure 9.6. 3D profile of fracture face of the broken rail.

9.2.2 Track/operation data and experimental methodology

From the above observation, the rail fracture failure was induced by a squat defect on the rail surface. It was a typical HH 60Kg rail manufactured in 1996, which is located at a transition from the tangent track at 14.94km to an 800m radius at 14.98km. The annual loading tonnage is 22.81 MGT. The concrete sleeper was installed in 2006. The schedule grinding interval of the interested rail is 15 MGT. However, due to short of grinding windows, it was out of the grinding cycle, corresponding to 9 MGT out of traffic loading since the last grinding. The squat defects of the interested rail section were successfully detected by the ultrasonic technique and continuously monitored the development. The development history from the first defect report to fracture failure was 1 year 2 months.
and 8 days. It must be noted that the rail fracture occurred just 10 days after the most recent inspection.

It is highly suspicious for a fast process of rail fracture failure induced by squats. Hence the detailed metallurgical examination and analysis were performed to determine the initiation and growth of the squats, particularly the specific cause of fast fracture failure.

9.2.3 Experimental procedure

Three regions, a squat broken section adjacent to Squats 2 and 3, the region near Squat 1 before rail broke, and the region near Squat 5 after rail broke, were chosen for metallurgical examination. Samples were sectioned through the longitudinal direction (parallel to rolling direction) to determine the crack growth and development behaviour inside the railhead. The structure and microstructure of the surface and subsurface layer were characterized by using light microscopy (Leica M205A) and (Leica DM6000), scanning electron microscopy (JEOL JSM-6490LV), micro-indentation hardness measurement (Vickers diamond tester with 500 g load, transmission electron microscopy TEM (JEOL ARM-200F) with accelerating voltage of 200 kV, working with SAED, BF and DF and EDS in TEM modes. Thin lamella sample was prepared for TEM characterisation by Focus Ion Beam (FIB) milling using FEI Helios NanoLab G3 CX instrument as shown in Figure 9.7.

![Figure 9.7 FIB process for TEM sample preparation.](image)

Figure 9.7 FIB process for TEM sample preparation. (a) The selected area of interest with the Pt layer deposited onto the area (b) milling both sides of the samples to produce a trench (c) bright-field TEM image after the final thinning and cleaning process.

9.3 Results

The cracks morphology and behaviour, the microstructure evolution of the surface layer in relation to the rolling direction, the characteristic of the material at the subsurface layer
of the rail at the broken section as well as at Squat 1 and Squat 5, corresponding the area before and after broken section were examined and discussed.

### 9.3.1 Squat at the broken location and the adjacent area

A shown in Figure 9.8, the typical V-shaped squat defects appeared on the rail surface. Three squats at the rail broken region were developed within a very short distance of 88 mm (also shown in Figure 9.2). It can be found that the Broken Squat and the adjacent Squats 2 and 3 were within the same level of development and showed almost similar size by comparing the superficial appearance, crack length, and growth angles of leading and trailing cracks.

![Figure 9.8 Surface cracking morphology at the Broken Squat and adjacent squats](image)

Figures 9.9 a and b show the crack growth behaviour and development pattern by observing the longitudinal section of adjacent Squats 2 and 3. For Squat 2 in Figure 9.9 a, the leading crack propagated down within an angle about 20° to the surface. When it reached a depth of 5 mm, the leading crack branched into to multi cracks; one developed in the horizontal plane and another grew down further to 6.5-7 mm in depth. Trailing crack developed within a shallow angle about 15 ° and 2 mm in depth under rail surface. For Squat 3 in Figure 9.13 b, the leading crack appears a deeper development compared with that at Squat 2. It developed down with an angle of about 23° and branched into two cracks at a depth of about 7 mm. The branching crack was turning down to 8.5 mm. The trailing crack grew with a shallow angle of 5° and 2 mm in depth.
The growth pattern of the above squat cracks is typical. However, the crack depth in the range of 6.5-8.5 mm can be considered as a serious stage, and the cracks already started to develop in the transverse plane as they are out of the layer of compressive residual stress.

