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Degradation Mechanisms of AA Conductor Strands in Air and

Jiregna Hirko Foggi

Degradation Mechanisms of AA 1350 Aluminium Alloy Conductor Strands in Air and Chloride Solution

Thesis for the Degree of Philosophiae Doctor

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Norwegian University of Science and Technology Faculty of Natural Sciences Department of Materials Science and Engineering



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Summary

The primary aim of this doctoral thesis is to investigate the degradation mechanisms contributing to the failure of the aluminium strand material (AA1350) used in steel reinforced aluminium conductors. These conductors are widely used in rural and marine environments due to their combined electrical and mechanical properties of the aluminium and galvanized steel core strands, respectively. The galvanizing is supposed to serve as a sacrificial anode for the steel core and mechanical barrier between the steel core and the aluminium strands to protect the strands against galvanic corrosion. Degradations due to fatigue and wear caused by wind induced vibration are the common causes of failure. Corrosion (galvanic, pitting and crevice) is also assumed to contribute. It was of interest in this thesis to investigate the significance of these mechanisms of degradations in the degradation of the aluminium strands of a conductor.

In the literature, investigation of the parameters of fretting degradation of aluminium strands has been based on bending fatigue test of the conductors in a laboratory air. The basic understanding of the parameters and mechanisms of fretting and wear has generally been restricted to dry laboratory environments. Studies investigating the effect of marine environments, especially the role of corrosion, are scarce. Therefore, the purpose of this thesis is to investigate the basic degradation mechanisms of the aluminium strand material used for conductor construction under controlled mechanical and electrochemical parameters.

The conductor samples investigated in this research were obtained from the utility companies through SINTEF Energy Research AS. The investigation was started with the failure analysis of samples collected from marine environments to see if this effort would reveal the relevance of mutual action of corrosion and wind induced vibration in the failure. Cracks appeared to initiate at slightly worn and corroded areas of the contact surfaces. However, it was not possible to distinguish between corrosion occurring before or after initiation of the cracks. The failure analysis suggested that fretting corrosion and sliding wear-corrosion were responsible for the degradation of the aluminium strands and indicated the need for further investigations in these areas as mentioned above. The crack initiation sites, which led to breaking of the strands, were corroded and fretted indicating that the two degradations contributed to failure. Analysis of the fracture surface suggested that the major portion of the fatigue life of strands was consumed by the crack initiation phase, which involved both fretting and corrosion. In addition, crevice corrosion and pitting corrosion degradations were observed. This

indicated that the protection provided by the galvanic coating was not sufficient to avoid corrosion of the aluminium strands. This indicated also the need for further analysis of degradation of the strands by fretting and wear under applied potential (enforced by cathodic protection) in chloride solution.

In the light of the failure analysis results above, fretting and sliding wear degradations were investigated in the laboratory in air and 3.5% NaCl solution for better understanding of the role of the electrochemical and mechanical parameters in the degradations involving mutual action of the two. The parameters for the fretting tests were a frequency of 1 Hz, normal force of 5 - 60 N and displacement amplitudes from $3 - 100 \,\mu\text{m}$ on a flat polished aluminium strand surface against alumina ball ("ball on flat") for 4200 s. Fretting of alike cylindrical strands were also performed and compared to that of ball on flat surface to investigate the effect of geometrical factors. The sliding wear test conditions were also the same as that of fretting tests except that the displacement and normal force were 5 mm and 2 N, respectively. The effect of applied potential was investigated in the NaCl solution by use of a potentiostat. The fretted and worn surfaces were investigated using scanning electron and optical microscopes. Material loss volume was estimated for the sliding wear tests using an optical 3D confocal microscope.

Investigation of sliding wear in NaCl solution at potentials in the range of -0.74 - 1.34 V_{SCE} showed that the electrochemical environment reduced the degradation by wear by acting as a lubricant. Direct contribution of corrosion to the material loss was not significant. The lubrication effect arose from the presence of the solution and the debris of wear and corrosion products, formed largely by the oxidation of the metallic debris into an interfacial layer. The mechanical processes were thus the dominating factors of degradation in the potential range of -0.74 - -1.34 V_{SCE}. However, the degradation mechanism changed considerably with the applied potential. In the potential range of -0.76 - -0.86 V_{SCE}, the metal surface is thermodynamically expected to be passive. However, this condition could not prevent mechanical wear dominated by delamination (shearing and tearing of the surface) and abrasive wear from fragments of debris. Above the pitting potential (-0.75 V_{SCE}), corrosion contributed to the degradation by anodic dissolution of the metal, mainly at the mechanically damaged areas. At more negative potentials down to -1.34 V_{SCE}, the wear mechanism was dominated by plastic deformation and fatigue wear. At potentials more cathodic (negative) to -1.34 V_{SCE}, the corrosion degradation became independent of mechanical wear. At OCP, sliding wear in NaCl solution caused cathodic polarization of the surface to about -1.32 V_{SCE}. This value was

determined by the cyclic depassivation, caused by localized exposure of the bare metal by rubbing, followed by repassivation by oxidation.

A fretting map, demonstrating the occurrence of stick, mixed stick slip and gross slip regimes within the range of normal force and displacement amplitude used, was established for both tests in air and NaCl solution as stated above. Physically, these regimes are associated with no degradation, cracking degradation and wear degradation, respectively. Of the contact parameters investigated in this work, the mixed stick slip regime was the dominant one, indicating susceptibility of the aluminium strands to cracking degradation (fretting fatigue). However, the boundary of the transition from the mixed stick slip to gross slip contact condition was shifted to smaller displacement amplitudes in NaCl solution compared to that in air, *i.e.*, the gross slip regime was wider in NaCl solution than in air for the same contact parameters. It is concluded, therefore, that the NaCl solution favours the gross slip rather than the mixed slip condition at the contact interface. This indicates the reduction of the tangential friction force under fretting in NaCl solution, such that fretting fatigue crack initiation takes longer time than in air. A tangential friction force arising from fretting in NaCl solution is about a third of that arising in air. The smaller tangential friction force in air is due to the lubricating effect of the NaCl solution, similar to that discussed for sliding wear above. This indicates that fretting fatigue life is expected to be much longer in marine environments than in air as far as fretting fatigue degradation is concerned.

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After my contract with NTNU ended on 09.01.2012, I started working for National Oilwell Varco Norway AS. This has enabled me to stay in Norway, acquire assets, learn and acquire skills in a dynamic culture and working environment with first-hand industrial experience and above all financial stability to complete this thesis.

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Glory to God for His mercy and help!

Jiregna Hirko Foggi

Submitted on the 28th of February 2020, Trondheim

Preface

This doctoral thesis is based on work carried out at the Department of Materials Science and Engineering, Norwegian University of Science and Technology (NTNU), under the supervision of Prof. Kemal Nisancioglu. Prof. Nuria Espallargas from Department of Engineering Design and Materials of NTNU and Researcher Steinar Refsnæs from SINTEF Energy have also contributed as co-supervisors. The work was funded by the Research Council of Norway, through SINTEF Energy Research AS under the project entitled "Condition Monitoring of Overhead Power Lines".

All experiments have been conducted by the author, except assistance in the operation of Electron Microprobe Analysis (EMPA) by Morten Raanes for identification of the alloy composition.

The thesis constitutes seven chapters. The first two chapters deal with general introduction and literature review on the main topics. The results are organised into three separate chapters on failure analysis of overhead power lines exposed to marine environments, sliding wear of the aluminium alloy, used as a conductor strand, in air and sodium chloride solution, and fretting of the aluminium alloy in air and chloride solution. The last two chapters present the summarizing discussion and conclusions of the work.

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1 Introduction

1.1 Background and Motivation

It is generally agreed that the failure of overhead power lines occur by degradation of the conductor at the suspension tower (Fig. 1.1) caused by wind induced vibrations and corrosion (in marine environments) [1, 2]. Economically, a conductor accounts for about 40% of the total investment in the power transmission and distribution systems [3] and its failure involves a significant economic loss and safety consequences. Moreover, since the load on the tower comes out of balance, other components may fail, ultimately leading to tower collapse. Therefore, understanding the failure causes and mechanisms of the conductors is of practical interest to mitigate the problem in design and operation.



Fig. 1.1 Power line components at the suspension tower [4].

Overhead power line conductors are typically made of ACSR (Aluminium Conductor, Steel reinforced) for use in marine environments and high capacity lines demanding high mechanical and

electrical functional requirements. It consists of stranded and galvanised steel core surrounded by one or more layers of AA 1350 aluminium alloy¹ strands [5]. The galvanised coating is used as a mechanical barrier of zinc between aluminium and steel strands and for cathodic protection against galvanic corrosion [6]. In these conductors, while most of the electric energy is transferred through aluminium strands, most of the mechanical load is carried by the steel core due to its higher axial stiffness as opposed to the more elastic and gradually creeping aluminium strands. Other standard conductor types, such as all aluminium conductors (AAC) and all aluminium alloy conductors (AAAC), are also used [7]. The service life of the ACSR conductor is determined by the failure of the aluminium strands.

The failure of a conductor due to reduction in fatigue strength of its strands by fretting has been well established, especially at suspension clamps [8, 9]. In a marine environment, the failure is also claimed to result from electrochemical corrosion [6, 10]. Localised corrosion forms such as pitting and crevice corrosion at the conductor-clamp and inter-strand interfaces, and galvanic corrosion of the innermost layer of aluminium strands adjacent to the steel core have been suggested as the causes of failure [1].

Fretting damage normally results from a cyclic minute relative displacement, also called "slip", at the interfaces between the conductor strands and clamp and adjacent strands due to wind-induced vibrations [11]. The conductor strands are also subjected to severe clamping pressure, tension applied during installation, and from the weight of the conductor itself at a suspension clamp. Consequently, the conductor-suspension clamp assembly (sketched in Fig. 1.2a) [12] has been recognised as the most susceptible location for fretting degradation and associated breaking of the aluminium strands. Detailed investigation of fretting damage of ACSR conductors at a suspension clamp showed that breaking of the aluminium strands occurs mainly at the mouth of the clamp (Fig. 1.2b) [13]. Especially, the portion of a conductor located between the contact edge of the keeper and the clamp body, where strand breaking is concentrated, has been recognised as the most susceptible zone to fretting fatigue damage [14]. Hence, several studies on failure characterisation of conductors have largely been concentrated on fretting damage investigations at the suspension clamps, which involve complex conductor construction, material properties, loading and other contact conditions.

¹ From here on referred to as "aluminium".



Fig. 1.2. Critical location of a suspension clamp. a) Schematic illustration of conductorsuspension clamp assembly showing the keeper edge (KE) and last point of contact (LPC) [12]. b) Distribution of failure of conductor strands from the clamp centre to the clamp mouth.

The load transferred to each contact interface from clamping and wind induced vibration varies from the inner to the outer layer aluminium strands. This causes variations in the relative slip amplitude occurring at the interfaces due to variation in the strain for different radial positions. In addition to fretting damages, localized corrosion attack may occur in the crevices of the strands and the contact interfaces, which may contribute to the damage processes and influence the service life. The issue has not been investigated much, and it is, therefore, the subject of this thesis.

Wind induced vibrations cause different fretting damages, such as fretting fatigue and fretting wear [15], in the conductor strands at the mouth of a clamp. The vibration of a conductor causes bending of the conductor with respect to the clamp, and this causes a cyclic relative slip at the interfaces of the strands in a contact. The relative slip in turn causes fretting fatigue or fretting wear, or sliding wear in special cases, depending on the magnitude of the relative slip amplitude at the contact interface. Accordingly, different contact interfaces experience different levels of slip amplitude which causes different fretting regimes characterized by distinctive surface damages observable by post-test metallographic examinations [16]. The types of fretting regimes developing at the interfaces and the role of corrosion on degradations related to these regimes have not yet been clarified, especially in relation to wear and crack initiation and propagation processes.

Lifetime modelling of conductors exposed to marine environments is a difficult task due to limited understanding of the tribological/tribocorrosion mechanisms involved in the degradation and failure of the strands. In view of this, investigating the failure mechanisms due to fretting and sliding wear behaviour of the aluminium strand material is of great significance in design, failure analysis and operation of such lines. Therefore, the study presented in this thesis is expected to lead to a better understanding of the degradation and tribological behaviour of the ACSR conductors for further use in the design of overhead power line conductors and their suspension clamps.

1.2 Objectives of the present study

The primary objective of the present study is to obtain a better understanding of the causes and mechanisms of degradation of the aluminium strands of ACSR conductors exposed to marine environment under dynamic load, causing interfacial slip (rubbing). The study will also investigate the damage mechanisms of the strands and correlation of these with known fretting damages, such as fretting wear and fretting fatigue. Fretting and sliding wear mechanisms of the aluminium strand material in air and chloride media will be investigated. Moreover, the reciprocal interaction of wear and corrosion will be studied. Electrochemical and corrosion behaviour resulting from the variation of the slip amplitude and applied potential will be characterized by using surface characterization techniques, material loss, hardness measurements, and electrochemical techniques to improve the understanding of the underlying mechanisms and the degradation rate controlling factors.

In the light of the challenges presented in relation to the lifetime modelling of overhead power lines exposed to marine environments, the objectives of this study are to:

- a) analyse the failure of ACSR conductors exposed to marine environments and identify the root causes of failure and the failure types,
- b) study the sliding wear degradation of the aluminium strands used in ACSR conductors in air and NaCl solution,
- c) investigate the fretting degradation of the aluminium strands used in ACSR conductors in air and NaCl solution, and
- d) investigate the significance of electrochemical corrosion in the foregoing mechanical degradation processes expected by use of advanced electrochemical and surface-analytical techniques.

In view of the objectives outlined above, the thesis is organized and presented in the layout shown in Fig. 1.3. The content of each section is briefly described in the next sub-section.



Fig. 1.3. Thesis layout, sections and interrelationships between sections.

1.3 Structure of the thesis

The results obtained in this work are organized into three sections (§3 through 5), which are written in the form of self-standing manuscripts intended for publication. Therefore, some level of repetition, especially in the experimental parts, has been unavoidable. §2 reviews background information about the overhead power line components and basic terminologies used in the application and presents a literature survey of the causes of conductor failure and investigations related to material, design and construction, corrosion, vibration, wear and fretting. Information about the effect of corrosion on wear and fretting (tribocorrosion) behaviour of passive metals, with emphasis on aluminium alloys, is also presented and discussed. §3 reports the important conductor degradation parameters and the failure mechanisms obtained through failure analysis on failed samples obtained from real service environment. Systematic investigation of the contribution of the mechanical and electrochemical factors contributing to conductor failure is carried out and documented to understand the significance of tribocorrosion in the degradation processes. §4 investigates the sliding wear behaviour of the aluminium strands (AA 1350) in air, tap water and 3.5% NaCl solution at open circuit potential (OCP) and controlled potentials. The damage mechanisms occurring in the test environments are investigated through surface characterization, wear loss measurement and electrochemical response of the surfaces. The fretting behaviour of the aluminium strand material in air and chloride solution is presented in §5. A fretting map illustrating the fretting regimes, where fretting fatigue or fretting wear is expected, is presented. A summarizing discussion of the findings throughout this thesis and suggestions for further work are given in §6. Finally, the main conclusions of the thesis are listed in §7.

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2 Literature Review

2.1 Introduction

This chapter reviews the factors and mechanisms limiting the service life of overhead power line conductors exposed to marine environments, with emphasis on the degradation of the aluminium strands. The review emphasizes on wind induced vibration and corrosion induced degradations after giving a brief background information on overhead power line components, theory of the wind induced conductor vibration, fretting damages, sliding wear and tribocorrosion.

In general, the chapter reviews the existing literature relevant for the research results to be presented in the subsequent chapters. Each of these chapters contains backgrounds and justifications directly relevant to the specific objectives discussed. Therefore, the intention here is to present a general background on the known degradation mechanisms to put the forthcoming chapters in proper perspective. Earlier work in the area is summarized and critically discussed to justify the main objectives and conceptual framework of this thesis.

2.2 Overhead power lines

The need for transmission of power over large distances at higher voltage increases continuously [1]. This has led to an increase in demand for a network of overhead and underground power transmission lines to transport and distribute electrical energy. Since underground transmission lines are far more expensive for maintenance and installation, overhead transmission lines are extensively used. In addition, overhead power lines are efficient to transport electric power over large distances at higher voltages to consumers [2] located in commercial, urban, rural, and industrial areas as shown in Fig. 2.1 [3]. The main components of overhead power transmission lines, also called overhead lines, are the supports (towers), foundations, conductors, and insulator strings which are judiciously designed and selected for sustainable operation.



Fig. 2.1. Basic components of a power system constituting components from power generation to consumption [3].

2.3 Overhead power line components

Functionally, the structural components of a power transmission line can be categorized into two basic units, *viz.* the conductor and the supporting structure including foundation, towers, and insulators. The conductor is responsible for conducting the electric energy whilst the supporting structure holds one or more conductors above the ground [4] (Fig. 2.2 [5]). The supporting structure interfaces with the conductor through the suspension system comprised of insulator strings and clamps. The insulator element is anchored to the tower cross-arms at one end and clamp fitted on the conductor at the other end, see Fig. 2.2.

In addition to the suspension clamps, other clamp types such as tension clamps, spacer clamps and dead-end clamps are also directly attached to the conductor. Furthermore, fittings used for aviation warning, vibration recording, connection, and vibration damping are usually fitted to a conductor. The clamps and fittings directly attached to the conductor are collectively called hardware. The contact and interaction of these hardware, especially the suspension clamp, with the conductor forms basis for mechanical and corrosion degradations, which eventually lead to conductor failure, as explained further down in this chapter. The clamps are commonly fabricated from cast aluminium

alloys or galvanized ferrous metals to reduce galvanic coupling effect in a mild atmosphere [6]. The function of these devices is briefly outlined in the following paragraphs.



Fig. 2.2. Typical arrangement of different components and parts on a power line [5].

Tension clamps (Fig. 2.3a) [7] are fittings used for tension fixing, rerouting and grounding a conductor. They are made of threaded "U" bolts and nuts, a keeper and a clamp body as shown in the figure. During assembly, the conductor is deformed and fitted to the shape of the clamps. Consequently, plastic deformation, static bending stress, and traction pressure is induced in the conductor strands.



Fig. 2.3. a) Tension clamp. b) Suspension clamp [7].

Suspension clamps (Fig. 2.3b) are used for supporting a conductor on tower at intermediate locations. They have two main parts in contact with the conductor surface: the keeper and the clamp body as shown in the figure. A conductor is clamped between the keeper and the clamp body by means of the U-bolts and nuts as shown in the figure [7].

Spacer clamps (Fig. 2.4 [8]) are components used to hold bundled conductors in a proper configuration. These clamps often incorporate a damping element and function as both spacer and damper. Hence, they are often called spacer-damper used to control the vibration amplitudes within an acceptable level to avoid contact between conductors in out of phase oscillation [9].



Fig. 2.4. A suspension clamp configuration for 6 conductors [8].

Vibration dampers (Fig. 2.5 [8]) are used to damp vibration induced conductor motions.



Fig. 2.5. Vibration dampers, spacer clamp, and suspension clamp fittings on conductors [8].

2.4 Conductor constructions and materials

The conductor is one of the key items of a power transmission line serving both electrical and mechanical functions [10]. Economically, it accounts for an investment cost of about 20 - 40% of an overhead power line[11]. Traditionally, hard drawn copper had been in use for conductor construction due to its conductivity and good tensile strength. However, the cost, scarcity, and relatively low strength-to-weight ratio of copper have led the utilities to look for alternative solutions to achieve optimum electrical and mechanical requirements. However, it is practically impossible to obtain these properties in a single material. So, on the basis of combined properties of conductivity, cost, tensile strength, weight, and availability, aluminium has become preferable to achieve an optimum transmission efficiency than copper [10]. Hence, aluminium based conductors of different design and size are extensively used by utilities nowadays.

Aluminium based conductors can be classified according to their constructions and the material constituents. They are usually made of round aluminium strands helically and concentrically twisted together or around a central core intended to provide reinforcement depending on the operational requirements [12]. The commercially available conductors, (categorized in Fig. 2.6) exist in the form

of all aluminium conductor (AAC), all aluminium alloy conductor (AAAC), steel reinforced aluminium conductor (ACSR), and aluminium alloy reinforced aluminium conductor (ACAR) [10-12]. There are also emerging aluminium based conductor reinforced with a core made of composite material called aluminium conductor composite core (ACCC) [13]. The cross-sectional arrangement of the aluminium and steel strands, for example, in 6/1 and 26/7 ACSR types is schematically illustrated in Fig. 2.7. The choice for a given application is determined by operational requirements. For example, while AAC is more suitable for lightly loaded districts, ACSR is preferable in heavily loaded locations to transfer electricity over long distance [14].

Two types of aluminium alloys are generally used in conductor construction. The first type is heat treatable medium strength aluminium alloy (AA 6201-T81) which is used to make AAAC and the core strands of ACAR. The second type, and which is more common, is a commercial purity aluminium alloy, commonly known as electrical conductor grade aluminium alloy designated as AA 1350. It is used as a strand in the AAC, ACSR and ACCC [12]. The required nominal chemical composition of the constituent elements for both alloys is presented in Table 2.1 [15, 16].



Fig. 2.6. Classification of overhead power line conductors.



Fig. 2.7. Contacts between ACSR conductor strands. a) 6/1 ACSR. b) 26/7 ACSR.

Table 2.1 Nominal chemical composition (%) of AA 6201-T81 and AA 1350-H19.

Element	Cu	Fe	Si	Mn	Mg	Zn	В	Cr	Ga	v	Others,	Others,	Al,
										+ Ti	each	total	min
AA 1350	0.05	0.40	0.10	0.01	-	0.05	0.05	0.01	0.03	0.02	0.03	0.10	balance
AA 6201	0.10	0.50	0.50 - 0.90	0.03	0.60 - 0.90	0.10	0.06	0.03	-	-	0.03	0.10	balance

In the composite conductors, the reinforcement core carries the greater portion of mechanical load whereas the aluminium strands carry the bulk of electricity. The reinforcement, especially in the case of ACSR, has two functions. Firstly, it gives higher tensile strength combined with smaller sag, hence possibility to use in long spans. Secondly, a smaller sag enables use of shorter towers and provide higher thermal rating. Furthermore, the galvanic coating on the core prevents rusting of the steel and galvanic corrosion of the aluminium strands in marine environments [12]. Therefore, ACSR conductors have been the most preferable conductors in the power transmission and distribution systems. Nevertheless, the degradation of the aluminium strands at the clamps and fittings, especially suspension clamps, has been the major failure concern [17].

2.5 Conductor-clamp contacts

Design and construction of a conductor and clamping system are essential factors in the degradations leading to failure. The materials involved in the contacts of an ACSR conductor depend on the number of layers of aluminium stands, which can be one or more (Fig. 2.7) depending on the required

capacity. Accordingly, the contacts in a conductor can be lateral, inner and outer, as shown in Figs. 2.7a and b. For single layered ACSR conductors, (Fig. 2.7a), while the inner contacts occur between the surfaces of the galvanized steel core and the aluminium strands, the outer contacts occur between the outer layer aluminium strands and the cast aluminium alloy clamps [18]. The lateral contacts occur between adjacent aluminium strands in the same layer. The same is true for multi-layered conductors (Fig. 2.7b) as well, except for additional inner contacts between identical aluminium strands lying helically and concentrically on each other. In general, the contacts in a conductor can be seen as strand-to-strand and strand-to-clamp as shown in Fig. 2.8.



Fig. 2.8. Contacts in a conductor at a clamp.

The contacting surfaces of the strands, called contact interfaces, experience clamping pressure from the installation as static compressive stress and plastic deformation of elliptical and rectangular shape on the inner (point) and lateral (line) contacts, respectively. The contacts start as a point or line and progressively increase in area as the clamping load increases. This process eventually creates elliptical plastic deformation marks at the point contacts and rectangular plastic deformation marks at the line contacts. From various conductor tests and failure analysis works, such plastic deformation marks are clearly visible on both outer (Fig. 2.9a) and inner (Fig. 2.9b) layers of strands [19]. Such permanent deformation influences the subsequent degradation by inducing initial residual stress and changing local mechanical property of the material. Hence, the tightening torque applied during installation is an essential factor in the degradation of a conductor.



Fig. 2.9. Contact surface of a failed ACSR conductor strands at a spacer clamp. a) View of the plastic deformation on the outer layer aluminum strands. b) View of the inner layer aluminum strands showing elliptical fretting marks (see arrows pointing at dark oxide) located at the end of the clamping region [19].

In general, the conductor - clamp assembly design, material properties and the environment establish the contact conditions, which constitute the essential parameters determining the contact damage [20]. The environment contributes to conductor degradation at the clamps in two ways. Firstly, it causes wind induced vibration on the conductor and leads to fatigue and wear degradations. Secondly, it causes corrosion damages in marine and industrial (polluted) areas. Therefore, the contacts in the conductor-clamp assembly are known as the critical locations where the failures are concentrated. Specifically, the segment between the last point of contact (LPC) and keeper edge (KE) (Fig. 2.10) has been identified as the location where conductor strands usually break, as presented in §1 [21, 22] under a wind induced vibration.



Fig. 2.10. Illustration of conductor and suspension clamp assembly, showing components, keeper edge (KE) and last point of contact (LPC). R is the radius of curvature of the clamp body.

2.6 Wind-induced conductor vibration

Wind induced vibration has been known as the root cause of fatigue failure of overhead power lines. Varney [23, 24] presented theoretical explanation of the vibration in terms of formation of turbulent eddy currents of wind blowing across the conductor and the armour rods at the clamps used for its protection. The vibrations responsible for the degradation of a conductor are classified as Aeolian vibration, galloping and subspan oscillations [25], depending on the motion resulting from the interaction between the conductor surface and the wind. Swaying wind, climatic conditions, and conductor configuration are the parameters which determine the mode of vibration. The Aeolian vibration and galloping affect all types of conductors, whereas the subspan oscillation influences only transmission lines with bundled conductors lying in the same horizontal plane [26]. Regardless of how they are formed, interact with the conductor and classified, all vibrations essentially trigger oscillatory bending or torsion of the conductor with respect to a clamp and lead to dynamic loading and interfacial displacement at the contacts interfaces of the strands, thus, creating fatigue and wear degradations of the strands.

According to Lindley's failure analysis report [27], aeolian vibration is a small slip amplitude conductor motion responsible for fretting at the inter-strand interfaces whereas the subspan oscillation is a very severe form of vibration causing bending of a conductor with respect to clamps and fittings, or even galloping. He also reported that the subspan oscillation mode of conductor vibration occurs at a frequency of about 1 Hz whereas the aeolian vibration mode involves a frequency of about 20 Hz. This agrees with another report showing a frequency range of about 3 - 150 Hz for aeolian vibration and 0.5 - 1.5 Hz for both subspan oscillation and galloping [28].

2.7 Loads acting on conductors

Conductors experience static and dynamic loads at the strand-to-strand and strand-to-clamp contacts. The sources of static loads are the axial tension from installation and span weight, bending load (due to change in curvature of the clamp), clamping pressure, and residual stresses from conductor fabrication [29]. These are normally defined by constructional details and do not require further tracking during the service life of a conductor. The dynamic load is caused by wind induced vibrations [11]. It is caused by environmental conditions, as discussed in the preceding section and can be predicted from statistical models and meteorological databases [4]. The action of both the dynamic and static loads generates a cyclic tangential force and relative slip on the contact surfaces of the strands. This leads to a severe form of degradation called fretting, which accelerates the fatigue failure of conductor strands [1].

2.8 Degradation mechanisms

The aluminium strands of ACSR fail due vibration and corrosion induced degradations, as summarized in Fig. 2.11. The figure shows that vibration induced degradations can be classified into fatigue and wear types. While the fatigue degradation can further be classified into high cycle and low cycle fatigue, the wear degradations can be classified as fretting wear or sliding wear. Similarly, the corrosion induced degradations documented so far are mainly localized corrosion (pitting and crevice corrosion), galvanic corrosion, and atmospheric corrosion, in the present case due to pollution.



Fig. 2.11. Classification of conductor degradation.

As generally known, stressed metallic structures exposed to corrosive environments experience premature breaking, a characteristic called stress corrosion cracking. Similarly, the fatigue resistance of metallic structures carrying fatigue load in a corrosive environment reduces significantly, a condition called corrosion fatigue [30]. Fretting is a surface degradation characterized by fatigue or wear depending on the contact parameters. While fatigue degradation deals with initiation and propagation of cracks in a structure, wear is associated with detachment of material from a surface in the form of debris. Both fretting wear and fatigue degradations are influenced by corrosion, a phenomenon called "fretting corrosion". It must be noted that the wear aspect is limited under fretting fatigue conditions due to a very small interfacial displacement at the contact. Therefore, corrosion may influence the degradation at the crevices forming at the periphery of the contact where cracks nucleate and initiate.

Fretting is a degradation phenomenon arising at the interface of two contacting surfaces in a periodic minute relative displacement, or "slip" under a normal force and a cyclic tangential force [31]. It occurs usually in such connections as clamped, bolted, riveted and press-fitted, which are not intended

to move against each other. If it is caused by and /or mutually acts with a bulk cyclic load, the phenomenon is called "fretting fatigue", in which fretting enhances crack initiation and growth and the bulk cyclic load controls the propagation. The factors influencing fretting are normal load (contact pressure), displacement amplitude, type of materials involved, surface condition, environment, and the dynamic loading amplitude. For additional details, the reader is referred to reference [32]. In general, fretting related degradations can be classified as fretting fatigue, fretting wear, and fretting corrosion.

Tomlinson [33] recognized for the first time, that the fretting damage occurs due to a minute relative slip, as small as a submicron. It occurs in a range of displacement amplitude bounded by lower and upper limits at a given normal load [34]. The degradation mechanism changes from fretting fatigue to fretting wear as the displacement amplitude increases [35]. Moreover, fretting induced degradations can be classified according to the mode of the interfacial displacement as tangential, radial, rotational, or torsional fretting as illustrated by the arrows in Fig. 2.12. In reality, the displacement may involve two or more of these modes. However, most studies on fretting are restricted to the effect of a single displacement mode, particularly tangential fretting [36].

The magnitude of the displacement amplitude dictates the degradation mechanism by influencing the stress and strain state on the contact surface. An increase in the displacement amplitude leads to a critical tangential stress which initiates surface cracks, which acts as a stress concentrator and reduce the fatigue strength of the material [37]. Further increase in the displacement overstrains the surface and causes wearing away of the generated cracks instead of growing them, hence reducing the stress concentration [38]. This process leads to a change in the surface degradation mechanism from high to low cycle fatigue and back to the high cycle fatigue mode with increasing displacement amplitude, as explained in relation to fretting regimes later. Above the upper limit of the displacement amplitude causing fretting, the degradation mode transforms from fretting wear to reciprocating sliding wear, as shown in Fig. 2.13 [39]. In effect, fretting wear differs from a reciprocating sliding wear by occurring at extremely small relative slip and very low sliding velocity. It also differs by the fact that the wear debris becomes trapped within the contact interface, whereas the debris is continuously eliminated in the reciprocating sliding wear. Moreover, the wear rate increases with increasing magnitude of the displacement amplitude under the fretting wear condition, whereas it remains nearly constant in the sliding wear condition, as shown in the figure. Therefore, the magnitude of the
displacement amplitude arising at the contact interface is a key parameter determining whether fretting fatigue, fretting wear or sliding wear is the prevailing mechanism [39].



Fig. 2.12. Schematic illustration of basic fretting modes, P is the normal load, and D is the displacement indicated in the direction of the arrows [36].

Conceptually, fatigue degradation of metallic structures is a fracture or cracking caused by a cyclic action of a load far less than the static strength of the material. The action of the cyclic load causes fatigue failure by nucleating and growing microscopically small cracks, which is followed by propagation of the macro cracks and ultimately complete failure of the structure. In other words, fatigue is a degradation process resulting from gradual accumulation of a damage involving crack initiation phase, propagation and eventual failure. The sum of the number of load cycles required for each of these phases gives a fatigue life of the structure. Technically, fatigue life of a structure can be classified into crack initiation and crack growth (propagation) periods as shown in Fig. 2.14 [40]. Since the structure cross section is severely diminished by crack growth at the final failure phase, the structure fails suddenly by static overloading, hence its contribution to fatigue life is marginal.

The crack initiation in the fatigue process is a surface phenomenon whereas the surface conditions may not be so relevant for the subsequent crack propagation [40]. Corrosion may contribute to both crack initiation and crack propagation. However, fretting is critical for the crack initiation phase but has little or no influence on the subsequent crack propagation. Generally, fretting damage is a surface phenomenon influenced by the condition of the surface [41]. The crack initiation phase ends when the microcrack growth becomes independent of the surface condition. It is, then, followed by the propagation phase which depends on the bulk material properties, i.e., resistance to crack growth is determined by the stress intensity factor, designated as K (Fig. 2.14).



