-TITANIUM ALLOYS -
IN-VITRO CORROSION AND WEAR WITHIN VARYING BIOLOGICAL ENVIRONMENTS

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ABSTRACT

Ti (cpi) and Ti-6Al-4V alloy have both been used as biomaterials for over two decades. They provided a successful alternative to the use of stainless steels/Co-Cr alloys due to their superior corrosion resistance and biocompatibility. However, recently there has been great concern over the biocompatibility of the alloy. Especially, if titanium is released into the tissue along with aluminium and vanadium which are both now believed to cause toxicity effects. Thus, several alternative alloys have been proposed with slightly varying compositions. The aim of this project was to try to evaluate how a variation in composition and phase structure might influence the corresponding corrosion/wear behaviour of such materials. Also, to try to understand how the passivity and breakdown characteristics of these materials varies within well defined and controlled, but varying biological environments. The corrosion and wear behaviour of these alloys was investigated, initially, within a simple phosphate buffered saline (PBS) solution and then within progressively more complicated environments containing albumin protein, serum proteins and even macrophage cells. The results of this study suggest that although the composition and phase structure plays an important role in influencing the corrosion resistance, other factors such as surface oxide properties and their interaction with the ions/organic species within the environment may be just as important a consideration. When considering the degradation of such materials due to wear, it was found that surface hardness change may be a useful parameter to take into consideration. The co-joint action of corrosion and wear was also studied to try to mimic the in-vivo situation where both degradation effects may be occurring at the same time or even along different time scales. Overall, it was found that alpha-beta/near beta alloys proved to possess the best combination in terms of corrosion and wear resistance and that both the alternative alloys tested within this test matrix were found to produce a more favourable response than Ti-6Al-4V alloy. Ti-13Nb-13Zr alloy seemed to produce the best results in response to any changes in the environment and even to a change in its surface roughness, however, there were one or two concerns raised as to the possible effects of tribological wear as a result of harder abrasive three body wear, while testing in a simple PBS environment. However, in the presence of proteins this alloy still produced better results compared to the other alloys. It is sometimes known for such materials to become worn during loosening of the implant in the in-vivo situation, which produces rougher surfaces in contact with the biological environment. Open circuit potential studies were performed as well as accelerated corrosion testing to investigate how changes in surface roughness might affect the corresponding passive stability and corrosion behaviour. Overall, it was found that Ti-13Nb-13Zr alloy produced better results in PBS solution, whereas, in proteinaceous solutions Ti-6Al-7Nb alloy produced slightly better results. The influence of human macrophage cells on the passive stability and corrosion resistance of these alloys was also studied. It was found that these cells interfered with the metal oxide/protein layer changing the corrosion behaviour of these materials and altering the results, suggesting that the corrosion data using simple PBS solutions should be used with caution when screening such materials during in-vitro studies to try to assess their suitability as biomaterials. It was found that the cells became activated to a different extent as a result of exposure to the environment itself and then following corrosion.

CHAPTER 1: TITANIUM ALLOYS AS BIOMATERIALS FOR ORTHOPAEDIC APPLICATIONS - AN INTRODUCTION

1.1 TITANIUM ALLOYS - A REVIEW OF THEIR PHYSICAL METALLURGY

An understanding of the physical metallurgy of titanium alloys is necessary to establish some evidence which promotes the use of titanium as a non-ferrous metal over its counterparts. This may also lead us to reason how and why it is increasingly becoming a metal of choice in commercial applications, especially in relation to the medical device industry.

Titanium has a fairly low density (60% that of stainless steel). It may be highly strengthened by alloying and deformation processing. It is non-magnetic and has good heat-transfer properties. Its coefficient of thermal expansion is a lot lower than that of steels and less than half that of aluminium. Titanium and its alloys have melting points higher than those of steels. The combination of high strength, stiffness, good toughness, low density and good corrosion resistance provided by various titanium alloys allows weight savings in aerospace structures and high-performance applications. The excellent corrosion resistance and biocompatibility coupled with high strengths make titanium and its alloys useful in biomaterial applications. Because of its unique corrosion behaviour, titanium is used extensively in prosthetic devices such as artificial heart pumps, pacemaker cases, heart-valve parts and load-bearing joint replacements [Boyer et al., 1985, Donachie, 1988, Jaffee and Promisel, 1970, Collings, 1984, Godard et al., 1967].

The physical metallurgy of titanium shows that it undergoes an allotropic transformation at about 885 °C, changing from a close-packed hexagonal (c.p.h.) crystal structure {ie. alpha phase} to a body-centred cubic (b.c.c.) crystal structure {ie. beta phase}. Alloying elements are known to, either, lower or raise this transformation. Depending on their microstructure, titanium alloys fall into five categories: alpha, near-alpha, alpha-beta, near beta or beta. Each of these denotes the general type of microstructure present after heat treatment and processing.

The beta phase is stable upto the melting point. It is possible to have alpha, beta, and mixed alpha-beta structures. The properties of the alloys can be adjusted by the choice of heat treatment. Microstructures in titanium alloys are usually developed by heat treatment and processing (working) followed by further heat treatment. Structural changes are achieved by the production of a high temperature beta-phase and by the transformation of beta on cooling. Typical alloys that are air cooled from the beta-phase field will exhibit a 'Widmanstatten' alpha-phase morphology [Boyer et al., 1985].

The alpha phase forms preferentially on the $\{100\}_{beta}$ matrix planes. This is often termed as 'basket weave morphology'. The sluggish nature of the diffusion reaction leads to retention of metastable beta between the Widmanstatten plates. More rapid cooling (eg. water quenching) can lead to the formation of a martensite form of the alpha phase (termed alpha') as with the ferrous system which is also a diffusionless process. Alpha' is an h.c.p. structure but with different lattice parameters from the equilibrium alpha structure. A well defined habit plane exists and alpha' has a high dislocation density. The martensite starting and finishing temperatures, M_s and M_f, for titanium alloys are alloy composition dependant. Alpha stabilizers increase M_s, while beta stabilizers decrease it. M_f behaves in a similar way to M_s. Alpha' does not experience the same hardness increase as martensite in the ferrous system. There is a possibility of having retained metastable beta in heat treated alloys if M_f is below room temperature or if diffusion controlled decomposition is sluggish. The metastable beta can be decomposed by re-heating to 500-700 °C. For titanium alloys this process is also termed 'AGEING'. The phases formed in the decomposition of the metastable beta influence the mechanical properties significantly.

The microstructure of titanium (cpi) depends on the specific annealing cooling conditions employed and the amount of cold working. Annealing at 880 °C in the alpha phase gives a large grain size and a low strength (TS = 280 MPa). A more significant influence on mechanical behaviour is brought about by dissolving interstitial solutes, eg. H, N, C and O in the metal. H content must be kept below 100 ppm otherwise titanium hydride can precipitate which lowers the toughness, although this has little effect on strength. O, N, and C all affect the strength.

For Ti-6Al-4V, consider the vertical section at 6% Al. The alloy is slow cooled from the beta-phase region and as a result the alpha-phase begins to form, below the beta transition temperature (~980 °C). A Widmanstatten microstructure develops. Upon quenching, the alloy forms alpha' martensite, but there is some retained beta at room temperature, since M_f is less than 0 °C. If the alloy is then annealed at 700 °C for 4 hours, and subsequently air cooled, a mixture of equi-axed alpha and beta grains are obtained. On further cooling, the beta grains transform to a mixture of globular retained beta and an even softer material (TS = 1050 MPa) is formed. To obtain higher strength, the alloy is then 'SOLUTION TREATED' by heating the alpha/beta alloy to the solution treatment temperature (950 °C) for 1 hour, producing a high ratio of alpha:beta. On quenching, the beta transforms to alpha' and retained beta. The alloy is then 'aged' at approx. 500 °C for about 6 hours. This allows precipitates to form in the alpha' and retained beta phases as they decompose to produce more stable phase mixtures (TS = 1120 MPa) [Jaffee and Promisel, 1970, Collings, 1984].

The substitutional alloying elements play an important role in controlling the microstructure and properties of titanium alloys (Boyer et al., 1985, Donachie, 1988 and Jaffee et al., 1970). Tantalum, vanadium and niobium are beta amorphous (ie. have similar phase relations) with body-centred cubic titanium. Titanium does not form intermetallic compounds with the beta isomorphous elements. Eutectoid systems are formed with chromium, iron, copper, nickel, palladium, cobalt, manganese and certain other transition metals. These elements have a low solubility in alpha titanium and decrease the transformation temperature. They are usually added to alloys in combination with one or more of the beta isomorphous elements to stabilize the beta phase and prevent or minimize the formation of intermetallic compounds, which might occur during service at elevated temperature. Zirconium and hafnium are unique in that they are isomorphous with both the alpha and beta phases of titanium. Tin and aluminium have significant solubility in both the alpha and beta phases. Aluminium increases the transformation temperature significantly, whereas, tin lowers it slightly. Aluminium, tin and zirconium are commonly used together in 'alpha' and 'near-alpha' alloys. In alpha-beta alloys, these elements are distributed approx. equally between the alpha and beta phases. Almost all titanium alloys contain one or more of these three elements because they improve creep strength in the alpha phase.

Many more elements are soluble in beta titanium than in the alpha. Beta isomorphous alloying elements are preferred since they do not form intermetallic compounds. However, iron, chromium, manganese and other compound formers are sometimes used in beta-rich alpha-beta alloys or in beta alloys, because they are strong beta stabilizers and improve hardenability and response to heat treatment. Nickel, molybdenum and palladium improve the corrosion resistance of unalloyed titanium.

Intermetallic compounds and secondary phases are formed in titanium alloy systems. The more important phases, historically, have been omega (w) and alpha''. The omega phase has not been proven to be consistent in commercial alloy systems using present day processing practice. Alpha'' has been considered to be a factor in some cases of stress-corrosion cracking. Most of the present interest in alpha'' centres on its use as a matrix for a high temperature titanium alloy.

Unalloyed titanium

Oxygen and iron are the primary variants in these commercially pure grades. The strength of titanium (cpi) increases with increasing the oxygen and iron contents.

Alpha and Near-alpha alloys

Unlike alpha-beta and beta alloys, alpha alloys cannot be strengthened by heat treatment. Generally, alpha alloys are annealed or recrystallized to remove residual stresses induced by cold working. Poor forgeability is manifested by a greater tendency for centre bursts or surface cracks to occur, which means that small reduction steps and frequent re-heats must be incorporated in forging schedules. Near-alpha or 'super' alpha alloys contain some retained beta phase, but, consist primarily of alpha and behave more like conventional alpha alloys than alpha-beta alloys.

Alpha-beta alloys

These alloys retain more beta phase after solution treatment than do alpha alloys, the specific amount depending on the quantity of beta stabilizers present and on heat treatment. Alpha-beta alloys can be strengthened by solution treatment and ageing. Solution treatment and ageing can increase the strength of alpha-beta alloys 30-50%, or more, over the annealed or over-aged condition. Response to solution treating and ageing depends on section size; alloys relatively low in beta stabilizers (Ti-6Al-4V, for example) have poor hardenability and must be quenched rapidly to achieve significant strengthening. As the content of beta stabilizers increases, hardenability increases. Alloy composition, solution temperature and ageing conditions must be carefully selected and balanced to produce the desired mechanical properties in the final product. Alpha-beta alloys, when properly heat treated, display an excellent combination of strength and ductility. They are also stronger than both the alpha or the beta single phase alloys [Boyer et al., 1985].

Beta alloys

Beta alloys are richer in beta stabilizers and leaner in alpha stabilizers than alpha-beta alloys. They are characterized by high hardenability with the beta phase completely retained on the air cooling of thin sections or water quenching of thick sections. Beta alloys have excellent forgeability, and in sheet form can be cold formed more readily than high-strength alpha-beta or alpha alloys. The chief disadvantage of beta alloys in comparison with alpha-beta alloys are high density, lower creep strength and a lower tensile ductility in the aged condition. They are metastable because cold work at ambient temperature or heating to a slightly elevated temperature can cause partial transformation to alpha. The principal advantages of beta alloys are that they have a high hardenability, excellent forgeability and good cold formability in the solution treated condition.