![Crack growth in the longitudinal section of (a) Squat-2 and (b) Squat-3.](image)

Figures 9.10a, b, c, and d show the microstructure evolution of the surface layer at the longitudinal sections of regions A, B, C, and D, which were outlined by a dashed box in Figure 9.8. The material at the topmost surface layer was severely exhausted due to rolling/sliding contact. It is interesting to note that the shear direction of the texture at the squats can be either in the opposite direction of travel as shown in the Figure 9.10 b,c or in the same direction of traffic as shown in the Figures 9.10 a, d, which is not reported previously. A possible explanation of this is that a revered shear cyclic loading may occur due to significant braking and/or sliding effort induced by the wheel on the rail. Nevertheless, severe plastic deformation is clearly present at the rail surfaces, as well as small WELs patches. Hence the initiation of rail surface degradation for the neighboring squats was driven by either rolling contact fatigue RCF or white etching layer.
Figure 9.10 Surface and subsurface characterisation across the longitudinal section of regions near the broken section showing microstructural evolution.

The rail was broken just after 24 MGT traffic load since the latest grinding. To identify the potential cause of the TD cracking and rail fracture failure at low tonnage, the detailed metallurgical examination was conducted along the crack month. As shown in Figure 9.11, a large abnormal microstructure, a martensite patch of 3.2 mm in length was found along the main crack of the Broken Squat. The top tip of the martensitic patch is 8.0 mm down from the rail surface, which is exactly corresponding with the starting point of TD cracking at stage 2 as shown in Figure 9.5. The formation of martensite structure in the pearlitic rail material is detrimental to the toughness and crack arrest capability of steel. The squat grew as normal under the influence of the stresses/strains caused by wheel and rail contact, but it turned down in the transverse direction at low tonnage due to the existence of extreme brittle martensitic patch at the crack front.
Figure 9.11 Pre-existed martensitic patch along the crack path of the broken section contributing to turning down into the transverse section.
Vickers microhardness with 500g loading was performed at the martensitic patch and adjacent bulk material. As shown in Figure 9.12, the average hardness of the martensitic region is about 852 HV, which is much higher than that of the pearlitic region (bulk material) at 368 HV.

Figure 9.12 Vickers hardness measurements of the pearlitic and martensitic regions along the broken crack (500g loading).

Figure 9.13 shows the detailed development of TD cracking. The interface between the pearlite and martensite remains bonded as shown in Figure 9.13 d. However, the martensitic patch was fractured and cracked through itself due to its high hardness and low fracture toughness as shown in Figure 9.13 b and c. The growth of TD crack in the martensitic patch was rapid in nature which resulted in the rail fracture failure at the low tonnage.
Figure 9.13 Development of TD cracking.

Figure 9.14 a-d shows bright field and dark field TEM images with SAED pattern, obtained from the interface area between bulk material and martensitic patch. A sharped transition can be clearly observed between pearlite and martensite phase structure with a full absence of any gradient structure. The microstructure of bulk material comprises a regular arrangement of ferrite and cementite lamella, with some random orientation of cementite near the interface region due to partially dissolved. In contrast, the microstructure of the martensitic patch appears to have structureless features, possibly due to the formation of fine martensitic crystal grains. The SAED pattern taken from bulk material confirmed the single-crystal ordination with a zone axis near to \{1-11\} ferrite (bcc) structure, as shown in Figure 9.14 c. Whereas, SAED pattern obtained from the martensitic region is composed of a polycrystalline ring, as shown in Figure 9.14 d. The SAED also shows the splitting spots in the (110) and (220) rings, which indicate that a tetragonal structure (bct) of the martensite has formed.
Figure 9.14 TEM micrographs with corresponding SAED pattern obtained from the interface region between the martensitic patch and bulk matrix, (a) bright-field image, (b) dark-field image, (c) and (d) SAED patterns of the selected area outlined by a circle in the (a).

The BF and DF TEM micrographs in Figure 9.15 shows the detailed microstructure of the selected area, outlined by the dashed box in Figure 9.14 a. The cementite with short-strip like particles with a different size was found through the ferrite matrix. It indicates that the cementite lamella has partially dissolved at the boundary layer. At the adjacent region, the plates like martensite crystals were observed near the interface boundary.
Figure 9.15 TEM micrographs of the selected area, outlined in Figure 9.14a, showing microstructure characteristic at interface boundary in high magnification (a) BF, (b) DF.

Figure 9.16 shows the microstructure of the martensitic region away from the interface layer. The BF image exhibits the existence of a fine twinned structure. The DF micrograph taken by using a diffraction spot of the matrix, outlined by the circle in the SAED image, shows a twin martensite layer with the existence of some bright ultra-fine particles between twins interfaces, which is most likely related to undissolved cementite particles. Splitting spots were also observed in the diffraction pattern.
Figure 9.16 High magnification TEM images taken from martensitic patch showing the formation of twined structure, (a) BF image, (b) DF image using diffraction spot of the matrix, outlined in the SAED image (c).