Fig. 2.13. Schematic illustration of the variation in fretting fatigue life and fretting wear rate with relative slip amplitude (Δ) ranges in different fretting regimes [39].



Fig. 2.14. Phases of fatigue life and related factors [40].

2.9 Fretting Damages

Fretting induces surface cracks at the edge of a contact surface and accelerates fatigue failure of structures. Fatigue degradation, in which fretting, generated by the fatigue load itself, causes crack initiation, is called fretting fatigue [42]. In effect, it changes a high cycle fatigue failure into a low cycle fatigue by accelerating the crack initiation period. Fretting fatigue differs from plain fatigue by the fact that the crack initiation always occurs at a contact surface by a minute cyclic relative displacement amplitude. Moreover, multiple cracks of different length and orientation may initiate in the contact region and grow competitively, of which one dominant (the longest) crack propagates and eventually causes fracture or failure. The crack initiation is influenced by the prevailing fretting regime, which depends on the displacement amplitude, in the contact (stick, mixed slip, and gross slip regimes).

Fretting is one of the surface effects influencing crack initiation process in which surface cracks incubate (nucleate) and grow to embryonic stage [43]. The cracks generated by fretting, which are inclined to the surface, act as stress concentrators and dramatically reduce the fatigue resistance of the material [37]. Crack initiation involves sub-processes such as cyclic slip, crack nucleation and micro crack growth, where fretting is a driving factor contributing to each of these processes [44]. There is no consensus about the factors which determine the total fretting fatigue life, e.g., whether it is dominated by crack propagation or crack initiation. For example, Goto and Endo [45] reported that the crack initiation period accounts for about 20 - 25% of the fretting fatigue life based on electrical resistance measurement and scanning electron microscope observation of a surface in a fretting fatigue life is an initiation controlled process. Nevertheless, there is general agreement that fretting is the critical factor limiting the fatigue life of a structure.

Fretting fatigue cracks initiate at the edge of the contact surface where the tangential force is highest [47]. The tangential force is the dominant factor for crack initiation and grow. Physically, cracks initiate at the boundary between the stick and slip contact surfaces of the mixed stick slip contact condition [48], which agrees with the theoretical analysis and experimental evaluation [39]. Large number of cracks initiate at the edge of the contact surface and grow oblique to the surface under the combined action of the cyclic tangential and bulk stresses. When the contribution of tangential force (fretting) to the crack growth ends, the crack forms a knee point and becomes nearly perpendicular to

the contact surface. At this stage, the crack propagates under the action of the cyclic fatigue load [45]. Furthermore, the higher the tangential force, the higher the reduction in fatigue strength of the material it causes and the lower the number of cycles needed for crack initiation.

Occurrence of fretting fatigue depends primarily on the magnitude of the displacement amplitude at the contact interface. Gao *et al.* [49] studied the influence of displacement amplitude on the fretting fatigue behaviour of several alloy steels. They found that the tangential stress varies in phase with the alternating internal stress in contacts subjected to relative displacement within the elastic range. Subsequently, proportional loading results at the contact surface. When the relative displacement amplitude was increased to the partial slip (mixed stick slip condition), the variation of the tangential force and alternating fatigue cracks. By further increasing the relative displacement, the authors noted rather deep wear scar and severe wear damage at the contact surfaces than crack formation. Similarly, the influence of displacement amplitude on the resulting fretting degradation was reported by Kennedy *et al.* [50] and Vingsbo and Söderberg [39] in the 1980s. Therefore, the magnitude of the relative displacement has been recognized as the critical parameter determining the type of fretting damage, *i.e.*, fretting fatigue or fretting wear.

The loads acting in a fretting contact induce overstress and overstrain depending on the magnitude of the displacement amplitude. Overstressing is associated with crack nucleation and initiation (fretting fatigue). Whereas, overstraining is pertinent to surface degradations involving subsurface microstructural transformation (strain hardening) and wearing of the contact surface (particle detachment) [51]. The wear process generates debris of metallic particles (which quickly oxidize) called "third-body" at the interface. Unlike sliding wear, the debris is accommodated at the interface under a fretting condition. Consequently, the two rubbing surfaces (also called "first bodies") become totally or partially separated and form a "three-body" contact. The third-body layer governs the subsequent degradation by influencing the electrical, rheological, mechanical and electrochemical properties of the interface. For example, the accumulation of oxide debris at the contact interface of aluminium strands causes very high electrical resistance thereby leading to arching and rise in temperature [34].

The formation of debris is the basis for the so called third-body concept of analysing wear and friction behaviour of fretting contacts, introduced by Maurice Godet [52]. According to this concept, the

generation and elimination of debris are competitive processes governing fretting fatigue and fretting wear degradations [53]. Based on the behaviour of the third-body, three essential stages of wear has been suggested. The first stage is where detachment of particles by the classical wear mechanisms (adhesive wear, fatigue wear, abrasion wear and tribochemical wear). The term tribochemical wear refers to a material removal process in which chemical or electrochemical process contributes to the wear degradation, nowadays called tribocorrosion [54]. In the second stage, the debris of metallic particles are trapped and circulate in the contact interface, building up and forming a powder bed (third-body), which reduces the wear, as discussed above. Finally, elimination of debris from the interface occurs, and this leads to a restart of the above process, causing a wear cycle [55]. It was reported, for aluminium alloys, that the wear process begins by adhesion [56] followed by the removal of material by delamination. The detached particles may be transferred to the first-body surfaces, ground down and oxidized, or they become loosely attached particles and flakes of oxide coated metal fragments [57, 58].

Fretting wear mechanism is a three-step degradation process, as summarized below [59]:

- 1. *Adhesion and metal transfer stage* in which the protective oxide-metal surface layer is dispersed by the wear process and results in adhesive contact. The process depends on the hardness of the oxide and metal. The adhesion coefficient is determined by the mechanical and physical properties of the contacting materials. However, the metal transfer and resulting plastic deformation is determined by the strength of the contact. This initial stage dictates the subsequent wear damage.
- Production of oxidized debris the surface deformed in the previous stage breaks down into debris oxidized in the process, which occupies the interface and reduces the direct metallic contact.
- 3. *Steady-state stage* in which the zones affected by the initial stage become disintegrated and dispersed, resulting in a mixture of debris and metal in the wear track.

Fretting causes changes in the mechanical and electrochemical properties of a surface. For example, Sauger *et al.* [60] reported that the debris is generated from a very thin superficial layer called tribologically transformed structure (TTS). The layer forms during the first few cycles of fretting on metallic surfaces including aluminium alloys [61]. Advanced surface characterization revealed that TTS is a deformed nanocrystalline layer with a chemical composition identical to the base material. On a fretted surface, it is located between the wear debris and plastically deformed layer on the base

metal, as illustrated in Figs. 2.15a and b. It is substantially harder than the base material. Formation of such deformed nanocrystalline layer on aluminium alloys, including those with high and commercial purity grade, by near surface plastic deformation or surface processing operations such as cold working and abrasion has been well recognized and characterized [62-64]. Moreover, it is known that such ultrafine grain layer significantly influences the electrochemical and corrosion properties of aluminium surface with subsequent heat treatment.



Fig. 2.15. Illustration of material transformation by wear process. a) Cross-sectional view of a fretted surface after chemical etching showing location of TTS. b) Axial secitional view of a fretted surface with oxide debris (unetched) [60].

Vingsbo and Söderberg [39] introduced the concept of fretting map which is used as a tool for investigation of the state of fretting contacts, based on the plot of tangential friction force against the slip "displacement amplitude" at the contacts. The study was conducted by imposing a constant displacement amplitude, normal force and frequency of oscillation and then measuring the tangential force in each fretting cycle. The tangential friction force, simply called "tangential force", is generated as a result of relative displacement at the interface and it is measured at a constant normal force in every fretting cycle. A plot of the tangential force vs. displacement amplitude results in linear, elliptical, or parallelogram shaped curves as shown in Fig. 2.16a. A plot of the normal load against the displacement gives a "fretting map". It reveals three essential fretting regimes, characterized by the domain of the displacement amplitude *vs*. the normal load resulting in a similar fretting curve.

This results in three fretting regimes known as the stick regime, mixed stick slip regime, and gross slip regime [39]. As can be seen from Fig. 2.16a, the corresponding fretting curve was a straight line for the stick regime indicating elastic accommodation of the imposed displacement, hence no slip at the contact interface. The fretting curve was characterized by a hysteresis loop of elliptical shape for the mixed stick slip regime and a deformed hysteresis loop with parallelogrammatic shape for the gross slip regime. The loop deformation for the gross slip regime was attributed to sudden drop in the tangential force above a certain displacement amplitude due to yielding of the contact surface. The imposed displacement amplitude is accommodated by elastic, elasto-plastic and plastic deformation of the contact surface in the stick, mixed stick slip and gross slip fretting regimes, respectively. The state of a fretting contact can be classified as "partial slip regime" or "gross slip regime" depending on whether the centre of the contact moves (slips) or not. This classification is called the fretting running status (cf. Fig. 2.16a) [65].

The curves generated at a given test condition may not be reproducible due to irreversible changes in the surface and hence the coefficient of friction during the test. Therefore, another fretting contact analysis approach, called "friction log" has been developed (Fig. 2.16b). This is a three-dimensional plot of tangential force *vs* displacement amplitude and number of cycles. It reflects the evolution of the fretting curves during a given fretting test thereby giving insight into the variation in the interface condition through the shape of the curves. Accordingly, the state of a fretting contact can be classified into partial slip, mixed, and gross slip regimes as illustrated in Fig. 2.16a. The mixed regime is associated with unstable friction log (variation of the fretting curves in shape from one to another), as schematically illustrated in. Fig. 2.16b [65]. Based on the evolution of the friction logs, the running condition (based on contact kinematics) and material response fretting map (based on surface degradation mechanisms) (Fig. 2.16d) were introduced [66]. The material response fretting map shows zones of normal force vs displacement domains where a fretted material is expected to show no cracking (stick), cracking (mixed stick slip) and wear (gross slip) used throughout this thesis.

The sample surfaces in Fig. 2.16c experienced limited damage by oxidation and wear in the stick regime. The mixed stick slip regime, also called the intermediate regime, showed marginal wear and oxidation, but cracking occurred at the edge of the contact. In contrast to the other mixed stick slip regimes, the contact surface subjected to the gross slip regime underwent significant wear degradation enhanced by oxidation and marginal surface cracking. This indicates that the friction logs or fretting curves can be used to predict the conditions in which wear or fatigue degradation is expected.



Fig. 2.16. a) Classification of fretting regime into partial and gross slip fretting regimes based on kinematics at the contact interface. b) Classification of fretting regime into partial slip mixed and slip regimes as per evolution in the friction logs. c) Fretting degradations in partial slip, mixed, and slip regimes. d) Running condition and material response fretting maps [65]. F_t - is tangential force, D - is relative slip (displacement), F_n - is a normal force and N - is the number of cycles.

2.10 Sliding wear

In tribology, sliding is a term used to describe a continuous (unidirectional) or reciprocating (back and forth) movement of two contacting surfaces in a relative motion. Unlike fretting wear and fatigue degradations, the literature available on sliding wear of overhead power line conductors is quite limited. A large body of literature on the subject is concentrated on aluminium-silicon alloys and aluminium based metal-matrix composites which are widely used in automotive and aerospace industries [67]. Furthermore, most of the studies have been performed by using unidirectional rotating pin on disk, which is not relevant for the type of relative sliding expected in ACSR conductor subjected to wind induced vibration. However, some of the results are reviewed below to highlight the relevant parameters and available methods.

A review [68] of sliding wear behaviour of aluminium-based composites in dry conditions indicates that adhesive wear and delamination wear are the two dominant wear mechanisms. The dynamics of the wear process changes due to the formation of stable third-body layer on the contacting surface, in which the fine equiaxed wear debris causes high wear resistance [68]. Hence, the material removal rate increases first and declines gradually. Further, the critical parameters affecting the wear regime are identified as applied load, sliding speed, surface hardness, reinforcement fracture toughness and surface morphology. Pits and subsurface cracks nucleated during the first 1-10 passes in the small region underneath the sliding surface, and these became deeper in the subsequent passes [69]. The depth of pits increased as the normal contact load and the coefficient of friction increased. The wear rate was controlled by either nucleation of pits or subsurface propagation of crack depending on the material and sliding conditions. Similar wear mechanism was reported for Al-Si alloys [70]. Kumar et al. [71] reported that addition of copper increased dry sliding wear resistance of aluminum. Further, the sliding distance, sliding speed and applied load were reported to influence the wear properties of these alloy types. Moreover, investigation of sliding wear behavior of aluminum pins (high purity aluminum, aluminum alloy 1100 and Al-l at.% Mg) sliding against rotating disks of M2 tool steel showed that material removal takes place predominantly by dragging out metal at the exit end of the wear track [72]. The test showed that the friction coefficient was identical for all the three pins in steady state wear condition and formation of subsurface deformation.

Sliding wear changes the surface and subsurface behavior and contributes to mechanical and electrochemical degradations. Rice, Newtony, and Waybe [73] identified three distinctive subsurface

characteristic zones for materials subjected to sliding and impact wear. These are the superficial zone with mixed composition, the next zone with refined grains and plastically deformed, and the base metal.

2.11 Tribocorrosion

Historically, the origin of tribocorrosion (tribo-electrochemistry) is associated with material surfaces coated with a passive film and destruction or modification of these surfaces by mechanical loading. Landolt *et al.* [74] defines tribocorrosion as "an irreversible transformation of a material resulting from simultaneous physico-chemical and mechanical interactions in tribological contact" in relative motion. More plainly, Mischler [75] suggested that tribocorrosion is a degradation of material resulting from mutual action of wear and corrosion. These definitions demonstrate that tribocorrosion constitutes two basic components: corrosion (electrochemical/chemical) and wear (mechanical) degradations as presented in Fig. 2.17 [76]. The mechanical component is usually sliding wear, fretting, abrasion, cavitation, tribo-oxidation, or solid particle erosion [76, 77].



Fig. 2.17. Principal concept and definition of tribocorrosion [76].

An essential feature of tribocorrosion is that the mechanical process affects the corrosion process and/or *vice versa*. Thus, the degradation is characterised by synergism resulting from coupled mechanical and electrochemical/chemical effects. Therefore, the total material loss may be larger or smaller than the mere addition of the individual components due to the synergy effect which can be positive or negative depending on the specific reaction and/or wear product generated in the contact. For example, it is positive when the material removed from the surface can be oxidized, form an oxide bed, and reduce further chemical dissolution or mechanical wear of the contact surface. It is negative when the detached particles act as hard abrasive particles and accelerate both electrochemical and mechanical wear degradation. The total material loss is the sum of material loss by wear without corrosion, the material loss due to corrosion without wear and synergy effect which accounts for a change in the rate of corrosion due to the presence of wear and/or *vice versa* [78].

According to Ponthiaux *et al.* [79] the mechanism of synergistic degradation of contacting surfaces can be explained by the following five mechanisms.

- 1. The generation of debris in a wear process accelerates or reduces the wear process,
- 2. Formation of galvanic coupling between the worn and unworn areas increases anodic dissolution of depassivated area of the surface in a corrosive environment.
- 3. Galvanic coupling between the two counterparts may occur.
- 4. As a result of formation of soluble corrosion products, the corrosive environment may become increasingly aggressive or the chemistry of the interface environment may be altered.
- 5. Strain hardening resulting from the mechanical loading in the contact area may alter the electrochemical kinetics and/or repassivation processes.

Characterization of tribocorrosion degradation necessitates multidisciplinary investigation approach by the complexity nested in it by synergism. It involves aspects of materials, mechanical/operational variables, solution/environment, and electrochemical parameters are the key factors affecting the tribocorrosion degradation of a sliding contact subjected to electrochemical control as illustrated in Fig. 2.18 [74, 76].

Nowadays, investigation of mechanical and electrochemical response of the surface can be carried out by using a tribometer facility enabling continuous measurement of the evolution of frictional force (tangential force) and electrochemical properties as schematically illustrated in Fig. 2.19 [76, 80]. The electrochemical unit interfaced with the tribometer constitutes a potentiostat along with a reference electrode, the counter electrochemical polarization studies. Therefore, a set of parameters

can be selected depending on the purpose of the investigation, a given set of technical data (mechanical and electrochemical) can be measured, online output parameters can be recorded and used to obtain results which can be analysed and interpreted. The general steps used in a tribocorrosion study is summarized and presented by Mathew *et al.* [76] as in Fig. 2.20.



Fig. 2.18. Schematic illustration of parameters influencing tribocorrosion behavior of sliding contacts under electrochemical control [74].



Fig. 2.19. Schematic illustration of experimental rig for study of tribocorrosion [76].

Tribocorrosion induced degradation is encountered and investigated in several applications which can fall under the categories of tribocorrosion in a living system and tribocorrosion in industrial systems [76]. A great deal of literatures published on tribocorrosion so far have been concentrated on investigation of materials used for medical implants, nuclear, aeronautic and automotive industries [81-84]. Information available on aluminium alloys is scarce. If any, they are mostly on aluminium alloys used in other applications than electric conductor grade aluminium alloys. Although this area is not widely investigated and documented for overhead power line conductors exposed to marine environments, it is clearly expected to occur in the mutual presence of corrosion and wear. Therefore, this review focuses on the tribocorrosion of passive metals with particular emphasis on aluminium alloys.



Fig. 2.20. Basic procedures in tribocorrosion study [76].

Fretting influences the electrochemical processes on aluminium alloys held in an electrolytic solution. For example, Bethune and Waterhouse [85] studied fretting corrosion (tribocorrosion) behavior of commercial purity aluminum and its alloys in an aerated 1% sodium chloride solution using a reciprocating tribometer set up under potentiostatic control. The open circuit potential was measured before the start of fretting and the same potential was maintained during the test. The study showed that fretting caused increase in the negative current until it achieved nearly a constant negative value as the fretting cycle continued as shown in Fig. 2.21. The fluctuation in the current was attributed to cyclic rupturing of the passive oxide film by rubbing. The negative current increased sharply but to a value less than the initial static current when fretting was terminated, indicating fast restoration of the oxide film.



Fig. 2.21. Potentiostatic fretting wear of commercially pure aluminum *vs*. time in 1% NaCl [85].

Galvanostatic polarization of aluminum surface, measured in 0.1 N sodium chloride solution with and without abrasion (by silicon carbide abrasive paper) [86], showed substantial decrease in the corrosion potential to a very negative value and significant increase in the corrosion current (about 30 times) in the presence of abrasion on the surface compared to without abrasion (Fig. 2.22). The figure shows that sliding (abrasion) significantly increases the electrochemical process on the surface of aluminum alloys. Waterhouse [86] used also the galvanostatic method to study the effect of pH on the fretting corrosion of aluminum alloys. He reported that the additional effect of fretting on corrosion in the low and high pH ranges is limited due to rapid corrosion of aluminum. However, fretting causes significant decrease in the surface potential in the intermediate pH range where the passive oxide is stable. In fact, the pH dependent variation of effect of fretting on the electrochemical process agrees with the well-known thermodynamic behavior (Pourbaix diagram) of aluminum in seawater. In effect, wear contributes by damaging the protective oxide film in the neutral pH range. Liu *et al.* [87] studied the tribocorrosion behaviour of AA 7075-T6 alloy in 3.5% NaCl solution using pin on disc tribometer at potentials in the range -1.68 to -0.18 V(Ag/AgCl), where -0.68 V(Ag/AgCl) was the open circuit potential. The wear rate increases with the potential in the anodic range. At the cathodic potentials and OCP, however, the wear rate was nearly constant. The measured current remained nearly constant during the test for all potentials except for -1.68 V (Ag/AgCl) where the current increased swiftly when the potential was imposed and became constant afterwards. It was negative and positive for the cathodic and anodic potentials, respectively. There was no noticeable change in the current transient at the start and termination of the sliding wear, suggesting that the electrochemical response was controlled by the applied potential but not wear in all cases. However, the morphological analysis of the surface outside the wear track showed pitting around the intermetallic phases at the OCP and applied cathodic potential except at -1.68 V (Ag/AgCl) where intergranular corrosion attack was reported. Although this work was on high strength aluminium alloy, it demonstrates the role of applied potential, mechanical wear, material composition and electrochemical response on the tribocorrosion of the alloy.



Fig. 2.22. Galvanostatic polarization of commercially-pure aluminium with and without abrasion wear in 0.1 N NaCl [86].

Mraied *et al.* [88] investigated the effect of Manganese concentration (5.2 and 20.5 at %) and microstructure on the tribocorrosion resistance of aluminium in a 3.5% sodium chloride solution. They reported that addition of proper amount of manganese substantially improves the tribocorrosion

resistance of aluminium by solid solution strengthening and microstructural refinement. The tribocorrosion resistance was due to formation of nanocrystalline and fully amorphous microstructures formed by the addition of 5.2 and 20.5 at% Mn, respectively. They found also that the material degradation by tribocorrosion of Al-Mn alloy increased with the applied potential. Mechanical wear dominated at the cathodic potentials. In contrast, at anodic and open circuit potentials, active corrosion of locally depassivated area of the wear track (by mechanical wear) resulted in accelerated material loss. Moreover, the authors noted that the higher the concentration of manganese the faster self-repair kinetics of the depassivated area. The work illustrates the significance of the alloying elements in aluminium in determining the tribocorrosion resistance of its surface.

Commercial purity aluminium alloys (AA 1xxx series) generally contain Fe-rich intermetallic particles as impurities [89]. These form local microgalvanic coupling with surrounding aluminium matrix and lead to localized corrosion in chloride solutions. In the microgalvanic cell, the secondary phase and the surrounding solid solution matrix act as the cathodic and anodic sites, respectively. Consequently, the matrix around the secondary phase particles corroded to form shallow pits. The pits passivated when the phases became undermined by corrosion [90]. However, the surface defects left behind may form the basis for crack initiation and increase the resistance to current transfer in the conductor strands. Therefore, the material composition and microstructure of the strands is crucial to the corrosion and fatigue degradations of the strands.

The electrochemical properties of freshly generated and bared aluminium alloy surface from either sliding on the surface or advancing crack where passivity is mechanically disrupted is of practical interest to understand the fretting corrosion and tribocorrosion characteristics of the aluminium strands subjected to Aeolian vibration in a marine environment. Potential transients for aluminium wires guillotined in buffered phosphate solution at pH 6.9 were used to investigate the behaviour of the freshly generated surface and its subsequent evolution with time [91]. The open circuit potential dropped instantaneously to a very negative value of about - $1.76 V_{SCE}$ and then recovered linearly with log time due to repassivation of the bared surface. Such potential drop, as low as to - $1.35 V_{SCE}$ was reported to result in cathodic corrosion, particularly alkaline pitting on the aluminium alloy surface [92]. Similarly, the sliding and fretting process on the Al alloy strands, which are expected to generate bare metal surface, have been reported to cause potential drop due to similar changes in the electrochemical process on the surface [85, 93].

Similar to metal working, sliding wear on various metallic surfaces showed formation of subsurface layers composed of fragmented metal and oxide debris [94]. These layers have different electrochemical and mechanical properties from the bulk aluminium. The sliding on the surface has been reported to increase the hardness of the surface in the wear track similar to the cold working process. Subsurface layers produced by wear in service are impossible to detect and take corrective measures before they contribute to further deterioration of the aluminium strands of a conductor.

2.12 Lubrication in aqueous environment

Seawater may act as a lubricant and reduce the wear and fatigue degradations [95]. Mindivan *et al.* [96] reported that the friction and sliding wear of aged aluminium alloys of type 2618, 7012, 7075, which were reinforced by 50 vol% SiC, was significantly reduced in water compared to that in air. Similarly, fretting investigation of a 0.64% C high tensile strength steel rope in an aqueous environment and air showed that the aqueous media lowered friction and wear by acting as a lubricant. According to the study, the liquid environment acts as a lubricant, a quality which is similar for all dilute solutions independent of the fretting process. The suggested mechanism was that the liquid environment separates the counterparts, thereby reducing adhesion, wear rate and coefficient of friction [97].

The lubricative effect of seawater has been reported on other materials. Evaluation of a sliding wear behavior a plasma sprayed-carbon nanotube-reinforced aluminum oxide coatings in seawater and dry air showed that the seawater solution reduced the coefficient of friction and wear damage [98]. Similarly, evaluation of friction and wear of diamond-like carbon films in air and water showed that friction and wear rate was much smaller in water than in ambient air environment [99]. The lubricating effect of aqueous solutions, compared to that in air to reduce friction and wear was also reported for fretting wear of bearing steel in air and aqueous environments [100]. Furthermore, investigation of the influence of aqueous environment on the fretting damage of cylindrical roping steel strands in air and NaCl solution showed that water has no effect on the contact conditions at smaller displacement amplitudes but acts as a lubricant for fretting at higher strokes [101]. The study also showed that the coefficient of friction in the gross slip regime and its transition boundary is significantly reduced in water compared to that in air. Similarly, Fan *et al.* [102] reported that seawater, compared to dry air, decreases the fretting wear behavior of a 304 stainless steel by strongly suppressing adhesion by its cooling and lubrication effect.

Generally, the available literature indicates that the marine environment reduces the wear and fatigue damages. It worth noting that friction and wear are not generic material properties and behavior but results of a tribo-system, explained "in terms of roughness effect, hardness effect, ductility effect, oxide film effect, reaction layer effect and transfer effect" as stated in the review by Kato [103].

2.13 Literature survey: Vibration induced degradation of ACSR conductors

Wind induced vibration inflicts fretting damages on conductor strands at their contact interfaces. Understanding and solving these degradations necessitates study of the fatigue and wear mechanisms, in particular *fretting fatigue and fretting wear* [27]. In extreme situations, the relative slip amplitude may exceed the range causing fretting wear and lead to a reciprocating sliding wear at the interface [39], though it is not as frequent as fretting wear. Furthermore, the contribution of corrosion to these degradations is unavoidable in marine environments. Nevertheless, studies combining the effect of these phenomena on overhead power line conductors is virtually unavailable in the open literature.

The range of displacement amplitudes leading to fretting fatigue or fretting wear could not be determined by the available investigation methods for the conductors. Studies of fretting wear behaviour of the material used for the strand in similar environments are also quite limited in the available literatures. Most studies have been focussed on fretting fatigue. However, studies on fretting [31, 35, 104] show that the damage mechanism transforms from fretting fatigue to fretting wear as the slip amplitude increases (cf. Fig. 2.13). The transition is influenced by factors such as material and environment besides the normal contact load and slip amplitude. Hence, a general background on these essential features are needed.

Historically, fretting damage investigations on overhead power lines have dealt with fatigue damages resulting from cyclic bending stress. Since vibration induced fatigue failure was noticed in the late 1920s [23, 24] field survey and laboratory investigations have been conducted in order to establish the theoretical basis, contributing factors, experimental methods to simulate the phenomena, and mitigation mechanisms. Monroe and Templin, [105] reported methods for the measurement of actual stresses in a vibrating conductor. The measured stresses were verified against analytically determined values in an attempt to establish a safe design limit. Furthermore, these authors reported that using dampers is viable to mitigate such failures.

In the late 1930s, Double and Tuck [17] showed that fatigue failure of aluminium strands by Aeolian vibrations was accelerated by "nicking", a term used to refer to fretting, between the strands caused by the clamping stresses, and to a lesser extent by cyclic contact (hammering) of the conductor surface with the clamp. They suggested that the influence of nicking can be alleviated by using clamps with larger radius of curvature at KE and LPC to avoid pivoting action when the conductor bends. In later work, clamp geometry was further established as one of the detrimental factors contributing to conductor damages. The suspension clamps were especially damaging due to mutual occurrence of vibration from either side [106]. Mathematical and experimental investigations of the effect of clamp geometry on the stresses and strains [107, 108] showed that suspension clamps with saddled curvature at the edge evenly distribute the bearing load and minimize the mean stress in each strand. Specifically, both the static and dynamic strain levels were increased substantially as the longitudinal radius of curvature of the clamp (Fig. 2.23) was decreased, hence decreasing the fatigue performance of the conductor [22]. Thus, the use of clamps with long curvature radius significantly reduced conductor fatigue damage.

The bending amplitude of a conductor caused by vibration is a key parameter closely related to its fatigue failure. Ramey and Silva [109] showed experimentally that amplitude reduction can substantially extend conductor fatigue life. Therefore, amplitude reduction techniques such as use of vibration dampers [110], and designing the clamps with large curvature radius at the edge to accommodate the bending curvature of the conductors [111] are used to attenuate the influence of vibration bending amplitudes. In this regard, relative slip between the outer layer strands and the clamp occurs only if the vibration induced bending amplitude curvature exceeds that of the clamp [112].

In view of the foregoing, it is now accepted that fatigue degradation of conductors is caused by windinduced vibration leading to periodic bending of the conductor against the suspension clamp. Consequently, the conductor strands undergo a minute periodic slip motion against each other or against the clamp at their contact interfaces. The relative slip is significant only at the mouth of the clamp where conductor strands often fail. The periodic slip induces shear stresses on the contact surfaces. Furthermore, it generates surface fretting damages due to the normal contact load from clamping and stranding of the conductor. It causes premature failure by accelerating the fatigue crack initiation stage by formation of surface cracks or flaws, some of which propagate into fatigue cracks



[42]. In summary, fretting transforms a high cycle fatigue process into a low cycle fatigue, which means reduced fatigue life.

Fig. 2.23. Illustration of curvature radius of a suspension clamp: (a) short radius and deep groove, (b) medium radius and shallow groove, (c) medium radius shallow groove, and (d) long radius and deep groove. R is the longitudinal radius of the clamp [22].

In further examination of the effect of bending amplitude on fatigue, Zhou *et al.* [18] conducted a laboratory bending fatigue test of ACSR conductors by applying amplitudes of 0.6 - 0.8 mm and a static axial load of 25% of rated tensile strength (RTS). The study showed that the main degradation processes occur at the clamp exit in the form of plastic flow, local wear, and local fatigue. According

to the finding, the crack initiation and the propagation processes are strongly dependent on the fretting regime and external loads, respectively. In this study, the fretting regime location was defined with respect to the center of the suspension clamp as stick, mixed and gross slip fretting regimes (Fig. 2.24a and b). As shown in the figure, the fretting fatigue life was substantially lower in the mixed regime. In each fretting regime, cross-sectional sample was investigated using optical microscope. The cross-sectional view showed cracking in the stick (Fig. 2.25a) and mixed regimes (Fig. 2.25b), whereas removal of material from the surface by wear occurred in the gross slip regime (Fig. 2.25c). The axial cross-sectional views demonstrate that oblique cracks grow and propagate on the contact surfaces. Similarly, Boniardi *et al.* [113] showed plastic deformation and cracking damages at the strand-to-strand contacts (Fig. 2.26). The cracks initiated on the contacting surfaces grew into the bulk and caused breaking.





Fig. 2.24. Variation of fretting fatigue life as a function of a) fretting mark area and b) distance from the suspension clamp center [18].





Fig. 2.25. Cross-sectional view of a conductor after bending fatigue test showing cracking in the sticking (a) and mixed (b) regimes and wear in the gross slip (c) regime [18].



Fig. 2.26. Broken conductor strands at the contact interfaces marked in white circle and arrows showing elliptical plastic deformation [113].

The superposition of bending and shear stresses with fretting creates favourable condition for crack initiation at the contact surface of a strand. Investigation of failed samples from service environment and bending fatigue tests in a laboratory rig (Fig. 2.27) [18] followed by metallographic examination confirmed that the cracks initiate at the fretted surfaces [114]. Thus, the most important question has been on the magnitude of slip amplitude and number of vibration cycles leading to the fatigue failure. The vibration parameters (number of cycles and conductor bending amplitude) can be measured using data recording systems developed for vibration monitoring [18, 115]. However, the magnitude of the relative slip and normal contact load arising at the contact interfaces as a result of conductor bending has been impossible to measure directly. This is due to the fact that the interface location is physically inaccessible. Instead, analytical assessment of fretting fatigue failure has been based on the conductor bending amplitude measured with respect to LPC [116]. The bending amplitude transforms to both bending stress and relative slip at the interfaces of contacting strands. Therefore, it has been used as a substitute parameter for determination of the fatigue inducing stress in practice [117].



Fig. 2.27. Broken conductor strands at the contact interfaces marked in white circle and arrows showing elliptical plastic deformation [18].

2.14 Literature Survey: Corrosion induced degradation of ACSR conductors

Aluminium based conductors are normally regarded as resistant to corrosion because of the natural oxide film spontaneously forming on the surface. The air-formed thin film is naturally colourless, transparent and firmly adherent to the metal. It thickens upon exposure to humid air and blackens due to absorption of carbon and other impurities. Hence the conductor surface darkens according to the level of pollution in the air. Nicholls and MacDonald reported that blackened conductors possess about 8 - 9% higher current rating than bright conductors and blackening is not necessarily an

indication of corrosion. However, the film is unstable in harsh climatic environments such as marine (coastal) and industrial areas [118]. The corrosion degradation depends on the corrosivity of the environment to which it is exposed. This is different for marine, industrial and onshore geographic environments. The corrosivity of a given environment is defined by a corrosivity index, discussed and referenced by Refsnæs *et al.*[119].