1.2 THE USE OF TITANIUM ALLOYS AS BIOMATERIALS IN ORTHOPAEDIC APPLICATIONS

Titanium has now been used as an implantable material for several decades, usually being used in the form of commercially pure metal or the Ti-6%Al-4%V alloy. These two materials have had an excellent reputation for corrosion resistance and biocompatibility over other alternatives, such as the stainless steel and cobalt-chromium alloys, presently available [Saul, 1994, Cahn et al., 1992, Boyer et al., 1985, Davidson et al.,1994, Speck et al., 1980, Aragon et al., 1972, Pilliar, 1991, Meullar et al., 1970, Clark and Williams, 1982, Torgersen et al., 1994, Silva et al., 1990, Rae et al., 1975, Hansson et al., 1983, Rosenberg et al., 1993, Scales, 1991, Poss et al., 1988]. This is why they have become materials of choice for many surgical devices. However, there is now considerable controversy over the biocompatibility of the alloy and concern over its use [Blumenthal et al., 1989, Jacobs et al., 1991, Williams D.F., 1977, Moran et al., 1991, Elo et al., 1972, Gabriel et al., 1994, Rogers et al., 1994, Korovessis and Repanti, 1994, Williams D.F., 1994, Browne and Gregson, 1994, Wapner, 1991, Maurer et al., 1994].

There are two main points to consider in this controversy:-

(a) Firstly, although titanium has an excellent corrosion resistance, if titanium is released into the tissue by the way of passive dissolution or perhaps by another process involving wear, the reaction may be quite different. The response could be either a mild one, producing a discoloration of the tissue to a severe one involving inflammation leading to problems such as ‘osteolysis’. Of greatest concern here is the release of titanium through an abrasion process, either, at the bearing surface of a joint replacement or at the interface between titanium and bone.

(b) Secondly, if deposits of titanium alloy particles are present in the tissue, then concentrations of aluminium and vanadium will also be present. This issue concerns the optimal alloying additions. The alloy referred to (ie. Ti-6Al-4V) has exceptionally good mechanical properties and it has generally been assumed that its corrosion resistance and biocompatibility are as good as the pure metal. The alloy does, however, contain both aluminium and vanadium which themselves have considerably different biological properties than titanium.

The toxic effects of these elements have been documented on numerous occasions. For instance, aluminium toxicity has been linked to 'Alzheimers dementia' and in renal dialysis patients (affected patients exhibit 'osteomalacia' with low bone formation and no fibrosis). Vanadium on the other hand, is suggested to interact with intracellular components and inhibits certain enzyme functions. Fibroblasts exposed to vanadium have shown gross necrosis with only cytoplasmic debris and pynocytic nuclei in contact with the metal [Scales, 1991, Blumenthal et al., 1989, Jacobs et al., 1991, Williams D.F., 1977, Gabriel et al., 1994, Wapner 1991, Maurer et al., 1994, Lalor et al., 1991, Black, 1988]. Thus, overall, it has been proposed that undesirable biocompatibility properties may arise through the presence of these elements and that better performance would be achieved through the use of alloys of different composition.

In the present study, a range of alloys were investigated, but in-depth studies were undertaken on two specific alloys because of their compositional contents (ie. Ti-6Al-7Nb and Ti-13Nb-13Zr). The Ti-6Al-7Nb alloy was developed using the alloying element niobium which replaced vanadium in the Ti-6Al-4V alloy, since both vanadium and niobium are beta amorphous (ie. have similar phase relations) with body-centred cubic (b.c.c) titanium, and thus, can both act as suitable beta stabilizers, suppressing the formation of the alpha phase which is a close-packed hexagonal (c.p.h.) structure [Boyer et al., 1985, Donachie, 1988, Jaffee and Promisel, 1970]. The use of niobium can produce ultimate strengths of upto 1000 MPa and above by either dispersion hardening or alloying without any great loss in ductility. Thus, after appropriate manufacture, niobium can be used in high-stressed implants despite its low ultimate strength (annealed approx. 300 MPa). Also, the fatigue behaviour of niobium does not change significantly in a corrosive environment as well as showing a very low notch sensitivity and tendency to crack propagation. These are beneficial properties for prostheses that are subjected to conditions in the body that can lead to implant failure that originates at the implant surface. The elastic modulus of niobium ($\sim$110 MPa) is closer to that of bone than cobalt-based alloys. As well as this, its excellent biocompatibility make it possible to design implants directly attached to the skeletal system. However, the friction behaviour of niobium makes it unsuitable for use at articulating surfaces.

Niobium and tantalum are covered by a dense protective oxide layer in biological and most other environments which makes them chemically inert and imparts excellent corrosion resistance. Apart from titanium- and zirconium-based materials, only tantalum and niobium can be combined with other metals without breaking down this passive film. The only reason why tantalum is more seldomly used, although it has very similar properties to niobium, is that its abundance on the earth is 5 ppm whereas niobium's is 65 ppm (similar to copper and tungsten). Therefore, the above properties are similar for tantalum which may also be considered as a useful surgical implant material [Saul, 1994, Streicher et al., 1991, Williams D.F., 1990 (a)]. Recent studies have also mentioned the excellent short- and long-term biocompatibility of niobium. There is no significant growth inhibition of fibroblasts and tight bone contact develops around intraosseous implants, lasting upto a year under both unloaded and heavily loaded conditions. Additionally, there are no reports of toxicity of tantalum and niobium in humans [Streicher et al., 1991, Williams D.F., 1990 (a)]. Unfortunately, the same cannot be said about aluminium, although, it continues to fulfil its role as a useful alpha phase stabilizer. It has significant solubility in both the alpha and beta phases of titanium. In general, increasing the grain size within the microstructure of a titanium alloy produces a small decrease, both, in tensile properties and ductility of the alloy. Fine equi-axed alpha grains may be produced by suitable heat treatment after cold-working and the only way in which the mechanical strength of a titanium-aluminium alloy can be significantly increased is by cold-working. Subsequently, alloys of titanium and aluminium exhibit a higher modulus of elasticity than either pure titanium or aluminium.

The other relatively "newer" material being presently considered as an alternative to Ti-6Al-4V is Ti-13Nb-13Zr, which once again relies on niobium as a beta phase stabilizer. The other alloying element, zirconium is one that is unique in that it is isomorphous with both the alpha and beta phases of titanium [Boyer et al., 1985, Donachie, 1988]. A combination of these two alloying elements has made it possible to develop a structure that is a "near" beta phase alloy, supposedly possessing a superior corrosion resistance over the alpha-beta phase alloys (ie. Ti-6Al-4V and Ti-6Al-7Nb), but one that still has enough alpha phase present in the final structure to provide the necessary mechanical strength. It has been proposed that Ti-13Nb-13Zr alloy is more favourable for orthopaedic implants than Ti-6Al-4V alloy because of its superior corrosion resistance and biocompatibility. Reasons for this superiority have included factual evidence to suggest that less metal ion release is likely to occur during spontaneous passivation of Ti-13Nb-13Zr alloy because the corrosion products of the minor alloying elements, niobium and zirconium are less soluble than those of aluminium and vanadium. Also, that the passive oxide layer is more inert consisting of a dense rutile structure providing greater protection to the underlying alloy [Kovacs and Davidson, 1993 (a), Kovacs and Davidson, 1993 (b)]. Ti-13Nb-13Zr has also been shown to have a lower elastic modulus (E ~ 79 GPa) compared to Ti-6Al-4V (E ~ 115 GPa) and the Co-Cr-Mo alloy (E ~ 227 GPa) which are currently used [Davidson et al., 1994]. The elastic modulus is a measure of the amount of energy needed to pull atoms apart, which will depend on the nature of the atoms in the material. It is an inherent property or a constant for a material. Each material will have its own elastic modulus and it is important that it refers to the ease with which a material deforms elastically. It has nothing to do with the strength of the material. Ceramics usually have very high moduli and are, therefore, very rigid. Metals generally have slightly lower moduli than ceramics [Williams and Cunningham, 1979].

This property may be an important factor when considering the loosening of a joint prosthesis due to fretting which may cause the release of significant amounts of metal into the tissue [Williams D.F., 1994]. High stiffness implants can also cause bone loss due to stress shielding and there is a growing concern over excessive bone resorption. High stiffness may limit the ability to transfer load to the surrounding bone in the proximal calcar region [Pilliar, 1991]. Thus, to improve load transfer, designs have been considered with less cross-sectional area to increase flexibility, but at the expense of fit and fill (ie. stability of implant with bone). Should stem loosening occur, then, stem stresses may exceed the relatively low fatigue strength of the alloy and lead to stem breakage [Davidson et al., 1994]. In any case, compared to other acceptable alloys, titanium alloys have lower moduli of elasticity, thus, theoretically (if not necessarily practically) reducing micromotion between femoral stem and bone. This suggests that such materials have certain advantages, but once loosening occurs, abrasion will lead to more rather than less metal ion release [Williams D.F., 1994].

1.3 THE CORROSION AND WEAR BEHAVIOUR OF TITANIUM ALLOYS

It has been shown for Ti-6Al-4V to corrode 176 times slower than Type 316L stainless steel within biologically simulated solutions [Aragon et al., 1972]. Because of the excellent corrosion resistance of titanium alloys, under normal physiological conditions, they should not breakdown without the presence of other accelerating factors, such as wear. However, there is still some release of the material into tissue due to passive dissolution. Thus, metal ions and metal-protein complexes could still cause tissue damage, and therefore, passive corrosion rates are still a significant factor as far as the corrosion behaviour of these materials is concerned [Browne and Gregson, 1994, Healy and Ducheyne, 1992].

There is no doubt that passive titanium relies upon a surface oxide film for its protection and that when it is breached or removed, corrosion ensues. It follows, that thicker films are more protective than thinner ones. Rest potentials are found to be dependent on the thickness of the passive film, having more electropositive potentials with increasing film thickness [Solar et al., 1979]. It is partly due to this phenomenon that much of the research which is presently being undertaken in this area of study, involves looking at novel techniques to grow passivating films onto such materials, in order to improve the corrosion resistance and biocompatibility, further still [Shirkhanzadeh, 1992, Liu and Burstein, 1995].

The above mentioned, protective crystalline oxide film that normally passivates titanium is believed to be 12-16 angstroms thick. After about 70 days it reaches 50 angstroms and after 4 years it grows to about 250 angstroms when exposed to air. In terms of composition, the film is known to vary along its thickness. Near the metal interface it consists of TiO and at the surface it is TiO_2, but in the interior it contains Ti_2O_3. The production of passivity is considered as the anodic process of oxide film formation, as a monolayer, from the thermodynamic and kinetic points of view. Readily adsorbable anions such as Cl^- hinder passivation and promote dissolution in terms of the relative adsorption of anions and water at the metal-oxide interfaces, as influenced by anion concentration and electrode potential [Williams D.F., 1990 (a), Godard et al., 1967, Hoar, 1967, Pan et al., 1994].

When comparing Ti (cpi) and Ti-6Al-4V alloy, the corrosion behaviour of Ti (cpi) has been found to be least affected by changes in surface roughness, whereas, Ti-6Al-4V alloy has been shown to become progressively more electro-positive when the potential was measured after it had been coarsely polished (600 grit emery paper) and fine polished (0.05 micron diamond paste), respectively [Jaffee and Burte, 1973, Solar et al., 1979]. As a means to explain how the surface roughness might affect the corrosion behaviour of such materials, a model of the passive film has been proposed [Solar et al., 1979]. This is to say that mechanical polishing in air produces a microscopically rough surface consisting of oxide needles at surface irregularities on top of a smooth planar oxide. However, because of the great difference in thickness of the two layers, the field across the needles will be negligible compared to the field across the planar oxide. Thus, if electrochemical reactions occur where film growth takes place, the net result will be a reduction of the effective area (or no. of active sites) of the surface.

In strongly reducing solutions, some oxide needles may dissolve increasing the effective area. Also, when the degree of mechanical polishing compound was varied, then, it was found that the effective area decreased as the size of the polishing compound decreased. This observation is consistent with the oxide needle model proposed because a finer abrasive would produce a greater number of scratches or microscopic irregularities on the metal surface, providing more places for the needles to form. This is the main reason for careful monitoring of surface roughness factors during surface preparation of specimens throughout this project.

Metallic wear is not known to be the main cause of failure of prosthetic devices in the short-term, however, in the long-term it may contribute to reasons for prosthesis removal [Dumbleton, 1978]. Thus, it is important to consider the effects of wear debris that may accumulate in the tissues adjacent to an implant, which can sometimes cause problems related to toxicity in the way of unfavourable inflammatory responses leading to loosening of the implant and eventually fracture of the prosthesis [Pizzoferrato et al., 1991, Williams D.F., 1990 (b), Nasser et al., 1990, Maloney, 1993]. When considering materials for the application of orthopaedic prostheses, it is not only important to consider properties such as corrosion resistance, but also those of wear behaviour. Subsequently, when evaluating 'new' materials it is important to study these two phenomena, both on their own and in a co-joint manner, especially since the in-vivo situation predicts that both corrosion and wear may be a cause for concern, either acting separately along different time scales or together at any one instant. This is to say that the implant may undergo degradation by the corrosion process alone, by the mechanical wear process alone, and by consecutive or simultaneous synergism of the two processes [Dumbleton, 1978, Buchannan et al., 1987, Angelini et al., 1991, Hong and Pyun, 1991, Chakraborty et al., 1991, Hedeyat et al., 1992, Cook et al., 1994, M^{c}Donald and Bloebaum, 1995, Buchannan et al., 1983, Thomsen et al., 1995, Kumar et al., 1990].