Compositional analysis by EDS mapping was performed on the interface region to determine the alloying elements, including carbon (C), oxygen (O), Silicon (Si), manganese (Mn) and Iron (Fe), at the pearlite and martensite phases. The EDS mapping in Figure 9.17 shows a uniform distribution of the elements at the different regions, with no obvious change in composition. Hence, the large martensite patches in the railhead were not formed by elemental segregation.
9.3.2 Squat characterisation before the broken section (Squat 1)

Figure 9.18 shows that squat-1, which is 187 mm away before the broken section, contained the leading crack with a length of about 21.5 mm and tailing crack with a length of about 11.6 mm. These two cracks developed on the running band with an angle of about 96°. The size of squat-1 appears similar to that of the Broken Squat and its adjacent squats (squats 2 and 3). The surface cracking pattern of squat-1 confirms that this squat can be classified as a moderate defect of the stage, like the squats at the broken section. A bright distinguished patch that was believed to be white etching layer was observed, as indicated by an arrow.

Figure 9.18 b is the longitudinal section of leading and trailing cracks, showing the propagation and growth pattern under the surface of the rail. The leading crack developed down with an angle of about 18° to the rail surface and branched into two cracks at a depth of 5 mm; one developed into the horizontal plane and another further turned down.
to 6.7 mm in depth. The trailing crack propagated under a shallow angle with the surface and reached a depth of 3.5 mm.

Figure 9.18 Squat-1 before the broken section, showing the characteristics of surface-breaking cracks in (a) and subsurface crack growth in (b)

Figure 9.19 shows the characteristic of the surface layer along the longitudinal section at the interested regions A and B as outlined in Figure 9.18a. In region A, a thin white etching layer was observed on the surface of the rail. A small crack was found associated with the white etching layer. Plastic deformation can be seen underneath WEL, confirming that WEL is mechanically induced. In addition, the crack path and the faces were fully oxidised. In region B, shown in Figure 9.19 b, the characteristic of the top surface layer confirmed the existence of a thin continues layer of white etching layer that
was the result of an accumulation of severe plastic deformation due to repeated dynamic load.

Figure 9.19 Surface and subsurface characterisation across the longitudinal section of regions A and B near Squat-1 (before the broken section) showing microstructural evolution and white etching layer formation caused by severe plastic deformation.

9.3.3 Squat characterisation after the broken section (Squat-4 and Squat-5)

As shown in Figure 9.20, the surface cracking morphology of Squat-4 and Squat-5 is typical, with the formation of V-shaped cracks on the running band area. The size of Squat-5 is slightly smaller than Squat -4. No other visible abnormal features were observed on the surface of the rail expect some fine grinding marks.
Figure 9.20 Squat-4 and Squat-5 after the broken section, showing the characteristics of surface-breaking cracks in (a); subsurface cracks growth for Squat -4 in (b) and for Squat-5 in (c).

Figures 9.20b and c show the crack growth behaviour of Squat-4 and 5. The leading crack in Squat-4 grew down with a shallow angle of about 16° to the rail surface. It branched...
into two subsurface cracks at depth 3 mm; one developed in a horizontal plane with a constant depth of about 3 mm, and the other turned down into the rail with an angle about 50° to a depth of 6 mm. The trailing crack propagated with a shallow plane to 2-2.5 mm in depth. For Squat-5, the leading cracks developed with an angle of about 15° and branched at a depth of about 3.4 mm. Tiny subsurface cracks were found to have branched from the leading crack and turned up to the rail surface.

The growth behaviour and development pattern for leading and trailing cracks of Squat-4 and 5 are similar to those observed at the broken section and Squat-1. The leading cracks in squats tend to turn down at a depth of about 6-8 mm. However, the cracks would not develop much quicker than normal unless a pre-existing defect, such as a martensitic patch, promotes the transition into a transverse plane.

Figure 9.21 shows the surface and subsurface characterisation through the longitudinal section at the regions A and B, outlined by a dashed box in Figure 9.20 a. There is a distinguished abnormal microstructure at the subsurface of the rail at region A, which shows a martensitic structure. Micro-hardness measurement at the region of abnormal structure confirmed that it has a high hardness value up to 842 HV which is consistent with the hardness of the martensitic patch found in the crack path of the broken section. The subsurface area also contained a high intensity of defects that appeared darker in the microstructure. The existence of martensite islands away from the broken section confirms that severe material impurity is popular in the rail originated from the rail manufacturing process before the installation.