Recently, study on an ACSR conductor exposed to atmospheric corrosion in the Mediterranean coast, polluted with sand blown from desert, SO₂ and CO₂ was investigated using mechanical testing, chemical analysis, and visual and microscopic examination [120]. It was found that galvanic corrosion caused depletion of the galvanic coating on the steel core and led to severe localized corrosion (pitting) of the aluminum strands. Pitting corrosion caused reduction in the cross-sectional area of the Al strands and induced local overheating in the conductor to the extent that the grease intended to protect against galvanic coupling of the inner layer aluminum strands with the galvanized steel core was evaporated and lost its functional property.

It is generally agreed [10, 28] that ACSR conductors are susceptible to galvanic corrosion resulting from electrolytic coupling of the dissimilar constituent metals (aluminum, zinc and steel) in the conductor when it becomes exposed to moisture with saline impurities, especially chloride ions. First, the zinc coating corrodes and protects the aluminium strands [121, 122]. After the galvanic coating corrodes away, the aluminum strands become sacrificial to the steel. Subsequently, the aluminum strands adjacent to the core corrode [123]. Bulging occurs by formation of corrosion products (also called "internal corrosion") and ultimately leads to failure. Galvanic corrosion of a conductor is also caused by installed fittings which are electrochemically more noble to the aluminum strands [124, 125]. For example, Pearson [126] reported that severe galvanic corrosion of the outer surface of a conductor was caused due to fitting of mid-span steel sleeve joints. Actually, this form of corrosion is nowadays rarely an issue as the majority of clamps and fittings are made of cast aluminum alloys.

Localized corrosion in the form of crevice corrosion [28] can also occur in the presence of chloride ions, enhanced by galvanic coupling discussed above. Crevice corrosion is known to occur due to penetration of moisture to the contact interfaces of the strands and insufficient oxygen availability, which causes oxygen concentration cell between the crevice and free (but moist) strand surface. Crevice corrosion occurs normally in the form of pitting inside the cavity. Ash *et al.* [28] stated that conductors experience locally accelerated corrosion due to small voltages developing between the strands. However, the mechanism and corrosion characteristics of such stray current phenomena have not been discussed [127].

Though the material loss by pitting corrosion is limited in quantity, its influence on the structural capacity is significant. Localized corrosion attack on the aluminium strands exposed to marine environments is pitting [128] which is associated with crevice geometries of the conductor and conductor-clamp or fittings. It arises from penetration of saline solution/moisture into the interstices of the conductor strands at the clamps and fittings. For example, Refsnæs et. al. [119] investigated the impact of a bird protection system on a transmission line exposed to corrosive environment. The result indicated that the bird protection system reduces the service life of a conductor by 12 - 22% [119]. Similarly, Caltz and Piotgieter [129] evaluated the corrosion degradation of an overhead power line exposed to service in marine atmosphere and found that the conductor was mainly affected by pitting corrosion of up to $5 \,\mu$ m deep pits. Forrest and Ward [130] also demonstrated that pitting is the cause of ageing of a conductor. These works show that pitting corrosion is a factor limiting the service life of a conductor. On the other hand, it is known that the corrosion mechanism of such commercial purity aluminium alloys is a shallow pitting which is related intermetallic particles [131]. Such pits may not be very sensitive to crack initiation, but still create the weak spot for initiation of fatigue cracks. These studies show that localized corrosion degrades the fatigue capability of aluminium alloys including the one used in overhead power lines.

The mechanism of localized corrosion has been known to be driven by trickling of electrolyte (wetted salt aerosols) into the narrow geometries (crevices) of the conductor strands with each other or the clamps and fittings. The stagnation of the electrolyte into the crevices, absence of mass diffusion into the crevice leads to depletion of dissolved oxygen, decreased pH (acidification) and enrichment of Cl⁻ inside the crevice causes depassivation of the aluminium surface, hence crevice corrosion. The pH is reported to drop to 4.1, which is acidic range and causes acidic dissolution [132]. This type of electrochemical process change is expected in the crevices of overhead power line conductors exposed to marine environments.

2.15 End of service life of ACSR conductors

Based on assessment of weight loss, tensile strength test, and electrical resistance measurement Forrest and Ward [130], reported that corrosion causes 0.4 - 0.5% on average deterioration in mechanical strength, per year, of ACSR conductors exposed to industrially polluted and coastal environments. The authors documented that industrial pollution related corrosion causes degradation on the outer surface of a conductor in the form of uniform corrosion whereas the corrosion in a marine environment causes mainly pitting (due to the crevice mechanism) and galvanic corrosion arising from electrolytic coupling of the aluminium, zinc coating and steel.

On the basis of the mechanical strength of the aluminium strands, the service life of a conductor is suggested to be 30% loss; or 12% loss in the average strength of the whole conductor [130]. This value is usually determined by the tensile test of a segment of a conductor compared to the fresh conductor state. Surface defects by corrosion, wear or fatigue damage can be reflected by the tensile testing of the individual strands [133]. The surface degradations caused by corrosion or wind induced vibration can also be reflected on the torsional ductility test of the conductor [134]. Monitoring and measuring the remaining service life of the ACSR conductor has been attempted by *in situ* detection of the loss of the galvanized layer on the steel core [124]. Evaluations of damage to different components of power lines have indicated that the conductor is the weakest line item which needs replacement first [135].

2.16 Discussion

Conductors serve both electrical and mechanical functions in a transmission system. The deterioration of these functions leads to failure. This can be caused by fatigue cracking, wear, corrosion or the combination of two or more of these. Since the intention of this study is to improve the understanding of the degradation mechanisms of the aluminium strands of ACSR conductors in a marine environment, an effort was made to summarize and critically evaluate the various degradation mechanisms already reported in order to justify the purpose of this thesis.

Most of the conductor damage investigations conclude that the service life is determined by fretting damage caused by wind induced vibration [114]. While the fatigue aspect has been well studied, the wear aspect seems less stressed although the earlier works [18, 27] seem to indicate that both fretting

wear and fretting fatigue are the mechanisms affecting the aluminium strands. The displacement amplitude determines whether fatigue or wear degradation occurs at the strand-to-strand and strandto-clamp contacts. In addition, trapping of saline aerosol, moisture, or wear debris at the contact interfaces is unavoidable in marine environments. Therefore, the combination of all these factors affects both the wear and fatigue processes taking place, *i.e.*, mutual occurrence of mechanical and electrochemical degradation is unavoidable. In these circumstances, the degradation mechanism becomes fretting corrosion or tribocorrosion in the event of fretting and sliding wear in a corrosive environment, respectively. The relevance of these mechanisms needs to be documented in more detail by failure analysis of samples recovered from marine environments and further investigations.

The range of displacement amplitudes leading to fretting fatigue or fretting wear could not be determined by the available investigation methods for the conductors. Studies of fretting wear behaviour of the material used for the strand in similar environments are also quite limited in the available literature. Most studies have been focussed on fretting fatigue tests in a laboratory air. Nevertheless, these studies show that the damage mechanism transforms from fretting fatigue to fretting wear and eventually reciprocating sliding wear as the slip amplitude increases. There is a limited possibility to control and apply a predetermined displacement amplitude to investigate the wear or fatigue degradation it causes. The transition from one fretting regime to another is influenced by factors such as material and environment besides the normal contact load and slip amplitude. Hence, a general background on these essential features is necessary.

As reviewed above, the galvanic coating protects the steel strands from rusting and apparently provides cathodic protection to the aluminium strands. The strands are often subjected to wear and fatigue degradations at the same time. This causes surface degradation which removes the protective oxide film and changes the electrochemical properties of the interfacing surfaces. In the case of aluminium, the corrosion potential is highly negative as can be learned from scratch tests in a chloride solution. Consequently, the potential established under this condition may keep the aluminium strands at very low potential and influences the corrosion and wear behaviour of the strand material. Despite the proximity in the galvanic series, the sacrificial effect of zinc to aluminium may be compromised in the presence of wear. Therefore, it is of interest to investigate in general the corrosion behaviour of the aluminium strands under wear at various imposed potentials to improve the understanding on the tribocorrosion behaviour of the material.

In addition to the forgoing, certain details about the tribocorrosion performance of the aluminium strands seem to be lacking in the earlier studies. As explained, the wear behaviour of a material is influenced by an environment. The environment can penetrate to the contact interfaces of the strands and influence the wear degradation by changing the property of the wear debris (third-body) generated or electrochemically dissolving the surface. Especially, the wear behaviour of the alloy in air, chloride free environment and marine environment were not documented in the open literature. Additional surface characterization and electrochemical investigation are therefore necessary to clarify the so-far unknown tribocorrosion mechanisms and wear characteristics of the aluminium strands in different environmental conditions such as air, tap water and chloride solution to enhance understanding about the degradation characteristics of the conductors in similar service conditions.

Although there are many full-scale fretting damage investigations, fundamental work on fretting and sliding wear of the strand material seems to be limited. Full scale investigations have been very useful in generating useful information on the response of the conductors to externally imposed loads for a given contact condition. The relationship between the imposed bending amplitudes and other parameters are monitored with respect to the number of cycles to failure of the strands and *ex situ* characterization of the contact surfaces. It is impossible to set up *in situ* tools such as open circuit potential measurement, polarization measurements, acquisition of current transients and friction monitoring, of overhead power line conductors in real service. Therefore, the fundamental investigations conducted at a controlled normal load, displacement amplitude and environment in the laboratory should be used to get an insight into the response of the material. Fundamental tests, which can determine the tangential friction force acting on the interface and the coefficient of friction and allow *in situ* monitoring of response of the surface to a corrosive environment, can be performed in a controlled laboratory condition. Such studies will enable to identify the fretting regimes, the characteristics of fretting degradation in each regime and the response of the strand material to different magnitudes of load and corrosive environment.

The purpose of the present study is therefore to provide details of factors causing the degradation of overhead power line conductors exposed to marine environment through identification and analysis of contributing factors. The failure analysis of the specimen is performed to find evidences of corrosion and mechanical degradations or to find the major failure types and characteristics. Moreover, the analysis is used as basis to sort out failure mechanisms causing and contributing to shortening of service lifetime of the conductors. Tribocorrosion and fretting corrosion characteristics

of the aluminium strands and the performance of the galvanic coating supposed to provide cathodic protection to the strands are addressed.

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3 Failure Analysis of Overhead Power Lines Exposed to Marine Environments

Abstract

Fretting and corrosion have often been claimed as the key failure mechanisms of overhead power line conductors exposed to marine environments. The failure occurs usually in the vicinity of the suspension clamps attached to the conductor. Fretting corrosion is the simultaneous action of corrosion and fretting caused by wind induced vibrations. However, such failure mechanism is scarcely investigated and documented in the open literature. Therefore, failure analysis was conducted on failed, steel reinforced aluminium conductors (ACSR) recovered from service in marine environments. The failed conductor-clamp assemblies were investigated for the causes of damage. Fretting, galvanic and localised corrosion were identified as possible factors contributing to failure. The interstices of the outer layer strands showed pitting and crevice corrosion despite cathodic protection expected by the galvanised steel core. The strand-to-strand and strand-to-clamp contact surfaces provided the sites for subsequent fretting corrosion damages, characterized by wear and fatigue. These processes leading to crack initiation and growth was suggested as the factors determining the service life of the conductors.

3.1 Introduction

Failure of overhead power line conductors exposed to marine environments, typically made of steel reinforced aluminium strands (ACSR), has been a problem for utilities in planning and operation [1]. The failure of these conductors is often attributed to damages inflicted on aluminium strands by wind induced vibration and corrosion at the clamps and fittings, especially the suspension clamps [2]. Regardless of the presence of integrated protection mechanism in the ACSR conductors, corrosion has been claimed to be one of the service life limiting parameters [3]. Most of the earlier studies of failure available on ACSR or other types of conductors have been limited to either corrosion or vibration induced damages. Those studies are briefly summarized below.

Failure analysis of ACSR conductors have been conducted in both laboratory and field. For example, fretting fatigue failure characterization [4], and the effect of clamp design on the fatigue failure [5, 6] have been largely investigated in the laboratory. Fricke and Rawlins [7] examined failed conductor strands from lines in service and verified by laboratory tests and concluded that all fatigue cracks originated at fretted sites of the strands. Furthermore, failure analysis and laboratory simulations of

all aluminium alloy (AAAC) [8] and ACSR conductors [9, 10] and study of fretting patterns in a conductor-clamp contact zone [11] have ascertained that the failures were caused by fretting at strand-to-strand and strand-to-clamp contacts under wind induced vibration. The vibration causes bending of the conductor at the clamp, which in turn creates relative slip at the strand-to-strand and strand-to-clamp contact surfaces at the clamp edges. Subsequently, wear and microcracks develop at the contact surfaces, reducing the resistance of the strands to fatigue loading (fatigue strength), which leads to premature failure.

Investigation of failed ACSR conductor at a spacer clamp [12] showed plastic deformation marks, fretting wear, traces of Al₂O₃ debris and silicon particles on the worn surfaces of external layer strands. The analysis further suggested that breaking of the strands occurred in the gross slip fretting regime. However, the internal layer strands showed elliptical plastic deformation marks at their contacts with the adjacent layer strands [10]. Presumably, fretting damages experienced on different layers of the conductor varied with location of the strands.

Fretting and corrosion, though rarely considered in combination in the failure analysis of overhead power lines, are believed to be the factors causing the degradation of the conductors serving in marine and coastal areas [3]. For example, study on ACSR conductors exposed to marine and industrially polluted environments showed that the rate of deterioration of mechanical strength of the aluminium strands was about 0.4 - 0.5% per annum [1]. The corrosion damages affected the inner and outer layer strands for exposure to marine and polluted environments, respectively. The degradation of the inner layer strands followed the depletion of the galvanic coating on the steel core [13, 14].

Historically, investigations on fretting corrosion have been concentrated on special applications such as medical implants, aeronautic, and nuclear industries. The open circuit corrosion potential (OCP) of the aluminium surface sharply falls in the presence of fretting wear due to continuous disruption of the naturally formed protective oxide film [15]. Corrosion also influences the wear process by changing the chemistry and composition of the contact interface. This suggests that fretting corrosion on ACSR conductors involves synergistic processes. Hence, the lifetime prediction methodologies based on corrosion (loss of galvanic coating) [13] or mechanical factors (fretting fatigue) [10] *per se* may be unrealistic.

While considerable experience has been acquired over the last several decades concerning the fretting and corrosion degradations of ACSR conductors, a satisfactory understanding of the mechanisms of failure considering the mutual action of corrosion and wind induced vibration is still lacking. Therefore, the objective of this chapter is to investigate the failure mechanisms of ACSR conductors exposed to marine environment through characterisation of failed sample surfaces with particular emphasis on identification of mutual action of corrosion and vibration induced damages.

3.2 Experimental

Failed ACSR samples with their respective clamps, obtained from utility companies in Norway, were investigated. The conductors investigated were 6/1 and 26/7 ACSR types with a commercial purity aluminium strands made of hard drawn AA 1350 alloy in full hard H19 condition, designated as AA 1350-H19 [16]. The conductors were reinforced with stranded and galvanized high strength steel core as described in detail in Section 2.4.

The conductors investigated were standard ACSR types for which the nominal chemical composition of the aluminium strand material was presented in § 2. The composition of the sample strand material, presented in Table 3.1, was determined by using an electron microprobe analyser (EMPA-JXA-8500F) on a selected cross-section which was metallographically polished through 1 μ m diamond paste. Almost all Fe precipitates as Al₃Fe intermetallic phases. Most of Si is in solid solution with the aluminium matrix [17]. The clamp components, keeper and clamp body are usually made from Al-17 % Si alloy [12].

Element	Cu	Fe	Si	Mn	Zn	Al
Wt. %	0.01	0.3	0.09	0.01	0.01	99.6

Table 3.1 Chemical composition of the AA 1350-H19 strand sample determined by EMPA.

The conductors were dismantled from the clamps, and the contact surfaces of the strands were investigated for wear, corrosion, and surface cracks visually and by optical microscopy. These surfaces were then cleaned from corrosion products with a standard solution of hot chromic phosphoric acid (ASTM G1-90). Further surface characterization was conducted using scanning electron microscopy (SEM) equipped with energy dispersion spectroscopy (EDS) capability.

The degradation of the galvanic coating on the steel strands was investigated using an optical microscope and EMPA-JXA-8500F instrument. The latter was used for mapping the elemental composition of corroded surfaces and polished cross-sectional samples. The elemental mapping was intended to determine the elements contained in the corrosion deposits on the surface.

3.3 Results

3.3.1 Conductor construction and failure location

The conductors investigated were classified as 26/7 (Fig. 3.1a and b) and 6/1 (Fig. 3.2a, b and c) ACSR types according to their construction. The first type was made of 2 layers of 26 commercial purity aluminium strands reinforced with seven galvanized steel wires. The second type constituted a layer of 6 aluminium strands reinforced by one galvanized steel strand, as explained in detail in chapter 2.

Most of the failures observed were at the clamps, especially suspension clamps (Fig. 3.1a [18]), which contribute to failure by introducing local contact loads and plastic deformation due to clamp assembly. Moreover, the clamps contribute to wear damage by causing material removal by rubbing against the conductor strands thereby significantly reducing the conductor cross-section, as shown in Fig. 3.2a or fatigue failure by causing fracture of all or some of the aluminium strands as shown in Fig. 3.2b and 3.2c as examples. In general, the main damages observed at the clamps are plastic deformation, cracking or wear. These appear to be coupled with corrosion in marine environments as presented in the following paragraphs.



Fig. 3.1. a) Suspension clamp-conductor components [18]. b) Tension clamp-conductor components. The diameter of the outer layer aluminium in b) strands is 3.71 mm



Fig. 3.2. Photographs of a) worn conductor, b) broken conductor and c) broken strands at both ends of the clamp (arrows). The diameter of the aluminium strands is 3.71 mm.

3.3.2 Plastic deformation

As explained in Chapter 2, clamping during installation caused plastic deformation of the conductor bundle at the suspension clamp (Fig. 3.3a) and tension clamp (Fig. 3.3b). The clamping caused stress at the contact surfaces and increase in gaps between the strands thereby causing crevice corrosion. The gaps facilitated ingress of moisture and corrosive media into the crevices between the inner layer strands (Fig. 3.3b), hence causing darkening and corrosion of the conductor surface at the interface with clamps. The bottom contact surface of the conductors with the clamps were more corroded than the top contact surface, indicating more accumulation of moisture in that part [3]. Clamping essentially generates the contact load, which is a parameter necessary for the surface fatigue and wear to occur under wind induced vibration at the contact interfaces.



Fig. 3.3. Photographs of plastic deformation of 26/7 ACSR conductors at a) suspension clamp and b) tension clamp, the latter showing corrosion products. The gap between the strands was increased as a result of clamping. The actual diameter of the aluminium strands was 3.69 and 3.71 mm on the outer and inner layer strands respectively.

3.3.3 Fretting corrosion

Fretting regimes

The contact surfaces between the inner and outer layer of the aluminium strands in a 26/7 ACSR conductor was investigated. The contact surfaces were changed from an initially barely point contact

to substantially larger elliptical plastic deformation marks (Fig. 3.4a). They showed neither material removal (wear), surface fatigue (cracking) nor corrosion related degradation of practical interest. Even the sparse corrosion deposit on the strands was absent on the contact surfaces. This indicates that the surfaces were in stick contact with the counterface. Hence no relative slip and access of moisture to the interface occurred. Therefore, the contact surface damage was purely mechanical plastic deformation induced by clamping. This type of fretting damage occurs in stick contact fretting regime, as discussed in §2.





Fig. 3.4. Macrographs of surface damages at the contact interfaces between aluminium strands with actual diameter of 3.69 mm. a) Stick contact fretting condition with sparse oxide debris and intact elliptical plastic deformation mark, indicating no relative slip. b) Partial slip fretting contact condition with stick contact at the centre (A) and region of relative slip around (B), containing corrosion products and wear debris. c) Gross slip fretting contact condition with wear at the centre, surrounded by oxide and wear debris around the contact edge. d) Gross slip fretting contact at lateral and inner contacts characterized by wear and plastic deformation (indicated by arrows).

Fig. 3.4b shows a contact surface damaged by plastic deformation at the centre (zone A) corrosion and wear at the peripheries (zone B). The damage consisted of stick contact at the centre (A) and surrounded by an area of slip contact. The neighbouring contact surfaces in the figure showed a similar type of degradation. Therefore, the contact interface marked in the figure was in the mixed stick slip fretting regime described in §2. While the slip part of the contact surface was worn and corroded, the stick part was intact except for plastic deformation indicating no access of corrosive media. Hence, the surface degradation was apparently by fretting corrosion in a mixed stick slip fretting regime. The corrosion products between the strands along the lateral contacts (shown by an arrow in Fig. 3.4b) indicate that crevice corrosion had occurred.

Some of the elliptical plastic deformation marks showed signs of wear degradations. As shown in Fig.3.4c, the whole contact surface (marked A) showed wear degradation and accumulation of corrosion product formed by oxidation of metal debris. Similarly, the nearby contact surface (marked as B) experienced degradation by wear, plastic deformation, and surface cracks. However, accumulation of debris and corrosion products at the edges was smaller than that of A, indicating that fretting damage varied axially on the same strand. Fig. 3.4d indicates gross relative slip on a 6/1 ACSR conductor both at the lateral strand-to-strand and strand-to-clamp contact surfaces. The damages on the contact surface with the clamp were in the form of plastic deformation and smearing of deformed metal at the contact with the clamp. Material removal occurred only at the lateral contact. The absence of debris and smeared metal at the contact edge indicates that the relative slip amplitude was large such that the debris could not be accommodated within the contact. Hence, the lateral contact interface may have experienced sliding wear.

A signature of corrosion attack was observed in the mixed stick slip and gross slip fretting degradations where the surfaces experienced rubbing. In addition, components subjected to mixed stick slip are susceptible to fatigue damage as described in §2. With an intention to investigate the role of corrosion in the damage, the surface morphology of a sample subjected to mixed stick slip fretting regime was analysed. The topography of the contact shows clear difference between the surfaces damaged under stick and slip (Fig. 3.5a). While the morphology of the surface in the stick part appears relatively smooth and intact, the slip part was lightly worn and pitted. A SEM secondary electron image of the surface within the slip zone showed that the material was etched (Fig. 3.5b). Observation of few pits (Fig. 3.5a) suggests that also corrosion may have contributed to material degradation by fretting is more critical in the mixed and gross slip fretting regimes than in the stick regime. In addition, corrosion may contribute in enhancing material removal (wear) and surface fatigue leading to breaking of the strands as presented in the following sections.



Fig. 3.5. Secondary electron SEM images of contact interface of a conductor strand, recovered after service in a marine environment, obtained after cleaning in hot chromic phosphoric acid bath. a) Image of partial slip fretting condition with stick contact at the centre and slip contact at the edge with wear and pits at edge and intact at the centre. b) Higher magnification Secondary electron SEM secondary electron image of zone A showing an etched area in the slip zone.

Fretting wear and corrosion

As presented in the preceding subsection the contact surfaces in a gross slip state showed degradations involving both fretting wear and corrosion. The material loss due to wear was so large that the cross section of the conductor was significantly reduced, indicating that the degradation was due to sliding wear as discussed earlier in the chapter (Fig, 3.2a). Normally, fretting wear becomes sliding wear beyond a certain critical relative slip amplitude. The difference between the two is determined based on the rate of wear volume, the sliding velocity and the resulting wear scar morphology, as presented in §2. Therefore, it is possible that the degradation displayed in Fig. 3.2a occurred by excessively large slip amplitude better characterized as a sliding wear.

Fig. 3.6a shows a fretted elliptical contact surface of a strand damaged by fretting wear and broken at the edge of the contact surface. The fracture location is not included in the image. The entire contact surface was under relative slip in the direction shown by the arrow in figure. This indicates that the contact surface was subjected to gross slip fretting regime. The rubbing deformed the surface, delaminated and pushed the material to the edge of the contact surface along the slip direction. Moreover, surface and subsurface cracks were found on the worn surface as shown in Fig. 3.6b. These cracks appear to be nucleated, propagated and interconnected in the rubbing process to cause delamination of metal as fragments and flakes. Therefore, the dominant wear mechanism was delamination. Other wear mechanisms such as deformation and smearing were also involved. The imprint of removed flake was left behind on the wear scar as shown in Fig. 3.6b. This shows that fretting wear and fatigue were competitive damage processes on the surface in which the latter followed and led to failure.



Fig. 3.6. a) Secondary electron SEM image of an elliptical contact surface exposed to gross slip fretting contact condition. b) Magnified view of the region marked A in (a) showing cracks and delamination caused by fretting in the wear track.

Fig. 3.7a shows the contact surface of a 6/1 ACSR conductor strand which was damaged by fretting wear and corrosion. The fretting damage was caused by rubbing of the surface against the cast aluminium alloy clamp. As a result, the strand surface was plastically deformed and worn away. In addition, the worn surface was covered by oxide layer and locally pitted and disintegrated (see Fig. 3.7b). Clearly, the surface degradation involved both wear and corrosion. In contrast to the topography of the worn surface with oxide, the morphology underneath the oxide layer was mainly damaged by delamination and abrasion (cf. Fig. 3.7c and d). It was abraded and fragmented tangentially parallel to the axis of the strand.





Fig. 3.7. Fretting wear - corrosion damage morphology on aluminium strand rubbed against the clamp. a) Image of worn surface damaged by wear and pit. b) Secondary electron SEM image of the worn surface showing broken oxide cake and pit marked A. c) Secondary electron SEM image of the worn surface after removing the oxide showing delamination and abrasive wear. The area marked B is magnified. d) High magnification secondary electron SEM micrograph of the area marked B in c) showing ridges of deformed metal.

Fretting fatigue and corrosion

Fig. 3.8a shows a broken strand of 26/7 ACSR conductor at a fretted contact surface. The strand was broken at the edge of fretted elliptical plastic deformation mark. The contact surface of the strand was damaged by plastic deformation, fretting wear in gross slip condition and corrosion. Fretting wear produced debris, which was smeared and adhered to the surface. There was also corrosion deposit outside the contact surface. After cleaning the oxide, the morphology of the contact surface showed plastic deformation, wear and superficial cracks (see Fig. 3.8b). The surface cracks were oriented transverse to the strand axis on the worn surface and axially at the contact edge (indicated in the figure). The wear was in the form of deformed and smeared metal flakes on the surface and imprints of removed flakes are noticeable on the image. Thus, the fretting wear degradation was in the form of delamination of metal flakes. The contribution of corrosion is unclear from the morphology. Corrosion occurred mainly in the form of etching as displayed in Fig. 3.8c. The etching was severe in the crack crevices.





Fig. 3.8. Fretted site on broken aluminium strand. a) Fatigue crack at plastically deformed and worn elliptical mark. b) Secondary electron SEM image showing plastic deformation near crack initiation site taken after cleaning oxide deposit in hot bath of chromic phosphoric acid. c) Secondary electron SEM micrograph showing voids and corrosion etch at the crack initiation edge. The diameter of the aluminium strand shown in a) is 3.71 mm.

A conductor strand broken under a mixed stick slip fretting condition is shown in Fig. 3.9. The contact surface was in stick contact condition at the centre with neither wear nor corrosion damage as shown in Fig. 3.9a. The area to the left of the stick contact towards the crack edge suffered corrosion and wear as evidenced by the surface cracks and corrosion deposit. However, the opposite side of the contact surface was relatively less worn and showed no visible cracks, illustrating variation in the degradation mechanism within the same contact surface.

Fig. 3.9b shows a SEM secondary electron image of the stand surface mentioned above after cleaning the corrosion deposit indicated in Fig. 3.9a. Different areas (indicated by arrows A, B, and C) of the sample were analysed in order to find the damage signatures produced by corrosion and fretting. SEM secondary electron image of the crack initiation site (arrow A) and the worn surface (arrow B) are shown, respectively in Fig. 3.9c and d that the surface was damaged by corrosion (etched). The surface outside the contact interface (arrow C) was abraded and bears transverse surface cracks (Fig. 3.9e). The abraded and cracked area was heavily etched by corrosion (Fig. 3.9f).









Fig. 3.9. Secondary electron SEM image of a) aluminium strand broken under partial slip fretting corrosion showing contact surface with slight wear, corrosion deposit and fracture, b) strand surface with corrosion deposit removed, showing crack initiation site (arrow A), wear track (arrow B), and abraded area beside the wear track and crack initiation site (arrow C), c) morphology of in the area indicated by arrow A, d) corrosion attack (etching) morphology of the worn surface indicated by arrow B, e) the surface indicated by arrow C, and f) abraded and cracked surface in e) showing etching inside the coalescing crack. The actual diameter of the aluminium strand shown in a) is 3.71 mm.

3.3.4 Fractography

Fig. 3.10 shows a 6/1 ACSR conductor with all aluminium strands broken at the worn surfaces. Rubbing of the strands against the clamp caused the wear damage. The degradation process substantially reduced the conductor cross-sections and led to their failure. This type of excessive wear can be beyond the range of fretting wear amplitude. Hence, it is a fretting wear process caused by large relative slip. The rubbing removed material from the contact interface and caused thinning of the strands. No accumulation of debris was observed at the edge of the contact indicating large relative slip amplitude on the surfaces. The fracture surface profile is not identical for all the strands. It is vshaped on some and slanted all the way on others. In all cases, the cracks initiated on the fretted surfaces and grew inclined at some angle to the contact surface. Such difference in crack profile suggests that crack advancing loads act in different direction at different depth in the strands. Furthermore, the fracture surface topography was different from strand to strand in the same conductor. Two particular cases were further analysed to explain the factors contributing to the crack growth and propagation.



Fig. 3.10. Photograph showing broken aluminium strands of lubricated 6/1 ACSR conductor at a suspension clamp. The strands were broken at surfaces worn by the clamp. The diameter of the aluminium strands is 3.71 mm.

In the first case, the crack surface profile was of nearly "V-shape" (Fig. 3.11a) with crack imitation and propagation from opposite surfaces (shown by arrows) and the final fast fracture phase at the centre. There were also cases where both the initial crack growth and the propagation phases were oblique to the strand axis all the way (cf. Fig. 3. 10). Fig. 3.11a shows that there were numerous subsurface cracks beneath the crack initiation edge and a major subsurface crack bordering the final fast ductile fracture. This type of crack initiation, growth and propagation has been reported in the literature [8, 12]. The fatigue crack striations common on plain fatigue crack propagations was not noticeable on these strands. This indicates that the crack growth and propagation path was controlled to a major extent by the same factors which initiated the crack, namely fretting fatigue.







Fig. 3.11. Fracture surfaces of aluminium strands. a) Quasi V-shaped fracture surface showing crack initiated from two sides (see arrows). b) Fractured strand crack surface divided into crack nucleation and initiation (Stage I), crack propagation (Stage II) and fast ductile fracture (stage III). c) High magnification secondary electron SEM micrograph of the fracture surface underneath the crack initiation showing subsurface cracks. d) Secondary electron SEM image of the surface inside the crack growth stage (Stage I) showing corrosion etching and subsurface cracks.

Fig. 3.11b shows another type of fractured surface of a strand with crack initiation, propagation and final fracture stages, designated as stage I, stage II and stage III, respectively. In stage I, the crack grew along the high shear plane (at about 45 degrees). It was a short crack growth stage until it reached the critical threshold necessary for the crack propagation to be overtaken by the bulk fatigue stress. The crack was initiated on the worn and corroded strand surface reported in Fig. 3.6. This stage ranges from the initiation point on the surface to the propagation phase where fatigue striations began. The striations were caused by cyclic stress from wind induced vibration. As presented in the previous section, there were numerous surface cracks initiated on the contact surface, but the one first reached the critical length and propagated to stage II presented below. As generally known, the third stage (stage III) is a rapid mechanical process. Hence, it is not relevant for service fatigue life prediction.

In stage II, the crack progressed for every cycle of vibration and stopped when there was no vibration. As shown in Fig. 3.11b, the distance between successive striations and thickness of the contours varied during crack propagation. The thicker striations were where the crack advance was temporarily halted due to vibration stoppage, whereas the thinner ones corresponded to crack advance during successive cycles. The striations have the crack initiation area as a common centre, and they increased in radius as the crack advanced. Each striation contour was a crack arrest location between successive vibration cycles during which there was no crack growth and electrochemical activity could take place at the crack tip. Though it is unclear how much time was elapsed under the crack arrest conditions, the crack advance appears to be by plain corrosion fatigue. The crack advance mainly depends on the cycles of vibration, which is usually less than 5 - 10 cycles per second as presented in Section 2.6, which gives enough time for corrosion to interfere in the crack advance. The fracture surface bears plastic deformation marks (marked in Fig. 3.11b) in all areas of the designated crack zones. These must be due to axial relative slip and collision of the broken halves during continued vibration.