Wearing has a deleterious effect on the corrosion resistance of such materials, because it hinders the formation of a continuous oxide layer and consequently the passivation of the surface [Angelini et al., 1991]. Removal of the surface in this way, can give rise to the effect of wear-accelerated corrosion [Buchannan et al., 1987, Brown, 1980]. However, if corrosion is known to be occurring and subsequently wear is then introduced in a more synergistic way as mentioned above, then, this synergism between erosion and corrosion can account for increased removal of material as part of the overall wearing of the implant [Chakraborty et al., 1991]. It has been shown that this co-joint action of corrosion and sliding wear is much more destructive than the action of either by themselves [Angelini et al., 1991, Hong and Pyun, 1991, Chakraborty et al., 1991, Hedeyat et al., 1992, Cook et al., 1994, Stott and Breakell, 1989, Meyer-Rodenbeck, 1992, Allen, 1981].

In order to relate to the in-vivo situation more closely, let us consider how the effects of corrosion and wear might influence the degradation of the hip endoprosthesis. While bone cement debris has been recognized as an important feature in aseptic loosening of a prosthetic component, the interest is now being focused on the impact of debris and wear-off particles from polyethylene and metal alloys. Since, an optimal lubrication cannot be generated yet, direct contact between the components produces wear particles, depending on the materials used. This strengthens the argument for using materials which are not only strong and corrosion resistant but also which reduce friction and wear. In this respect, it becomes necessary to look at properties such as hardness which is a measure of the resistance to abrasion of a material [Williams and Cunningham, 1979, Dupuis, 1996].

1.4 THE BIOLOGICAL CONSEQUENCES OF CORROSION AND WEAR OF TITANIUM AND TI-6AL-4V ALLOY

If we want to develop the next generation of implantable devices that will offer greater efficiency and reliability, it is necessary to demand high standards of biocompatibility from the materials we use. More recently, 'biocompatibility' has been defined as the exploitation by materials of the proteins and cells of the body to meet a specific performance goal [Ratner, 1993]. Thus, it is important to consider the consequences of using titanium and its alloys as biomaterials in terms of corrosion and wear.

Ti-6Al-4V alloy is a commonly used biomaterial and although it is known to have a superior corrosion resistance compared to other alternative materials such as Co-Cr based alloys and stainless steels, its biocompatibility has recently raised doubts. Even in the absence of corrosion, titanium, aluminium and vanadium ions are undergoing passive dissolution. It has been suggested that ions released from this alloy over long periods of time into the implant interface in contact with bone could interfere with the normal osteoid mineralization and remodelling processes of bone in that region. In turn, this would result in subsequent loosening of the implant [Blumenthal et al., 1989].

The toxicological effects of elevated concentrations of titanium in serum is unknown, but, it is often regarded as an inert biocompatible material. Epidemiological studies have found that inhalation of titanium-containing dust is relatively benign, but reports have suggested the association of titanium particles with pleural disease, granulomatous disease of the lung, and malignant tumours of the lung. In addition, certain titanium compounds have been demonstrated to have a suppressive effect on platelets [Jacobs et al., 1991]. Titanium ions which accumulate in local tissues after long periods of time, give rise to dark tissue staining with unknown consequences [Browne and Gregson, 1994]. These black-stained periprosthetic tissues showed macrophages and multi-nucleated giant cells that contained metal particles [Korovessis et al., 1994].

Aluminium is a transportable element, capable of reaching distant parts of the body, and is known to influence some forms of bone disease and neurological disorders (eg. Alzheimer's dementia). Vanadium is considered to be an essential element in the body, but may become toxic at excessive levels. It has been suggested that it interacts with intracellular components and inhibits certain enzyme functions [Williams, 1977, Browne and Gregson, 1994]. Fibroblasts exposed to vanadium have shown gross necrosis [Wapner 1991, Browne and Gregson, 1994, Black, 1988, Pilliar, 1991].

When the titanium alloy specimen together with its oxide film is placed in the serum, a variety of mechanisms occur, each of which will have an eventual outcome on the dissolution process. Further oxidation of the oxide film occurs involving metal atom diffusion from the metal oxide interface to the oxide bio-liquid interface. Concurrently, dissolution of the oxide is controlled by a number of factors:

a) the nature and strength of the oxide bond,
b) oxide thickness,
c) the defect structure of the oxide (vacancies, interstitial elements etc..),
d) degree of ordering within the oxide,
e) anisotropic behaviour due to preferred dissolution sites on specific crystallographic planes.

The oxide layer is highly polar and attracts water soluble molecules. Two types of hydroxyl groups are thought to exist on the titania surface. One OH group is bound to one Ti^{4+} site (terminal OH), the other to two such sites (bridged OH). Bridged OHs are strongly polarized by the cations and are expected to be acidic in nature and exchangeable with cations. As oxidation of the alloy takes place, the oxide thickens as a result of which the net potential difference across it decreases and the driving force is reduced. This may be a factor in the improved dissolution resistance exhibited by the thicker oxides. Other processes occurring include diffusion of mineral ions or atoms (eg. Ca and P) from the bioliquid into the oxide. Biomolecules may also adsorb onto the oxide, the degree of adsorption being dependent on various factors such as geometry, surface roughness and type of molecule. Some biomolecule desorption and exchange occurs at the surface until a stable layer is formed.

The titanium alloy implant spontaneously passivates to create an adherent titanium oxide surface film. This film is initially less ordered and this results in higher titanium ion mobility. Additionally, the alloy contains a small proportion of aluminium (6%), but aluminium oxide, Al_2O_3, has a highly negative free charge of formation, higher than that of titanium dioxide, TiO_2. Thus, there is a greater driving force for formation of Al_2O_3 and studies have confirmed its existence in the outer layers of the oxide which forms on Ti-6Al-4V. TiO_2 has a high dielectric constant, whereas, Al_2O_3 has a much lower dielectric constant and thus a lower isolating effect exists. A greater driving force for ion migration combined with a smaller more mobile aluminium ion may result in the proportionately higher release of aluminium from the substrate.

Rutile is more dense and has a closer packed structure with fewer paths for ion diffusion than anatase titanium dioxide. The more ordered and dense rutile structure will therefore have improved dissolution resistance, although the defect structure remains important. TiO_2 is a non-stoichiometric n-type semiconductor oxide with an effective formula of TiO_{2-x}. The predominating effects in this oxide are oxygen vacancies, the concentration of which will be influenced by impurity elements incorporated as a result of the interaction between the oxide surface and bioenvironment as well as other elements in the parent alloy. TiO_2 surfaces bind with cations, particularly polyvalent cations, as a result of the electrostatic interactions between titanium-linked oxygen ions on the implant surface and cations. Calcium ions are also attracted to the oxide covered surface by electrostatic interaction with negatively charged oxygen species [Browne and Gregson, 1994].

The dissolution process can be described by two phases. Initial adsorption of polar species and desorption of dissolution products resulting in a high initial release rate. This is followed by a build up of a hydrous boundary layer along with protein adsorption, and a consequent reduction in the transport for the dissolution product from the oxide substrate [Browne and Gregson, 1994].

Corrosion is a strong mediator of systemic effects. This is because a large proportion of corrosive products are soluble, and thus, pass through the circulatory systems of the body, to all remote sites. Corrosion proceeds through four principle mechanisms:-

1. Ionization that releases metallic ions directly.
2. Oxidation which releases metal oxides.
3. Hydroxylation which releases metallic hydroxides.
4. Reaction forming either complex inorganic or organometallic ions.

The fourth mechanism of corrosion is the one that concerns us most with respect to systemic effects. It is now well recognised that both specific and non-specific binding to serum proteins produce a wide range of potentially biologically-active organo-metallic compounds [Ling, 1984].

Low rates of corrosion have been observed in the absence of protein, which may be due to a protective oxide layer or there may be no mechanism by which the oxidation of the metal can be coupled to the reduction of oxygen. Certain metals are notable for their tendency to form stable complexes with potential ligands, such as proteins. Thus, a protein in solution could abstract already oxidized metal from a protective oxide layer, rendering the underlying material liable to attack. Alternatively, it is possible for the protein to act as a catalyst, oxidizing the metal with internal disulphide bonds subsequently reformed by oxidation with oxygen [Clark and Williams, 1982].

Wear processes may also act to increase corrosion rates. The mechanism is that of physical destruction of the passive film on the metallic surface. As the surface re-passivates, there is a competition between passivation and corrosion, resulting in elevated metal release. This is a form of the well recognised corrosive-wear. Disruption of the passive film may also lead to presentation of a softer material to the opposing polymer surface. This is clearly the case in the wear of Ti-6Al-4V alloy. Here the surface film is essentially titanium dioxide and is far harder and more abrasion resistant than the underlying unoxidized alloy. This is the apparent explanation for the severe wear rates seen for this alloy in-vitro and in-vivo [Ling, 1984].

The debris generated by the wear process initiates a severe macrophagic response in the surrounding tissues, which in time results in osteolysis, loss of normal body architecture, and implant failure [Nasser et al., 1990]. Total joint prostheses differ fundamentally from other types of implants because of the presence of wearing surfaces that lead to the production of small particles. As a result of particle formation, the exposed surfaces of material in contact with tissues are considerably increased. The smaller the particle size, the greater the surface per given weight, with faster solubility and greater local exposure [Escalas et al., 1976]. During in-vitro studies, it has been found that both polyethylene and Ti-6Al-4V alloy particles have the ability to induce the release of mediators (ie. tumour necrosis factor [TNF], interleukin-6 [IL-6], and prostaglandin E_2 [PGE_2]) known to be involved in bone resorption [Haynes et al., 1993, Rogers et al., 1994, Horowitz and Purdon, 1995]. Multiple reports have demonstrated that titanium alloy is not acceptable as an articulating surface because it has a poor resistance to abrasion. It can result in severe metallosis in the periarticulate tissue, leading to progressive osteolysis and early failure of the arthroplasty [Evans et al., 1993]. Black-stained periprosthetic tissues showed macrophages and multinucleated giant cells that contained metal particles [Korovessis et al., 1994, Breen and Stoker, 1993].

During a recent study, it has been shown that as well as the macrophage reaction to titanium debris, there is also a very large T-lymphocyte response. This would not be expected if the loosening of the prosthesis were due to a simple mechanical failure or to a giant cell reaction to polyethylene wear debris. The presence of T-lymphocytes and the absence of accompanying B-lymphocytes or plasma cells suggests that an immunological reaction is occurring in tissues adjacent to the prosthetic joint. The complete absence from the tissues of aluminium and vanadium, both present in the titanium alloy requires explanation [Lalor et al., 1991]. Other studies have demonstrated that the levels of metals detected in tissues adjacent to implants are not in the same proportions as those in the parent alloy [Vernon-Roberts and Freeman 1976], due perhaps to their different solubilities. Animal studies have shown that vanadium is very soluble and is cleared from the circulation through the kidneys very quickly while titanium is insoluble and remains in the tissues. Aluminium is more soluble than titanium and may be transported away from the site [Brown et al., 1989].

To summarize, it seems that the corrosion and wear of titanium and its alloys are of great concern when considering the biological consequences of particulate matter and ions present in the body both near the site of the implant and far away from it.

1.5 IN-VITRO CORROSION AND WEAR TESTING

In-vitro testing has made it possible to evaluate test materials within well defined, but slightly varying environments, in order to establish a better scientific understanding of the passivation, re-passivation and breakdown characteristics of the alloys during accelerated corrosion and wear. Using such methods, it was possible to study the effect of protein adsorption and the presence of proteins as biological macromolecules interacting within the corrosive environment.

Laboratory based corrosion testing is useful in:-

(i) Evaluating and selecting materials for a specific environment or a given definite application.

(ii) The evaluation of new or old materials or alloys to determine the environments in which they are suitable. The information obtained aids in the selection of materials to be tested for a specific application. This category could also include the effects of changes in the environment, such as additions of inhibitors or de-aeration on the corrosion of metals and alloys.

(iii) Control of the corrosion resistance of the material or corrosiveness of the environment. These are usually routine tests to check the quality of the material [Fontana, 1986, Holland, 1992].