Figure 9.21b in region B shows that the inclusions exist near the surface layer. The SEM micrograph of the inclusion does not exhibit a clear microstructure. The inclusions were thought to be iron oxides formed during steel production.

It is interesting to note again that the severe plastic deformation oriented in the opposite direction of travel was observed.
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Figure 9.21 Surfaces and subsurface characterisation across the longitudinal section of regions A and B near Squat-4 and Squat-5, respectively; showing microstructural evolution of rail away from the broken location. Inclusions like a martensitic patch and iron oxides inclusions were observed.

9.3.4 Fatigue crack growth experiment

In order to examine whether crack propagation is substantially affected by the amount of anisotropy existed in rail material, the fatigue cracks growth experiment was performed on the samples obtained from the railhead section of the broken rail. For comparison, different groups of samples were taken from a different rail which characterised by similar chemical and mechanical properties but no-impurity in the material.

The specimens for crack growth tests were cut from a depth of 5-15 mm to the rail surface based on the harness measurement, which can ensure the specimens are within the head hardened zone and away from any microstructural change and crack at the surface and subsurface. Figure 9.22 shows the hardness distribution values (HV10) across the railhead of the two different rails examined.
Figure 9.22 Cross-section of railhead showing the hardness distributions across the depth, as well as samples zone position. (a) Broken railhead, (b) Normal railhead

The dimension of the specimen for the crack growth test is shown in Figure 5.23. The experiment was carried out on an Instron 8801 hydraulic tensile and fatigue testing machine.

Figure 9.23 Specimen for fatigue crack propagation testing.

Unidirectional cyclic stress with a $\sigma_{\text{max}} = 70\%$ of yield stress and $\sigma_{\text{min}} = 0$, stress ratio $R = 0$ and frequency of 2Hz were used in this experiment. The length of crack growth was measured by using an advanced camera after every 200 cycle intervals until the fracture failure. The test was repeated 5 times to ensure accurate results.

Figure 9.24 shows the relationship between the growth of crack length and the number of cycles. It can be seen that the crack growth for the two different rails is significantly varied. The crack developed quickly, and the figure life was very short for the broken rail which was suffered by material impurities such as martensitic patches and inclusion. The fracture failure suddenly occurred without any sign of warning. In contrast, the crack for
the normal rail propagated regularly. The corresponding stereo micrographs show the fracture surface of two different rail samples.

![Diagram showing crack length development for two different rail samples.](image)

**Figure 9.24** Crack length development for two different rail samples.

SEM was performed on the crack propagation surface to examine the fatigue performance of the broken rail steels. Figure 9.25 characterises the crack initiation near the open notch. On the bottom of this image, there is a smoother zone which suggests the crack was nucleated (initiation zone). From a high magnification image, abnormal objects with different sizes can be clearly seen in the structure, suggesting that inclusions contributed to accelerating crack growth in the broken rail specimen.
9.4 Discussion

A detailed metallurgical examination was conducted on a broken rail, which was caused by a moderate squat defect, in terms of mechanisms of squat initiation, transverse direction (TD) cracking development and fracture failure of the rail. The rail was broken after 24 MGT traffic load since the last grinding 1 year and 15 days ago. It was 9 MGT tonnages out of the grinding cycle. Even the squat was at the moderate stage of the development, the rail fracture occurred just 10 days after the most recent inspection.

The results found that there were pre-existing abnormal microstructure, brittle martensitic islands, randomly distributed in the pearlitic rail steel. A martensitic patch with a size of about 3.2 mm was found at a depth about 8 mm from the rail surface, located at the turning down point of TD cracking. The squat was grown reacting to stress fields due to the contact but was turned down in the transverse direction reacting to the bulk stresses and the existence of a large martensitic patch at the crack front. The brittle and hard nature of this martensitic patch was the root cause to promote TD cracking and consequent rail fracture failure. The growth rate of TD crack in the martensitic patch was high in nature, resulting in the rail fracture failure at the low tonnage.

The abnormal microstructure was distributed randomly in the rail material, not only found at the crack growth path of the Broken Squat but also the regions away from the Broken Squat. The existence of the martensitic islands is significantly detrimental to the rail toughness, rolling contact fatigue resistance and crack propagation resistance. As there is no evidence of alloying composition differences between pearlitic and martensitic phases,
the formation of the martensitic phases is considered due to the un-proper heating treatment during the rail manufacturing process.