Fig. 3.11c shows SEM secondary electron image of the fracture surface beneath the fretted surface at the crack initiation edge in stage I, marked A in Fig. 3.11a. There were subsurface cracks underneath the fretted surface, about 50 μ m deep, in a plane parallel to the contact surface. The subsurface cracks are oriented transverse to Stage II fracture surface. Subsurface crevices were formed and widened as a result of joint mechanical and corrosion degradations (Fig. 3.11d), causing etching of material from the surface and structural weakening of the strand. The fractured surface and subsurface cracks were etched by corrosion indicating that the corrosion occurred after the crack surfaces were formed. Also, the corrosion media penetrates to any newly opened crack surface and oxidizes it.

3.3.5 Corrosion damages

Localized corrosion

Corrosion deposit was observed in the crevices between the outer layer strands. Fig. 3.12 shows corrosion products along the lateral contacts of the outer layer strands of a 26/7 ACSR conductor at the clamp exit. The conductor was dismantled, and the oxide deposit was cleaned before SEM characterization of the corroded area. Longitudinally elongated pits were found as shown in Fig. 3.13a. The pits were associated with noble Fe-intermetallic phases as indicated in the figure. The Al matrix was attacked around these phases forming some of the pits. The mechanism of formation of these types of pits have been known to be related to micro-galvanic coupling between the noble intermetallic phases and the surrounding active aluminium matrix [19]. Higher magnification imaging

inside the pit, shown by arrow A in Fig. 3.13a, revealed crystallographic attack, typically observed on Al alloys in chloride media [20], as shown in Fig. 3.13b. SEM secondary image of the cross-section of the strand showed that the pitting corrosion attack inside the crevices was deeper at the intermetallic particles (Fig. 3.14). However, not all the pits could be related to the intermetallic phases. Most pits were formed in the crevices between the adjacent strands and affected by the local geometrical and environmental factors.



Fig. 3.12. Corrosion deposit between crevices of adjacent 26/7 ACSR conductor strands at the clamp edge. The actual diameter of the aluminium strands is 3.71 mm.



Fig. 3.13. a) Secondary electron SEM micrograph of aluminium strand surface after cleaning the corrosion deposit in a hot bath of chromic phosphoric acid (G1-90) showing crevice corrosion attack. b) Secondary electron SEM micrograph of (zone A) showing pitting corrosion attack.



Fig. 3.14. Secondary electron SEM cross sectional image of surface damaged by crevice corrosion and wear showing pitting corrosion attack around noble Al₃Fe intermetallic phases.

Galvanic corrosion

In ACSR conductors, the galvanic coating on the steel core acts as a sacrificial anode and a barrier between the aluminium and steel strands. The coating undergoes galvanic corrosion in providing cathodic protection to the steel core, as discussed in §2. Accordingly, corrosion products could be observed between the coating and the inner layer of aluminium strands, as shown in Fig. 3.15 for 26/7 ACSR. The corrosion deposits were nonuniformly distributed along the strands.



Fig. 3.15. Corrosion deposit between aluminium strands and the galvanized steel core, caused by galvanic corrosion. The diameter of the aluminium strand is 3.69 mm.

Fig. 3.16 shows a micrograph of the cross-section of the galvanized steel core for the above conductor sample. The coating was corroded and the corrosion product was cracked on its surface. However, most of the metallic layer remains on this micrograph. There were also areas where the zinc coating was depleted, and the steel substrate was locally detected. Fig. 3.17 shows such an analysis in plan view of an area (Fig. 3.17a) partially covered with corrosion product. Elemental mapping indicated significant concentration of Al (Fig. 3.17b), low concentration of oxygen (Fig. 3.17c) and Fe (Fig. 3.17d), and absence of Zn (Fig. 3.17e) in the marked area. The colour codes used for the element concentrations are listed in Fig. 3.17f. Contrary to the foregoing, substantial concentration of Fe and an appreciable concentration of oxygen was measured outside the marked region, as shown in Fig. 3.17c and d. High concentration of iron and low concentration of zinc outside the marked region suggests that the Zn layer was corroded there. Outside the marked area, the surface showed small concentrations of zinc and aluminium, respectively.



Fig. 3.16. Backscattered electron image of cross-sectional view of a galvanised steel strand showing the steel substrate and the zinc coating. The coating is oxidized at its surface in the form of zinc oxide flake.

The high concentration of Al coupled with low oxygen concentration suggests that the galvanised surface was covered by metallic aluminium at the contact interface, which must have resulted from wear. In addition, the apparent absence of zinc together with the low concentration of oxygen, suggests that the zinc coating did not heavily corrode, but covered by the metallic aluminium, in that area. Absence of zinc inside the marked region can be due to corroded Zn or mechanically rubbed Al, which covers the Zn layer underneath. If Zn were corroded, Zn-rich corrosion products would be expected to remain. It would also be expected to detect a higher concentration of the iron element underneath. Therefore, the result suggests that the zinc coating was screened by the rubbing against
in Al strands' layer. This conclusion is supported also by the cross-sectional analyses discussed above. Although Zn degradation may have occurred locally, the remaining Zn was probably still protective of the steel core against galvanic coupling with the aluminium strands.



Fig. 3.17. a) Wavelength-dispersive spectroscopy (WDS) elemental maps of a selected area on galvanized steel strand for b) Al, c) O, d) Fe, and e) Zn. F) Colour bar indicating the concentration level of each element in atom percent.

Material microstructure

Fig. 3.18 shows the microstructure of the aluminium alloy strand obtained by using low magnification optical microscopy. The grains were made optically distinguishable by anodizing in the conventional manner [21]. It shows that the microstructure of the grains of the strand are axially oriented and elongated along the axis of the strand (shown by the arrow). Such orientation of the grain results from the drawing process used in the production of the strand. The grains are thin and axially elongated. As presented in the earlier section, the corrosion attack followed the grain orientation with some grains possibly more active than the others. The tiny pores on the surface were the locations where intermetallic particles were removed by the oxidation process during the anodizing.



Fig. 3.18. Optical image of microstructure of a polished and anodized AA 1350 aluminium alloy strand showing elongated and thin grains in the axial direction.

3.4 Discussion

3.4.1 Tribocorrosion

Damaged ACSR conductors recovered from marine environment were investigated for degradation causes. Plastic deformation, fretting corrosion, and localized corrosion were specific degradation processes found on the contact surfaces of aluminium strands. This supports and adds up to the assertion that fretting related failure of the strands occurs at the contact interfaces due to reduction in fatigue strength [7]. In addition, the galvanized coating on the steel core strands showed some corrosion. Nevertheless, the corrosion damage caused no obvious galvanic contact of the aluminium and bare steel strands. This may indicate proper cathodic protection of the steel core, as well as protection against extensive pitting of the aluminium strands, whenever electrolyte penetrated between the conductors.

Degradation of the aluminium strands of the conductors analysed occurred at the clamps. Specifically, at the strand-to-strand (Figs. 3.4 and 3.7) and strand-to-clamp (Fig. 3.2) contact interfaces, in agreement with earlier reports [4, 8, 9]. It has been well established that the contact interfaces of the strands and the clamping region (Fig. 3.2a, b and c) are the most susceptible to fretting fatigue, fretting

wear [22-24] and corrosion degradations. This is primarily, due to the constraint imposed by the clamps leading to fretting damage under wind induced vibrations. In addition, conductors are subjected to static load comprising self-weight, tension due to installation, and clamping pressure. Moreover, the clamps create favourable condition for retention of corrosive media to cause crevice corrosion (Fig. 3.3a and b) particularly in the bottom of the conductor. In addition, clamps interact with the outer surface of the conductor causing fretting wear (Fig. 3.2a) and fretting fatigue failures (Fig. 3.2a and b) [3].

Clamping pressure generates normal contact force at the contact interfaces of the strands. This load causes plastic deformation. The resulting damage depends on the interfacial contact condition, material behaviour and environment [25]. For example, the damages inflicted because of variation in the normal contact force and relative slip amplitude have been well established in the form of fretting regimes constituting the fretting map [26]. These regimes represent different forms of fretting corrosion damages, namely fretting fatigue and wear-corrosion degradations (Fig. 3.4a, b and c).

The degradation morphologies observed at the contact surfaces of strands appear to correspond to the fretting regimes generally classified as stick, mixed stick slip and gross slip [26] where fretting damages were categorized with respect to their morphologies. Accordingly, the strand surfaces subjected to stick regimes of fretting experienced only plastic deformation, shown in Fig. 3.4a. Such surfaces in stick contact preserve their passivity against corrosion. The vibration amplitude was also not large enough to produce relative slip at the contact interfaces. Typically, the failure mode which may occur in this type of contact is high cycle fatigue (HCF) described in §2.

The surface morphology shown in Fig. 3.4b corresponds to the description of fretting contact in a mixed stick slip fretting regime where the contact surface partly experiences relative slip while the centre remains still in stick contact due to elastic accommodation of part of the imposed amplitude. Wear and corrosion degradation occur at the rim of the contact surface subjected to relative slip. Subsequently, the slip area of the contact showed etching and pitting (Fig. 3.5a and b) and accumulation of corrosion products (Fig. 3.4b). It is also the case that fretting contacts in this regime predominantly cause fretting fatigue failure [27], especially in the added presence of corrosion (see Fig. 3.9a). Corrosion appears to contribute to crack initiation in the form of pitting and to its growth by oxidizing freshly exposed metal at the crack tip.

The surface degradations observed in Fig. 3.4c and d correspond to a typical fretting damage occurring in the gross slip fretting regime in which the entire interface experiences relative slip. This type of degradation results when the relative slip amplitude exceeds the critical threshold causing mixed stick slip fretting condition [28]. The degradation is mainly in the form of material removal from the surface, *i.e.* wear, which theoretically removes the surface cracks initiated before they grow and lead to fracturing. The fretting wear process caused the accumulated debris of corrosion products at the interface. Some of the fretting wear damages (Figs. 3.2a and 3.4d) may be in the category of reciprocating sliding wear, which occurs at larger slip amplitudes, and causes thinning of the conductor cross-section as described in §2. This may be clarified by studying the transition of fretting wear to reciprocating sliding wear using wear coefficient and wear volume studies besides morphological studies [29].

The wear damage in general occurred due to the rubbing of the strands against the clamps (Fig. 3.7a and 3.10) or alike strands in contact (Fig. 3.6a). The wear mechanism seems to be delamination, by which the surface is deformed, and subsurface cracks are developed in the rubbing process, causing the flaking away of metal fragments from the surface (Figs. 3.6 and 3.7). In addition, abrasive wear mechanism seems to have contributed to some extent due to plastic deformation and fragmentation of metal from the surface acting as abrading particles (Fig. 3.7c and d).

Breaking of the strands occurred due to cracks initiated at the contact surfaces subjected to fretting contacts in a mixed stick slip condition (Figs. 3.9a) and gross slip regimes (Figs. 3.8a and 3.10), which also involved corrosion. Initiation of cracks leading to fracture is common on contacts exposed to the mixed stick slip fretting regime, whereas breaking under the gross slip fretting regime is less common. Azevedo and Cescon [12] reported breaking of strands caused by gross slip fretting damage at a spacer clamp. This suggests that both the fretting wear and fatigue-corrosion may cause breaking of the conductor strands. This necessitates the understanding of fretting corrosion degradation mechanisms and determination of the time needed for the cracks to initiate and grow to the critical length at a clamp.

Corrosive etching (Fig. 3.8c, 3.9c and 3.9d) of the fretted surfaces may have occurred after the conductor was taken out of service. Micropits may not survive the deformation of the surface during the fretting processes. However, the crystallographic pitting corrosion morphology on the slightly fretted surface (Fig. 3.9e and f) suggests that corrosive media might have penetrated into the surface

cracks. Otherwise, the role of corrosion in the presence of rubbing may be limited to the passivation of the metal surface freshly exposed by wear. Thus, the sequential effect of corrosion and fretting on a surface is of open discussion. It is suggested that the fretting damages the surface and creates favourable condition for corrosion by exposing an active fresh metal surface that spontaneously oxidizes whether it is on a surface or at the crack tip. The brittle oxide formed at the crack tip easily breaks and contributes to further crack propagation. Corrosion may also create weak spots (stress raisers) on a strand where fatigue cracks initiate under wind-induced vibrations.

3.4.2 Lifetime estimation

All fracture surfaces observed in this part of the work grew to the critical length at about 45° to the plane of contact [30]. Beyond the critical length, cracks continue to propagate inclined (Figs. 3.10), in a v-shape (Fig. 3.11a) or nearly perpendicular to the strand surface (Fig. 3.11b). The proportion of the tangential and axial stresses determine the direction of crack growth [31]. The inclination of the crack decreases with relative increase in the tangential stress, suggesting significant contribution of the fretting load. The fretting fatigue life is the sum of the time needed for crack initiation, growth and ductile fracture (Fig. 3.11b) of which the ductile fracture phase (stage III) is very fast and hence irrelevant.

The crack propagation period during Stage II by corrosion fatigue can be estimated from fatigue striations marks (Fig. 3.11b) by approximating the frequency of vibration and the number of striations contours, though this type of fracture surface profiles are rarely encountered as reported in the results section. If each striation mark results from a cycle of vibration, it is possible to approximate the duration of the crack propagation by estimating a reasonable frequency and number of vibration cycles corresponding to the number of the striations. By looking at the striations in the figure, it is impossible to argue that the crack propagation period in stage II would be a considerable portion of a conductor strand service life, which is normally about 40 - 50 years. Therefore, the service life is controlled by the crack initiation phase on contact surfaces and growth to the critical length by the combination of static and dynamic loads and corrosion.

3.4.3 Corrosion

Crevice corrosion damage in the crevices formed between the strands (Fig. 3.12) and between the clamp and the strand (Fig. 3.2a and b) was caused by penetration of moisture into the crevices. In

addition, it seems that crevice corrosion occurred in the mixed stick slip areas in fretting contact (Figs. 3.5, 3.4b and c). There was pitting attack on the strand surfaces, which seems to be caused by microgalvanic coupling between the noble Fe containing intermetallic phases distributed in the aluminium matrix (see Figs. 3.13a and 3.14), leading to dissolution of the metal around the phases [19]. Although these pits are relatively shallow and oval shape, they may accelerate fatigue crack initiation [32]. Furthermore, the thermo-mechanical processing of the strands during production seems to be related to the axially elongated corrosion attack (Fig. 3.13a) following the grain orientation displayed by the microstructure (Fig. 3.19). These suggest that conductor construction, material composition, the production of the strands may contribute to the mechanical and electrochemical degradations.

One possible reason for the occurrence of crevice and pitting corrosion, is the insufficient protection provided to the aluminium strands by the Zn layer on the steel core. It requires adequate electrolytic coupling between the Zn layer and the local crevices. Periodic and continuous wetting of the gaps between the strands and the Zn layer was probably not sufficient for this requirement of effective cathodic protection. In any event, the main purpose for galvanizing the steel core is probably for the protection of the steel core itself.

There was no considerable degradation of the zinc coating on the steel core. The signs of electrochemical (Fig. 3.15) and mechanical degradations (Fig. 3.16) observed local depletion (damages) which may not affect the sacrificial capability of the coating significantly, as the remaining layer can still give the cathodic protection. Therefore, the galvanized zinc coating was not the cause of failure for the conductors investigated.

3.5 Conclusions

- Contact surface morphologies corresponding to the stick, mixed stick slip, and gross slip fretting regimes were observed on the contact interfaces of the strands at the clamp exit. The degradations observed in each of these regimes were plastic deformation, fatigue cracks, and fretting wear.
- Fracture of the strands occurred at the contact surfaces which experienced mixed stick slip and gross slip fretting condition.

- Fretting wear and fatigue were the main degradation mechanisms of the aluminium strand samples obtained from power lines failed during service in marine atmospheres.
- Crevice corrosion was observed at the interstices of the outer layer strands with the clamps and adjacent strands.
- Pitting corrosion occurred in the areas affected by crevice corrosion and associated with the Fe containing noble intermetallic compounds distributed in the aluminium matrix.
- Contribution of the corrosion observed on the failure of the aluminium strands could not be assessed.
- Fractographic analyses suggested that the major part of the degradation period of a conductor strand is related to processes leading to crack initiation and growth to critical length. The contribution of crack propagation phase beyond this critical length to the service lifetime is marginal.
- Zinc coating on the steel core appears to have protected the steel core effectively, as its main expected function. The coating does not stop localized corrosion on the outer layer aluminium strands due to lack of sufficient electrolytic coupling.

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4 Sliding wear of AA 1350 strands in air and chloride solution

Abstract

The sliding wear behaviour of AA 1350 aluminium alloy was investigated in air, tap water, and 3.5% NaCl solution. The tests in the NaCl solution were carried out at open circuit and potentiostatically at potentials applied in the range -1.46 V_{SCE} to -0.62 V_{SCE}. The current density and coefficient of friction were measured continuously during the test in chloride solution. Work hardening, loss of the material by wear and surface morphology were investigated at the end of the tests. The wear degradation of the strands was significantly higher in air than in aqueous environments (tap water and chloride solution). This is attributed to the lubricating effect of the aqueous environment. The destruction of the protective oxide film by the sliding wear at open circuit potential causes cathodic polarization of the surface to a substantially negative mixed potential due to increased rate of oxidation and reduction reactions on the depassivated areas. In the wear track, the wear degradation was dominant over that of corrosion. Although small relative to wear in the degradation process, corrosion causes rapid oxidation of any bare metal exposed to the solution. At cathodic potentials higher than -1.34 V_{SCE} and anodic potential higher than -0.74 V_{SCE}, the mechanical wear does not influence the electrochemical process though wear is still the dominant degradation mechanism. Application of more negative and positive potentials than these limits caused alkaline and acid corrosion, respectively. In view of the dominating role of sliding wear in chloride solution under applied potential, degradation rate control is not possible by electrochemical means.

4.1 Introduction

Commercial purity aluminium alloys, particularly AA 1350, are widely used as conductor strands as reviewed in §2.4. The aluminium strands constitute one or more helical and concentric layers of aluminium strands around a stranded and galvanized central steel core in steel reinforced aluminium conductor (ACSR), which is typically used in marine environments. Galvanizing of the steel core provides corrosion protection and serves as a mechanical barrier between the aluminium and steel strands [1, 2]. Failure of these conductors usually occurs due to degradation of the aluminium strands at the clamps and fittings [3] by wear [4] as discussed in §2. Corrosion is suggested also to contribute to the failure [3], as discussed in §2. As reported further in §3, degradation of the strands in marine environments has been suggested to involve a mutual action of corrosion and wear and fatigue degradations.

As explained in §1 and 2, wind induced vibration leads to conductor degradation by causing cyclic bending of the conductor with respect to a clamp. This creates a relative slip (rubbing) at the contact interfaces of the conductor strands, described in §2.4, and leads to fretting fatigue and wear degradations depending on the magnitude of the relative displacement arising at the interface [5-7], which determines whether the degradation mechanism becomes fretting fatigue, fretting wear or sliding wear, as reported in §2.8. However, it is difficult to quantify this parameter due to lack of access to the contact interface. In some cases, the arising wear degradations are characterized as sliding wear at small relative displacement amplitudes as reported in §3. Since the degradation occurs in marine environments, it is also necessary to investigate the effect of electrochemical corrosion [8] on the wear degradation of the aluminium strand material.

The galvanized zinc coating on the steel core strands affects the electrochemical conditions on the of the aluminium strands as it is intended also for cathodic protection against localised corrosion of the aluminium strands. The deterioration of the coating is often used as the beginning of the end of service life of ACSR conductors used in marine environments [9, 10]. Interaction of the electrochemical and mechanical factors determining the conditions for the degradation of the strands is not well known. Hence, studies combining mechanical wear and corrosion at a possible range of potentials are needed to clarify the tribocorrosion degradation mechanisms of the strand material.

The open circuit potential of a commercial purity aluminium in neutral NaCl solution drops to a very low negative value in the presence of wear, which removes the passive oxide film protecting the surface and exposes the bare metal to the electrolyte [11, 12]. At such negative potentials, the surface can undergo cathodic corrosion [13, 14]. Cathodic corrosion can also occur locally by chemical dissolution of the oxide film on the aluminium matrix, caused by increased pH due to evolution of hydrogen gas at the cathodic sites Fe - rich intermetallic particles, which are more noble than the surrounding aluminium matrix (the anode). This leads to microgalvanic coupling between the intermetallic particles and the aluminium matrix in the presence of corrosive environment, hence leading to formation of small pits around the particles [15-17]. Information on the possible interaction between wear and such localized corrosion phenomena is also scarce. Therefore, the purpose of the present study is to investigate the wear degradation of alloy (AA 1350) in NaCl solution at the relevant potentials from -1.46 to -0.62 V_{SCE} either impressed or prevailing at the open circuit potential. Furthermore, the wear behaviour of the material in air, and wetted surfaces in tap water and NaCl solution is also investigated.

4.2 Experimental

4.2.1 Materials

The strand material was the drawn-hardened aluminium alloy AA 1350 whose chemical composition was presented in §3.2. The specimens were embedded in epoxy and metallographically polished to 1 μ m diamond paste to expose a flat rectangular area of 79.5 mm² (3.71 mm in diameter and 25 mm in length).

4.2.2 Galvanic corrosion test

In a preliminary study, galvanic corrosion (in the absence of wear) between a galvanized steel strand surface and an AA 1350 alloy strand was investigated. This was intended to quantify the couple potential and the galvanic current occurring between the two metals in NaCl solution. The specimens were degreased in acetone prior to exposure to 3.5% NaCl solution, which was maintained at 25 ± 1 °C and continuously stirred at a fixed rate by use of a mechanical stirrer. The galvanic contact was provided and monitored by a G750 Gamry potentiostat operated in the zero-resistance ammeter mode. The aluminium and zinc coated steel electrodes had an equal exposed area of 81 mm². The galvanic couple potential and the galvanic current were continuously measured for 2 h.

4.2.3 Potentiodynamic polarization data

Electrochemical polarization data were obtained in a neutral chloride solution in the potential range $-0.900 - -0.150 \text{ V}_{\text{SCE}}$ at a rate of 0.1 mV/s in anodic direction by use of a G750 Gamry potentiostat. The sample geometry used is as described in §4.2.1.

4.2.4 Tribocorrosion Test Procedure

A reciprocating tribometer described in reference [18] was used for the tests (Fig. 4.1). The tribometer was equipped with a spring regulated normal force capability in the range 1 - 70 N and sliding frequency range of 1 - 8 Hz. The apparatus included tribocorrosion software (Tribocorrosion software V01.12.06 R01) to perform the potentiostatic wear tests. It was a reciprocating ball-on-flat tribometer with a classical slider-crank mechanism driven by a DC motor. The instrument produced a reciprocating motion enabling an alumina ball (4.76 mm diameter) to rub against the sample surface. The normal force was applied through a compressive spring, and the tangential frictional force was measured using transducers. As shown in the figure, the sample, used as the working electrode, along

with the counter electrode (made of coiled platinum) and the reference electrode were immersed in the electrolyte. The equipment was capable of measuring the number of cycles, frequency and the electrochemical parameters (current and potential) continuously during the test. The potential was measured with reference to a standard saturated calomel electrode (SCE), which has a standard potential of 0.2441 V vs. the standard hydrogen electrode (SHE).



Fig. 4.1 Schematic illustration of the experimental set up used for tribocorrosion tests [18]

All experiments were conducted at a temperature of about 25 ± 1 °C. Before each test, the alumina ball and the sample surface were cleaned first in acetone and then in ethanol. The normal force, frequency of oscillation, and the displacement amplitude used in each experiment is 2 N, 1Hz, and 5 mm, respectively.

At this normal load, the average contact pressure induced on the sample was calculated based on the mechanical properties of the two bodies and presented in Table 4.1. This loading condition was chosen to obtain the wear on the surface without inducing damage in the bulk of the material. Calculation showed that the loading causes a maximum shear stress of 210 MPa, which is higher than the ultimate shear strength of the material, at a depth of 0.017 mm below the contact interface. Hence, the loading is expected to cause wear degradation on the sample surface upon rubbing.

Material	Young's modulus (GPa)	Poissons's ratio	Yield Strength (MPa)	Ultimate Strength (MPa)	Initial Hertzian contact pressure (MPa)
AA 1350	70	0.33	165.5	186.1	640
Alumina	300	0.21	665		

Table 4.1 Mechanical properties of the substrate metals and the calculated contact pressure against alumina ball (Ø4.76 mm) under a normal load of 2 N.

After immersion in 3.5% NaCl solution, the samples were held at open circuit for 240 s, and the corrosion potential was recorded. A selected potential was then applied for the next 240 s, during which the current on the specimen was recorded. At the end of this period, sliding was started and continued for the next 4200 s. The applied potential was maintained for the next 240 s after sliding was stopped. After switching off the applied potential, the corrosion potential was measured for another 240 s. At the end of the test, the sample was rinsed in distilled water and dried. The total sample area exposed to the solution was about 79.5 mm² including about 5.0 mm² wear track area. The applied potentials were selected at $0.12 V_{SCE}$ intervals in the range $-1.46 - -0.62 V_{SCE}$. A fresh sample was used for each potential step. The test was performed also under open-circuit conditions. The OCP was measured for 240 s before the rubbing was started and after it was stopped. It was monitored for 4200 s while rubbing was applied.

In addition to the above, sliding test in a laboratory air was conducted to simulate wear degradations in ambient atmospheric conditions. Moreover, wet condition in 3.5% NaCl solution (unlike the immersed condition above) was simulated by continuously dripping the solution on the sample surface and absorbing it by tissue paper held beside the wear track. The wet test was also performed in tap water to simulate chloride-free environment. The wet tests in water and chloride solution will be designated as "wet water" and "wet chloride" solution, respectively, throughout this chapter.

4.2.5 Post-mortem surface characterization

The sample was removed from the rig, washed, rinsed and dried for post mortem investigation of the morphology by a Zeiss Ultra 55 field emission gun scanning electron microscope (FEG-SEM). The corrosion products were cleaned from some of the samples by stripping in a standard hot chromic-phosphoric acid bath (ASTM G1-90). Wear volumes were estimated by using an optical 3D confocal

microscope (IFM Alicona) on stripped wear tracks. Micro-hardness was measured inside the wear track, after cleaning off the oxide, using Vickers indentation at a load of 0.05 kg applied for 15 s, according to ISO6507.

4.3 Results

4.3.1 Potentiodynamic polarization of AA1350 aluminium strand

Fig. 4.1 shows the polarization data for a polished and as ready for sliding wear test aluminium strand sample obtained at a scanning rate of 0.1 mV/s from point A to C in 3.5% NaCl solution.



Fig. 4.2. Polarization curve for AA 1350 aluminium alloy strand at a potential sweep rate of 0.1 mV/s in 3.5% NaCl solution. The samples were prepared as described in §4.2.

In the range between the corrosion and pitting potentials, the aluminium strand behaves as passive a small average current density of order 10^{-3} mA/cm². At about -0.750 V_{SCE}, which is the pitting potential, the current density increases sharply due to breakdown of the passive oxide film, thereby causing pitting [19].

4.3.2 Galvanic Corrosion

The couple potential of the galvanic cell described in section 4.2.2 is shown in Fig. 4.3a. It was quite steady during the duration of the experiment, stabilizing at -1.034 V_{SCE} after 2 h exposure. The value is close to the corrosion potential of Zn in sea water [20, 21]. Since zinc is more active than the aluminium alloy at this temperature, it controls the couple potential. The galvanic current density increased initially to a maximum of 2.70 μ A/cm² and then declined gradually to a steady-state value of 2.47 μ A/cm² after 5500 s, as shown in Fig. 4.3b.



Fig. 4.3 Measured a) couple potential and b) current density between galvanized steel and aluminium strands during the galvanic corrosion test as a function of immersion time in 3.5% NaCl solution. The exposed area of each sample was 0.81 cm².

4.3.3 Triboelectrochemistry

The following subsections report the series of results obtained for each selected consecutive triboelectrochemical step applied to a given specimen. These steps can be summarized as follows:

Wear at OCP (Fig. 4.5)

- OCP in the absence of sliding wear
- OCP with sliding
- OCP after interruption of sliding

Wear in the presence of potentiostatic polarization

- Potentiostatic current after application of potential (no sliding; A to F in Fig. 4.6 and 4.7)
- Potentiostatic current with sliding wear (F to I in Fig. 4.8 and 4.9)

Potentiostatic polarization after termination of sliding (I to J in Fig. 4.9 and 4.10)

OCP after termination of sliding and potentiostatic polarization (J to K in Fig. 4.11)

Wear at OCP

Before the onset of sliding, the open circuit potential (OCP) of the sample loaded and immersed in the NaCl solution was nearly constant at - 0.76 V_{SCE} [19] as shown in Fig. 4.4. When rubbing was started after 240 s, the OCP decreased rapidly to a substantially low value of about -1.38 V_{SCE} as shown in Fig. 4.5. The OCP recorded during the sliding wear corresponded to the entire surface exposed to the solution, constituting the worn and unworn parts of the surface with an area ratio of about 0.20. The decrease of the OCP is due to exposure of active bare metal to the solution as a result of destruction of the protective oxide film by rubbing, which is a well-known behaviour for a damaged aluminium surface in NaCl solution [22].



Fig. 4.4 Open circuit potential of AA 1350 alloy recorded in 3.5% NaCl solution.



Fig. 4.5 Sharp fall of OCP at the onset of sliding to a very low negative value and evolution of the potential during sliding. The test was conducted by rubbing an AA 1350 alloy sample against alumina ball (4.76 mm diameter) in a 3.5% NaCl solution for 4200 s at F = 2 N, f = 1 Hz and displacement amplitude of 5 mm.

As the sliding continued, the potential increased slightly in the anodic direction and then slightly decreased in an undulating behaviour towards a nearly steady-state value of -1.32 V_{SCE} . The high-frequency oscillations were related to activation and repassivation of the surface in the wear track. The undulating behaviour was long term changes occurring in the wear track, such as accumulation and removal of oxidized metal fragments and debris. The potential measured during the rubbing was a mixed potential determined by the kinetics of corrosion of the active surface exposed in the wear track and reduction of dissolved oxygen and evolution of hydrogen on the whole exposed surface. When the sliding was stopped, the potential increased sharply in the positive direction. This indicates that the wear track repassivated rapidly with time, but at a rate slower than the initial drop when the rubbing started.

Wear in the presence of potentiostatic polarization

The OCP measured prior to the potentiostatic polarization and onset of sliding reproduced that shown in Fig. 4.4. When the potential was applied after 240 s at OCP (point A in Fig. 4.6; still no sliding), the current density increased in the negative direction for all potentials, as shown in Fig. 4.6 except at -0.62 VSC (which is above the pitting potential). The potentiostatic current density transients before the onset of sliding, exhibited very small negative levels of order -10⁻⁶ A/cm² for potentials -0.74 V_{SCE} and -0.86 V_{SCE}, indicating essentially passive surface conditions. This confirms the passivity potential range obtained by the potentiodynamic curve presented in Fig. 4.1. The negative current density transients increased to a maximum (labelled B) with increasing negative applied potential to -1.22 V_{SCE} and decreased again to a lower steady-state value, which also depended on the applied potential (decreasing with decreasing negative applied potential as indicated by arrow C). At -1.34 and -1.46 V_{SCE} (Figs. 4.6 and 4.7, respectively), the negative current density transients attained a maximum as in the case of the more positive potentials (labelled B), decreased to a minimum (labelled D on the figures) and then increased monotonically with time (arrow E). At -0.62 V_{SCE} , which is more positive than the pitting potential, the current density increased significantly in the positive direction from the instant the potential was applied (Fig. 4.7), indicating anodic dissolution of the material.



Fig. 4.6 Potentiostatic current densities recorded on AA 1350 alloy surfaces (for 240 s) held at the specified potentials (SCE) in 3.5% NaCl solution from the moment the potentials were applied (A) to the onset of sliding wear (F). The labels B and D represent the maximum and minimum negative current densities, respectively. Arrow C indicates decreasing trend of the negative current densities with decreasing negative potential. Arrow E indicates monotonically increasing negative current density.

Fig. 4.8 shows the current transients obtained during 4200 s after the onset of sliding wear on the samples still polarized potentiostatically as discussed above. Total current is reported in this figure for 79.5 cm² total exposed area including the wear track area because it was impossible to distinguish between the active and passive areas of the worn surface exposed during the sliding test. When sliding was started (labelled F), the currents at all potentials rose sharply in the positive direction except for -0.62 and -1.46 V_{SCE}, which responded marginally (see Fig. 4.9). After the initial rise (Fig. 4.8), the currents decayed to lower, more passive, values with time, except for -1.34 V_{SCE}. The anodic peak at the start of sliding and the steady state value attained later increased with the applied potential. The current surges, labelled H on the curves at -0.74 and -0.86 V_{SCE}, must be due to destruction of the passivating layer in the wear track, exposing bare metal to the solution, analogous to that at the start

of sliding. The sudden increase in the current was more often on $-0.74 V_{SCE}$, which corresponds to a slightly higher value than the critical pitting potential for this alloy in the test solution. At -0.86 to $-1.10 V_{SCE}$, the net current transients decreased to steady state values, which became smaller with decreasing applied potential, but they remained positive throughout the sliding period. The current transient decayed from initially positive to a small negative steady state value at $-1.22 V_{SCE}$. At more negative potentials, the current transients were negative throughout the sliding period (see Figs. 4.8 and 4.9 for -1.34 and $-1.46 V_{SCE}$, respectively). The decrease of the current transients (or increase of the negative currents) with decreasing applied potential indicates increasing contribution of the reduction reaction to the net current.