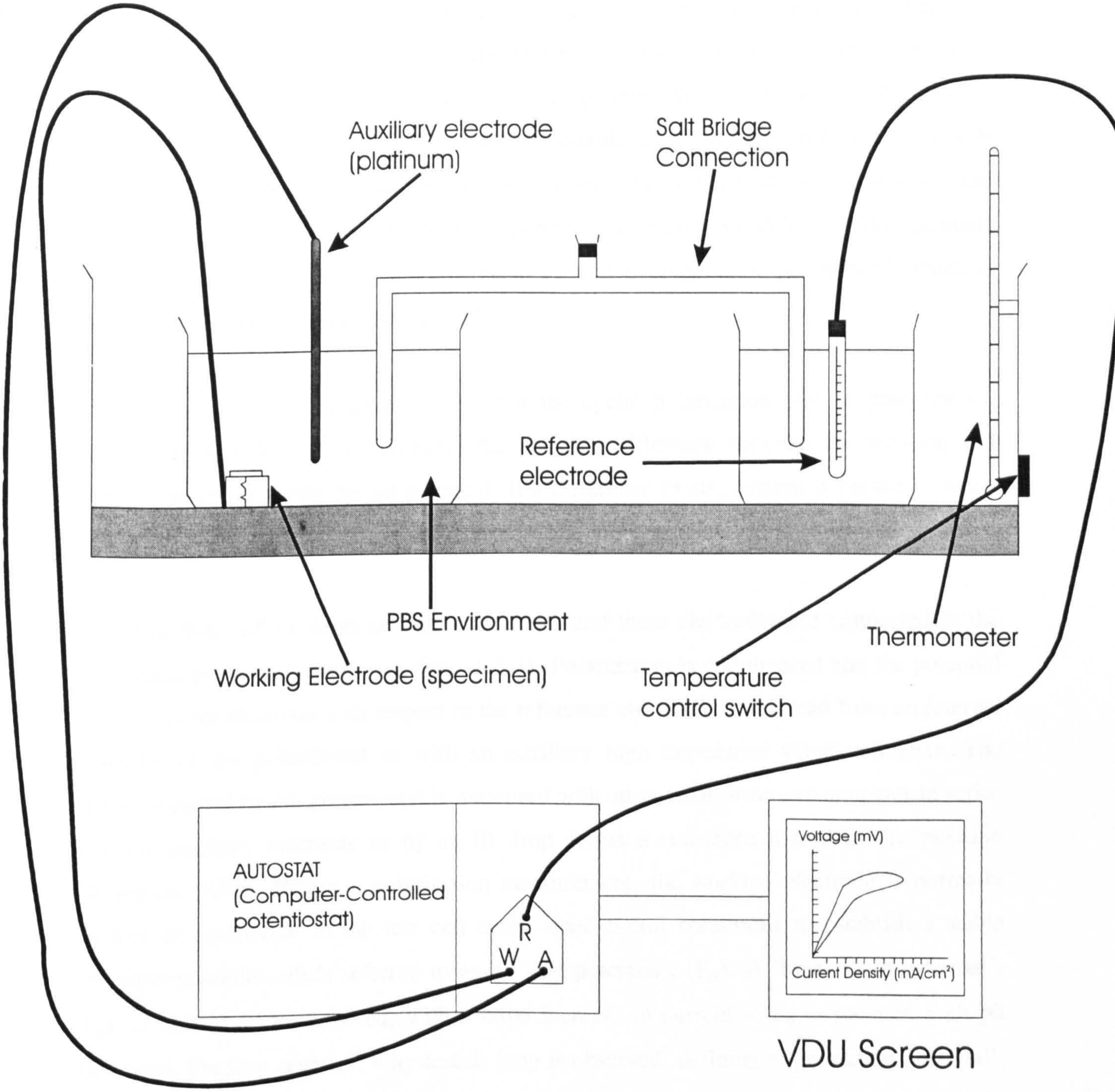
As far as the testing of materials is concerned, flat samples are usually preferred because of easier handling and surface preparation. Also, a small sheet-metal specimen is easy to weigh and to measure. However, if pitting corrosion is expected, a small specimen can have a low probability of developing a single pit. Furthermore, specimen size should be related to the volume of electrolyte (A minimum of 250 ml of testing solution for each 6.3 cm^2 of specimen area has been suggested [Shrier, 1976]). If the ratio is too large, then the electrolyte may become quickly exhausted due to the build up of corrosion products in the electrolyte. These products can either accelerate or inhibit corrosion. A large number of specimens is better than one large one, since more replicates can give a more reliable measure of corrosion.

The weight loss or gain of the specimens may also be used as a valid measure of the corrosion products attached to them following corrosion. However, to measure the weight of the metal, it is necessary to remove the corrosion products first. This is not easy without removing the material as well. A corrosion test can reveal that a metal is particularly sensitive to a particular ion. However, it cannot really be used to predict the behaviour of the service environment [Ailor, 1971]. It is often known for weight gains to take place following corrosion, rather than weight loss, as a consequence of corrosion products (such as metal hydroxides and metal-protein complexes) which form at the surface, having a compounding effect on the weight measurement of the specimen [Fontana, 1986, Ailor, 1971, Cohen, 1962].

Most electrochemical corrosion test methods are based on the recording of polarization curves, preferably determined by potentiostatic or potentiodynamic techniques. The application of an anodic current or an anodic potential is used to intensify the corrosive environment. Though, galvanic methods, based on the application of a constant anodic current, have been used extensively, potentiostatic methods (ie. the application of a constant anodic potential), are becoming more important because they are sounder in principle [Wranglen, 1985].

In a potentiostatic step method, the potential is changed in arbitrary fixed increments at fixed time intervals with the current being recorded just prior to the potential change. Potential steps between 5 and 100 mV are taken in 0.5 to 5 minute (or longer) intervals. In the potentiodynamic technique, the potential is continuously varied at some fixed rate whilst simultaneously recording the current flow. Unless the rate of potential change or scan is infinitely slow, polarization curves cannot accurately reflect the electrode kinetics. Potentiodynamic polarization curves, whilst being less precise than potentiostatic curves, are produced more rapidly and are more useful for comparative purposes [Wranglen, 1985]. Potentiodynamic polarization is a widely used in-vitro method for general corrosion testing. Although, knowledge about corrosion has increased considerably over the years, it is still necessary to actually test new alloys and known alloys if they are to be used in "new" surroundings. Such polarization studies are usually conducted using a ‘potentiostat’ set up alongside the electrochemical cell and computer controlled visual display unit (Fig. 1.5.1).

Figure 1.5.1 : Schematic layout of laboratory corrosion testing apparatus



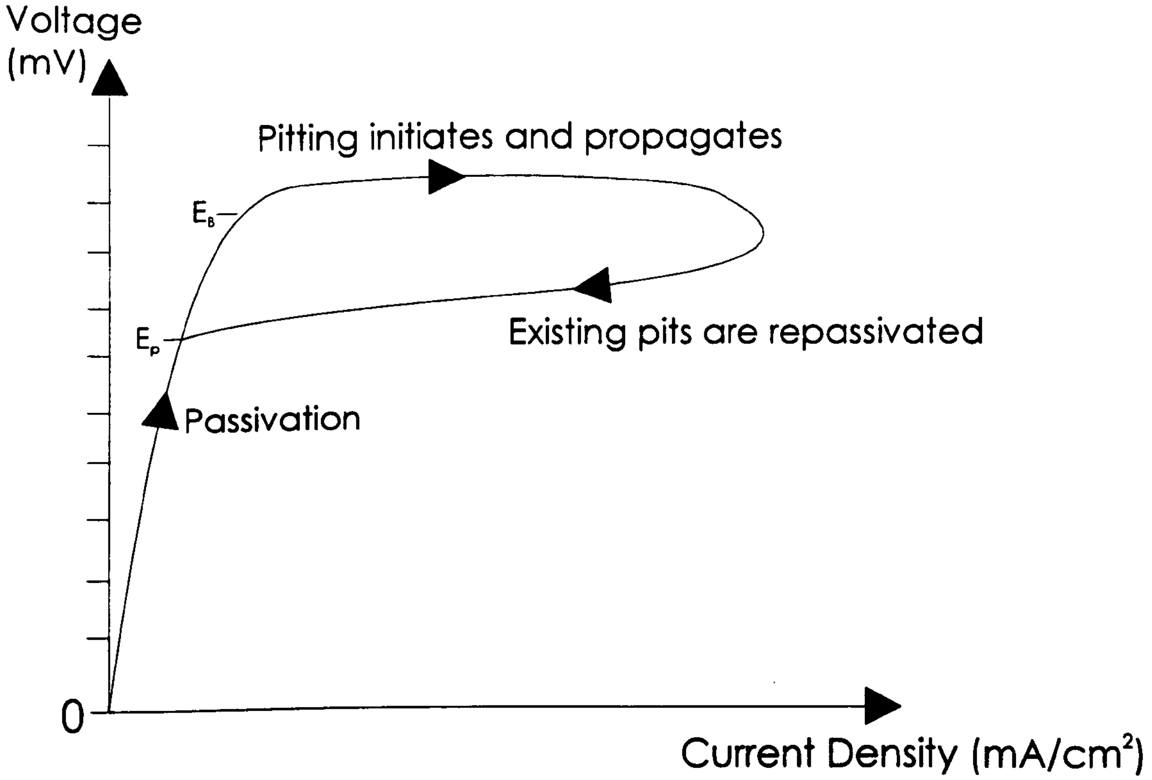
A metal establishes a potential with respect to a solution which cannot be measured in absolute terms. The potential difference between it and another electrode can be measured. Furthermore, changes in potential difference can be related to the metal electrode under investigation, if the other electrode does not change (ie. is a reference electrode). There are several reference electrodes which are constant provided no current is drawn from them. Therefore, they are used with millivoltmeters of high impedance. A metal in contact with a solution containing the metal ions at unit activity establishes a fixed potential difference with respect to every other metal in the same condition. The full list of these comprises the electrochemical series of normal electrode potentials of metals in which all the potentials listed are actually differences between the metal and hydrogen, the potential of which is arbitrarily designated as zero [Scully, 1966].

The potentiostat control loop obtained from the cyclic polarization plot of potential vs. current density (Fig. 1.5.2) compares the potential difference between the working and reference electrodes with the set potential. If a difference exists, current is passed between the working and auxiliary electrodes to restore the W.E. potential to the required value.

The electrolyte cell contains all three electrodes and these electrodes are connected to the appropriate potentiostat terminals (Fig. 1.5.1). Polarization is commenced and the potential of the working electrode with respect to the reference electrode can be read from an internal voltmeter of the potentiostat or with an auxiliary high impedance valve voltmeter. The current supplied by the potentiostat is measured with an internal meter, an ammeter in series with the auxiliary electrode or by an IR drop across a resistance placed in this position [Wranglen, 1985]. Before a polarization measurement, the working electrode is normally allowed to equilibrate in the test cell under open-circuit conditions to establish a stable corrosion potential, often referred to as its 'rest potential', (E_R). A 'breakdown potential', (E_B), is known to be observed, with a large increase in current at E_B, associated with pit formation. On scan reversal, a hysteresis loop is observed, defining a 'protection potential', (E_P). E_P is the potential at which the current density on the negative sweep from the maximum anodic potential equals the current density in the passive region (ie. the point at which the hysteresis loop is complete, or, rest potential is reached) [Cogan et al., 1994]. The small area within the hysteresis loop and the small difference between the breakdown and protection potentials (E_B - E_P) are good qualitative indications of resistance to pitting and crevice corrosion [Ling, 1984, Cahoon et al., 1975, Wilde, 1972].

It has been shown that the protection potential (E_P) is not a unique parameter. It is known to vary in a semi-logarithmic way with the amount of localized attack induced by polarization; that is the chemistry changes in the growing pit caused by the hydrolysis of corrosion products and chloride ion build up in the cavities. On this basis, it may be concluded that E_P data (however obtained) cannot be used alone as criteria for protection against the propagation of pre-existing pits or crevices. Based on cyclic polarization procedures, a method of assessing the susceptibility and resistance of an alloy to pitting or crevice corrosion is suggested. It was shown that, for alloys which initiate pits during anodic polarization, the magnitude of the 'difference potential' (E_B-E_P) is related to the crevice corrosion weight loss. For alloys that have no critical protection potential, the presence of hysteresis during cyclic polarization with an artificial crevice indicates susceptibility to crevice corrosion [Wilde, 1972].

Figure 1.5.2 : A typical cyclic polarization plot of potential vs. current density



It has been known for considerable differences to be obtained in the value for E_B, for the same material. This is partly due to variations in surface preparation, for instance, it has been found for higher values of E_B to be associated with specimens following electropolishing, for example [Sutow et al., 1976]. This suggests that variations in surface roughness may cause considerable differences in values of E_B, and thus, should be monitored very carefully during surface preparation. As far as rest potential (E_R) is concerned, it has been found for this value to be dependent on the thickness of the passive film, displaying more positive values with increasing film thickness for the same material. Effective area can be thought of as a measure of the number of active anodic areas. Thus, E_R becomes more positive as the film becomes thicker and the number of anodic areas decreases [Solar et al., 1979].