The microstructural examination was also conducted on other squats located before and after the broken section to validate the above findings. The results showed that severe plastic deformation was present in the running band, but the material flowed in the opposite direction of travel. The interesting phenomenon about the orientation of subsurface texture might be caused by a reversed cycle of strain, from braking/sliding actions. White etching layers were found on the rolling contact area caused by severe plastic deformation.

The morphology and dimension of the series of squats formed on the broken rail are nearly identical. The typical V-shape surface-breaking cracks were located on the running band with clear visible leading and trailing cracks. They are in a moderate stage of squat development crack.

The crack growth behavior along longitudinal direction appears a typical pattern. The leading crack tends to turn down when they grew to a depth of about 6–8 mm under the rail surface. However, the cracks would not develop further at an unusually fast speed until brittle fracture failure at lower tonnage without the influence of pre-existing larger defects in rail material, such as martensitic patch, inclusion, and oxide.

In summary, the initiation of rail surface degradation was caused by rolling contact fatigue and white etching layer induced from severe plastic deformation. The squat was grown reacting to stress fields due to the contact but was turned down in the transverse direction reacting to the bulk stresses and the existence of a large martensitic patch at the crack front. The growth rate of TD crack in the martensitic patch was high in nature, resulting in the rail fracture failure at the low tonnage. The risk of transverse breaks in rail rises as the crack penetration reaches 8 mm in depth.

9.5 Conclusion

A squat broken rail fractured at unusual lower tonnages was investigated to determine the mechanisms of squat initiation, transverse direction (TD) cracking development and fracture failure of the rail. Understanding the root cause of the squat induced broken rail
is highly important to provide a proper maintenance strategy and prevent this danger rail failure in the future. The important findings are summarized as follows:

1. The rail fracture failure was induced by squat defects on the rail surface. The rail was broken after 24 MGT traffic load since the last grinding 1 year and 15 days ago. The squat defects of the interested rail section were successfully detected continuously monitored by the ultrasonic technique. Although the squats are at a moderate stage of development, the rail fracture suddenly occurred just 10 days after the most recent inspection.

2. The abnormal microstructure, brittle martensitic patches, has been found in the rail material, which is detrimental to the toughness and crack growth resistance. The squat grew in reacting to stress fields due to the contact but turned down in the transverse direction in reacting to the bulk stress and the existence of a martensitic patch at the crack front. The growth of TD crack in the martensitic patch was rapid in nature, resulting in the rail fracture failure at the low tonnage.

3. The abnormal microstructure was distributed randomly in the rail material, which is significantly detrimental to the rail toughness, rolling contact fatigue resistance and crack propagation resistance. As there is no evidence of alloying composition differences between pearlitic and martensitic phases, the formation of the martensitic phases is considered due to the un-proper heating treatment during the rail manufacturing process.

4. Rail material at surface and subsurface was plastically exhausted due to rolling/sliding contact. The existence of white etching layers caused by severe plastic deformation was also confirmed. The initiation of rail surface degradation is thought to be driven by rolling contact fatigue and the formation of WEL.

5. The leading cracks of the squats were propagated down into the rail matrix at an acute angle to the surface about 15°-23° in the direction of traffic. When the cracks reached a depth of 6-8 mm, they turned down further at the angle about 50° -60°.
The risk of transverse breaks in rail rises when the crack penetration reaches 8 mm in depth with the influence of severe material impurity in rail steel.

6. The fatigue crack growth experiment demonstrated that the propagation of cracks was substantially affected by the amount of anisotropy that existed in the rail material.
10. Conclusions and Recommendations

10.1 Conclusions

Squat on the rail surface is now a serious problem to cause rail degradation and occurs in many railway networks across the world. The ex-service rails containing various stages of squats and WELs have been comprehensively investigated by field examination; advanced experimental techniques and high-resolution metallurgical analysis. The root causes of squat formation and the mechanisms of squat initiation and subsequent crack propagation on the rail surface were discussed in detail.

Detailed conclusions have been presented in Chapters 3, 4, 5, 6, 7, 8 and 9. The general conclusions are described as follows:

1. Current work confirms that there are a correlation between squat formation and the occurrence of WELs on the rail surface. Synchrotron XRD results obtained from squat defects and WELs at the vicinity regions of squats are consistent, which can be considered as a solid evidence to prove the relationship between squats and WELs. Occurrences of cracks extending from WELs down into the rail in squat regions were confirmed by microstructural observations, which is consistent with the premise that cracks in squats may originate from an extension of cracks in WELs.