Fig. 4.7 Potentiostatic current densities, recorded on polished AA 1350 alloys held at -0.62 V_{SCE} and -1.46 V_{SCE} for 240 s in 3.5% NaCl solution, from the instant the potentials were applied (A) to the onset of sliding (F). The specimens were held at OCP for 240 s prior to application of the potential. The labels B, D and E represent the maximum, minimum, and increasing trend, respectively of the negative current density at the cathodic potential.

Fig. 4.9 shows the current density transients for the applied potentials at -1.4 and -0.62 V_{SCE} before the onset of sliding (A to F), under sliding wear (F to I), after sliding was terminated (I to J), after the current was turned off to attain OCP condition (J to K) and the end of the experiment (K). The current density is based on the total exposed area of the sample. The current density transients were not affected by the onset (arrow F) or termination of sliding (arrow I). The current densities increased continuously in the negative and positive directions, respectively until the potential was turned off (J). This indicates that the electrochemical process on the surface of the specimen was not affected by the mechanical wear at such highly negative and positive potentials, respectively and beyond.



Fig. 4.8 Potentiostatic current transients recorded from the start (F) to interruption (I) of sliding wear on AA 1350 alloy held at various applied potentials (SCE), specified in the figure, for 4200 s in 3.5% NaCl solution, applied normal force of 2 N, displacement frequency and stroke of 1 Hz and 5 mm, respectively. Prior to the start of sliding, the samples were held for 240 s at OCP, followed by 240 s at their respective potentiostatic potentials, as shown in Figs. 4.6 and 4.7. Sporadic current surges occurred after reaching steady-state for -0.74 and -0.86 V_{SCE} (labelled H). The current levels increased in the negative direction with increasing negative potential. After the sliding was stopped, the samples were held at the same respective potentials for 240 s and then at OCP for another 240 s (Fig. 410.).



Fig. 4.9 Potentiostatic current density transients on AA 1350 alloy in 3.5% NaCl solution from the moment the potentials specified in the figure were applied (A), sliding wear was started (F), sliding was interrupted (I), the potential was removed (J) and the experiments were stopped (K). The specimens were held at OCP before the potentials were applied at A and after the potentials were removed at J (plotted in Fig. 4.11). The sliding wear test was performed at a force of 2 N, frequency of 1 Hz, displacement stroke of 5 mm.

Potentiostatic polarization after termination of sliding

Fig. 4.10 shows potentiostatic current transients measured for 240 s after sliding was terminated at point I in Fig. 4.8, with the potential still applied. The current at each potential shifted in the negative direction when sliding was interrupted. The current shift became smaller with increasing negative potential; and it is then interpreted as decrease in the anodic contribution from oxidation of bare metal exposed to the solution during the last sliding pass. At -0.74 V_{SCE}, the current dropped from 125 μ A to 75 μ A when sliding was stopped and then increased gradually in the positive direction. It was significantly higher than the value before the onset of sliding when the surface was still passive (cf. Fig. 4.6). Since the potential is slightly above the pitting potential, corrosion, which had initiated by wear before the interruption of sliding, continued to propagate after sliding was stopped. At -0.86 V_{SCE}, which is well below the pitting potential, the current stabilized at 12.5 μ A, which was also appreciably higher than that before the onset of sliding. Although most of the surface must have passivated, it is possible that few pits continued to propagate at occluded sites. Pit propagation is known to occur at potentials more negative than the critical pitting potential under such conditions [23]. At -0.98 V_{SCE}, the steady-state positive current of 4.5 μ A was also significantly higher than the negative value before the start of sliding, although it was more typical of a passive surface. At -1.1 V_{SCE}, the current transient dropped from a positive value of 12.5 μ A (under sliding) to a negative value of -12.5 μ A when sliding was terminated, indicating passivation of the anodically active sites in the absence of wear. The current values obtained for the more negative potentials were all negative and more or less of the same order of magnitude as the values before the termination of sliding, indicating the increased importance of the reduction rate in relation to the oxidation rate. The significance of these results as far as the effect of wear on the surface will be discussed later in the chapter. The currents became zero as expected when the applied potentials were interrupted (J), and the samples were subsequently allowed to corrode freely until the end of the experiment (K).



Fig. 4.10 Current transients measured for 240 s at the specified applied potentials on worn AA 1350 alloy in 3.5% NaCl solution from the moment the sliding wear (cf. Fig. 4.8) was interrupted (I) until the current was also interrupted (J). The specimens were then held at open-circuit condition until the end of experiment (K).

OCP after termination of sliding and potentiostatic polarization

Fig. 4.11 shows open circuit potential transients recorded from the termination of potentiostatic polarization (arrow J) to the end of the experiments (K). It should be recalled that, each specimen was successively 1) subjected to the test solution at OCP for 240 s (\$4.3.3), 2) potentiostatically polarized for 240 s at a selected potential (\$4.3.3), 3) subjected to sliding wear at the same potential for 4200 s (\$4.3.3), and then 4) kept at same potential for 240 s prior to start of this step which lasted for 240 s.



Fig. 4.11 Open circuit potential transients recorded in 3.5% NaCl solution on specimen surfaces subjected to prior experimental steps with and without sliding wear and potentiostatic polarization at the potentials specified in the figure (SCE). The open circuit potential transients were recorded from the moment applied potentials were interrupted (J) to the end of the experiment (K).

The open circuit potential increased gradually with time in the anodic direction for all potentials at and below -0.98 V_{SCE} applied in the previous step (§4.3.3). In contrast, for samples previously exposed to potentials in the range -0.86 - -0.62 V_{SCE} , the open circuit potential transients dropped sharply at the outset and then continued to decrease monotonically in the cathodic direction. The steady-state values could not be obtained because of early termination of the test at K. The values are determined by chemical and mechanical processing of the surface in the previous steps, as will be discussed later in the Discussion §4.4.

4.3.4 Friction coefficient

Fig. 4.12 shows the coefficient of friction measured simultaneously for selected samples studied under the conditions discussed above, as specified in the figure and its caption. Due to the large scatter in the data (specially the test in air), the averaged results and standard deviation are presented in Table 4.2. The reported coefficient of friction is the average over the entire sliding duration The COF was lowest in water and highest in air. The COF values in NaCl solution, in wetted and immersed conditions, were higher than those in tap water. The COF in NaCl was slightly higher for the immersed than the wetted case. The smaller COF for the tests in water and NaCl solution compared to that in air is presumably an indication of the lubricating effect of the aqueous environment at the contact interface. A higher scatter was observed for the tests in air.



Fig. 4.12 Evolution of coefficient of friction with time obtained from sliding wear of AA 1350 alloy against an alumina ball (4.76 mm diameter) in air and 3.5% NaCl solution at different applied potentials (SCE). The test was performed at a force of 2 N, frequency of 1 Hz, and 4200 s duration. The average coefficient of friction and standard deviation in each case are presented in Table 4.1 for the test conditions considered.

The average COF value appeared to be highest at -1.34 V_{SCE} for the results obtained under the potentiostatic conditions specified in Table 4.1. A plateau value of about 1.1 was attained in the range -1.1 to -0.86 V_{SCE} before the COF decreased again with further increase in the applied potential in the anodic direction.

Table 4.2 Material loss in volume and average COF (± standard deviation) measured due to sliding
wear of the specimen under different test conditions (F = 2 N, $f = 1$ Hz, $t = 4200$ s, alumina ball
diameter = 4.76 mm).

	40 ⁻² 90 - 117	Coefficient of friction	Material loss volume (x10 ⁷ μm ³)
	Air	1.2 ± 0.4	27 ± 2
	H_2O^1	0.8 ± 0.1	6 ± 1
Test environment	NaCl ¹	0.9 ± 0.1	7 ± 3
	OCP	1.1 ± 0.1	9 ± 1
	-1.46	1.4 ± 0.1	11 ± 1
	-1.34	1.7 ± 0.1	11 ± 0
	-1.22	1.4 ± 0.1	8 ± 1
	-1.10	1.1 ± 0.1	7 ± 0
	-0.98	1.2 ± 0.1	7 ± 0
	-0.86	1.1 ± 0.1	7 ± 1
Applied potential (SCE)	-0.74	1.0 ± 0.1	8 ± 1
in NaCl solution	-0.62	0.8 ± 0.1	9 ±1

¹ The test environments specified as H_2O and NaCl were wet conditions described in § 4.3.5 (last paragraph). The electrochemical tests in NaCl solution were conducted on wetted specimen surfaces.

4.3.5 Material loss

The average material loss in terms of volume on each sample after the sliding tests and oxide cleaning procedure reported above is presented in Table 4.2. Under potentiostatic conditions, minimum wear loss occurred in the applied potential range of -1.10 - 0.86 V_{SCE} which corresponds to the potential range in which the surface is passive. The volume loss increased with change of potential both in the negative and positive directions. Comparatively, the material loss in air (27 x $10^7 \mu m^3$) was significantly higher than the rest of the test conditions, which can be attributed to the lubrication effect of the aqueous environments, as evidenced by the low COF values shown in Table 4.2. Material loss was slightly higher for the tests in chloride solution (immersed) than in tap water, indicating the contribution of corrosion in the more aggressive chloride environment.

4.3.6 Morphological Investigation

Morphology of surfaces tested in air

Fig. 4.13a shows secondary electron SEM micrograph of uncleaned wear track of the specimen tested in air. The surface was plastically deformed and sheared into flakes delaminated from the surface, called third body particles, which are randomly dispersed on the surface and visible inside the cracks transverse to the sliding direction (indicated by arrow). These third body particles may be oxide particles or fragments of metallic particles coated with oxide which can interfere in the wear process and influence the wear mechanism. As can be observed from the image, the dominant wear mechanism was delamination. In some areas of the wear track, the flakes were broken down into particle fragments. On other areas, however, the torn pieces were still attached to the surface. The damage on the surface confirms that the normal force chosen causes yielding and disintegration of the surface as expected in §4.2.4.

Fig. 4.13b shows the micrograph of the specimen surface presented in Fig. 4.13a after cleaning in a hot bath of chromic phosphoric acid. It shows plastically deformed metallic flakes and particles adhered to the surface, metallic ridges, abrasion marks and cracks transverse to the sliding direction, indicated in the figure. The flake surfaces appear to be extruded in the sliding direction. The disappearance of the particles dispersed on the surface (shown in Fig. 4.13a) after cleaning indicates that they were oxide debris. Moreover, abrasive wear was evidenced by the scratch lines parallel to the sliding direction (shown by arrow).





Fig. 4.13 Secondary electron SEM image inside the wear track of AA 1350 alloy after sliding test against an alumina ball (4.76 mm diameter) in air a) showing particles of metal and oxide distributed on the surface degraded by surface cracks transverse to the sliding direction b) Micrograph of the surface after cleaning in a hot bath of chromic phosphoric acid showing abrasion marks, metallic particles and flakes adhered to the surface.

Morphology of the wear track tested in tap water

The secondary electron SEM micrograph (Fig. 4.14a) of the sample tested in tap water (wetted surface) shows debris accumulated at the edges of the wear track. The wear track was covered with a dense layer of debris (see Fig. 4.14b), which creates a layer of third body soaked in tap water during the test. The morphology underneath the debris is shown in Fig. 4.14c after cleaning in a hot bath of chromic phosphoric acid thereby stripping the oxide. The surface was coarsely roughened by plastic deformation caused by the cyclic rubbing. The filling of the voids and occluded area by debris may have acted as lubricant and reduced the COF and material loss as shown in Table. 4.2.





Fig. 4.14 Secondary electron SEM image of AA 1350 alloy surface wear track after sliding test against an alumina ball (4.76 mm diameter) in a drop of tap water (wet condition). a) General view of the wear track with debris deposited at the end tips and inside the wear track covered with oxide. b) Micrograph of the wear track at higher magnification showing surface covered by dense debris. c) Micrograph of the surface after removing the debris, showing plastically deformed and roughened surface with mechanical pits and crevices.

Morphology of the wear track tested at open circuit in NaCl solution.

The morphologies of the samples tested in the NaCl media (wet and immersed) were quite similar. Hence, the results presented are limited to the immersed test conditions. Fig. 4.15a shows the image of the wear track with debris, accumulated mostly at the ends of the stroke and the side edges, indicating that the debris was removed mostly in the sliding direction. The wear track was covered by a layer of debris, shown in Fig. 4.15b, which was broken down and dispersed at some areas of the wear track and still dense at other sites. Therefore, the debris covering the wear track was not dense like the wet in tap water. The disintegration of the debris layer may have been caused by both the rubbing and corrosion processes. The debris on the surface acted as the third body and influenced both the wear and corrosion processes as indicated by fluctuation of the open circuit potential during

rubbing (cf. Fig. 4.5) and reduction in the coefficient of friction (see Table. 4.2). In the sliding process, the debris was soaked in the electrolyte and probably smeared on the surface as a layer.



Fig. 4.15 Secondary electron SEM micrograph of AA 1350 alloy specimen surface after sliding test in a 3.5% NaCl solution at OCP. a) Overview of inside and outside of the wear track with wear debris accumulated at the edges. b) High magnification image of the surface inside the wear track showing layer of debris. The test was performed at a normal force 2 N, a frequency 1 Hz, displacement amplitude of 5 mm and duration of 4200 s.

After cleaning the debris in hot chromic-phosphoric acid bath, typical micropitted morphology of an aluminium alloy, formed by immersion in NaCl solution, was revealed outside the wear track (Fig. 4.16a). Pitting occurs around the noble Fe-rich intermetallic phases, commonly present in commercially pure aluminium alloys [24]. Removal of the debris in the wear track by treatment in chromic-phosphoric acid shows that it was mainly composed of aluminium oxide. The morphology underneath the debris (Fig. 4.16b) was severely deformed into flake like topography with sheared and flattened like features. The surface appeared to be cyclically deformed by the sliding wear and detached as thin metal flakes. Unlike outside the wear track, there was no micropitting observed inside the wear track, due to the cyclic deformation of the surface.





Fig. 4.16 Secondary electron SEM image of the wear track of AA 1350 alloy after sliding test against an alumina ball (4.76 mm diameter) in a 3.5% NaCl solution at OCP. a) Surface outside the wear track showing micropit formation around the intermetallic phases. b) The wear track after cleaning in hot chromic phosphoric acid showing plastic deformation, roughening and ridging on the surface.

Morphology outside the wear track at applied potentials in NaCl solution

Fig 4.17 shows SEM secondary electron images of surface outside the wear track obtained from specimens tested through the sequence outlined in §4.3.3 after stripping in hot chromic phosphoric acid. At -0.62 V_{SCE}, which is above the pitting potential [19], there was no corrosion attack on the surface outside the wear track (Fig. 4.17a). However, the edge of the wear track, marked on the figure, was heavily corroded. The corrosion mechanism seems a general crystallographic etching related to the mechanically damaged areas. At -0.74 V_{SCE}, no considerable corrosion attack was found on the surface outside the wear track other than slight pitting (Fig. 4.17b). The surface polarized potentiostatically at -0.86 V_{SCE} did not show any significant sign of corrosion attack (Fig. 4.17c). With the applied potential increasing in the negative direction up to -1.34 V_{SCE}, shown in Figs. 4.17d through 4.17g, the surface showed increasing degree of micropitting. At -1.46 V_{SCE}, the number and size of alkaline pits increased towards the scalloped shape of an alkaline-etched (cathodically-corroded) surface (Fig. 4.16h) [24].








Fig. 4.17 Secondary electron SEM micrograph of surfaces outside the wear track after polarization at the potentials (SCE) specified in the figure, according to the test sequence presented in §4.3.2 and Figs. 4.8 and 4.9 after stripping the corrosion products. The sliding test was conducted at a normal force of 2 N, frequency of 1 Hz, and 5 mm displacement in 3.5% NaCl solution.

Wear track morphology under applied potentiostatic potentials

Fig. 4.18 shows the low magnification images of the wear tracks taken after the test sequences described in §4.3.3. At the anodic potential (-0.62 V_{SCE}), the edge of the wear track and mechanically defected areas outside the wear track were heavily corroded as indicated in Fig. 4.18a. No significant corrosion products or worn out debris could be detected possibly due to dissolution in locally acidified solution. The corrosion attack was significantly reduced when the potential was decreased to -0.74 V_{SCE} (Fig. 4.18b), which is slightly above the pitting potential of the alloy [19]. The layer of debris forming in the wear track was partially exfoliated (delaminated) for applied potentials down to -0.98 V_{SCE} (Fig. 4.18c and d). At more negative potentials, the layer of debris remained dense on the surface of the wear track and piled up at the peripheries of the wear track (Fig. 4.18e - h).







Fig. 4.18 Low magnification secondary electron SEM images of wear tracks tested at the potentials (SCE) specified in the figure according to the procedures described in §4.3.3 and presented in Figs. 4.8 and 4.9. The areas marked A were examined at higher magnification.

Fig. 4.19 shows SEM secondary electron images of the areas marked A in Fig. 4.18. At -0.62 V_{SCE} , which is above the critical pitting potential [19], the debris was dissolved and the metal underneath was attacked by corrosion (Fig. 4.19a). The corrosion attack decreased with decreasing potential from heavy and dense attack at -0.62 V_{SCE} to fewer pits at the edges at -0.74 V_{SCE} (Fig. 4.19b). At potentials from -0.74 - 0.98 V_{SCE} , the wear track was covered by thick layer of debris (oxide cake) which was partially exfoliated (marked in the figure) (Fig. 4.19c and d) as described in the preceding paragraph. The exfoliated areas showed cracks and abrasion marks, indicating that the cause was mechanical sliding. The sharp increase in the current transients, marked H in Fig. 4.8, during sliding wear may have been caused by the exfoliation of the layer of debris thereby exposing a bare fresh metal to the electrolyte. At potentials from -1.10 to -1.46 V_{SCE} , the wear tracks were covered by a dense layer of debris (Fig. 4.19e-f) except for a few micropits (Fig. 4.19f and h) and abrasion trace shown in Fig. 4.19f. Such abrasion may have been formed by entrapping of fragmented particles at the contact interface. At -1.34 V_{SCE} (Fig. 4.19g), the surface was decorated by dispersed white

particles which are probably salt particles left behind due to improper cleaning/contamination of the sample at the end of the experiment.









Fig. 4.19 Secondary electron SEM images of as tested wear tracks (Marked A in Fig. 4.18) of AA 1350 4200 s at various applied potentials (SCE) in 3.5% NaCl solution (see §4.3.3 for more detailed description of the electrochemical test sequence) at 2 N normal load, 5 mm displacement amplitude and 1 Hz frequency. The areas marked B in figures a to f (slightly magnified) and the whole area of figures g and h are presented in Fig. 4.20 after cleaning in hot chromic-phosphoric acid bath.

Wear track morphology-stripped sample

Fig. 4.20 shows SEM secondary electron images of wear track areas marked B in Fig. 4.19 except those which are at one to one scale, after stripping in hot chromic phosphoric acid solution. The removal of the debris in the reagent indicates that the debris constitutes corrosion deposit and possibly mixed with metallic particles. As the figure suggests in general, the wear track surface beneath the debris was plastically deformed and roughened by the wear and cyclic deformation.











Fig. 4.20 Secondary electron SEM images of stripped wear tracks of AA 1350 tested as described in Fig. 4.19. a) Crystallographic corrosion attack on the worn surface, especially the scratched sites. b) Light pitting attack on a surface which was deformed and ridged. c) and d) Mechanically deformed, sheared and fragmented surface in the wear track without indication of corrosion attack. e) and f) Mechanically worn surface morphology with ridged and deformed metal without indication of significant corrosion attack. g) Wear track morphology showing metallic deformation and ridges of metal caused by wear with some localized corrosion attack. h) Micrograph showing mechanically ridged and deformed surface with uniform corrosion attack (etching).

At -0.62 V_{SCE} , the wear track edge and surface were extensively attacked by crystallographic corrosion attack (Fig. 4.20a), especially at the scratched sites and the edge. The attack became smaller and more localised as the potential was decreased to -0.74 V_{SCE} as shown in Fig. 4.20b. The corrosion attack at these anodic potentials may have occurred after the sliding was terminated, *i.e.*, when there is no deformation destroying the crystallographic facets or etching feature. At potentials in the range -0.86 - -1.22 V_{SCE} , no considerable corrosion attack was detected (Fig. 4.20c-f). Possible cathodic pitting was impossible to observe in the presence of sliding wear due to continuous deformation of

the surface. In view of this, the crystallographic etching attack formed on the specimen tested at -1.34 V_{SCE} (Fig. 4.20g) was apparently caused by contamination with the NaCl solution as described in the preceding subsection. However, it is clearly visible that the degradation morphology varied with the applied potential. At -0.86 and -0.98 V_{SCE} (passivity range), the degradation occurred by plastic deformation and formation of cracks transverse to the sliding direction which lead to detachment of material from the surface in the form of fragments and delaminated flakes (see Fig. 4.20c and d). At more negative potentials down to -1.35 V_{SCE} , the degradation occurred in the form of large flakes and smaller fragments of particles. Nevertheless, the surface wear occurred by plastic deformation and breaking of the surface into debris (cf. Fig. 4.20e, f and g). The scratch line shown in Fig. 4.20f was revealed after cleaning the debris on the surface (cf. Fig. 4.19f). This must be caused by scouring of the surface by a detached particle trapped at the interface.

Fig. 4.20h shows the morphology of the wear track tested at -1.46 V_{SCE} . The surface was plastically deformed. However, it was not as rough as in the case of more positive potentials due to the fact that it was etched uniformly (Figs. 4.20h and 4.21) over the entire surface because of cathodic polarization at a highly negative potential. As presented earlier, the transient current was not influenced by the sliding wear (Fig. 4.9), indicating that the metal surface was actively etched from the moment the potentiostatic polarization was applied to until it was terminated. During the sliding, any feature of corrosion etching on the surface could be easily deformed and disfigured. Hence, the corrosion etching formed on the wear track must have occurred after the sliding was terminated and the specimen was held under potentiostatic polarization and subsequently at OCP in the test solution. Therefore, the rough plastic deformation on the surface is attributed to the wear process whereas the microscopic etching morphology inside the wear track (Fig. 4.21) corresponds to cathodic etching.



Fig. 4.21 High magnification secondary electron SEM image of the worn surface shown in Fig. 4.20h tested as described in Fig. 4.17. The metal surface was etched by corrosion on microscale.

4.3.7 Work hardening

Table 4.3 shows the results of average microhardness values measured inside the wear tracks formed in the specified test conditions. The hardness of the unworn surface of the sample outside the wear track was also determined. As shown in the table, the sliding caused an increase in hardness of the surface regardless of the test conditions and the potentials applied. In particular, the hardness of the surface tested in air increased significantly (85.6 MPa) compared to that of the unworn surface ($42 \pm$ 5 MPa). In comparison, there was little difference in the hardness of the surfaces tested in wet and immersed conditions, although the values are a bit higher than the hardness of the unworn surface. Hardness of the wear track formed in chloride solution (immersed) was slightly higher than the wet chloride, which in turn was slightly higher than wet water, as reflected also by the COF values (Table 4.2) and wear loss results.

		Hardness (MPa) ²
	Air	89.6 ±4
	H_2O^1	47.6 ± 1
Test environment	NaCl ¹	50.5 ± 1
	OCP	52.8 ± 2
	-1.46	48.2 ± 5
	-1.34	49.4 ± 6
	-1.22	53.9 ± 7
	-1.10	52.9 ± 2
	-0.98	50.6 ± 4
	-0.86	54.2 ± 3
Applied potential (SCE) in NaCl	-0.74	52.2 ± 6
solution	-0.62	51.6 ± 7

Table 4.3 Vicker's hardness measured inside the wear track under different test conditions (F = 2 N, f = 1 Hz, t = 4200 s, alumina ball diameter=4.76 mm)

The microhardness values measured in the wear track after the sliding tests in the potential range - $0.62 - -1.46 V_{SCE}$ and OCP in 3.5 % NaCl solution are also shown in Table 4.3. The hardness was measured after stripping the corrosion products from the surface. The hardness in the wear track was about 51 ± 2 MPa, independent of the applied potential within the limits of the scatter, as a result of work hardening [25].

4.4 Discussion

4.4.1 Comparison of wear in air and aqueous environment

An important result of this work is the finding that sliding wear is the dominating mechanism in the degradation of alloy AA 1350 in the environments investigated. In particular, degradation in air is significantly higher than that in aqueous environment under otherwise identical experimental conditions. In air, degradation was characterised by severe plastic deformation, damage in the form of grooves along the sliding direction, shearing of the surface transverse to the sliding direction, formation of craters by delamination and fragmentation of the deformed surface into particles and flake-like metallic debris (Fig. 4.13a and b).

In contrast, the wear degradation in tap water and NaCl solution at OCP and applied potentials were milder, indicating lubricating effect provided by the aqueous medium. The plastic deformation was minimal. Oxidised debris and surface film were formed, and these reduced direct contact between the specimen and alumina ball, changing the two-body contact into a three-body contact, reviewed in §2. The debris soaked in the aqueous medium functions as a mixed lubricant, providing hydrodynamic lubrication by the aqueous medium and boundary lubrication by the colloidal debris, as proposed by Jahanmir *et al.* [26]. Consequently, transfer of material from the specimen onto the alumina ball becomes reduced unlike the sliding in air where strain-hardened material transfer to the alumina ball accounts for higher wear degradation, according to the findings of Jia *et al.* [27] on bronze–graphite composite wear in water and air. In fretting at larger amplitudes as well, according to Sato *et al.* [28], the corrosion products formed under fretting in seawater reduce wear damage by acting as a lubricant, reviewed in §2.12.

There was no considerable difference in the volume of material lost by sliding wear (cf. Table 4.2) in tap water and NaCl solution at OCP and applied potentials in the range of -0.74 to -1.35 V_{SCE}. As explained earlier, the contribution of corrosion to material loss in these test conditions were by oxidation of any exposed bare metal resulting in oxide growth, which accounts for only marginal material loss in relation to that related to wear. Although the material loss values listed in the table appear to be slightly less for the sliding in tap water and at potentials more positive than -0.98 V_{SCE} in chloride solution in relation to the other experimental conditions, the differences are very small and within the error margins. However, the increased material loss at the applied anodic (-0.62 V_{SCE}) and cathodic (more negative than -1.35 V_{SCE}) potential indicates significant contribution of corrosion, although material loss by sliding wear is still the dominating factor.

4.4.2 Significance of OCP during sliding in chloride solution

In the absence of sliding, the OCP stabilizes at about -0.760 V_{SCE} , which is the expected value for the type of alloy investigated [19]. It is normally close to the critical potential for passivity breakdown, above which pitting corrosion occurs. However, onset of sliding wear at OCP caused a significant and rapid surge of the OCP in the negative direction, resulting in values undulating in the range -1.22 to -1.35 V_{SCE} (cf. Fig. 4.5). The measured OCP is still a mixed potential determined by the reduction processes, especially water reduction, on the bare metal exposed by sliding wear and oxidation of the bare metal. Thus, continuous exposure of bare metal, although for a brief period locally, favours increased rate of water reduction, which results in significant increase in the local pH, which in turn

restricts repassivation and also increases the rate of metal oxidation (cathodic corrosion). These phenomena pull the OCP to appreciably negative values, which is a cathodic potential range as can be seen from the polarization curve (Fig. 4.2). Consequently, the degradation (inside and outside the wear track) resulting at OCP was similar to those obtained from sliding wear at potentiostatically controlled potentials in the range of -1.22 - -1.34 V_{SCE} (cf. Figs. 4.15, 4.16, 4.17f and g, 4.19f and g, and 4.20f and g), to be discussed further below in the next subsection.

The initial surge at the outset of the first cycle is caused by the alumina ball ploughing into the surface by exposing highly active bare metal to the solution inside the wear track, initiating the cathodic corrosion outside the wear track as described above. Moreover, the surface undergoes severe plastic deformation (Fig. 4.16b). With subsequent sliding, the deformed layer detaches into fragments and forms a pile of debris at the edges, especially at the ends of the stroke (cf. Figs. 4.14a, 4.15a, and 4.18b-h). The wear debris is progressively detached from the surface. Part of the debris remains in the wear track, forming debris bed (Figs. 4.15, 16a and b), separating the contacting surfaces in the manner discussed in the previous subsection.

The undulation of the mixed potential value (see in the figure) during the sliding can be attributed to random changes in the active and passive areas related to mechanical passivity breakdown and the local electrochemical reaction kinetics in the wear track. The rapid increase in the potential in the anodic direction upon termination of the sliding, although much slower than the negative surge at the outset, is due to recovery of the surface after severe wear damage. It is known from study on aluminium surface scratched in a neutral aqueous solution that the potential rises first sharply as the surface repassivates and then at a much slower rate as the pH recovers locally at the debris packed affected sites by a diffusion controlled process, influencing the electrochemical reaction rates and thereby the OCP [29].

4.4.3 Current transients under applied potential prior to sliding

The potential range (-1.46 to -0.62 V_{SCE}) selected for the electrochemical polarization work includes values well outside the expected OCP range with or without sliding. These values were included for the sake of completeness and scientific curiosity, not knowing whether stray DC currents between the high AC power strands can originate in practice to justify the investigation of the extreme potential values. Figs. 4.6 and 4.7, showing the electrochemical behaviour without sliding, was obtained for comparison with the corresponding data in the presence of sliding. Moreover, the results are relevant

for determining the behaviour of the surface outside the wear track, which was exposed to the same potential as the wear track or imposed the same OCP, which was determined by the conditions existing in the wear track during sliding at open-circuit conditions. The cathodic behaviour of aluminium alloys in aqueous solutions is well-known and extensively discussed in the literature [13, 30-32]. It was invoked in the previous section to explain the negative OCP values observed during sliding.

The net current transient was negative at all cathodic potentials, before the onset of sliding wear (Figs. 4.6 and 4.7). The increasing cathodic net current densities in the negative direction with decreasing potential indicates that the reduction reaction rate was increasing. However, the increase in the current to a negative peak and its subsequent fall back to a minimum indicates passivation of the cathodic sites. After falling to the minimum, the currents stayed nearly constant for potentials anodic to -1.34 V_{SCE} . These steady-state current densities increased with increasing potential in the negative direction.

In the present case without sliding, the noble intermetallic particles are the dominating cathodic sites. The reduction reaction, dominated by water reduction, causes local increase in the pH near the surface of the intermetallic cathodes, which in turn destabilize the passivating oxide on the surrounding aluminium matrix, and causes pitting around the particles. The particles are eventually excavated from the surface by pitting. The pit repassivates due to ensuing reduction in the rate of water reduction and decreasing pH, causing the decrease in the cathodic current. The steady-state current density is determined by the degree of passivity obtained on the sample surface, determined by the applied potential, the final surface pH, and the extent to which fresh particles from the bulk of the aluminium matrix are exposed due to the cathodic pitting.

4.4.4 Corrosion outside the wear track during sliding

The applied potential or measured potential (OCP) are uniform over the entire exposed surface (whether sliding wear is applied or not). Therefore, the electrochemical behaviour discussed in the previous section in the absence of sliding and the corrosion morphology obtained outside the wear track in the presence of sliding are identical. Thus, the surface outside the wear track, subjected to cathodic potentials in the range of interest, developed pits around the Fe-rich intermetallic particles, with sizes increasing from a small fraction of a μ m to several μ m are formed by polarization in the range -0.86 to -1.35 V_{SCE}, as shown in Fig. 4.17, according to the mechanism discussed in the previous

subsection. This type of pitting is therefore superficial and does not cause significant damage or perforation, except at potentials more negative than $-1.35 V_{SCE}$, at which the alkalinisation may spread over the whole surface as fresh Fe-rich particles are continuously exposed, and the corrosion takes a form similar to that obtained by alkaline etching. At potentials more positive than the critical pitting potential of the alloy in chloride solution *ca.* $-0.76 V_{SCE}$, the corrosion mechanism occurs by acidic hydrolysis, according to the well-established electrochemical theory for this type of pitting [33-35].