To summarize, the potentiodynamic polarization technique for corrosion studies has certain advantages:-

(i) When set up, the method is fast and easy to run.

(ii) Measurements are reproducible and can be made with adequate accuracy and precision.

(iii) Several different parameters (anodic charge, rest potential, breakdown potential, passivity range, pitting corrosion) can be derived from the same run.

Some of the limitations of such a technique are:-

(i) They are only short time tests (ie. accelerated) which means a drawback in correlation with the long-term effect in-vivo.

(ii) Cannot be used to find the reduction in strength due to a certain type of corrosion.

(iii) The results of potentiodynamic polarization testing depend entirely on the oxidation and reduction reactions taking place during the corrosion, propagating a transport of charged particles (ions and electrons) available for measurement. If the corrosion reaction involves a release of non-charged particles, these will escape measurement, and the result may thus be misleading in assessing the corrosion sensitivity and quantification of released substances [Holland, 1991]. In addition to the above limitations, generally, predicting the corrosion behaviour of active-passive alloys can be very difficult, because there are several mechanisms by which the breakdown of passivity might occur. These include penetration of the passive film by aggresive anions (eg. Cl^-), anion adsorption leading to an increase in the dissolution rate of the passive layer, mechanical breakdown of the passive film due to electrostatic forces, a reduced dissolution rate due to the interaction of proteins inhibiting corrosion by binding onto sites prone to anodic dissolution and an increased dissolution rate due to the same organic species causing metal cations to be dissolved more readily into solution, to form protein/metal complexes at the metal surface causing increased preferential pitting corrosion.

Wear measurement is usually more significant when considering the long-term performance of joint prostheses and the effects of accumulation of wear debris in tissues. Wear measurement may also be required for any new material combination introduced for a joint prosthesis. Study of the material usually takes place in the laboratory and the wear behaviour is normally evaluated over a variety of conditions. Material tests may be carried out on simple test machines but the penalty is that there may be difficulty in extrapolating the results to the end-use. Testing of a device is more costly than the testing of a material, since evaluation must be done on a joint simulator. Laboratory wear testing machines are normally employed to screen materials, although, attempts have been made to predict wear rates of devices in-vivo from such simple in-vitro tests. The advantages of such machines are that they are simple, cost-efficient and can be operated under a constant load. On the other hand, the disadvantages are that simplicity usually means that only some aspects of the in-vivo situation are reproduced. Cost increases rapidly if more faithful reproduction of conditions are required [Mishra and Davidson, 1992, Semlitsch et al., 1992, Dumbleton, 1978, Angelini et al., 1991, Chakraborty et al., 1991, Hedeyat et al., 1992, Cook et al., 1994, Galante and Rostoker, 1973, Stott and Breakell, 1989, Lilley et al., 1992, Poggie et al., 1994, Albert et al., 1994]. To evaluate the effect of wear on the specimen, it may be possible to either undergo measurements of weight change to evaluate the volume loss of material, or, measure the change in surface roughness due to the wear scars produced [Lilley et al., 1992].

1.6 THE EFFECT OF A CHANGE IN pH AND THE PRESENCE OF PROTEINS ON THE CORROSION AND WEAR BEHAVIOUR OF TITANIUM ALLOYS

Before addressing exactly how the corrosion and wear behaviour of titanium alloys varies within well defined, but varying biological environments, it may be necessary to establish some basic principles of how the corrosion process occurs in terms of the physical chemistry involved:-

Corrosion is an electrochemical process involving the movement of electrons and positive ions.

If metal ‘M’ is placed in an aqueous electrolyte, metal ions will tend to go into solution, where they have a lower energy. There are always two reactions that occur, the anodic reaction which involves the oxidation of the metal to its salt, whereby the loss of ions results in an excess of free electrons in the metal:

$$M \rightarrow M(n+) + n(e^-) \qquad ....(1.1)$$

and the cathodic reaction, in which the electrons so generated are consumed. The precise cathodic reaction to occur will depend on the nature of the electrolyte, but two of the most important reactions in aqueous environments are the reduction of hydrogen:

$$2H^+ + 2e^- \rightarrow H_2 \qquad ....(1.2)$$

and the reduction of dissolved oxygen:

$$O_2 + 4H^+ + 4e^- \rightarrow 2H_2O \qquad ....(1.3)$$

in acidic solutions or

$$O_2 + 2H_2O + 4e^- \rightarrow 4OH^- \qquad ....(1.4)$$

in neutral or basic solutions.

Thus, if both the anodic and one of the cathodic reactions takes place, then corrosion occurs, and metal and hydroxyl ions react to give a metal hydroxide as the corrosion product in this case. Ideally, alloys can be chosen such that in a particular environment the cathodic reaction does not take place. However, even with corrosion-resistant alloys the equilibrium can be upset and reactions take place. One such situation is when crevices arise on the surface such as a ‘pit’.

Within a crevice, the amount of oxygen dissolved in the fluid will slowly be depleted, but a time is reached when the cathodic reaction can no longer take place in a pit because of the lack of oxygen. However, there is still sufficient oxygen over the remainder of the surface to allow the reaction to be sustained, such that a small cell of anodic area in the pit and a cathodic area surrounding it, is established. Further dissolution of the metal within the anodic area (ie. within the pit) can therefore occur as a self-perpetuating reaction [Williams D.F., 1979]. In all corrosion processes the rate of the anodic or oxidation reaction must equal the cathodic or reduction reaction. This is a basic principle of metallic corrosion. It also explains how variations in the local environment can influence the overall rate of corrosion by influencing either the anodic or cathodic reactions; the whole corrosion process can be arrested by preventing either of these reactions.

From a thermodynamic point of view, let us first consider the anodic dissolution of a pure metal in a solution of its salt as shown above by the anodic reaction. The metal consists of positive ions closely surrounded by free electrons.

When the metal is placed in a solution there will be a net dissolution of metal ions, since the Gibbs free energy (δG) for the dissolution reaction is less than for the replacement reaction. This leaves the metal with a net negative charge, thus, making it harder for the positive ions to leave the surface, increasing the δG for the dissolution reaction. There will come a point when the δG for the dissolution reaction will equal the δG for the replacement reaction. At this point a dynamic equilibrium is reached and a potential difference will be set up across the charged double layer between the metal and the electrolyte. This potential difference is characteristic of the metal and can be measured against a standard reference electrode. When this potential is measured against a standard hydrogen electrode in a 1N solution of its salt at 25°C, it is defined as the 'standard electrode potential' for that metal (Figure 1.6.1) [Fraker and Griffith, 1985, Marek, 1990].

The position of a metal in the electrochemical series primarily indicates the order with which metals displace each other from compounds, but it also gives a general guide to reactivity in aqueous solutions. Those at the top are the noble, relatively unreactive metals, whereas, those at the bottom are more reactive.

Figure 1.6.1 : The Electrochemical Series

METAL	POTENTIAL (V)
gold	1.43
platinum	1.20
mercury	0.80
silver	0.79
copper	0.34
hydrogen	0.00
lead	-0.13
tin	-0.14
molybdenum	-0.20
nickel	-0.25
cobalt	-0.28
cadmium	-0.40
iron	-0.44
chromium	-0.73
zinc	-0.76
aluminium	-1.33
titanium	-1.63
magnesium	-2.03
sodium	-2.71
lithium	-3.05

In general, the greater the reactivity, the more ions will go into solution before equilibrium is reached, and thus, the greater will be the negative electrode potential. Subsequently, the more reactive metals will have increasing negative values and the more noble metals will have increasing positive values [Williams D.F., 1979].

However, this series is of limited value in measuring the susceptibility of metals to corrosion, since some metals are protected by an oxide layer (eg. titanium) and are far less susceptible to corrosion than the series predicts. Also, because now further mechanisms would be required to upset the environment, which would be unrelated to the intrinsic reactivity of the metal. Oxide films are present on the surface of such metals as either coherent or incoherent (uneven porous layer) layers. Coherent films are known to act as a barrier to further transport of oxygen and metal ions. Not only will this oxide film prevent further oxidation, but may also protect the underlying metal from attack in corrosive environments. Such a metal is said to be "passivated". There are other reasons why the above guide to the corrosion resistance of metals may be considered over-simplified. In the situation described, a homogeneous pure metal existing within a changing environment, in which an equilibrium is reached with no further net movement of ions, seems rather unrealistic. In other words, the corrosion process takes place only transiently, but is effectively stopped once equilibrium is reached.

In reality, we usually have neither entirely homogeneous surfaces or solutions, nor complete isolation of the metal from other parts of the environment, and this equilibrium is easily upset. If the conditions are such that the equilibrium is displaced, the metal is said to be polarized and there are several ways in which this can happen. Two main factors control the behaviour of metals in this respect and determine the extent of corrosion in practice. The first is concerned with the driving force for continued corrosion (that is, the reasons why the equilibrium is upset and the nature of the polarization) and the second concerns the ability of the metal to respond to this driving force. Therefore, it is self-evident that if either the accumulating positive metal ions in the surrounding media or the accumulating electrons in the metal are removed, the net balance between the dissolution and replacement of the ions will be disturbed; The equilibrium is established precisely because of the imbalance of charge, so that if the latter is disturbed, so must the former. The result will be continued dissolution as the system attempts to achieve this equilibrium, in other words, sustained corrosion. An electron sink in contact with the metal or a dynamic medium will both achieve this [Fraker and Griffiths, 1985, Marek, 1990].

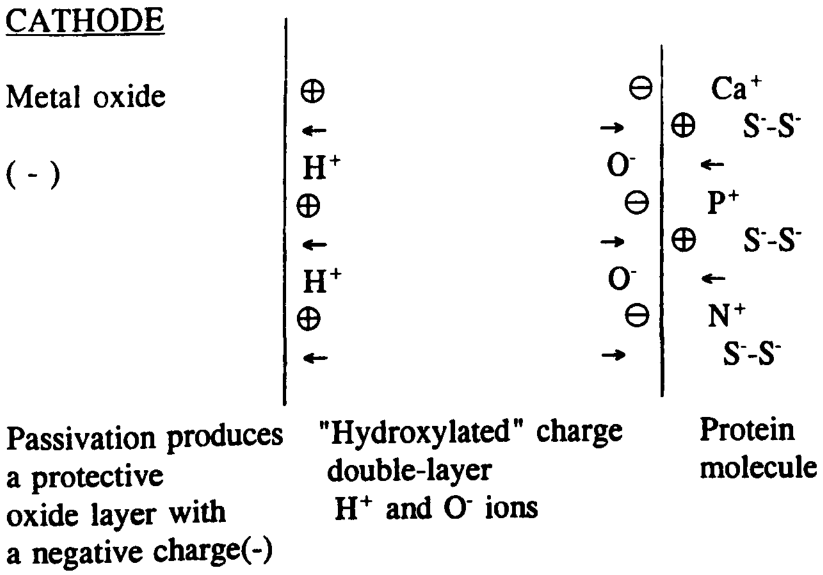
The normal pH of the body fluids is about neutral in the range 7.2-7.4. However, when an implant is inserted, the pH can shift to acidic values to about 5.2 at the injured sites and returns to normal in a few days. Thus, it becomes important to investigate slight variations in pH when undertaking controlled studies involving PBS solutions [Ban et al., 1991]. Previous studies have shown that when the pH was lowered, during polarization studies on Ti-6Al-4V, a slight increase in the corrosion current was noted, but with no change in the breakdown potential [Speck et al., 1980]. It has also been found that with pH levels greater than 5, the potential was seen to increase with time, indicating a decrease in the number of anodic sites. This decrease was explained as an overall thickening of the passive film, or to the healing of defects in the film. The decreasing of rest potential with time, observed when pH was less than 5, was explained as an increase in the active areas, probably due to some thinning or remodelling of the passive film [Solar et al., 1979].

Let us now consider how the presence of proteins within the corrosive environment might affect the corrosion behaviour of metals. It is well known for such biological macromolecules to affect the corrosion characteristics of different materials, especially metals, and it is known for their presence to, either, inhibit or accelerate corrosion. In this way, they are known to behave differently with one metal to the next. Their role in a corrosive environment is governed by many factors, such as the surface chemistry of the metal, protein adsorption characteristics, interaction of protein molecules with other ions present in the electrolyte solution to produce organic complexes, and the transport of anionic and cationic charges around and away from the local environment.