2. There are two distinguishable types of WELs based on different operational conditions. At the heavy braking rail regions, WELs consists of martensite and retained austenite; and undeformed pearlite exists at the transition zone between the WELs layer and steel matrix. This WEL was induced by a combination of temperature and pressure changes. This type of WEL is termed as TP-WEL. At the steady traffic regions (low braking utilisation), WELs contains nanocrystalline plate martensite, fine ferrite and fine fragmented cementite, which is caused by rolling contact fatigue. This type of WEL is termed as SD-WEL.

3. A new formation mechanics of WELs on the rails under the effect of the leaking current (arching phenomenon) has been proposed and investigated. The samples
of rail material obtained from the ex-serviced rail were examined under different leakage current conditions by a Gas Tungsten Arc Welding (GTAW) machine. Both arcing induced WEL and pre-existed WEL comprise similar microstructure, composed by fine martensitic units (fine plates and very few lathes), undissolved cementite lamella, remnant pearlite lamella and some grains of retained austenite. In addition, the undeformed pearlite structure occurs in the transition region between WELs and matrix for both WELs, confirming the absence of shear deformation. The brown etching layers (BELs) were formed in parallel to arcing induced WEL for the case of high energy input. It was found that BEL was formed as a result of the tempering process of the pre-generated WEL.

4. High Pressure Torsion (HPT) experiment was performed on the pearlitic rail steel to investigate whether pure severe plastic deformation results in the formation of a white etching layer. HPT was used because the loading condition is similar to that of the rail-wheel contact. WEL induced by HPT experiment is consistent with that taken from the ex-service damaged rail. This finding provides evidence that some type of WEL formed on the rail surface is caused by severe plastic deformation due to rolling/sliding contact loading.

5. Gleeble thermal-mechanical study was performed to examine the potential formation of the white etching layer (WEL) on rail steel material by a combination of thermal and mechanical processes. It has been found that WELs can be formed below critical pearlite to austenite transformation temperature under a combination of thermal input and high contact pressure.

6. A localized section of rail associated with water dropping from an air-conditioning system located in a tunnel roof represents an ideal controlled location to examine the role of water entrapment on crack growth. Water is not an essential condition for the formation of WELs. However, crack growth is significantly influenced by the water entrapment inside the crack faces. The presence of oxides associated with an absence of shear deformation on the cracks faces reveals that crack growth was driven by Mode I loading (tensile opening) associated with the mechanism of hydraulic entrapment of water. Stress corrosion cracking was
involved in both crack growth and the observed formation of secondary cracks. Current work provides strong evidence that the presence of water plays a significant role in crack growth and development under the service condition.

7. A case of squat-induced rail fracture failure reveals that the cracks can turn down and grow into the parent rail on a transverse plane. However, the abnormal microstructure in rail steel, such as brittle martensitic patches and inclusions due to the un-proper heating treatment during the rail manufacturing process, results in the rail fracture failure at the low tonnage.

10.2 Future work

- It is highly recommended to conduct on-site monitoring for the evolution of white etching layers and the formation of squat defects. Track monitoring will introduce a quantitative evaluation of defect generation and white etching layer formation over a long-term track.

- In this study, the formation mechanism of white etching layers was examined by conducting experiments on the small-sized samples taken from ex-service rails. The future work is recommended to conduct the experiments by using a full-size wheel/rail testing facility which can provide high correspondence results to the real application.

- White etching layers are now popular existed on the rail surface. It is highly recommended to examine the particular conditions for subsequent squat formation, such as WELs thickness, WELs length, microstructure, and morphology. In general, why and how some WELs can induce squat; why some WELs are harmless.

- Brown etching layers have been found as a result of the tempering process of the pre-generated White etching layers. However, the detailed metallurgical examination is still lacking.
• In this study, a metallurgical examination on the ex-service damaged rail taken from the track section associated with water dropping from tunnel roof confirmed that the presence of water plays a significant role in squat cracks propagation and growth. It is recommended to examine the influence of wet and dry conductions on squat growth and development in rails. The experiment should be conducted on rails containing squats at an early stage of development, and by using a full-size wheel/rail contact testing.

• It has been found that there are more squat defects existed on Head Harden rail (HH) in comparison to Standard Carbon rail (SC). An explanation from the microstructure point of view is missing.
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