4.4.5 Corrosion and wear inside the wear track during sliding

The current transients increased sharply when sliding was started and decayed to steady-state positive values for cathodic potentials more positive than -1.22 V_{SCE} (Fig. 4.8). This shows that the anodic partial current component increased due to oxidation of the depassivated area in the wear track and dominated over the cathodic partial current throughout the sliding period. The decaying of the current transient during sliding can be attributed to the formation of oxidized debris partially covering the surface as a third body, as discussed earlier and hindering complete depassivation by the subsequent sliding cycles. The passivation of the cathodic intermetallic particle sizes also causes the decrease in the cathodic partial current contributions. The decrease in the steady-state current density with increasing applied potential in the negative direction is also due to increase in the rate of water reduction, considering that the measured current density is a net value determined by the rates of oxidation and reduction processes.

At -1.22 V_{SCE} , the anodic partial current exceeded the cathodic partial current at the outset of sliding. However, it became dominated by the cathodic partial current shortly as reflected in the negative net current transient. This is due to further increase in the rate of water reduction by increasing polarization of the surface in the negative direction, as discussed in the preceding paragraph. The partial current densities due to oxidation and reduction become nearly equal at about -1.22 V_{SCE}, which is not too different from OCP measured during sliding in the absence of potentiostatic polarization, as expected.

At potentials cathodic to $-1.22 V_{SCE}$, the currents remained negative throughout the sliding period due to increased cathodic reaction with decreasing potential. Like in the case of more positive potentials, the negative net current increased sharply in the anodic direction at $-1.34 V_{SCE}$ due to increased anodic contribution by oxidation of aluminium at the depassivated surface at the onset of sliding. It started increasing in the negative direction with increasing rate of water reduction between points F and G in Fig. 4.8 for the applied -1.34 V_{SCE} and between A and B for -1.46 V_{SCE} in Fig. 4.7 with exposure of bare metal by wear as discussed for the foregoing cases. However, the negative current density for these potentials start decreasing after forming maxima at points G and B, respectively, which can be attributed to increasing resistance to flow of current by formation of oxidized debris at the interface (Fig. 4.19g). The influence of sliding wear on the corrosion process disappeared when the potential was decreased to -1.46 V_{SCE}. This shows that there exists a threshold potential between -1.34 and - 1.46 V_{SCE} (Fig. 4.8) at which the electrochemical process becomes independent of the mechanical wear. This threshold seems to coincide with -1.35 V_{SCE} as identified earlier [36] in the absence wear. Thus, the transition seems to be independent of the wear process.

At more negative potentials, the protective oxide film becomes unstable by increase of pH at the surface due to increasing rate of hydrogen evolution, leading to uniform etching of the surface as shown in Figs. 4.20h and 4.21 and discussed in §4.4.4 above. The continuing increase in the negative current regardless of the onset or termination of sliding (Fig. 4.9) indicates that the electrochemical process was not influenced by the presence of mechanical wear.

4.4.6 Electrochemical behaviour after interruption of sliding

The decrease in the net current upon termination of sliding (Fig. 4.10) indicates a decrease in the anodic partial current contribution due to surface passivation and termination of formation of fresh bare surface. However, at potentials more negative than -0.98 V_{SCE}, the steady-state current reached is determined by the applied potential, as was the case initially before the onset of sliding under potentiostatic polarization (§4.3.3), with the difference that the corresponding negative current densities are significantly higher after the termination of sliding (cf. Fig. 4.6 and 4.10), which can be attributed to increased surface area by corrosion and wear. Another factor can be the prevailing alkalinity of the debris covered surface in the absence of convection mixing caused by sliding, especially with increasing applied potential in the negative direction. Positive net currents prevailing after the sliding was terminated at potentials more positive than -0.98 V_{SCE} indicate that the anodic process initiated by sliding continued after the termination of sliding. This is analogous to propagation of localized corrosion forms, such as pitting, after the potential drops to values more negative than the pitting potential [23]. At -0.74 V_{SCE}, which is slightly more positive than the critical pitting potential, continuing increase in the positive net current is attributed to continuing pit propagation as suggested by the corresponding SEM micrographs in Figs. 4.19b and 4.20b.

The open circuit potential (Fig. 4.11) after interruption of sliding and current (Fig. 4.10) is difficult to interpret because of the different steps in the previous electrochemical and mechanical treatment of the specimens. The transients in Fig. 4.11 show recovery of the OCP for each case from the applied potential to a steady value, which seems to lie at about -0.95 V_{SCE} , which is a bit more negative than the initial -0.76 V_{SCE} of the undisturbed surface. A more negative OCP would indicate reduced cathodic area or increased resistance to the reduction processes. The decrease of cathodic area outside the wear track by removal of the Fe rich intermetallic phases may be part of the explanation. The cathodic intermetallic particles may also become buried under the debris bed, even buried under the deformed metal matrix smeared over during sliding, as Fig. 4.19 may indicate.

The potential of the surface was substantially influenced by the action of prior polarization and sliding wear presented in §4.3.3. As explained in the preceding paragraphs, the surface (inside and outside the wear track) experienced acidification, passivity, alkaline pitting and uniform etching with decreasing potential. In addition, the surfaces inside the wear track (Figs. 4.18 and 4.19) was covered with oxidized debris. Therefore, open circuit potential transient reflects the electrochemical process (repassivation) of the surfaces subjected to the conditions described. The trends are difficult to interpret because of the following factors. At potentials more positive than the pitting potential, the surface is forced to corrode by pitting (Step 2). The solution becomes acidified due to pitting corrosion. The surface is deformed and worn while the surface chemistry and morphology changes continuously in an unknown fashion during sliding under potentiostatic conditions in Step 3. During the subsequent step of continuing potentiostatic polarization in the absence of sliding, the surface morphology and microstructure formed by tribocorrosion disappears, especially with increasing applied potential in both positive and negative directions with respect to the OCP of the original polished surface (-0.765 V_{SCE}). The OCP measured afterwards is affected by these new conditions created. Since it is a mixed potential, it is difficult to interpret without additional polarization data simulating the conditions obtained. The same is true also for surfaces polarized at highly negative potentials. In this case the surface is alkalinized, which causes dissolution of the film and etching of the surface (cathodic corrosion). It is hoped and expected, however, that important morphological features due to sliding remain on surfaces exposed to benign conditions close to passivity region of the surface, which should lie in the potential range $-0.80 - -0.90 V_{SCE}$ [13, 14].

4.5 Conclusions

The sliding wear of AA 1350 alloy strand was investigated in air, tap water and NaCl solution, the latter at OCP and under controlled potentials from -1.46 to -0.62 V_{SCE} . The following conclusions are drawn from the study:

- Degradation of the AA 1350 alloy by sliding wear was significantly higher in dry air than in aqueous environment (NaCl solution and tap water), which provided lubricating effect to the surface. Therefore, aluminium strands in power lines are expected to be more resistant to sliding wear in wet than dry environments, given that the other controlling factors, such as the applied force and the sliding distance and frequency are the same.
- Mechanical wear component was the dominating factor controlling the degradation of the aluminium strands under immersed conditions and the mechanical parameters used in this investigation. Cathodic protection, such as by the Zn coating on the steel core, does not provide a significant protection against degradation by sliding wear.
- At potentials well above the critical pitting potential in the anodic direction (such as -0.6 V_{SCE}) and well more negative than -1.3 V_{SCE} in the cathodic direction (such as -1.46 V_{SCE}), electrochemical etching of the surface becomes significant and independent of mechanical wear. However, mechanical wear remains as the dominating degradation mechanism also for these cases.
- The sliding wear of AA 1350 alloy at OCP causes cathodic polarization of the surface to a
 substantially low steady state mixed potential of about -1.3 V_{SCE}. This is related to the
 exposure of bare metal by wear and resulting significant increase in the oxidation and
 reduction rates on the bare surface generated. However, mechanical wear still dominates the
 degradation process.
- Delamination is the dominating wear mechanism in air.
- The mixed potential established by the galvanic coupling of the aluminium strands and the galvanized steel core maintains the potential at about -1.034 V_{SCE} in the absence of wear. At this potential, the aluminium strands may experience cathodic corrosion (pitting). However, this is a superficial attack around the cathodic intermetallic particles. The attacked area repassivates effectively under the potential provided by the galvanic Zn layer after the particles are carved out of the surface.

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5 Fretting of AA 1350 alloy strand in air and chloride solution

Abstract

Degradation of AA 1350 aluminium alloy used for a conductor strand was investigated by fretting a flat polished specimen surface against an alumina ball in air and chloride solution. The displacement amplitude and normal force ranges used in the tests were 3 to 100 µm and 5 to 60 N, respectively. Fretting regimes (stick, mixed stick-slip and gross slip) were identified and a fretting map was developed for air and NaCl solution by analysing the friction logs and characterization of fretted surfaces. Fretting in the stick regime did not give any degradation. In the mixed stick slip and gross slip regimes, the degradation mechanism was by cracking (fretting fatigue) and wear, respectively. The presence of NaCl solution favoured slip at the contact interface, resulting in a larger gross slip regime than in air. Nevertheless, mixed stick slip was the dominant fretting regime for both test environments. Therefore, fatigue was more relevant than wear for the degradation of the strand material. The tangential force amplitude in air was about three times of that obtained in chloride solution for contacts characterized by mixed stick slip, suggesting longer fretting fatigue life in wet and marine environments than in dry air. Similarly, the cylindrical strand surfaces subjected to mixed stick slip contact condition showed substantially higher tangential force than that of a flat polished specimen surface fretted against alumina ball in air, indicating that the former geometric contacts are less resistant to fretting fatigue degradations. The electrochemical factors acting simultaneously with fretting were not a primary factor in causing degradation. The study leads to conclude that a chloride solution environment is beneficial in reducing the fretting fatigue and wear degradations of the material by acting as a lubricant.

5.1 Introduction

Fretting is the major cause of failure of aluminium strands used in ACSR conductors commonly used in marine environments, as documented in Chapter 3. As defined in §2.8.1, fretting is a phenomenon resulting from extremely small oscillatory relative displacement arising at the interface of contacting surfaces expected to be motionless. In ACSR conductors, fretting occurs at strand-to-strand and strand-to-clamp contact interfaces (§2.5), exposed to wind induced vibrations, frequently Aeolian vibration [1, 2]. In addition to fretting, corrosion is involved in the degradation in marine environments, hence causing fretting corrosion degradation as documented in Chapter 3.

As described in §2.9, Vingsbo and Söderberg [3] were the pioneers to present fretting damage in terms of a fretting map defined by the relationship between the tangential force and displacement amplitude arising at a contact interface for low carbon steel in a laboratory air. The fretting map constitutes the stick, mixed stick slip, and gross slip regimes. It was also explained that the stick regime occurs when the displacement imposed is within the elastic deformation limit of the surface and subsurface regions of the specimen, and hence low degradation by fretting is expected in this regime. In the mixed stick slip regime, the imposed displacement is accommodated by elasto-plastic deformation of the contact surface, and the regime is characterised by slight wear associated with cracking at the edge of the contact. In the gross slip regime, however, the imposed displacement is accommodated by plastic deformation and relative slip. It is characterised by wear degradation and limited crack formation due to removal of initiated cracks by wear.

Experimental observation [3] demonstrated that the characteristic shape of the transient tangential force *vs.* displacement in a fretting cycle, termed as "fretting curve or fretting loop", enables to identify the prevailing fretting regime in a contact. Accordingly, a closed loop (straight line), hysteresis loop (elliptical curve), and deformed hysteresis loop (parallelogram shaped curve) represent the stick, mixed stick slip and gross slip regimes, respectively. The plateau of the tangential force in the deformed hysteresis loop under the gross slip regime is attributed to complete relative slip between the counter parts, hence friction and wear are expected.

Essentially, the centre of the contacting surfaces is in stick contact under both stick and mixed stick slip regimes. However, the edges of the contacting surfaces undergo relative slip of different magnitudes, which is generally negligible for the stick regime but considerable in a mixed stick slip regime. Because of this, the two regimes are grouped together and named partial slip regime in later studies and the gross slip regime is often called slip regime, (see Fig. 2.16a). Accordingly, the transition from partial slip regime to slip regime is not direct. However, a regime called mixed regime or an intermediate regime prevails between these two, in which fretting crack initiates [4, 5].

The shape of a fretting curve may vary during a fretting test due to change on the contacting surfaces. In view of this, the fretting curve for the mixed regime changes in shape from one form to another while

fretting is going on, *i.e.*, fretting regime transformation occurs at the same displacement amplitude and normal force [4]. Such changes are determined by plotting and analysing the evolution of the fretting curves over the entire test duration instead of a single fretting cycle analysis. Zhou *et al.*, [6] developed a 3D presentation called a friction log (tangential force *vs.* displacement *vs.* number of cycle) to describe a fretting condition in a given number of cycles. Accordingly, the plot of the tangential force *vs.* displacement for a given cycle can become closed, elliptic or parallelepiped in shape (cf. Fig. 2.17b). Such principle, based on kinematics of the contact interface, has been used to predict the type of fretting damage in a contact and establish a fretting map, called running condition fretting map (cf. Fig. 2.17a). Similarly, a fretting map, called material response fretting map (cf. Fig. 2.17b), was established based on the degradation mode of the contact surface, *i.e.*, no cracking, cracking and wear (cf. Fig. 2.17d) [5, 6] as reviewed in §2.9.

Although the displacement at the contact interface is a critical parameter for fretting degradation in a conductor, it has been virtually impossible to determine in a clamped conductor [7]. Moreover, limited information is available on the influence of corrosive environment on the fretting regimes and fretting corrosion of the aluminium strands. Therefore, basic fretting investigation of the strand material in air and chloride solution is necessary to understand degradation behaviour under such conditions. Hence, this chapter is intended to establish a fretting map (identify the fretting regimes) in air and 3.5% NaCl solution and characterize the degradation in each regime by analyses of the friction logs, the electrochemical response, and surface degradation morphology. Moreover, the study compares tangential force transients generated from fretting in air and NaCl solution. In addition, the effect of contact surface geometry on fretting fatigue degradation will be evaluated by comparing the tangential force from fretting of ball - on - flat and identical cylindrical strand surfaces in air.

5.2 Experimental details

5.2.1 Materials

Mechanical properties and chemical composition of the material (aluminium alloy AA1350) were presented in §3.2.1. The sample preparation procedure was described in §4.2.1. The counter material used for fretting tests was a spherical alumina ball (10 mm diameter, SWIP AG Brügg, G10 AFBMA finish).

5.2.2 Experimental apparatus and test conditions

Fretting experiments were conducted in air at a temperature of 24 ± 1 °C and relative humidity of $47 \pm 7\%$. Moreover, fretting experiments were performed in a 3.5% NaCl solution using the apparatus described earlier [8]. The contact configuration consisted a stationary alumina ball specified above, rubbing against an underlying flat AA1350 alloy surface. For the test in chloride solution, all potentials were measured with respect to a standard saturated calomel electrode (SCE).

Tangential force (F_t) and normal forces (F) were measured by a Kistler 9251A pre-loaded quartz force sensor whereas the vertical pin advancement was measured by a Keyence LC2420 distance metering with a resolution of 0.1 μ m. The data acquisition system of the test rig captured data 1500 times per second. These data were averaged for each relevant experimental parameter and recorded simultaneously. The system further calculates the coefficient of friction at the midway of the displacement, D, as a ratio of the tangential force to the normal force.

During the experiments, the normal force was applied through a loading spring, and a friction sensor monitored the resulting tangential force. The apparatus measured the relative displacement at the contact interface by means of a sensor placed along the displacement direction. The recorded data included the normal force, tangential force, displacement at the specimen surface, fretting time, potential, and current. The evolution of tangential force versus displacement (F_t - D) was simultaneously calculated from the data by a computer.

5.2.3 Test procedure

Three types of experiments were performed. The first set of experiments were intended to identify the fretting regimes and hence carried out at normal forces (F) in the range 5 - 60 N (corresponding to a maximum Hertzian contact pressure range of 535 - 1224 MPa) and displacement amplitudes (D), which is the maximum applied displacement, from 2 - 100 μ m. The frequency (f) of oscillation was 1 Hz in all experiments. Each experiment was conducted at constant displacement amplitude (both in air and NaCl solution) lasted for about 3600 s. The average and transient OCP in a fretting cycle were recorded and analysed. Selected fretting curves and surface micrographs were analysed to present the degradations occurring in the fretting regimes.

The second category of experiments were performed in chloride solution by varying the displacement amplitude at the same normal force. The evolution of the average OCP and transient tangential force with the changing the displacement amplitude were presented and analysed. This experiment was intended to reflect realistic situations where the displacement amplitude at the contact interfaces of the strands in a conductor varies with climatic condition in marine environments.

The third class of experiments were potentiostatic fretting in a 3.5% chloride solution at -1.10 V_{SCE} . The normal force and the displacement amplitude were maintained at 5 N and 100 μ m, respectively to keep the contact in the gross slip fretting regime. The mean current and the transient current in the fretting cycles was recorded and analysed.

Finally, fretting of two alike cylindrical strands was conducted in in air at a normal force - displacement amplitude combinations of 50 N - 40 μ m and 40 N - 100 μ m. The angle between the axes of the strands was aligned to about 30 - 40 degrees. The friction logs and the fretted surface morphologies were studied. The samples tested were in the as-received condition, i.e. no metallographic or chemical preparation of the surface was adopted prior to the tests. The aim of this experiment was to find the relationship between the fretting of ball-on-flat geometries to the condition of the actual cylindrical contacts.

5.2.4 Surface characterization

The same procedures and methods (instruments) described in §3.2.3 were used for the surface morphology characterizations.

5.3 Results

5.3.1 Friction logs

Fig. 5.1 shows three-dimensional (3D) friction logs describing fretting processes as a function of tangential force, displacement and number of cycles. The evolution of the shape of the loops with number cycles indicates tribological changes in the contact interface. The figures show that the amplitude of the tangential force in air for the specified normal force and displacement amplitude was about three times

that in NaCl solution. The tangential force was directly proportional to the dynamic displacement during each cycle (back and forth) in the case of fretting at low displacement amplitude (4 µm). Consequently, the fretting loops were all straight lines, regardless of number of cycles, making up a two-dimensional (planar) friction log (Fig. 5.1a), for both fretting in air and NaCl solution. This characterizes a stick regime as explained in §2.8. When the displacement amplitude was increased to 35 µm, the fretting loops became elliptical for each cycle, hence the friction log became like ellipsoidal prism (Fig. 5.1b) for both test conditions. At 60 μ m, the profile of the friction log (Fig. 5.1c) started as a deformed parallelogram for the test in air and became elliptical hysteresis shortly with increasing number of cycles. This indicates that the prevailing fretting regime at the contact interface was mixed stick slip. In contrast, the fretting loops obtained in NaCl solution began as a parallelogram and transformed gradually into elliptical hysteresis as the number of cycles increased (Fig. 5.1c). It indicates that the contact condition started with gross slip and gradually tended to the mixed stick slip regime. When the displacement amplitude was increased to $80 \,\mu\text{m}$, the curves for air behaved similar to those at $60 \,\mu\text{m}$ except for larger hysteresis for the latter (cf. Fig. 5.1d). However, the friction log shape started as a slightly deformed parallelepiped for the test in NaCl solution, suggesting that the contact was in the gross slip regime, tending toward the mixed stick slip regime.

As can be observed from the Fig. 5.1c and d, the shapes of the friction logs were distorted during the first few cycles, probably related to the difference in the initial condition of the contact surfaces. They attained a steady shape with increasing number of cycles. To study these changes more closely, the friction log data were simplified to two-dimensional tangential force *vs*. displacement plots corresponding to a specified time interval (cycles) during the fretting experiment, as shown in Fig. 5.2 - 5.7. Such 2D cross plots allow a closer look at the state of the contact at regular intervals, e.g., presentation of several fretting curves obtained at a given normal force but different displacements in the same graph. Such plots simplify comparison of the fretting loops obtained in air and NaCl solution. Moreover, it enables identifying the fretting regimes and transitions from one to another, as will be shown in the following sections.



Fig. 5.1 Friction logs obtained from fretting in air and NaCl solution at the normal forcedisplacements of (a) $20 \text{ N} - 4 \mu \text{m}$, (b) $20 \text{ N} - 35 \mu \text{m}$, (c) $20 \text{ N} - 60 \mu \text{m}$, (d) 20 N - 80 at a frequency of 1 Hz for 3600 s. *Note*: The number of cycles axes factor is 60x.

5.3.2 Fretting regimes in air

The three typical fretting curves, namely linear, elliptical and parallelogrammatic shaped (Figs. 5.2 and 3) were obtained from the fretting experiment depending on the magnitude of the normal force and displacement (contact conditions) at the interface [9]. These curves were obtained by plotting transient tangential force *vs.* displacement data in a 2D plane as explained above. Each curve in the figures represents a fretting cycle (back and forth displacement) and contains two, usually symmetric halves, which correspond to the forward and backward displacements. As stated in the introduction, the profile
of the curves provides information about the kinematic state (state of movement) at the contact interface and expected surface degradation.

At a given normal force, the shape of the fretting curves varied with increasing displacement amplitudes as shown in Fig. 5.2a. It is a closed curve (straight line) at 3 μ m, elliptical hysteresis at 8 μ m and parallelogrammatic (or deformed hysteresis) at displacement amplitudes greater than 50 μ m. This indicates that the state of the contact at the interface of the alumina ball and the specimen surface was in stick, mixed stick slip and gross slip conditions respectively, as already known [9] and reviewed in Section §2.9.

The fretting curves were obtained at 5 N and different displacement amplitudes in air (Fig. 5.2a). For low displacement amplitudes, 3 and 4 μ m, the fretting curves were linear. This indicates that the contact interface was in stick contact, *i.e.*, the displacement amplitude was with the elastic deformation limit of the surface and subsurface region of the specimen. When the displacement amplitude was increased to 8 μ m, the fretting curve became elliptical, indicating that the mixed stick slip regime began to appear. The hysteresis of the curves increased as the displacement amplitude was increased to 15, 20 and 50 μ m. This shows that the fretting regime changed from stick to mixed stick slip contact condition when the displacement was increased from 4 to 8 μ m. In this regime, the displacement amplitude exceeds the elastic deformation limit of the surface and subsurface region of the specimen and results in plastic deformation of the surface (see the nonlinear parts of the curves). Therefore, elasto-plastic deformation is the displacement accommodation mechanism [9].





Fig. 5.2 Tangential force transients for different imposed displacements (specified in the figure), measured 30 min after the start of fretting in air at a normal force of (a) 5 N, (b) 15 N, (c) 20 N and a frequency of 1 Hz.

At 60 μ m displacement amplitude and higher, the tangential force initially increased linearly and then flattened as the displacement amplitude was increased and reached a nearly constant value at 100 μ m. This indicates that transition from mixed stick slip to gross slip regime occurred as the displacement amplitude was increased from 20 to 50 μ m. The fretting curves were deformed parallelograms at displacement amplitudes of 50 and 80 μ m, indicating that gross slip contact condition prevailed. In the gross slip contact condition, the tangential force increased proportionally with the displacement (the linear part of the curves) until the elasticity limit of the specimen and testing system is reached. The surface then deformed plastically (nonlinear part of the curve), and this led to a complete relative slip of the contact surfaces as evidenced by the extended plateau of the tangential force at 100 μ m. The plateau of the tangential force indicates yielding of the specimen surface and start of gross slip. It becomes extended with increasing displacement amplitude. Overall, the results demonstrate that the fretting regime was changed from stick to mixed stick slip and then gross slip as the displacement amplitude was increased at the normal force of 5 N.

Figs. 5.2b and c show the fretting curves obtained at a normal force of 15 and 20 N, respectively, for the displacement amplitudes shown in the figure. At 15 N (Fig. 5.2b), the mixed stick slip regime began to appear at a displacement of 12 μ m. The slight hysteresis indicates that the transition from stick to mixed stick slip regime occurred at a displacement slightly less than 12 μ m. The area of the hysteresis increased as the displacement amplitude was increased to 80 μ m, but the tangential force did not form a plateau, *i.e.*, it was still increasing with the displacement amplitude. This indicates that the prevailing regime in the contact was mixed stick slip. At 100 μ m, the plateau of the tangential force was formed, indicating that gross slip regime prevailed in the contact. This shows that transition from mixed stick slip to gross slip regime occurred as the displacement amplitude was increased from 80 to 100 μ m. As shown in Fig. 5.2c, only a stick and mixed stick slip regime contact conditions prevailed at all displacement under a normal force of 20 N.

Fig. 5.3 shows evolution of fretting curves at a displacement amplitude of 100 µm for normal force in the range 5 to 50 N in air, taken 30 min after the start of fretting, except for the one at 15 N, which was plotted from data recorded 10 min after the start of fretting (earlier than the others). However, since the steady-state condition was reached within the first 5 min for the fretting tests in air, as can be observed from the friction logs in Fig. 5.1, the plot after 5 min does not alter the conclusion for the gross slip regime. At 5, 10 and 15 N, the fretting curves were shaped as deformed parallelograms, indicating that the contact condition was in the gross slip regime. At 20 and 25 N, the amplitude of the transient tangential forces increased significantly, and the plateau signifying the gross slip regime disappeared. This demonstrates that the contact condition changed from gross slip to mixed stick slip regime and shows that the change in the fretting regime was accompanied by substantial increase in the tangential force amplitude. The hysteresis of the elliptical fretting curves became increasingly narrow with increasing normal force, as can be observed from the trend of that curves at 25 to 50 N. This indicates that the prevailing regime tended toward the stick regime with increasing normal force at all displacements. Moreover, it is clear from the figure that the gross slip regime occurred only for loads up to 15 N. Transition to mixed stick slip regime appeared when the normal force was increased from 15 to 20 N.



Fig. 5.3 Tangential force transients for different normal forces and constant displacement (100 μ m) recorded 30 min after the onset of fretting in air. The curve recorded after a relatively shorter duration is marked by *. The shape of the curves changed from parallelogram to elliptical hysteresis with increasing normal force, indicating change in the fretting regime from gross to partial slip between 15 and 20 N.

5.3.3 Fretting regimes in NaCl solution

Fig. 5.4 shows the fretting curves obtained in NaCl solution at a normal force of 15 N and displacement amplitudes specified in the figure. It shows linear and slightly open fretting curves for the displacement amplitudes 6 and 12 μ m, respectively. This indicates that transition from the stick to mixed stick slip regime occurs at a displacement amplitude slightly higher than the former. At displacement amplitudes of 40 μ m and higher, the curves became parallelogram shaped indicating occurrence of gross slip in the contact surface. The shape of the curves changed with time as illustrated for the 40 μ m curve, for which the pink and red curves correspond to the fretting cycles taken 3 and 25 min after the onset of fretting,

respectively. The change with time (fretting cycle) of the shape of the loops may indicate that the data point may be close to the border with gross slip regime [10]. In fact, transient behaviour was observed for all fretting curves obtained from tests in gross slip conditions in NaCl solution. Therefore, the data reported for normal force *vs* displacement at $15 \text{ N} - 40 \mu \text{m}$ combination ended up in the gross slip regime although the state of the contact, i.e., the shape of the fretting loops continued to change with time, as will be discussed further below.



Fig. 5.4 Tangential force versus displacement transients obtained at the imposed displacements (specified in the figure), measured 20 min after the start of fretting in NaCl solution at normal force of 15 N. The curves for 40 μ m displacement are different due to difference in the start times of recording after the onset of fretting (red for 3 min black for 25 min).

5.3.4 Influence of aqueous solution on fretting

From the analysis of fretting regimes in air and chloride solution in the preceding subsections, it appears that the test environment influences the displacement at which transition from mixed stick slip to gross slip regime occurs at a given normal force. This is clarified by comparing the evolution of fretting curves with the number of cycles obtained from fretting experiments at 20 N (the same normal force) and displacement amplitudes of 35, 60, 80 and 100 μ m. In each case, fretting loops taken 2, 5, 10, 20, 30, 40, 50, 55, and 59 min after the onset of fretting were analysed with an intention to explore the variation of the contact conditions during the tests. The friction logs for the tests at 35, 60 and 80 μ m displacements were presented in Fig. 5.1.

Fig. 5.5 shows the fretting curves obtained from fretting tests at the time intervals indicated in the preceding paragraph at a normal force of 20 N and displacement of 35 μ m in air and chloride solution. The curve obtained in air was elliptical 2 min after the onset of fretting, whereas the one obtained in NaCl solution was a deformed parallelogram. However, the fretting process reached a steady-state condition within 5 min and all the curves became elliptical throughout the test duration in both cases. This shows that the prevailing regime was mixed stick slip in both cases. Though both belong to the same fretting regime, the tangential force amplitude generated from the test in air was about three times that in NaCl solution. It is known that a fretting fatigue life decreases with increasing tangential force [11]. The hystereses are wider in NaCl solution than in air. Therefore, the difference in the test environment influenced mainly the tangential force arising at the contact interface, thereby influencing the fretting fatigue life. The fact that the curves were elliptical in the steady-state suggests that the transition to gross slip regime occurs at a larger displacement amplitude for the same normal force, as shown in the next paragraphs.



Fig. 5.5 Fretting curves obtained in air and NaCl solution at a normal force of 20 N, displacement of 35 μ m and frequency of 1 Hz, measured after the starting periods indicated in the figure.

Figs. 5.6a (in air) and Fig. 5.6b (in NaCl solution) show the evolution of the fretting curves with time at $60 \mu m$ displacement amplitude and normal force of 20 N. In air, the curve was a deformed parallelogram (after 2 min, curve A). It evolved to a curve with a small plateau after 5 min (curve B) and then achieved a mixed stick slip in the steady-state condition (overlapped curves) with time (indicated by arrow) within about 5 min. This indicates that the contacting surfaces were initially in a gross slip contact (during the first few minutes), and then transformed into mixed stick slip in the steady-state condition.





Fig. 5.6 Evolution of fretting curves with time obtained at a normal force of 20 N and 60 μ m displacement in (a) air and (b) NaCl solution. (c) Fretting curves obtained from tests in NaCl solution at increased normal force (25 N) and 60 μ m displacement. (d) Fretting curves obtained at increased displacement (80 μ m) and 20 N normal force. The plotted data were taken after the time intervals indicated in the figures (see the arrow) on flat and polished specimen fretted against 10 mm diameter alumina ball.

In NaCl solution, the fretting curves were all parallelogram shaped from the outset and changed gradually with time throughout the test duration (Fig. 5.6b), indicating that the contact condition was in a gross slip. On the subsequent fretting cycles, the transient tangential force amplitude increased by a factor of about 3 (arrows B) and the plateau (horizontal part of the curves decreased (arrow A), and the curves became increasingly elliptical. This indicates the contact condition was slowly transforming to a mixed stick slip regime (Fig. 5.6b). The shortening of the plateaus with time indicates that the relative slip at the interface decreased continuously. The change in the shape of the curves suggests that the contact was changing from gross slip to mixed stick slip with increased number of fretting cycles, which is the critical regime for crack nucleation and initiation [4]. Therefore, the fretting regime at a normal force of 20 N and displacement amplitude of 60 μ m gradually tended toward a mixed stick slip condition. Thus, the normal force - displacement combination results in a contact condition close to the boundary of the mixed stick slip and gross slip regime.

The fretting curves obtained from test in NaCl solution at 25 N - 60 μ m is shown in Fig. 5.6c, *i.e.*, the normal force was increased from 20 to 25 N at the same displacement amplitude as above (60 μ m). Prior to the test, the specimen surface was fretted at 15 μ m and same normal force for 34 min. Since the prior fretting was at a displacement amplitude causing a mixed stick slip contact condition (cf. Fig. 5.6c), its influence on the subsequent fretting is considered as marginal due to the fact that such fretting condition may not considerably alter the condition of the contact surface. The first three curves obtained over a period of 10 min were deformed parallelograms, indicating that gross slip regime prevailed as can be observed from the peak (arrow B) and plateau of the tangential force. In the later fretting cycles, the extended plateau of the tangential force decreased with time, as indicated by arrow A. Nevertheless, the transient tangential force decreased past the peak at B and then increased with time as the curves with increasing noise (arrow C). The noisy segment of the curves decreased with time as the curves tended to become elliptical. Compared to fretting at 20 N - 60 μ m (Fig. 5.6b), the rate of transition from gross to mixed stick slip regime was much faster as can be seen from the spacing of the curves. Therefore, transition from mixed stick slip to gross slip regime at 60 μ m occurred at a normal force between 20 and 25 N.

Fig. 5.6d shows fretting curves obtained at the time intervals shown in the figure after the onset of fretting at 20 N - 80 μ m in NaCl solution, *i.e.*, increased displacement amplitude from 60 to 80 μ m compared to

the ones described in the preceding two paragraphs. As shown in the figure, the fretting curve was a deformed parallelogram at the beginning (see the curve taken after 2 min. The arrow direction indicates increasing number of cycles) and it became a typical curve of a gross slip regime as of 5 min. However, the last two curves (after 50 and 60 min) were slightly distorted at the plateau of the tangential force indicating the gross slip regime was still there. Therefore, all curves indicate that the fretting contact was in the gross slip condition. This clarifies that the normal force and displacement amplitude combination (20 N - 80 μ m) causes a gross slip contact condition. Therefore, the transition from mixed stick slip to gross slip regime occurs in the range of 60 - 80 μ m displacement amplitude, but closer to 60 μ m as argued above.