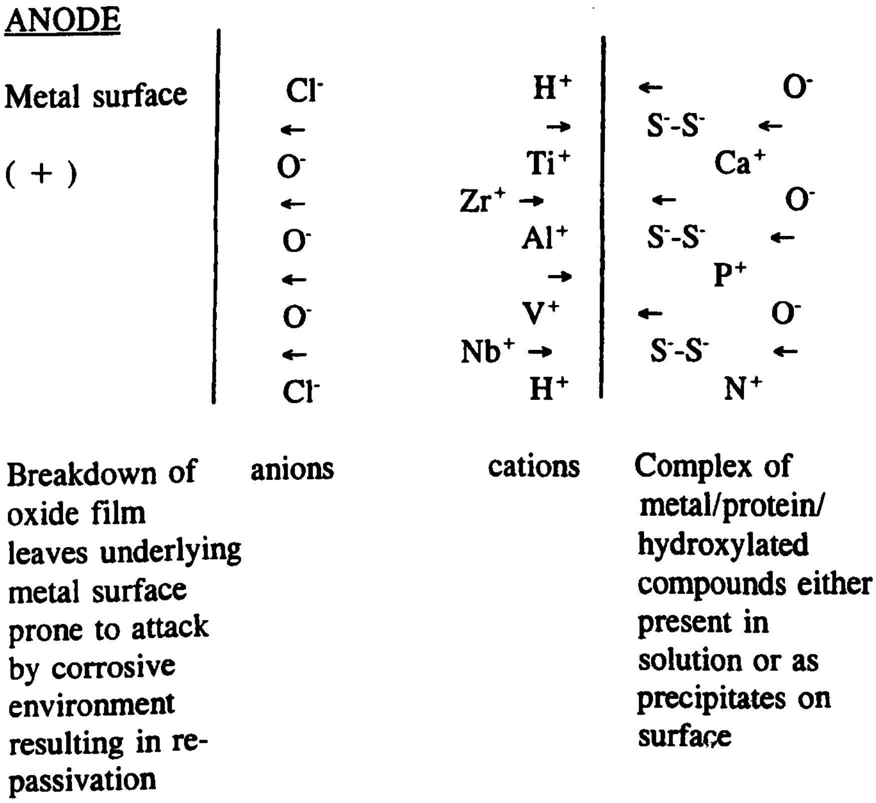
Before going into further detail of precisely how these interactions occur, it may be necessary to understand more comprehensively, how corrosion takes place in terms of charge transfer at the metal/electrolyte interface. Let us consider an accepted theoretical model for the charge distribution in terms of a charge double-layer structure (Fig. 1.6.2).

Figure 1.6.2 : Theoretical model of charge double-layer surrounding metal

Before corrosion



During corrosion



In this case, a double-layer is present due to hydroxylated ions at the metal surface and charge transfer can occur within the layer depending on whether the metal oxide is stable acting as a cathode or whether it is passivating upto and beyond removal of its protective oxide layer (ie. as an anode). In the absence of proteins, when a metal oxide surface breaks down (at breakdown potential, E_B), re-passivation begins to occur spontaneously, and as a result of this the underlying metal, which acts as an anode, will simply attract anions in the form of OH^- and Cl^- from the environment and due to re-passivation, it will eventually reach what is known as its protection potential (E_P). Thus, a material's ability to re-passivate quickly with respect to its breakdown is known to be a measure of its pitting corrosion resistance.

Now, let us consider the more complicated situation in which biological macromolecules such as proteins are present in the environment. Although, the presence of these organic species will not cause a completely new corrosion mechanism, they can influence the rate of corrosion by interfering in some way with the anodic and cathodic charge transfer. In doing so, they may accelerate the corrosion behaviour of some materials or may inhibit it for others.

There are six ways in which this might occur [Clark and Williams, 1982, Browne and Gregson, 1994, Williams D.F., 1986, Ban et al., 1991, Williams R.L., 1986, Cogan et al., 1994] :-

1. The biological molecules could upset the equilibrium of corrosion reactions by consuming one or other of the products of the anodic or cathodic reactions. This would upset the charge double-layer and allow further dissolution of the metal to occur.

2. The stability of the oxide layer depends on the electrode potential and the pH of the solution. Proteins often participate in electron-carrying roles and thus affect the electrode potential, and bacteria can alter the pH of the local environment through the generation of acidic metabolic products (It is well known that when an implant is inserted, the pH shifts to acidic values of about 5.2 at the injured sites and then returns to normal after a few days).

3. The stability of the oxide layer also depends upon the availability of oxygen. Adsorption of proteins onto the surface of materials could limit the diffusion of oxygen to certain regions of the surface. This could cause preferential corrosion of the oxygen deficient regions and lead to the breakdown of the passive layer.

4. Adsorbed protein could act as a catalyst, oxidizing the metal with internal disulphate groups, and form a metal-protein complex with disulphide bonds being subsequently reformed by oxidation with oxygen. Repeated activity of this nature is most likely to lead to cross-linking and precipitation of the protein on the surface.

5. Protein adsorption onto the surface may result in the formation of a protein layer that inhibits corrosion by binding to sites prone to anodic dissolution resulting in a reduced dissolution rate.

6. The cathodic reaction often results in the formation of hydrogen, and the build up of hydrogen tends to inhibit the cathodic reaction and restrict the corrosion process. However, biological species in the vicinity of the implant, in particular bacteria, are capable of using up hydrogen and thus allowing the corrosion to proceed.

It has been shown that the presence of wear during corrosion acts to speed up the re-passivation of the alloys accelerating the corrosion behaviour, as mentioned earlier. When considering the action of wear in an environment containing protein molecules it is important to take into consideration the lubricating effect they have on the wear behaviour [Kumar et al., 1990]. Furthermore, in the presence of corrosion, it might be possible to presume that if such protein molecules are attracted due to a larger driving force than that present during simple adsorption, then it should follow that such materials will also be more effectively protected by the presence of protein molecules, and thus, produce less severe wear scars due to enhanced lubrication.

1.7 GENERAL AIMS OF THE STUDY

The overall aim of the study was to investigate the co-joint action of corrosion and wear of titanium alloys in simulated physiological environments.

This involved a range of materials:-

1. Ti (cpi): commercially pure grade metal.
2. Ti-8Al-1Mo-1V: Near alpha phase alloy.
3. Ti-3Al-2.5V: Alpha-beta mixed phase alloy.
4. Ti-6Al-4V: Alpha-beta mixed phase alloy.
5. Ti-6Al-7Nb: Alpha-beta mixed phase alloy.
6. Ti-13Nb-13Zr: Near beta phase alloy.
7. Ti-15Mo: Beta phase alloy.

The materials test matrix was intended to represent a wide range of alloys with slightly varying compositions, but also taking into account a slight change in phase structure. Following an initial study concerning the corrosion and wear behaviour of all the above alloys in a simple physiological solution (ie. phosphate buffered saline), three alloys were selected focusing on those titanium alloys used within the *in-vivo* situation. The intention was to thoroughly investigate the behaviour of a small group of alloys within several biologically relevant environments. It was decided to focus on Ti-6Al-4V alloy, since this was an alloy presently being used, primarily, in orthopaedic applications where the study of corrosion and wear would be of paramount importance. Also, since this alloy was known to be causing concern to the scientific community in respect of its potential degradation consequently leading to problems of particulate debris and toxicity within the tissue.

Since two alternative alloys (ie. Ti-6Al-7Nb and Ti-13Nb-13Zr) had already been proposed as alternatives to the Ti-6Al-4V alloy, these were chosen for comparison with Ti-6Al-4V alloy. The three alloys selected were very similar to each other in terms of the phase structure. All three contained an alpha stabilizing element as well as one other beta stabilizing element, thus, the alloys were made up of a mixed phase structure. Also, research on relevant literature indicated that very little data was presently available on these materials, none of which had tested all three alloys within the same test matrix.

The variety of environments used within the project to test these alloys were:-
(i) The inorganic electrolyte, phosphate buffered saline (PBS) solution - at three different pH levels.
(ii) A simple protein, namely albumin in the PBS solution - at different concentrations of the protein and three pH levels.
(iii) A complex protein, namely foetal calf serum (10%), in the PBS solution.
(iv) Human macrophage cells in an environment containing PBS and foetal calf serum (10%).

Thus, the structure of the thesis involves an account of the behaviour of titanium alloys within progressively complicated biological environments. A summary of the work undertaken and its aims will be discussed in the next few paragraphs.

Chapter 2 involves a study of all six alloys listed earlier, including all of the constituent metals. The aim of this work was to obtain some general scientific data on the corrosion behaviour of a diverse range of titanium alloys, allowing a direct comparison to take place, within a relatively simple biological solution (ie.PBS). The results obtained emphasized the use of specific alloying elements, the use of certain phase structures of titanium alloys, and how these two parameters directly influenced the corresponding corrosion behaviour of the alloys. The pH of the solution was also altered slightly to examine the physical nature of each material in terms of exposure to slightly more acidic or alkali conditions. Surface roughness and hardness measurement were also undertaken to evaluate the mechanical integrity of the surface of each material in response to corrosion.

Following the initial corrosion studies as discussed above, chapter 3 involves work which investigated the co-joint action of corrosion and wear of all six alloys as well as pure titanium metal. The aim of this work was to provide a set of data concerning the wear behaviour of the alloys as well, which might relate more closely to the breakdown of these materials in the *in-vivo* situation. In this respect, two types of wear degradation were investigated. Firstly, the influence of wear to the surface prior to any ensuing corrosion, and secondly, a study of both wear and corrosion occurring simultaneously within the same environment. The wearing apparatus used was of a simple design, in order to reproduce a set of comparative data as opposed to a complicated simulation of, for example, a hip joint replacement. Thus, the set of data produced allowed for a direct comparison to take place of the corrosion and wear characteristics of all the alloys tested. Therefore, the results were intended to provide a set of data of significant scientific importance as opposed to a promotional comparison of a group of commercially available materials.

The work in the second and third chapter, then, provided a foundation of scientific experiments and data allowing an opportunity to exist for further study using the testing environment set up, so far. Further work, following on from this would provide an in-depth understanding of how certain alloys might react to slight changes within the surrounding environment. This allowed the possibility of studying how, by progressively complicating the biological environment, a theoretical explanation might be proposed, to describe how the changing parameters within the environment directly influence the behaviour of the alloys.

Chapter 4 concerns the corrosion behaviour of the three alloys (ie. Ti-6Al-4V, Ti-6Al-7Nb and Ti-13Nb-13Zr) within a PBS environment containing a single protein (ie. Albumin) followed by an environment that contains all the serum proteins (ie. foetal calf serum (10%)). Thus, the effect of proteins on corrosion was investigated, realizing that previous work by others on proteins, only involved metals or just Ti-6Al-4V alloy. Any results obtained during the study would provide a new set of data and allow a further understanding of proteins influence the corrosion behaviour of these alloys. The majority of the work that has already been done regarding the exact role of proteins within the corrosion environment still remains controversial. It seems that the role of proteins on corrosion depends on numerous factors including protein surface charge, metal electrode surface charge, metal surface chemistry, corrosion reactions between organic and non-organic species within the environment, and the transfer of charge across the metal/oxide/protein/biological environment interface.

Chapter 5 continues to use the more complicated environment of albumin and foetal calf serum additions to PBS. Only this time, just as the work in chapter 3 followed on from chapter 2, once again the influence of wear prior to corrosion and corrosive wear, was investigated. Previous studies suggested that proteins were known to either accelerate or inhibit corrosion behaviour depending on the protein itself and the material being tested. Proteins can also influence wear by having a lubricating effect. This study, investigated the co-joint action of corrosion and wear taking the lubrication of proteins into consideration.

Wear can cause a change in surface roughness which is also known to influence corrosion. Thus, the work that followed in chapter 6 contained a study of the rest potential (E_R) of the alloys at varying surface roughness produced by changes in initial polishing procedures. It has often been suggested in previous work that when endoprostheses have been examined following failure, the surface roughness of the metal surfaces were very coarse. An experimental study was initiated looking at variations of surface roughness. Only this time, as well as undertaking corrosion studies, it was proposed that measurements of rest potential (E_R) and breakdown potential (E_B) also be noted for differences. The times taken for rest potential to be reached was also undertaken. It was hoped that this might provide additional information relating to the stability of the alloys under varying conditions of preparation and failure. For instance, if it was found that a particular alloy became very unstable due to a coarse surface finish, which during a smooth finish seemed quite stable, then, this might have possible implications in accelerating corrosion following abrasion of the surface or poor handling during surgery. The changes in breakdown potential would also provide useful information on the stability of the passive oxide film on the alloys leading to eventual breakdown and failure, especially when looking at slight changes in the environment (ie. when testing in PBS and then in PBS + serum).