Fig. 5.7 shows fretting curves obtained at a normal force of 20 N and displacement of 100 μ m in air (Fig. 5.7a) and NaCl solution (Fig. 5.7b). In air, the curve was deformed hysteresis loop until 2 min (curve A) and then became a deformed parallelogram with a small plateau (arrow B) during the next 3 min (5 min after the onset of fretting), indicating that gross slip regime was prevailing in the contact during this period. In the subsequent fretting cycles (as of 10 min), the curves showed no plateau and the fretting process reached a steady state where the amplitude of the tangential force stayed at about \pm 34 N. Clearly, the prevailing fretting regime in the contact was changed from gross slip to mixed stick slip in the steady state.

In contrast to the fretting in air, the fretting curves obtained in NaCl solution (shown in Fig. 5.7b) were all shaped as deformed parallelogram. This shows that gross slip occurred at the contact interface for all the fretting cycles. Similar to the previous cases, the hysteresis of the fretting curves decreased slowly and the plateau of the tangential force became smaller with the number of cycles (as shown in the figure by the vertical arrow), while the tangential force amplitude increased. The results clarify that the 20 N - $100 \,\mu$ m combination causes mixed stick slip and gross slip regime under fretting in air and NaCl solution, respectively.



Fig. 5.7 Evolution of fretting curves with time, obtained from fretting of flat AA1350 alloy against 10 diameter alumina ball at frequency of 1 Hz, normal force of 20 N and displacement of 100 μ m. Test environments were air (a) and 3.5% NaCl solution (b).

5.3.5 Fretting map

Fig. 5.8 shows a fretting map established from the fretting regimes identified in the preceding sections for the tests in air and NaCl solution. It shows a stick, mixed stick slip and gross slip regimes separated by approximate boundary lines, labelled A, B and C. The boundary between the stick and mixed stick regimes is the same for the tests in air and NaCl solution as no considerable difference was found from the tests. However, the boundary between the mixed stick slip and gross slip regimes was different for the tests in NaCl solution (line B) from that in air (line C).



Fig. 5.8 Fretting map of flat and polished AA 1350 alloy fretted against 10 mm diameter alumina ball in air and 3.5% NaCl solution. The stick, mixed stick slip, and gross slip regimes are indicated. Line A represents the boundary between stick and mixed stick slip regimes in air and NaCl solution, whereas B and C approximate the boundaries between the mixed stick slip regime and gross slip regime in NaCl solution and air, respectively.

For the tests in air, the transition from mixed stick slip to gross slip regime was at the displacement ranges of 20-50 and 80-100 μ m at the normal forces of 5 and 15 N (Fig. 5.2a and b), respectively. The transition from gross slip regime to mixed stick slip regime was in the normal force range of 15 - 20 N (Fig. 5.3) at 100 μ m displacement amplitude. However, the gross slip regime did not prevail at all displacements for a normal force of 20 N and higher (cf. Figs. 5.1d, 5.3 and 5.7a).

For the tests in air and NaCl solution, the transition from stick to mixed stick slip regime occurred at small displacement amplitudes for all normal forces tested. The gross slip regime is confined to a small triangular region at the bottom right corner of the map i.e., at normal forces up to 20 N and displacement amplitudes from about 50 - 100 μ m for fretting in air. The mixed stick slip regime constitutes the major portion of the fretting map for both test environments. Moreover, only stick and mixed stick slip regimes prevailed at normal forces greater than 20 N for both tests in air and NaCl solution. It is clear from the fretting map that the area for the gross slip regime in chloride solution is larger than that in air. In other words, there are normal force and displacement amplitude combinations causing a mixed stick slip regime in air result in a gross slip contact condition when applied in NaCl solution.

In NaCl solution, the transition from mixed stick slip to gross slip regime was observed at 15 and 20 N and corresponding the displacement amplitudes of about 40 (Fig. 5.4) and 35 μ m (Figs. 5.1c and 5.6b), respectively. Likewise, the stick regime prevailed at displacement amplitude ranges of 6 - 12 and slightly higher than 4 μ m at the normal forces of 15 and 20 N, respectively (Figs. 5.4 and 5.1a). Moreover, it was evident from Figs. 5.1b, 5.5, 5.6d, and 5.7b that mixed stick slip regime prevails at 20 N normal force and 100 μ m displacement amplitude. However, at both 80 and 100 μ m displacement amplitudes, the contact condition belongs to the gross slip regime at 20 N (Figs. 5.1d and 5.6d and 5.7b). The transition in the fretting regime occurred when the displacement amplitude was increased from 35 to 60 μ m, close to the latter, as indicated in Fig. 5.8.

5.3.6 Effect of displacement amplitude in NaCl solution

Fig. 5.9a shows a friction log obtained from fretting at a normal force of 5 N by successively increasing the displacement amplitudes from 4, to 6, 10, 20, 15 and 100 μ m on the same sample in NaCl solution. That means, the sample surface was fretted for a pre-determined number of cycles at each displacement

amplitude. The sample was held in NaCl solution and the potential was measured for about 1500 s before the start of fretting. Fig. 5.9b shows selected fretting curves from the friction log in Fig. 5.9a at each displacement amplitude after a certain number of cycles. The shape of the fretting curves varied with the displacement amplitudes (indicated in the figure). At low displacement amplitude (4 µm), the fretting curves were all closed as expected, indicating that the contact was in stick regime. However, the fretting curves became elliptical hysteresis with increased displacement amplitude, resulting in ellipsoidal friction log signifying the occurrence of mixed stick slip contact condition. At 20 µm displacement amplitude (indicated by arrow B in Fig. 5.9a), the fretting curves started as a deformed parallelogram and became elliptical hysteresis gradually as indicated by arrow B in Fig. 5.9b, in a manner similar to the one presented earlier (cf. Fig. 5.6b), indicating a possible regime transition from gross slip to mixed stick slip. At 100 µm displacement amplitude, the fretting curves were all parallelogram shaped, indicating that the contact interface was in a gross slip fretting condition. The friction log evolution indicates that varying the displacement amplitude at the contact interface caused change in the fretting regime from stick to mixed stick slip and further to gross slip regime with increasing displacement amplitude. Nevertheless, the curves obtained on the surfaces subjected to prior fretting may not be the same as those on a fresh surface and so do the displacement amplitudes at which regime transition occurs.





Fig. 5.9 a) Friction log obtained fretting 10 mm diameter alumina ball against AA1350 alloy in a 3.5% NaCl solution at a normal force of 5 N and successively changing displacements specified in the figure at a frequency of 1 Hz. (b) Selected fretting loops from fretting at each of the displacements in (a). Several fretting loops plotted at 20 μ m imposed hysteresis show reduction in hystereses reduced with increasing number of cycles along arrow B. *Note*: The number of cycles axis factor is 60x in graph a).

5.3.7 Electrochemical response

Effect of fretting at variable displacement on open circuit potential

Fig. 5.10a shows the successively varied displacement amplitudes used in the fretting experiment at normal forces of 5 N (for which the friction log was shown in Fig. 5.9a) and 15 N. For the test at 15 N, the successive displacement amplitudes applied were 80, 60, 40, 50, and 6 μ m as shown in the figure. And the corresponding number of fretting cycles were about 1200, 2400, 1200, 5400, 1200 cycles, respectively, *i.e.*, the specimen was exposed to 11 400 cycles at the end of the test. The cycles are

converted to time, and the time axis shown in the figure is presented in minutes. Fretting (at 15 N) started with a displacement amplitude (80 μ m, labelled A) which causes gross slip contact condition (cf. Figs. 5.4a and 5.8), succeeded by fretting at smaller displacement amplitudes and ended after fretting at the lowest displacement amplitude 6 μ m, labelled B. In contrast, fretting at 5 N started at a low displacement amplitude (labelled C), which gives a stick regime (cf. Fig. 5.8) and successively increased to higher displacement amplitudes and terminated after fretting at 100 μ m (labelled D) which led to a gross slip contact condition (cf. Fig. 5.9a and b). The initial displacement amplitudes (labelled A and C) used in the tests were performed on a freshly polished surface. Therefore, the friction logs recorded for these displacement amplitudes can be mapped into the fretting regimes (in NaCl) identified in Fig. 5.8. Fretting curves obtained at the subsequent displacement amplitudes, however, cannot be used in the fretting map established due to the change in the surface condition by the prior fretting step.

The open circuit potential measured was nearly constant during the fretting experiments described above and in Fig. 5.10a, including significant changes in the displacement amplitude and normal force. It was nearly equal to the value without fretting at the beginning and end of the experiments at about -0.78 V_{SCE}, as seen in Fig. 5.10b, and termination of fretting (Fig. 5.10b). Fig. 5.11 gives a closer look at three cases of fretting cycles selected from the above data, in which the tangential force and displacement corresponding to stick (Fig. 5.11a), mixed stick slip (Fig. 5.11b) and gross slip (Fig. 5.11c) fretting regimes.

The stick regime corresponds to a normal force of 15 N and displacement amplitude of 6 μ m (labelled D, in Fig. 5.11a). The OCP was constant within a few mV at -0.778 throughout the fretting cycle. The mixed stick slip condition depicted in Fig. 5.11b was obtained at 40 μ m displacement amplitude and a normal force of 15 N. The OCP was still constant at about -0.76 V_{SCE}, with small minima of about 10 - 20 mV at the ends of the stroke (A and C) where the displacement was reversed. The decrease in the OCP corresponded to the deviation of the tangential force from direct proportionality with respect to time (point B), i.e., when the plastic deformation of the surface began. In other words, the potential decreased when the specimen surface started to deform plastically with increasing displacement amplitude. In this regime, exposure of fresh Al surface to the NaCl solution is expected to occur only in the slip part, which corresponds to the periphery of the contact surface as the central area remains in stick contact. The gross slip regime in Fig. 5.11c corresponds to 5 N normal force and 100 μ m displacement amplitude. The OCP

transients became slightly larger, not exceeding 10 mV above the average -0.78 V_{SCE}. OCP decreased to a minimum, as described in Fig. 5.11b, when the tangential force reached its maximum limit, and gross slip commenced at points B and D. At these points, the surface yielded and particle detachment is expected due to the relative slip. Therefore, breaking of the oxide film and exposure of bare metal is expected, leading to reduction in the potential. Moreover, the potential decrease occurred as the ends of the strokes (points A and C) were approached and direction was reversed. After decreasing at the onset of gross slip and direction reversal points, the potential increased slowly with the displacement.



Fig. 5.10 (a) Displacements used in fretting of alumina ball with AA1350 alloy specimen in 3.5% NaCl at 5 and 15 N. Fretting was started at displacement labelled A and ended at the displacements labelled B. (b) Mean OCP measured before fretting was started, during fretting and after termination of fretting.







Fig. 5.11 Evolution of transient OCP, tangential force and displacement in a single fretting cycle in 3.5% NaCl solution at 15 N normal force and 1 Hz frequency at displacements of (a) $6 \mu m$ (the stick regime), (b) 40 μm (mixed stick slip regime), and (c) 5 N - 100 μm (gross slip regime).

Fretting corrosion at constant displacement amplitude

Fig. 5.12 shows the transient response of OCP, tangential force and interfacial displacement in a single fretting cycle (from A to C and back to A) at 15 N normal force, corresponding to the gross slip regime (cf. Figs. 5.4a, 5.8 and 5.10a). From point A to B in Fig. 5.12a, the tangential force increased proportionally with the displacement and reached its peak at B, which was the elastic limit. Beyond this limit (from B to C), the surface underwent plastic deformation and slip to accommodate further increase in the displacement, while the tangential force flattened out as a result of yielding and plastic deformation [9]. As shown in the figure, the applied displacement amplitude ($80 \mu m$, A to C or C to A on the horizontal axis) was accommodated partly by elastic deformation (A to B and C to D) and then plastic deformation and slip (B to C and D to A). The points designated as B and D, on each half cycle, correspond to the onset of friction and slip; hence, the surface yields and plastic accommodation occurs [12].

As can be seen in Fig. 5.12b, the OCP was nearly constant (about -0.76 V_{SCE}) during the elastic accommodation of the displacement between the points A and B. As the yield point B was approached, the OCP surged to a negative peak of -0.79 V_{SCE} at the onset of slipping. As the slipping continued, the transient potential increased gradually, reached a peak and then decreased slightly to another minimum at the end of the stroke (point C) where the displacement direction was reversed. In the reverse direction, the OCP trend repeated in the reverse order on the graph. Although the changes observed in the OCP during a displacement cycle were small, they could be regarded as signatures of different events occurring during the cycle in terms of both type of regime and intensity of wear. The variation in OCP in the fretting cycle is related mainly to the transition between the elastic and plastic response of the surface to displacement. Because the OCP variation during the fretting cycle was small and similar in each cycle, the average OCP evolution during the test was not affected by fretting to a considerable extent. In contrast to this result, however, the presence of the NaCl solution, in relation to the laboratory air, had a significant effect on the fretting processes in influencing the transition from mixed stick slip to gross slip regime as presented in §5.3.4.

In the mixed stick slip regime (cf. Fig. 5.8, at 20 N normal force and 35 μ m displacement amplitude) as well, the evolution of the average OCP was affected by neither the start nor termination of fretting. Typically, the potential remained at about - 0.76 V_{SCE} throughout the test duration. Similar trend was

observed for fretting at stick contact condition as can be observed from fretting at the first displacement amplitude applied at 5 N in §5.3.7 (cf. Fig. 5.10a and b).



Fig. 5.12 (a) Evolution of transient tangential force and displacement, and (b) transient open circuit potential in fretting cycle at a gross slip contact condition obtained from fretting a flat AA1350 alloy against 10 mm diameter alumina ball in a 3.5% NaCl solution at normal force of 15 N and 80 μ m displacement at a frequency of 1 Hz. The data plotted was taken 20 min after the onset of fretting.

Fig. 5.13 shows the transient OCP, tangential force and displacement evolution obtained from a single fretting cycle at 20 N - 35 μ m (mixed stick slip regime (cf. Figs. 5.1b and 5.5) in chloride solution. As shown in the figure, the transient OCP decreased slightly when the transient tangential force became nonlinear and the ends of the stroke were reached (A and B).



Fig. 5.13 Evolution of transient OCP, tangential force and displacement in a single cycle of fretting taken 25 min after the start of fretting at normal force and displacement combination of 20 N - 35 μ m (mixed stick slip regime) in a 3.5% NaCl solution.

As presented earlier, the nonlinearity of the tangential force is caused by displacement accommodation by plastic deformation. As already known [12], the slight potential response was clearly related to the plastic accommodation part of the displacement and direction reversal. The potential trend in the fretting cycle was similar to the one observed for the specimen subjected to prior fretting at different displacement amplitudes (see Fig. 5.11b). In both cases, the potential trend is related to the displacement accommodation mechanism of the specimen surface.

In a stick contact condition where penetration of the electrolyte to the contact the interface was limited and there is no considerable relative slip at the interface, no change in transient potential measured during a fretting cycle is expected, as already observed on prior fretted surface in the same regime (cf. Fig. 5.11a). Therefore, there was no need to include the plot of transient potential evolution in this regime.

Fretting under potentiostatic condition.

This experiment was performed at a at a normal force of 5 N and 100 μ m displacement amplitude (Fig. 5.8) in NaCl solution, gross slip contact condition, to probe further the electrochemical response under the mutual action of fretting and corrosion. In the preceding section, it was observed that the OCP response was insignificant. It was of interest to investigate the electrochemical response under conditions similar to cathodic protection thought to be provided by the zinc coating on the steel core.

The sample was held in NaCl solution for about 650 s to stabilize the OCP at about -0.760 V_{SCE} before a cathodic potential of -1.10 V_{SCE} was impressed as shown in Fig. 5.14. When the potential was applied, the negative current increased suddenly to a negative peak (labelled B). The negative current transient decreased to a minimum peak (labelled C) in the positive direction and then increased again until fretting was started at D. At the onset of fretting, the negative current decreased sharply in the positive direction to a smaller value (still negative), and then decayed monotonically with time to a more negative steadystate value. The trend in the current decay without fretting, which is shown by the dashed curve, as it is speculated to be the extrapolation of the continuous curve before the start of fretting and lying parallel and slightly more negative to the curve in the presence of fretting. The mean current transient, which increased (about 0.4 μ A) by fretting, is considered as practically insignificant if the speculated trend without fretting is not too much in error. Therefore, the contribution of fretting to the potentiostatic current increase (in the positive direction) was not substantial, while it was more recognizable in comparison to the mean OCP evolution reported above.



Fig. 5.14 Mean current recorded during fretting under potentiostatic condition in 3.5% NaCl solution (gross slip fretting regime) at a normal force of 5 N, displacement of 100 μ m, frequency of 1 Hz and applied potential of -1.10 V_{SCE}. The current was recorded after the potential was applied at A, fretting was started (D) and end of measurement (E). Labels B and C designate the maximum and minimum negative current values.

Fig. 5.15 shows the negative current transient obtained from a single fretting cycle (taken 40 min after the onset of fretting) obtained from the test described in the preceding paragraph. While the transient tangential force varied proportionally (A to B) with the displacement in the elasticity range (elastic accommodation of the displacement), the negative current transient increased slightly (more negative). The current began to surge with the start of plastic accommodation of the displacement and reached the peak when the maximum tangential force was achieved (point B). As stated earlier, point B on the tangential force curve corresponds to yielding of the specimen surface and start of relative slip. At the time when the tangential force transient decreased slightly and continued as extended plateaus towards C (the ends of the stroke), the transient current decreased slightly and continued constant for the rest of the displacement. In magnitude, the increase or decrease in the current value was insignificant. This shows that the current response was associated with transition of elastic to plastic accommodation of the displacement like in the case of potential transient reported earlier.



Fig. 5.15 Current transient and tangential force in a single fretting cycle obtained 40 min after the onset of fretting in Fig. 5.14 under fretting at potentiostatic condition in 3.5% NaCl solution (gross slip fretting regime) at a normal force of 5 N, displacement of 100 μ m, frequency of 1 Hz and applied potential of -1.10 V_{SCE}.

5.3.8 Fretted Surface Morphology

The surface morphologies obtained from the fretting regimes identified in §5.3.5 is presented in this section. For the contact conditions in the stick regime, there is neither real fretting nor penetration of the electrolyte to the interface. Therefore, surface responses of samples tested in NaCl solution under stick regime fretting was excluded.

Morphology of surfaces fretted in air

Fig. 5.16 shows optical images of specimen surfaces fretted against alumina ball for 3600 s in air at 5 N and displacement amplitudes corresponding to the stick, mixed stick slip and gross slip regimes as identified earlier (cf. Fig. 5.8). Fig. 5.16a shows the morphology of the surface tested under fretting in the stick regime. As can be observed, the deformation mark was only the trace of indentation caused by the normal force applied, which was nearly circular shaped. This indicates that the applied displacement amplitude was totally within the elastic deformation limit of the specimen surface and subsurface region and the test system, as generally known [12]. When the displacement amplitude was increased (to 20 µm plotted in Fig. 5.1a), crack initiation associated with slight wear was observed near the periphery of the contact as shown in Fig. 5.16b. Metal was pushed out and heaped at the edges. A large central area of the contact surface showed no visible signs of wear. This morphology is, therefore, identified as exposed to the mixed stick slip fretting regime, consisting of typical slip (near periphery) and stick zones (near centre) on the contact surface. Fig. 5.16c shows the morphology of a surface fretted in the gross slip contact condition (at 100 µm displacement amplitude), characterized by major wear degradation. Wear debris was produced and accumulated both inside and outside the contact surface, visible as black debris. There was material detached from the specimen and transferred and adhered to the alumina ball as observed in the form of dark deposit on the surface in the optical image (shown in Fig. 5.16d).







Fig. 5.16 Optical image of flat AA1350 alloy surfaces fretted by 10 mm diameter alumina ball at a normal force of 5 N and frequency of 1 Hz in air. a) 3 μ m displacement at stick regime, b) 20 μ m displacement at partial slip regime, c) 100 μ m displacement at gross slip regime, and d) dark deposit of debris adhering to the alumina ball used to obtain (c).

Fig. 5.17 shows a SEM image of a surface fretted in air for 3600 cycles in the stick regime (20 N - 4 μ m, shown in Fig. 5.8). The friction log of the test was presented in Fig. 5.1a. The contact surface was plastically deformed and material was displaced in all direction from the centre and heaped at the edges similar to the one at a normal force of 5 N and 3 μ m displacement (cf. Fig. 5.16a). The contact surface was neither worn nor cracked. This shows that the imposed displacement was within elastic deformation limit.



Fig. 5.17 Secondary electron SEM micrograph showing surface morphology of a surface fretted in the stick contact condition for 3600 s. Test parameters: F = 20 N, $D = 4 \mu m$, f = 1 Hz. Flat AA1350 fretted against 10 mm diameter alumina ball in air.

Fig. 5.18 shows a SEM image of a surface subjected to 3600 fretting cycles in the mixed stick slip regime $(20 \text{ N} - 35 \mu \text{m})$ in air, as indicated by the friction log presented in Fig. 5.1b and fretting curve shown in Fig. 5.5. The centre of the surface (marked by contour) showed neither wear nor crack, but it was only plastically indented and deformed by the initial loading, indicating that it was in stick contact with the alumina ball throughout the test duration. As indicated in the figure, crack was initiated at the edge of the stick domain of the contact surface. The surface outside the contour was slightly worn (shown in the figure) by the relative slip. The displaced metal was nudged and piled towards the ends of the stroke.



Fig. 5.18 Secondary electron SEM image of typical fretted morphology resulting from fretting in air under mixed stick slip contact condition for 3600 cycles. The central area (marked) was in stick and the surrounding area was in slip. A crack was initiated at the boundary between the slip and the stick areas. Test parameters: F = 20 N, $D = 35 \mu \text{m}$, f = 1 Hz. Materials: flat AA1350 alloy specimen and 10 mm diameter alumina ball.

Fig. 5.19 shows the SEM image of a surface at the same normal force and increased displacement (60 μ m) (cf. Fig. 5.6a) in air, which was categorized into the mixed stick slip regime in the fretting map (Fig. 5.8). Plastically deformed metal was pushed out of the contact surface and piled at the ends of the stroke. A smooth surface, which was in contact with the alumina ball, was plastically deformed and flown out to the edge (which is called plastic flow of metal). Small cracks were formed on the fringes of the flown out metal as shown in the figure. Moreover, fretting caused slight detachment of particles combined with transverse cracks (arrow A) at the central area of the contact surface. The slip part of the contact surface was receded inwards as marked in the figure. As presented earlier in 5.3.4, the shape of the fretting loops indicated that the state of the contact involves a gross slip (at the early stage of fretting) and slip

for most of the steady state fretting condition (cf. Fig. 5.6a). This indicates that the contact surface was subjected to a mixed fretting regime, which is critical for crack initiation as reviewed in §2.8.



Fig. 5.19 Secondary electron SEM image showing fretted surface with transverse cracks, deformed and pushed out metal from the contact, and cracks (arrow A) on the boundary between slip and stick zones (arrow B). Particles appeared to be detached from the central area and a partly which was in stick contact. Test parameters: F = 20 N, $D = 60 \mu \text{m}$, f = 1 Hz, and 3600 cycles. Flat AA1350 fretted 10 mm diameter against alumina ball in air. Displacement-parallel to the dimension bar.

Fig. 5.20 shows SEM images of a surface fretted in air for 3600 cycles at 20 N - 80 μ m, for which the friction log was shown in Fig. 5.1d. Although the evolution of the friction log in the steady state was more like those for mixed stick slip regime, i.e., elliptic in shape as presented in §5.3.4 (cf. Fig. 5.1d), the observed morphology shows that there was gross slip on the surface. Like in the case at 60 μ m displacement amplitude (shown in Fig. 5.6a), the curves during the early stage were parallelogram shaped indicating that the contact was in a gross slip condition. Metal pushed out from the contact surface was plastically deformed and formed a bulge at the ends of the stroke as shown in Fig. 5.20a, like the previous cases. Cracking was observed near the centre. The smeared morphology and the absence of metallic

debris in the central worn area indicate that the area was exposed to gross slip even if debris formation was minimal (Fig. 5.20b). Though the metal surface was plastically deformed and smeared over the surface (instead of being detached), the morphology indicates the presence a gross slip. This disagrees, however, with the shape of the steady state fretting loops indicating elliptic profile with a large hysteresis (cf. Figs. 5.1c) which indicates a mixed stick slip contact condition.



Fig. 5.20(a) Secondary electron SEM micrograph a surface fretted at a mixed stick slip contact condition for 3600 cycles in air. Test parameters: F = 20 N, $D = 80 \mu m$, f = 1 Hz. Flat AA1350 fretted 10 mm diameter against alumina ball in air. (b). Marked area A in Fig. 5.20a at higher magnification, showing the worn surface with transverse cracks.

Fig. 5.21 shows SEM pictures of a surface fretted in air at a normal force of 20 N and displacement amplitude of 100 μ m for 3600 cycles (see the fretting curves in Fig. 5.7a). The degradation of the contact surface was similar to that observed above for 80 μ m displacement amplitude (Fig. 5.20). The worn surface and pushed out metal were cracked and more fragmented (Fig. 5.21a) than that in the previous example (Fig. 5.20). However, no cracking was observed inside the contact surface. Higher magnification view of the worn surface (Fig. 5.21b) shows significant flake formation by cyclic plastic deformation. Edges and imprints of removed and loosely attached flakes are visible on the surface as indicated in the figure. The relative slip thus caused material removal in the form of metallic flakes and debris of metallic particles. The resulting major surface degradation thus observed is a typical characteristic of the gross slip regime. Though the morphology of the contact surface clearly indicates that fretting wear was the dominating degradation mechanism, the fretting curve evolution in Fig. 5.7a indicated that the degradation began as gross slip during the first 10 min and transformed into mixed stick slip for the rest of the test duration. This contradiction will be discussed later.




Fig. 5.21 a) Secondary electron SEM image of typical fretted surface morphology resulting from fretting in air under partial slip contact condition for 3600 cycles. The surface was slightly worn. The boundary between the deformed and pushed out metal was fragmented. Inside the wear track was debris particles, metal flakes and traces of detached flakes. b) Detail of A showing wear damage in the form of flakes and wear debris. Test parameters: F = 20 N, $D = 100 \mu \text{m}$, f = 1 Hz. Materials: flat AA1350 alloy specimen and 10 mm diameter alumina ball.

Fig. 5.22 shows a SEM image of surfaces fretted in air at 40 N and displacements of 80 μ m (Fig. 5.22a) and 100 μ m (Fig. 5.22b) for 3600 cycles, belonging to the mixed stick slip regime according to the fretting map presented in Fig. 5.8. The contact surface was plastically deformed, and the deformed metal was pushed out and accumulated at the ends of the stroke. The pushed out metal was smooth and showed no signs of fragmentation or cracks. As shown in the figures, the contact surfaces constituted two zones, *viz.* the central stick zone (inside the contour inFig. 5.22a) and the surrounding slip zone (slightly worn). The morphology of the surface tested at the larger displacement appeared to show relatively limited crack initiation (Fig. 5.22b). These observations indicate that the contact condition leads to failure by fretting fatigue degradation.



Fig. 5.22 Secondary electron SEM images obtained from fretted surface in a mixed stick slip contact condition after 3600 cycles in air. In both cases, the surface was plastically deformed, and metal was pushed out of the wear track. Moreover, the central areas of the contacts were in stick contact and the surrounding areas (outside the initiated cracks) were in a slip. F = 40 N, f = 1 Hz. (a) $D = 80 \text{ }\mu\text{m}$. (b) $D = 100 \text{ }\mu\text{m}$.

Morphology of surfaces fretted in NaCl solution

Fig. 5.23a shows a SEM image of a surface fretted at 15 N and sequential displacement amplitudes of 80, 60, 40, 50, and 6 μ m in 3.5% NaCl solution (Fig. 5.10a). As described in §5.3.7, it was subjected to displacement amplitudes causing gross slip (cf. Fig. 5.12), mixed stick slip (cf. Fig. 5.11a), and stick (cf. Fig. 5.11a) contact conditions. As shown in the figure, the metal surface was deformed and nudged to the ends of the stroke (parallel to the scale bar), creating a bulge like morphologies formed during the test in air. The peripheral area of the contact surface was largely covered by thick and intact layer of debris, indicating that fretting at 80 μ m did not cause disruption of the debris layer covering the surface, hence limited exposure of bare metal. In contrast, the debris layer at the centre of the contact area was disrupted by the subsequent fretting cycles at reduced displacement amplitudes. The thick and intact debris layer was flaked (indicated in the figure) at some areas and disintegrated at others. The presence of debris might have prevented exposure of bare metal to the electrolyte and hence prevented change in the measured average open circuit potential, presented in Fig. 5.11b.





Fig. 5.23 a) Secondary electron SEM micrographs of a surface fretted at a constant normal force and variable displacement in a 3.5% NaCl solution. The horizontal distances C – C and B – B designate 80 and 60 μ m displacements, respectively. The debris covering the surface was intact near the periphery but disrupted and flaked at much of the central area (b). b) Detail of A showing disintegrated layer debris on the surface. Test parameters: specimen/counterpart AA 1350 alloy/10 mm diameter alumina ball, f = 1 Hz, F = 15 N and sequentially imposed displacement of 80, 60, 40, 50, and 6 μ m. The number of the fretting cycles at the displacements listed were 1200, 2400, 1200, 5400, 1200 respectively. At the normal force, the displacements caused gross slip (at 80, 60, and 50 μ m), partial slip (at 40 μ m) and stick (6 μ m) contact conditions (described earlier).

Morphology of samples fretted at constant displacement amplitudes

Fig. 5.24 shows a SEM image of a surface obtained after 3600 fretting cycles in a 3.5% NaCl solution at a displacement amplitude of 35 μ m and normal force of 20 N. The specimen was let to corrode in the chloride solution for several hours as loaded after fretting was stopped to get insight into the effect of corrosion on the surface. As presented earlier, mixed stick slip regime was prevailing in the contact at these test parameters (cf. Fig. 5.5). The morphology of the surface shows that the central area (stick zone) was neither corroded nor worn whereas the surrounding peripheral area (slip zone) was corroded and

produced corrosion debris visible on the surface. This indicates that there was no access of electrolyte to the stick part whereas the slip part was damaged by localized corrosion associated with fretting damage.



Fig. 5.24 Secondary electron SEM image of flat AA 1350 alloy surface fretted against 10 mm diameter alumina ball in a 3.5% NaCl solution under partial stick slip contact condition for 3600 fretting cycles. The area inside the contour was in stick and the surrounding area was in a slip contact condition. The surface was let to stay in the solution for several hours as loaded after fretting was stopped to get the corrosion effect pronounced. Test parameters: F = 20 N, $D = 35 \mu \text{m}$, f = 1 Hz.

Fig. 5.25 shows a secondary electron image of a surface morphology obtained by fretting at 20 N - 80 μ m in a 3.5% NaCl solution. As presented earlier (cf. Figs. 5.6d and 5.1d), the test parameters cause gross slip fretting regime. As shown in Fig. 5.25a, the surface was plastically deformed, nudged out and bulged at the edges but showed no sign of wear. However, it was locally etched by corrosion as shown in the magnified area marked A in Fig. 5.25b. The etching could not occur during fretting due to continuous slip on the surface. Thick debris layer covered the fretted surface and particles compacted into the surface. At the centre of the contact, there was an abrasion line indicating gross slip. The debris layer was locally broken (Fig. 5.25a). However, the average OCP did not change, indicating continuing protection of the underlying metal against corrosion, while degradation continued by gross slip.



Fig. 5.25 a) Secondary electron SEM images of typical surface morphology obtained from fretting in gross slip contact condition in 3.5% NaCl solution at a normal force of 20 N and displacement of 80 μ m. Metal was plastically deformed and pushed to the edges of the stroke and covered with debris layer. The contact area was also covered with debris. b) Detail of A showing an area affected by corrosion etching on the worn surface.

5.3.9 Fretting of cylindrical strands

Tests on cylindrical AA 1350 alloy strands fretted against one another were conducted in air to get an insight into how the fretting results obtained from ball-on-flat experiments are related to those of the actual strands under mixed regime conditions. This justifies the significance of the fretting map presented in \$5.3.5 for studying degradation mechanisms of the actual strands. The fretting test of cylindrical strands were performed at the normal force-displacement combinations of 50 N - 40 µm, 40 N - 100 µm and frequency of 1 Hz. These test data sets belong to the mixed regime of the fretting map developed based-on-ball on flat fretting tests (cf. Fig. 5.8). The friction logs (Fig. 5.26) obtained for the first one hour of the experiment also confirms that the state of the contact between the strands were indeed mixed stick slip regime as evidenced by the elliptical shape of the fretting curves for both cases. Obviously, the tangential force amplitude was much higher for the fretting at 40 N - 100 µm which is attributed to the increased displacement amplitude though the normal force was less. The hysteresis of the curves were reduced in size during the first few minutes and then the fretting process reached a steady state condition in which neither the amplitude of the tangential force nor the shape and size of the curves changed with time. Therefore, nucleation and initiation of fatigue crack is expected at the edges of the contacts as reviewed in \$2.9.



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Fig. 5.26 Friction log obtained from fretting of alike AA 1350 alloy cylindrical strands of 4.75 mm diameter in air at the normal force and displacement specified in the figure and frequency of 1 Hz. Only data recorded during the first hour was plotted as the fretting process achieved a steady state during the first 5 min.

Fig. 5.27 shows the shape and size of fretting loops obtained from fretting experiments on cylindrical strands against each other air and alumina ball-on-flat AA 1350 aluminium alloy fretting experiments in air and NaCl solution. The plot was based on the same normal force, displacement amplitude and data measured 30 min after the onset of fretting. It shows that the highest tangential force amplitude (44 N) occurred in the fretting of cylindrical strands in air. And the lowest tangential force amplitude (18 N) resulted from fretting of flat specimen against an alumina ball in NaCl solution, respectively. Comparatively, the tangential force amplitude resulting from fretting of flat specimen against the alumina ball in air was in between the two (31 N). However, the prevailing regime at the contact was the same in all cases, *i.e.* mixed stick slip regime. In particular, the hysteresis in the fretting curves of the flat and cylindrical specimens tested in air were wery small compared to the one tested in NaCl solution. This indicates that the surfaces tested in air were more in the sticking tendency than the one in NaCl solution.



Fig. 5.27 Tangential force *vs.* displacement obtained from fretting of flat samples (tested in air and 3.5% NaCl solution) and cylindrical strands tested in air. All data are for steady state condition, typically obtained after 30 min after the onset of fretting at a normal force of 40 N and displacement amplitude of 100 μ m at a frequency of 1 Hz.

Fig. 5.28a shows secondary electron image of the fretted surface of a cylindrical strand fretted against an identical strand at a normal force of 40 N and displacement of 100 μ m. It shows a nearly elliptical wear scar (marked) left behind by the fretting. Fig. 5.28b shows the detail of area marked A in Fig. 5.28a. It shows a crack on the surface, created, as already known and expected, in a mixed stick slip fretting regime. Although there was a crack, there was substantial particle detachment from the surface. The particle detachment was apparently caused after the test during mechanical separation of the adhered cylindrical specimens. The magnified area B in Fig. 5.28a is shown in Fig. 5.28c. Ductile fracture morphology formed by detachment of material was observed which appears to be related to separation of the samples as they adhered together after the fretting was stopped. The delaminated zone was apparently created by gently pulling apart the specimens. Therefore, the surface morphology along with the friction log confirm that the degradation mechanism was by formation of cracks at the edge of the contact surface.





Fig. 5.28 Secondary electron SEM micrograph showing wear morphology of a cylindrical strand after 5 hr fretting with identical strand at a normal force of 40 N, displacement amplitude of 100 μ m, frequency of 1 Hz in air and angle about 45 ° between the axes of the strands. (a) The contact surface (marked area) was plastically deformed and fragmented. (b). Magnified view of area marked B in Fig. 5.27a, illustrating metallic fragmentation and cracking on the surface. (c)

Magnified view of A showing delaminated (flaked) area left behind after separating the samples apart. The sample pairs adhered together, and they had to be separated manually after the test.

Fig. 5.29a shows the contact surface of the strand fretted in air at a normal force of 50 N and displacement of 40 μ m, which also belongs to a mixed stick slip regime (cf. Fig. 5.26). The contact surface was elliptical in shape with stick contact at the centre and cracks initiated at the edges. It was plastically deformed and showed no significant wear (material detachment). As shown in Fig. 5.29b, the detail of area marked A in Fig. 5.29a, the edge of the contact surface was slightly worn (see the detached particle in the figure) and crack was initiated which is in fact expected for contacts subjected to mixed stick slip regime [4].





Fig. 5.29 Secondary electron SEM micrograph showing wear morphology of a cylindrical strand after 10 hr fretting with identical strand at a normal force of 50 N, displacement of 40 μ m, frequency of 1 Hz in air, and angle between axes of the strands was about 45 degrees. (a) Elliptical contact surface (marked) with plastically deformed metal pushed out to the edges and the centre appears only plastically deformed. (b). Magnified view of area marked A in Fig. 5.28a illustrating metallic debris and cracks at the edge of the contact and debris.

5.4 Discussion

This study was intended to determine the fretting regimes of a fretting map and identify regimes where no degradation, crack and wear degradations are expected for a flat and polished AA 1350 alloy fretted against an alumina ball in air and NaCl solution. Moreover, it was aimed to elucidate the effect of the test environment on the fretting regimes and investigate the electrochemical response of the surface in a fretting corrosion condition. It was also aimed to determine the influence of contact surface geometries on the fatigue degradation of the strands.

A running condition fretting map characterizing the regimes, where cracking and wear degradations can occur or not, was developed for both in air and NaCl solution (Fig. 5.10). In the fretting map, the mixed

stick slip regime was dominant, *i.e.*, the major degradation mechanism under the conditions investigated was fretting fatigue for both dry air and chloride solution. Moreover, it was found that the domain of mixed stick slip regime for fretting in NaCl solution was narrower than that in air due to broader gross slip regime for the former (Fig. 5.8) as a result of shifting of the boundary between the mixed stick slip and the gross slip regimes. This indicates further that the NaCl solution favours relative slip by providing a lubricating effect.

It was found that fretting does not cause appreciable change in the average OCP (per cycle) in any of the fretting regimes investigated in NaCl solution. However, the transient OCP, at the instant it was measured in a fretting cycle, depended on the type of displacement arising at the interface, which in turn depended on the fretting regime. No remarkable variation in the transient OCP was observed during a fretting cycle in the stick regime. However, it decreased slightly for displacements leading to plastic deformation and yielding under the mixed stick slip and gross slip contact conditions. Overall, it was found that the average OCP per cycle remains nearly constant during elastic deformation. It surges in the negative direction to a minimum when the surface yields (gross slip occurs) and decreases with increasing slip and shows minima at the ends of the strokes (Figs. 5.11b and c, and 5.12b). While decrease of the OCP under plastic deformation is attributed to yielding of the surface, the decrease at the ends of the stroke appears to be due to momentary pause in the relative displacement. In all cases, these negative surges or minima must be related to local film breakdown, while the increase in the positive direction and maxima are signs of passivation. The negative surges indicate "cathodic control" of the surface, i.e., increased anodic oxidation, at sites where the film is damaged, which cannot easily provide the electric charge needed by the reduction process because the necessary cathodic areas (usually the heavy metal rich intermetallic phases) are not instantaneously available. In contrast, passivation causes "anodic control", in which the reduction process does not decrease as rapidly as the oxidation process that passivates the surface. These instantaneous electrochemical changes, which are suggested to occur by the changes observed in the instantaneous OCP measurements, are nearly reproducible during each cycle, returning to the starting values at the end of each cycle. This explains why the average OCP per cycle is nearly constant. It indicates further that the surface recovers its original electrochemical properties at the end of each cycle, independent of the ongoing fretting. In general, the OCP remains near the well-known pitting potential of the aluminium alloy during fretting [13]. Hence, the mechanism of corrosion, at the sites where the protective oxide is damaged by wear, must be similar to the well-accepted electrochemical

mechanism of pitting corrosion [14]. Although material degradation by pitting is not significant in relation to degradation by fretting, small changes in the OCP, which can easily be recorded during wear in chloride solution, give useful information about the mechanisms involved in the mechanical part of the degradation process.

The tangential force generated during fretting in air increased and reached maximum as the contact condition changed from gross slip to mixed stick slip regime (Fig. 5.3). This was seemingly an effect of plastic straining and adhesion of the surface. The occurrence of the highest value of the tangential force in the mixed stick slip regime may be the cause for crack initiation and accelerated fatigue failure in the regime [12]. In NaCl solution, the tangential force amplitude was reduced by a factor of about one-third of that in air (Fig. 5.1). The tangential force generated by fretting is a critical parameter for a fretting fatigue life [11]. Therefore, the reduced tangential force during fretting in NaCl solution suggests that the service lifetime of overhead power lines exposed to marine environments is expected to be much longer than those exposed to dry air environments. This is in agreement with the findings of Takeuchi *et al.* [15] on the fretting fatigue investigation of austenitic steel used for propeller blade in air and seawater. The reduction of the tangential force in NaCl solution may be caused by the lubricating effect of the aqueous environment [16], the corrosion products and wear debris, which hinder adhesion between the contacting surfaces and facilitate relative slip, thereby hindering crack initiation [17].

Fretting map (Fig. 5.8) was constructed based on ball-on-flat surface contacts though the contacting surfaces of the strands are mostly cylindrical in reality. The influence of geometry and environment on the fretting fatigue life was probed by comparing the amplitudes of the tangential forces generated from fretting on identical cylindrical strands and ball-on-flat contacts in air and chloride solution (Fig. 5.27). The results showed that the tangential force arising from cylindrical strands in a fretting contact were by far larger than those obtained from ball-on-flat in air. The crack initiation and growth period on contacts subjected to fretting decreases with increasing tangential forces [11]. This implies that the rate of fretting fatigue of identical cylindrical strands is shorter than those of ball-on-flat (both in air and NaCl solution) configuration. In view of this, the differences in the fretting scar morphologies between the ball-on-flat and cylindrical contact surfaces (Figs. 5.28 and 5.29) can be attributed to the differences in the contact geometries.

In mixed stick slip contact condition (cf. Figs. 5.2a, b and 5.4a, b), the displacement amplitude causes partly elastic and partly plastic (elasto-plastic) deformation as evidenced by the hysteresis loops of elliptical fretting loops. The hysteresis in the curves indicate that the fretting cycle induces permanent plastic deformation (strain hardening) of the contact surface where the centre and the peripheral areas became in stick and slip contact, respectively. Consequently, surface cracks initiated at the boundaries between the slipping and sticking parts of the contact surfaces (Figs. 5.18, 5.19 and 5.22a and b). The surface degradations observed in this regime were in agreement with the earlier theories of fretting fatigue crack initiation mechanism [6, 12]. In this regime, cracking (fretting fatigue) occurs where the fretting curves are elliptical in shape or change from elliptical to parallelogram, or vice versa, as described in §2.9. In this regard, the present results for air did not show any change in the shape of the fretting curves from that of one regime to another except for the initial fluctuations (cf. Figs. 5.5, 5.6a, and 5.7a). In general, the fretting curves started with a deformed hysteresis corresponding to mixed stick slip or gross slip curve, became narrower, and attained steady state shortly after (in about 10 min). This may be due to the disappearance of the initial surface condition after the steady state condition was attained. Moreover, material may be transferred from the specimen to the alumina ball (Fig. 5.16d) as the fretting continued and hence increased the tendency for adhesion. The change in the property of the contacting surfaces, thus, causes change in the fretting curves shown in Figs. 5.6a and 5.7a. In contrast to the tests in air, the steady-state condition was not achieved in the duration of tests in NaCl solution (cf. Figs. 5.6b, c and d and 5.7b). The delay or absence of the steady state condition may be due to short test duration or the presence of the chloride solution causing continuous change of the contact interface. The shape of the fretting curves was continuously changing from that gross slip to mixed stick slip fretting regime. However, there was no clear indication of crack initiation on the fretted surface in NaCl solution (cf. Figs. 5.23 to 5.25). In this regard, further research based on long duration experiment may be needed to substantiate the fretting fatigue crack initiation and wear in relation to the evolution with time of the shape of fretting curves. Notably, the general trend of the fretting curves and fretting regime description based on tests in air may not be directly applicable for fretting investigations in aqueous solutions.

In the mixed stick slip regime, the mixed potential transient showed slight decrease with increased plastic deformation (where the fretting curve is nonlinear) and reached a small minimum at the ends of the strokes (Figs. 5.11b). This was apparently due to plastic deformation of the surface leading to breaking of the oxide film in the part of the contact surface experiencing slip (Figs. 5. 11b). The decrease in the

mixed potential transient with increasing plastic deformation indicates that the mixed potential drop magnitude increased with increasing plastic deformation. It is also clear from the results that the corrosion attack occurs in the slightly worn are of the contact surface where crack can possibly initiate (Fig. 5.24) after fretting stops and the interface remains exposed to corrosive environment. Regardless, of this suggestion, the crack initiation under fretting in the mixed stick slip regime in NaCl solution seemed suppressed unlike the tests in air.

In the gross slip regime, the mixed potential transient in a fretting cycle depends on whether the applied displacement amplitude causes elastic or plastic deformation or vielding (Figs. 5.11b and c, 5.12, 5.15a and b). When the applied displacement amplitude falls below the elastic displacement limit of the subsurface, detachment of material from the surface and subsequent exposure of bare metal to the electrolyte is limited. Hence, no considerable decrease in the mixed potential transient. When the applied displacement caused plastic deformation of the contact surface, however, the mixed potential transient decreased slightly with increasing displacement. The minimum mixed potential achieved at the plateau (transition from elastic to plastic deformation) indicates that the potential dropped due to yielding of the surface thereby exposing bare metal to the electrolyte [12]. The insignificant decrease in the mixed potential under this condition shows that the only possible corrosion degradation taking place at the contact interface is a localized corrosion, namely pitting (see Fig. 5.25). The reason for the slight recovery of the potential with increased displacement was probably due to passivation of the bare metal exposed and absence of formation of new bare metal by the slip. Moreover, the minimum is achieved as the displacement approaches the ends of the stroke (point C), which causes temporary stop due to reversal in the displacement direction, which in turn causes temporary repassivation. Therefore, the mixed potential transient response of the surface was associated with the extreme stretching and yielding of the material surface. Essentially, the nearly negligible response of the mixed potential transient in the stick and mixed stick slip fretting conditions may be due to the fact that the imposed displacement amplitudes were below the threshold at which the material surface yields. [18]. Overall, the observations of the mixed potential response of the surface in different fretting regimes suggest that the disruption of the oxide film and reformation is an essential factor during fretting in an electrochemical environment rather than plastic deformation of the surface, in agreement with the suggestions of Taylor and Waterhouse [19].

Generally, for this work, the friction log and morphological results indicate that the contact surfaces experience fretting fatigue in the mixed stick slip contact condition (cf. Figs. 5.16a and 5.17). The stick regime showed neither surface crack nor wear of practical significance. The degradation of contact surfaces subjected to this regime is less critical compared to that of mixed stick slip regime as reviewed in §2.9 [3]. The degradation in the mixed stick slip regime occurred by crack initiation at the edge of the contact surface (Figs. 5.22) in agreement with the well-established theory and principle [12]. It is generally known that material removed on the surface subjected to this regime is virtually negligible [6, 12]. Corrosion attack occurred in the slip part of the contact surface (Fig. 5.24), hence coinciding with the area damaged by fretting and contributing to fretting fatigue degradation. Therefore, the contribution of corrosion to the degradation is insignificant in relation to fretting fatigue due to the fact that the wear aspect is limited in the mixed stick slip regime. In contrast, the presence of corrosion can be beneficiary to reduce the amplitude of the tangential force arising on the contact surface, hence creating more resistance to fatigue. For contacts subjected to gross slip condition, the degradation mechanism is characterised by fretting wear (Figs. 5.20 and 5.21) [3]. The wear mechanism was mainly by plastic deformation, where the deformed material was pushed out to the edge of the contact surface than being removed as debris.

5.5 Conclusions

- A fretting map representing the stick, partial slip and gross slip regimes was established for AA 1350 alloy in air and in 3.5% NaCl solution in water. The fretting map enables to characterize the expected degradation, such as fatigue or wear, based on the contact conditions (fretting regimes).
- During fretting wear in chloride solution, contribution of electrochemical corrosion to degradation is not significant. However, electrochemical parameters, such as OCP, if measurable during fretting tests in chloride solution, may be useful in indirect tracking of the mechanical factors, which may be important in understanding the fretting mechanism.
- The fretting map indicated that the mixed stick slip was the dominant contact condition which caused the major degradation in both air and NaCl solution at the range of normal force and displacement amplitudes investigated in this work. The degradation was correlated with crack initiation. In the stick regime, neither wear nor crack initiation was observed.

- The degradation rate due to fretting in NaCl solution was significantly lower than that in air. This is related to the shift observed from the gross slip regime dominating in NaCl solution to the mixed stick slip regime in air under otherwise identical test conditions. The tangential force arising from mixed stick slip fretting condition in air is about three times that arising in chloride solution. This in turn is attributed to the lubricating effect of the chloride solution, caused by the corrosion products and wear debris produced during the test, which reduce the adhesion between the contacting surfaces. This suggests that the fretting fatigue life of the strands can be expected to be much longer in marine environment than in dry air environment.
- The tangential force amplitude resulting from fretting of identical cylindrical strands was substantially higher than that resulting from ball on flat fretting contacts. Thus, the fretting fatigue life of identical cylindrical strands must be substantially shorter than ball on flat surfaces in a fretting contact. This suggests that the contact surface geometry plays a significant role in the degradation of the contacting surfaces subjected to fretting.

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6 Discussion

This chapter summarizes the important results of this thesis and discusses the degradation causes and mechanisms responsible for the failure of aluminum strands used in ACSR conductors exposed to marine environments, suggests possible measures for reducing degradation of conductor strands and increasing service life and recommends further work for improving the available know-how in the area.

6.1 Summary of main results

In this study, failure analysis was conducted on failed ACSR conductors recovered from marine environments. Fretting and sliding wear behaviour of the aluminium strands used in the ACSR conductors were also investigated in air and chloride solution in the laboratory to understand the degradation mechanisms of the material in these conditions. Main results are summarized as follows:

(1) The contact damages identified by analysis of failed aluminium strands in service correspond to plastic deformation, cracking and wear. These results corresponded well with the typical contact conditions predicted by the fretting map constructed by using standard fretting tests in the laboratory.

(2) The service life of the aluminium strands of the overhead power line conductors exposed to marine environment is controlled by the fretting fatigue crack initiation process. The mixed stick slip fretting mode is the dominating mechanism for this process.

(3) Under the mutual action of corrosion and mechanical degradations, the mechanical degradation by fatigue and wear is the controlling factor in determining the failure rate of the aluminium strands.

(4) Localized corrosion did not appear to have a significant role in determining the fatigue rate although pitting and crevice corrosion were observed in the crevices between the conductor strands at the clamp mouth.

(5) Delamination is the dominant wear mechanism for fretting and sliding wear in air whereas the deformation and fatigue wear are the main mechanisms in wet contact conditions.

(6) The boundary between the mixed stick slip and gross slip regimes shifts towards the former in a chloride solution compared to that in air. This is caused by the increased tendency to slip in the chloride solution in relation to that obtained by exposure to air alone.

(7) Exposure to chloride solution, in comparison to exposure to air, significantly reduces the tangential force arising from fretting (the lubrication effect) and, thereby, delays crack initiation.

(8) Reduction in the tangential force is suggested to be caused by reduction of adhesion at the contact interface due the formation of a debris layer, consisting of metallic and oxidized Al particles and the chloride solution.

(9) At small displacement amplitudes, fretting fatigue is the dominant degradation mechanism for both fretting in air and chloride solutions.

6.2 Significance of aqueous environment and corrosion

Observation of localized corrosion near and at the zones of failure in service gave the impression that these corrosion phenomena had a direct influence on the degradation by wear, as well as the initiation of cracks by fretting fatigue and their propagation, which cause failure of the aluminium strands exposed to marine environments in §3. However, the experimental results reported in §4 and §5 indicated that the degradation by wear and fretting fatigue was significantly smaller in chloride solution than in air due to the lubricating effect of the solution (§4 and §5). This agrees with the earlier observation of reduction of wear and friction in aqueous environments [1, 2]. The lubrication effect is indicated by the reduction of the tangential force compared to that in air in a fretting condition (§5). The reduced tangential force causes reduced rate of crack initiation. As a result, the crack initiation phase takes longer period in marine environments, and hence results in longer fretting fatigue life [3].

The lubrication is presently suggested to be caused by the formation of a surface layer consisting of a mixture of the solution, wear debris and corrosion products, called the third body, which prevents direct contact between the rubbing surfaces and reduces adhesion and material transfer at the interface. As reviewed in §2, the presence of such lubricant at the contact interface may reduce the first and most important stage of fretting degradation, adhesion and material transfer, which has significance for the subsequent processes. The wear mechanism, which is deformation and fatigue wear in the chloride solution, thus becomes different from that in air, which is delamination and abrasion.

In addition to its inhibitive role, corrosion may be utilized in monitoring the mechanical degradation processes in the presence of a marine environment. As shown in §4 and §5, the monitoring can be implemented by use of a reference electrode as the sensor to measure the corrosion potential (OCP). This is a mixed potential determined by the oxidation and reduction processes controlled by the localized mechanical degradation processes. The values measured in sliding wear (§4) and fretting (§5) experiments were quite different, indicating certain variations in these two degradation mechanisms. Degradations by fretting and sliding wear differ by the fact that the former is attributed to cracking whereas the latter is attributed to removal of material from the surface in the form of fragments of debris. These phenomena occur in different ranges of displacement amplitudes which affect the surface response. The OCP values measured as low as -1.34 V_{SCE} indicated exposure of bare aluminium locally by mechanical destruction of the passivating oxide film in the case of aluminium alloy surface sliding against an alumina ball (§4). The OCP measured in the range - 0.70 - 0.78 V_{SCE} during the fretting experiments (§5), which were significantly more positive than those for the sliding wear experiments, indicated electrochemically more passive surface, possibly caused by the limited destruction of the oxide film at the contact interface. The small changes caused in the OCP by changes in the fretting conditions, as discussed in §5, can have physical significance, which was not investigated any further in the present work. The method can possibly be developed further for use in the field in the presence of wet marine environment acting as the necessary electrolyte. However, this work has also shown that controlling this potential for reducing electrochemical corrosion (e.g., cathodic protection) is not possible since electrochemical processes do not contribute to a significant material loss in the degradation process.

6.3 Fretting fatigue as the dominating failure mechanism

A fretting map was established for the aluminium strand material in air and chloride solution, presented in §5. The map shows the combinations of normal force, displacement amplitude, frequency, and environment which cause fretting fatigue or fretting wear degradations [4]. The map shows that the lowest fretting fatigue life is expected when the aluminium strands are subjected to the mixed stick slip contact condition, which favours crack initiation (§5). Similarly, maximum wear degradation is expected when the strands are exposed to the gross slip fretting contact conditions [5]. Fretting wear degradation of the material was quite limited at the displacement amplitudes investigated. Increasing the displacement amplitude to higher values could have given more pronounced effect of the fretting wear conditions both in air and chloride solution. Within the practically reasonable and possible displacement amplitudes expected from wind induced vibrations, a displacement amplitude up to 100 μ m was suggested as reasonable [6], and it was used in the tests performed in §5. However, the fretting wear results obtained in §5, did not show a morphology similar to those obtained in §3. This may be due to the differences in the geometry of the rubbing counterparts (e.g., strand and strand or strand and cast aluminium alloy clamp in §3 and alumina ball on a flat and polished aluminium strand in §5), and the environmental conditions to which the samples have been exposed. Similarly, the dominant mixed stick slip regime of the fretting map implies that the fretting fatigue degradation is the dominant failure mechanism [2] of the strands for the displacement ranges investigated.

Only a small domain of the fretting map (\$5) belongs to the gross slip fretting regime, as a normal force smaller than 25 N and displacement amplitudes in the range 50 - 100 µm, in agreement with the studies of Ma *et al.* [7]. This indicates that the major degradation of the conductor strands occurs by fretting fatigue both in air and marine environments. However, the tendency for mixed stick slip contact condition is smaller in chloride solution than in air, as reflected by the shifting of the boundary between the mixed stick slip and gross slip regimes of the fretting map. This indicates that a marine environment influences the fretting regime and favors relative slip. Although crack initiation could not be demonstrated, the elliptical fretting curves (\$5) show that the degradation eventually occurs by fatigue in chloride solution. Thus, it appears that the presence of chloride solution delays the crack initiation phase of fatigue by stabilizing the mixed stick slip fretting contact condition.

6.4 Failure mechanism of overhead power line conductors in marine environments

The results of failure analysis showed that the presence of both corrosion and wind induced vibration degradations, namely fatigue and wear, are the causes of degradations (§3). Thus, the degradation mechanisms were described as fretting fatigue-corrosion, wear-corrosion and localized corrosion in §3. Breaking of strands was attributed to the fretting contacts associated with the mixed stick slip fretting regime (§3) as identified by observation of the contact surfaces in accordance with well-known principles [4, 8]. The presence of corrosion deposit at the periphery of the contact surfaces and pitting corrosion on the worn surfaces indicate additional degradation by corrosion which might have occurred together with fretting or in the aftermath. Hence, fretting fatigue-corrosion and wear-corrosion were suggested as the

degradation mechanisms responsible for the failure of the aluminium strands. However, the extent to which corrosion and fretting contribute to the degradations was not possible to identify from the failure analysis. In general, fretting corrosion damages were identified and classified into the well-known fretting regimes, viz. stick, mixed stick slip and gross slip fretting regimes based on the fretting damage characterizations [4]. However, it was impossible to differentiate whether the wear damages were caused by sliding or fretting from worn surface analysis.

Crevice corrosion damages associated with neither wear nor fatigue were found at the clamps and interlayer and inter strand contacts in §3. The crevice corrosion at the inter-strand interfaces was due to trickling and accumulation of moisture and corrosive media into the contact interfaces [9, 10]. The crevice corrosion attack on the outer layer strands and the clamps indicate that the galvanized steel core was not giving complete protection against localized corrosion. Therefore, the risk of localized corrosion seems to be always there on the conductors. The finding of the failure analysis agrees with other investigations on conductors exposed to marine and coastal areas by corrosion [10-12].

Conductor degradations occur at the clamps and fittings (especially suspension clamp) by the action of wind induced vibration and localized corrosion in marine environments. Based on fractographic analysis of fractured surfaces recovered from failed samples, the service life of conductors appears to be controlled by crack initiation and growth to critical length which is caused by fretting. The suggestion is supported by other studies concluding that about 90% of the fretting life of a component subjected to fretting is determined by the crack initiation [13, 14].

The crack initiation sites on the strands were damaged by both fretting and pitting corrosion. As described earlier, pitting does not occur in the presence of wear/deformation on a surface but together with fretting in stick and mixed stick slip condition due to the crevice effect of the contact and level of the potential. It can also initiate in the post fretting stage due to surface defects on passive aluminium strands as observed in §4 after the sliding wear was terminated and potentiostatic potential was switched off and the sample was held at OCP in NaCl solution. Pitting was also found to initiate on a fretted surface after termination of the experiments as seen in §5. So, any viable model for lifetime prediction should take sequential and simultaneous action of fretting and pitting corrosion into consideration. Pitting is complex

to model as the initiation and propagation [15] are far from clear let alone in conjunction with fretting. Likewise, fretting is also a new stream of tribological research for which analytical model is not fully developed yet. A fatigue crack initiation from corrosion pits [16] caused by fretting cracks may fit well with the assumption of semi-elliptical model of pitting and lifetime prediction [17]. Hence, this thesis has been concentrated on the basic issues of fretting and wear in the presence and absence of corrosion to clarify the way forward for lifetime prediction. The results are expected to enhance understanding on the sequences of degradation of the aluminium strands used in overhead power lines serving in marine environments.

The fretting wear aspect was limited by the range of displacement amplitude applied in the experiments. Increasing the displacement amplitude to higher values than 100 μ m could have enabled to exhaustively investigate the fretting wear degradation in air and chloride solution and mark transition from the fretting wear to sliding wear [18]. Such experiments could better be conducted in a test rig with a capability of both fretting and sliding wear tests. Thus, the transition in the fretting regimes from stick, mixed stick slip to gross slip regime and further into sliding wear can be clearly described. Lack of such laboratory facility has limited the results reported in this thesis to fretting degradations at low amplitudes and sliding wear at a displacement amplitude of 5 mm. The rigs used for the tests were also different in the minimum normal force applied, displacement amplitude range and radius of the alumina ball used. Hence, it was impossible to run a continuous test in which transition from fretting to sliding wear degradation can be assessed.

6.5 Suggestions for Further Work

- Post-mortem analysis of failed conductors was conducted on a small group of samples obtained from a real service environment after being stored for some period. Hence, characterization of a new batch of conductors, which are protected against further corrosion during storage and transport, is recommended.
- The fretting damage investigations in air and chloride solution were conducted at displacement amplitudes up to 100 µm. The results showed that fretting wear was the dominating failure mechanism in this limited domain of the displacement amplitude applied for the normal forces considered. However, sliding wear was observed at the displacement amplitude of 5 mm used in

Chapter 5. Therefore, further work is needed to clarify the transition from fretting wear to sliding wear for the same normal force and range of displacement amplitudes from 100 μ m to 5 mm. Such investigation can also clarify the variation of both the tangential force transient and electrochemical behavior of the surface at different displacement amplitudes.

- The fretting and sliding wear experiments presented in this study were based on ball on flat contact configuration. In reality, the strands are subjected to internal load (tensile and bending), carry an electrical current and operate at a significantly higher temperature. Fretting and sliding wear studies taking these parameters into account would reflect a more realistic degradation mechanism.
- Investigation of wind induced degradation and corrosion should be conducted in a controlled laboratory, where the clamping condition can be simulated, and artificial vibration can be applied. The marine environment can be mimicked by exposing the conductor-clamp assembly to a regulated chloride solution. Such set ups can reflect the influence of electrochemical on the tribological condition can be well reflected.
- The conductor strands carry high voltage electric current. As a result, the temperature of the strands can increase to significantly higher values than the ambient laboratory temperature used in this work. Further research is needed to clarify how the added presence of alternating current and increased temperature affect the degradation mechanisms.

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7 Conclusion

- Analysis of samples obtained from aluminum conductor strands, which had failed during service in marine environments, suggested that fretting and sliding wear degradations were the main causes for their degradation. These degradations are essentially caused by windinduced vibration which inflicts interfacial slip at the contact surfaces of the strands.
- Laboratory investigation on the AA1350 alloy strand used for the conductor strands showed that the wear degradation was faster and more extensive in air than in chloride solution, indicating the lubricating effect of the marine environment. Thus, the conductor strands are expected to be less resistant to wear degradation in dry air environment than humid, wet and marine environments in service. The lubrication effect is caused by formation of a layer of metallic and oxide particles of wear and corrosion debris in the chloride solution which transfers the load across the contact interface and reduces adhesion by direct contact between the rubbing surfaces.
- The fatigue life of the aluminum strands is determined by the initiation of the fretting fatigue crack, which is very slow in relation to the subsequent mechanisms, namely crack propagation by fatigue and ductile fracture phases of fatigue process.
- A fretting map, correlating the normal force to the displacement amplitude, is constructed for investigation of fretting degradation of the aluminum strands in air and chloride environment. The map shows the combination of normal force and displacement amplitude causing a stick, mixed stick slip, and gross slip contact conditions in chloride solutions and air. In terms of material response, these regimes are associated with no degradation, cracking degradation and wear degradation, respectively. The map verifies that fretting fatigue, which is associated with the mixed stick slip contact condition, is the dominant degradation mechanism for the aluminum strands under the conditions investigated in the present work.
- Delamination is the dominant wear mechanism for both fretting and sliding wear degradations in air whereas plastic deformation is the dominant wear mechanism in chloride solution.
- The tangential force at a given displacement amplitude for fretting in the mixed stick slip contact condition in air is about three times that in chloride solution, indicating fretting in air causes a significantly higher risk of fretting fatigue failure than in the marine environment, hence, a longer fretting fatigue life under fretting in the marine environment.

- The tangential fretting force generated from fretting of cylindrical strands is larger than that of fretting by an alumina ball on a flat polished aluminum strand surface in air, suggesting that the cylindrical strands are more susceptible to fretting fatigue than that of flat surfaces against an alumina ball. This indicates that strands with flat surface are more resistant to fretting fatigue than those with cylindrical surfaces.
- On a fretting map, the boundary between the mixed stick slip and gross slip regime in chloride solution shifts in favor of the latter due to the lubrication of the interface.
- The major degradation of the conductor strands is caused by vibration induced wear and fatigue under a mutual action of corrosion and wear/fatigue. Hence, cathodic protection is not beneficial against wear and fatigue damages acting reciprocally with corrosion.
- Under the combined action of corrosion and sliding wear, the major material removal occurs by the mechanical wear. The corrosion contribution comes mainly by passivation of the bare metal exposed by the mechanical wear, which is insignificant. Moreover, the corrosion products occupying the wear track hinders further material degradation by reducing the direct contact between the counter parts and acts as a lubricant.
- Under sliding wear at controlled potentiostatic potentials in NaCl solution, repassivation of the wear track cannot occur at potentials more negative than -1.34 V_{SCE} and more positive than -0.74 V_{SCE} because of alkaline etching by water reduction (hydrogen evolution) and pitting by acidic hydrolysis, respectively. Under such conditions, the effect of sliding wear on the corrosion degradation was negligible.