Finally, the work in chapter 7 involves the addition of human macrophage cells to the environment, since these cells are known to play a major role during the inflammatory response to foreign materials, and a lot of work has already been done on their activation as a measure of the biocompatibility of materials. They have been found to be present in the surrounding tissue of metal implants, but their exact role with metal ions during corrosion and wear is still not known. This study was aimed at providing evidence for the expression of cytokine enzymes which are known to be released during activation of the cells. This should also indicate the relative toxicity of any particular alloy with respect to the metal ions it releases both during passivation and corrosion.

CHAPTER 2 : IN-VITRO CORROSION STUDIES OF SEVERAL TITANIUM ALLOYS IN PHOSPHATE BUFFERED SALINE (PBS) SOLUTION

2.1 INTRODUCTION

Titanium alloys are suitable for implants because they combine good mechanical properties with a relatively inert behaviour, in both, static and dynamic conditions. Unfortunately, metal corrosion results in toxic corrosion products which once released from the implant to the surrounding tissues, may give rise to biocompatibility problems [Williams, 1987]. This fact could become a limiting factor when considering applications with, for instance, increased lengths of service. Therefore, for long-term implants, new metallic implants with improved corrosion resistance are sought. The need for appropriated metallic materials for long-term implants is nowadays the driving force for the development of alternative materials with improved corrosion resistance in body fluids [Escudero and Gonzalez-Carrasco, 1994].

The biocompatibility of metallic implant materials is closely related to their corrosion behaviour. Corrosion and surface film dissolution are two mechanisms for introducing additional ions into the human body. It has been concluded that the biocompatibility is determined not only by corrosion products, but also by exchange currents and reaction products of different redox processes involving tissue compounds [Kovacs, 1992]. Therefore, electrochemical methods are generally included when testing metallic biomaterials in synthetic environments. It has been shown that the rating of metallic implant materials based on *in-vitro* electrochemical measurements of current densities is in good agreement with their biocompatibility rating from *in-vivo* experiences [Pan et al., 1994]. The potentiodynamic polarization corrosion test supplies detailed information which indicates the passive range and sensitivity to pitting corrosion as discussed earlier in section 1.5. An analysis of the variance of the results shows the possible differences between the alloys tested [Holland, 1991].

During this study, cyclic polarization curves were analysed to compare and evaluate the 'pitting corrosion resistance' of several titanium alloys varying in composition and phase structure. The materials used in this study were obtained from various sources. Ti(cpi), Ti-6Al-4V, Ti-8Al-1Mo-1V and Ti-15Mo from IMI Titanium (U.K.) Ltd., England; Ti-6Al-7Nb from Sulzer Medical Technology Ltd., Switzerland; Ti-13Nb-13Zr from Smith and Nephew Richards Inc., U.S.A.; and Ti-3Al-2.5V from Aerospace Metals (Bury) Ltd., England;

The corrosion characteristics of some of these materials have been studied, previously (especially Ti(cpi) and Ti-6Al-4V). However, such studies have only analysed the corrosion behaviour of these materials within a limited potential range, usually upto approx. 2000 mV using de-aerated solutions, at which point the materials have been shown to breakdown. Also, any 'novel' material which has been investigated, has been presented simply as an alternative to the Ti-6Al-4V alloy, with an improvement in corrosion resistance [Scales, 1991, Ungersbock et al., 1994, Semlitsch et al., 1992]. In some cases, such alloys have been shown to display a better corrosion and wear resistance following treatment with an appropriate surface treatment prior to testing [Kovacs and Davidson, 1993, Babu et al., 1995].

In this study it has been possible to compare the corrosion resistance of a variety of alloys, including two recently developed novel ones (ie. Ti-6Al-7Nb and Ti-13Nb-13Zr). All materials were investigated using a larger potential range (between 0 and 5000 mV) in order to outline the breakdown characteristics, more informatively, and an aerated electrolyte solution was used as opposed to a de-aerated one to model the availability of oxygen in the physiological environment.

2.2 MATERIALS AND METHOD

All materials were received in the annealed condition, in the form of rods, which varied between 1 and 1.5 cm in diameter. The test specimens were then cut up to form discs of approx. 0.5 cm thickness. A wire lead was attached to the back of each disc, using a small amount of conducting paste. Following this, each specimen was mounted into a thermosetting resin mould, in the form of a 3 cm diameter rod. A large portion of the surface of the specimen was covered with an insulating lacquer varnish, leaving only a given specimen area exposed to the test solution. The specimen area was varied between approx. 0.1 - 0.5 cm^2. Each specimen was then polished down to a 1200 grit specification. Finer polishing was also undertaken to produce a "mirror-like" finish, using a 6, 1 and 1/4 micron diamond paste, respectively.

Finally, the specimen (anode) was placed into the electrolyte solution facing an auxiliary electrode (cathode) which had a surface area 4 or 5 times greater than the specimen. Surface roughness measurements were undertaken for each specimen before and after corrosion using a surface profilometer (a 'Taylor-Hobson Surtronic 3' profilometer was used). Hardness measurements were also undertaken using a 'Vickers hardness indentation machine'. For each specimen, an average value was obtained, using at least five individual readings.

Cyclic polarization curves were obtained using a potential range of between 0 and 5000 mV (an 'AUTOSTAT' computer controlled potentiostat was used). The potential was increased at a rate of 200 mV/min, starting with the rest potential (E_R). Electrode potentials were measured against a saturated calomel reference electrode. To ensure reproducibility, each material in a given solution, was subjected to triplicate polarization measurements. In some cases, it was necessary to undergo four or even five replicate tests to produce an acceptable degree of consistency. The electrolyte solution used in these tests was a phosphate buffered saline (PBS). The solution was made up by adding pre-packed tablets of a defined concentration to a relative amount of distilled water, which produced a 0.1M buffered solution at a pH of 7.4. The pH was then altered, as required, by adding measurable amounts of hydrochloric acid or sodium hydroxide. The solutions were maintained at 37°C throughout the tests using a suitable water bath.

2.3 RESULTS

The corrosion characteristics of each alloy tested and each constituent metal present in all the alloys, showed a breakdown of between 3000 and 4000 mV, very similar to the breakdown of commercially pure titanium (cpi), which produced a breakdown at approx. 4000 mV. To investigate the effect of using an aerated electrolyte solution as compared to a de-aerated one, similar tests were performed using a de-aerated solution. It was found for the materials to be breaking down at between 2000-2500 mV instead of around 3000 mV when using the aerated solution. The breakdown of these materials in de-aerated solution was comparable to the results obtained by other workers [Davidson et al., 1994, Jaffee and Burte, 1973, Speck and Fraker, 1980, Solar et al., 1979, Aragon and Hulbert, 1972].

Table 2.3.1 displays the surface roughness and hardness values of all six alloys alongside those of Ti (cpi). Figures 2.3.1 and 2.3.2 represent the data from table 2.3.1 in the form of histograms. Generally, it was found for harder materials to produce the finest surface finish following polishing. The results also indicated that alloys with mixed alpha/beta phase structures produce the most favourable mechanical properties as opposed to those with a single phase structure. These simple tests highlighted the importance of such properties in relation to the physical metallurgy of a titanium alloy, and thus, the reason for choosing alpha/beta alloys as load-bearing joint replacement materials.

Table 2.3.2 shows how the corrosion resistance of each material varies with finite changes in the pH of the environment (ie. pH = 5 and 9), as well as, outlining the specific phase structure of each alloy tested. Figure 2.3.3 represents the data from table 2.3.2 in the form of a histogram. This information gives a quantitative account of the ‘pitting corrosion resistance’ of each material following corrosion at pH=7.4. This has been measured as a difference between the breakdown and protection potentials (E_B-E_P). Generally, the lower the value, the more corrosion resistant the material. Although, all the materials tested were found to display an excellent corrosion resistance, the data obtained allowed a comparison showing only slight differences between each alloy. However, it became clear from the results, that Ti(cpi) produced by far the best corrosion resistance.

Tables 2.3.3 and 2.3.4 give a quantitative account of how the surface roughness and hardness values of each material were found to have changed following corrosion at the three different pH levels (ie. 5, 7 and 9). This data proved useful in reflecting how resistant such properties were to even slight changes in the pH of the environment. Figures 2.3.4 and 2.3.5 give a pictorial representation of this data in the form of histograms.

Table 2.3.5 provides similar information to that in Table 2.3.2, only in this case, each constituent metal composing the alloys was investigated alongside one another to directly evaluate how each metal component may or may not influence the overall corrosion characteristics of any particular alloy. Additionally, whether the corrosion behaviour of any particular metal correlates with its respective constituency within a particular alloy. Figure 2.3.6 displays this data in the form of a histogram.

Table 2.3.1 : Surface roughness and hardness values following initial polishing

MATERIAL	Surface roughness (μm)		Surface hardness (VHN)	
	Ra	$\overline{Ra}$	VHN	$\overline{VHN}$
Ti (cpi)	0.15	0.17	110	120
	0.16		118	
	0.19		132	
Ti-8Al-1Mo-1V	0.06	0.07	320	332
	0.07		336	
	0.07		340	
Ti-3Al-2.5V	0.10	0.10	300	311
	0.10		307	
	0.20		326	
Ti-6Al-4V	0.01	0.02	310	322
	0.02		318	
	0.03		338	
Ti-6Al-7Nb	0.01	0.02	338	349
	0.02		353	
	0.02		356	
Ti-13Nb-13Zr	0.05	0.06	346	353
	0.06		354	
	0.07		359	
Ti-15Mo	0.10	0.14	255	271
	0.16		266	
	0.16		292	

Figure 2.3.1: A histogram of the mean surface roughness values (Ra) of the alloys following initial polishing

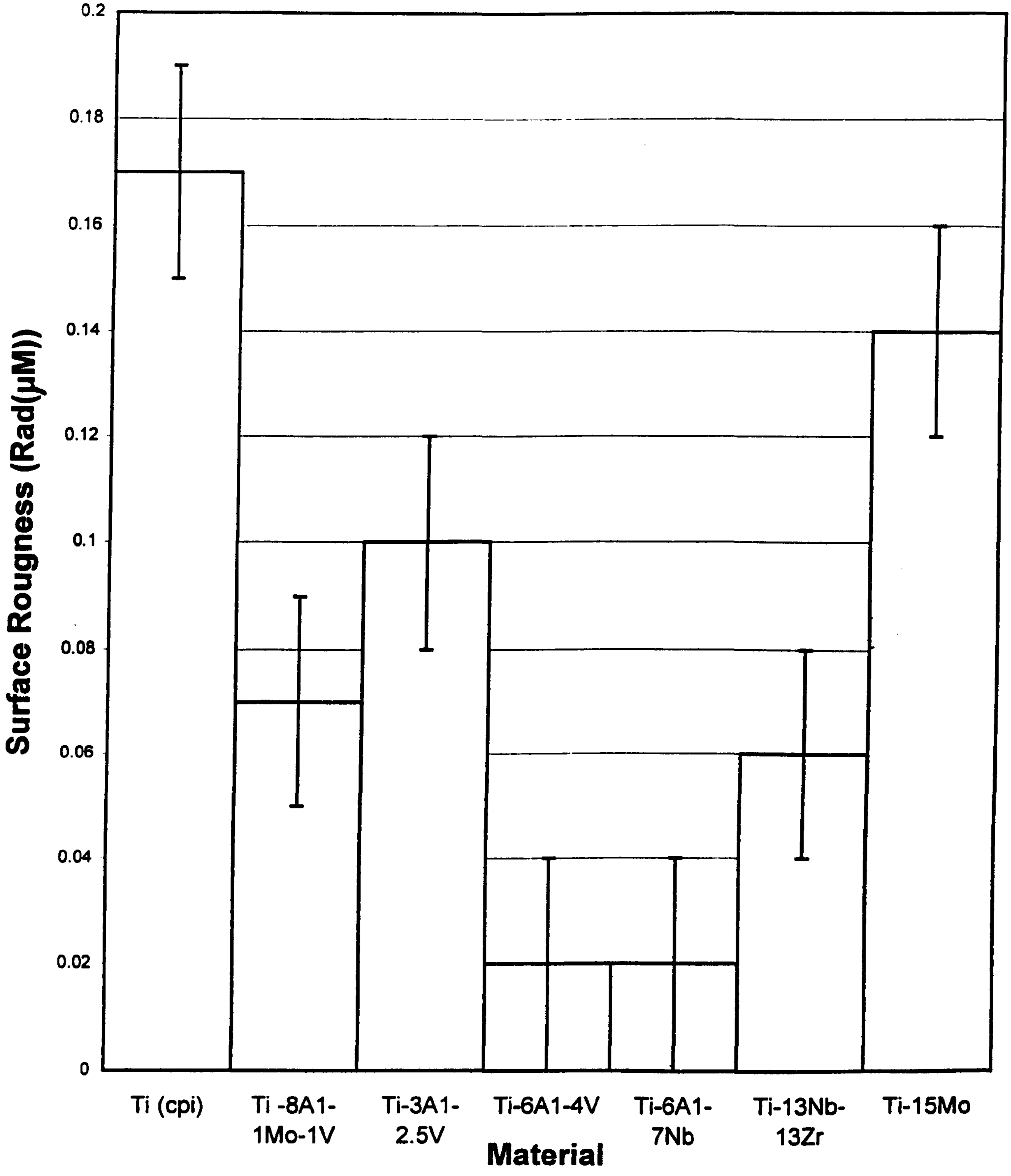


Figure 2.3.2: A histogram of the mean surface hardness values (VHN) of the alloys following initial polishing

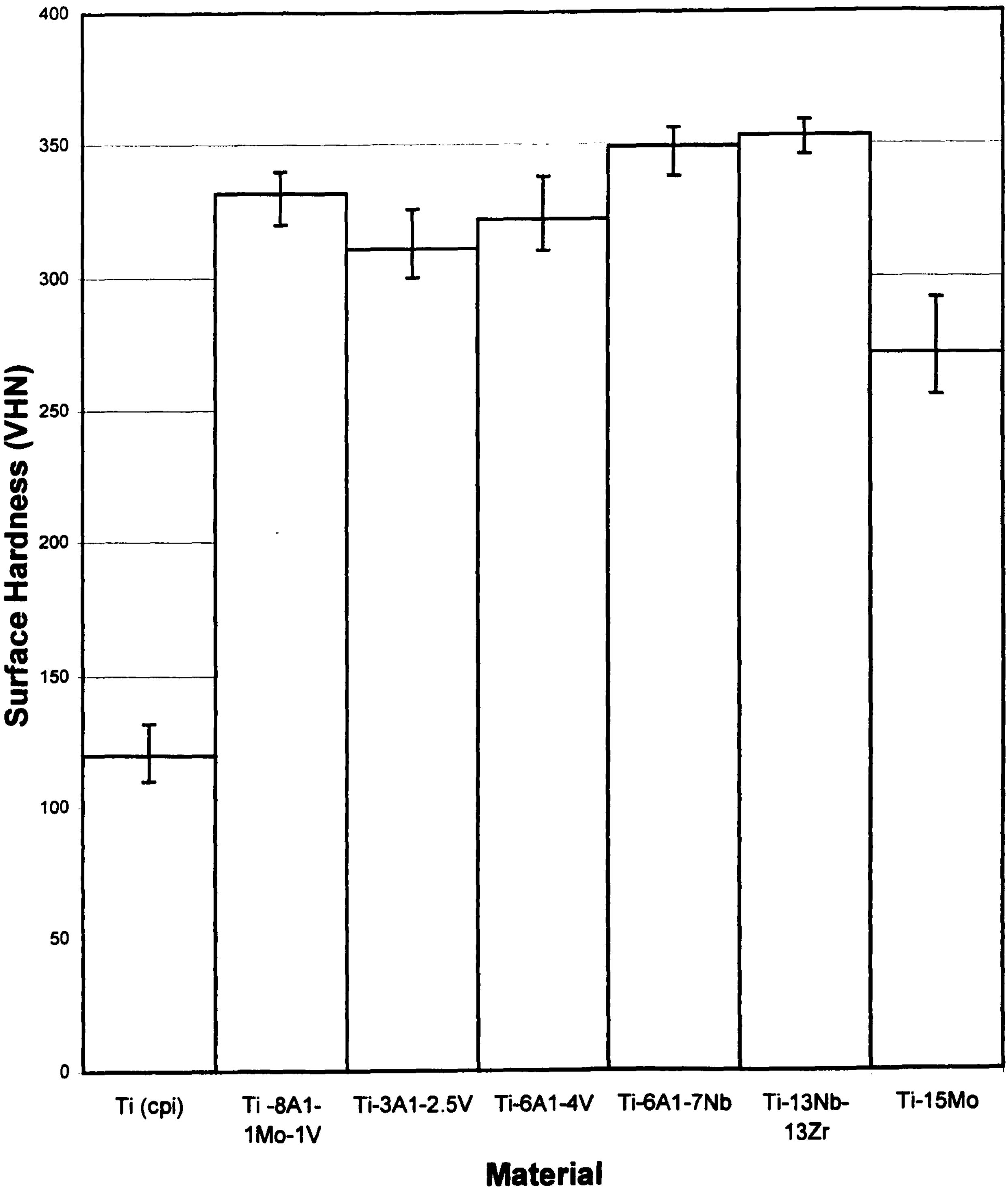


Table 2.3.2 : Pitting corrosion resistance (E_B-E_P) of the titanium alloys at varying pH levels

MATE-RIAL	PHASE	E_B-E_P (mV) at pH=5.0	$\overline{E_B\text{-}E_P}$ (mV) at pH=5.0	E_B-E_P (mV) at pH=7.4	$\overline{E_B\text{-}E_P}$ (mV) at pH=7.4	E_B-E_P (mV) at pH=9.0	$\overline{E_B\text{-}E_P}$ (mV) at pH =9.0
Ti(cpi)	pure metal	200	300	200	450	500	650
		300		550		700	
		400		600		750	
Ti-8Al-1Mo-1V	near alpha	1600	2050	3250	3450	4050	4100
		2150		3500		4100	
		2400		3600		4150	
Ti-3Al-2.5V	alpha-beta	2900	3025	4200	4300	3350	3700
		2975		4300		3650	
		3200		4400		4100	
Ti-6Al-4V	alpha-beta	1900	2600	3350	3850	3850	3950
		2700		3900		3900	
		3200		4300		4100	
Ti-6Al-7Nb	alpha-beta	2300	2500	2900	3025	3850	4000
		2400		3075		4050	
		2800		3100		4100	
Ti-13Nb-13Zr	near beta	3100	3150	2800	2900	2300	2650
		3150		2900		2700	
		3200		3000		2950	
Ti-15Mo	beta	3100	3250	2800	3000	4050	4250
		3300		3000		4300	
		3350		3200		4400	

Figure 2.3.3: A histogram displaying the corrosion resistance (E_B-E_P) of the alloys at varying pH levels

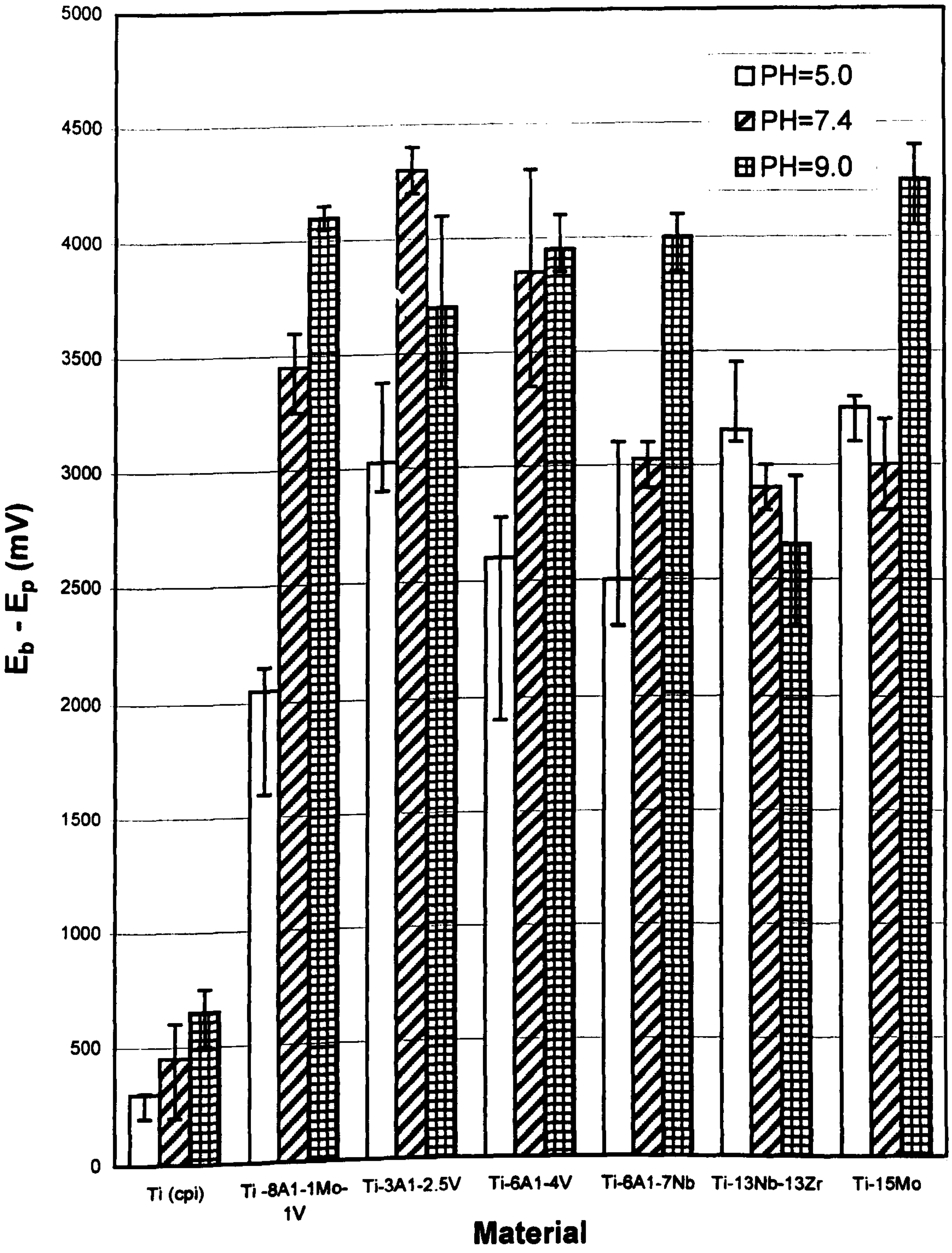


Table 2.3.3 : Changes in surface roughness following corrosion at varying pH levels

MATE- -RIAL	Change in surface roughness (microns)					
	At pH=5.0		At pH=7.4		At pH=9.0	
	Ra	$\overline{Ra}$	Ra	$\overline{Ra}$	Ra	$\overline{Ra}$
Ti(cpi)	0.09	0.10	0.19	0.27	0.17	0.21
	0.09		0.22		0.18	
	0.12		0.40		0.28	
Ti-8Al-1Mo-1V	0.02	0.04	0.02	0.03	0.03	0.06
	0.04		0.02		0.07	
	0.06		0.05		0.08	
Ti-3Al-2.5V	0.19	0.23	0.13	0.19	0.01	0.01
	0.22		0.16		0.01	
	0.28		0.28		0.02	
Ti-6Al-4V	0.24	0.27	0.02	0.02	0.09	0.16
	0.27		0.02		0.17	
	0.30		0.02		0.22	
Ti-6Al-7Nb	0.11	0.15	0.01	0.02	0.06	0.08
	0.16		0.02		0.09	
	0.18		0.02		0.09	
Ti-13Nb-13Zr	0.27	0.44	0.01	0.01	0.03	0.04
	0.43		0.01		0.04	
	0.62		0.02		0.05	
Ti-15Mo	0.01	0.01	0.29	0.41	0.02	0.05
	0.01		0.42		0.04	
	0.02		0.52		0.09	

Table 2.3.4 : Changes in surface hardness following corrosion at varying pH levels

MATERIAL	Change in surface hardness (VHN)					
	At pH=5.0		At pH=7.4		At pH=9.0	
	VHN	$\overline{VHN}$	VHN	$\overline{VHN}$	VHN	$\overline{VHN}$
Ti(cpi)	8	14	11	19	9	14
	12		17		11	
	22		29		22	
Ti-8Al-1Mo-1V	110	122	6	7	126	147
	124		12		152	
	132		15		163	
Ti-3Al-2.5V	16	19	64	77	101	110
	17		82		108	
	24		85		121	
Ti-6Al-4V	92	106	8	12	153	162
	111		10		166	
	115		18		167	
Ti-6Al-7Nb	129	141	37	43	139	141
	142		39		139	
	152		53		145	
Ti-13Nb-13Zr	167	189	72	91	185	199
	199		93		205	
	201		108		207	
Ti-15Mo	98	110	60	76	100	114
	109		80		119	
	123		88		123	

Figure 2.3.4: A histogram showing the mean change in surface roughness following corrosion at varying pH levels

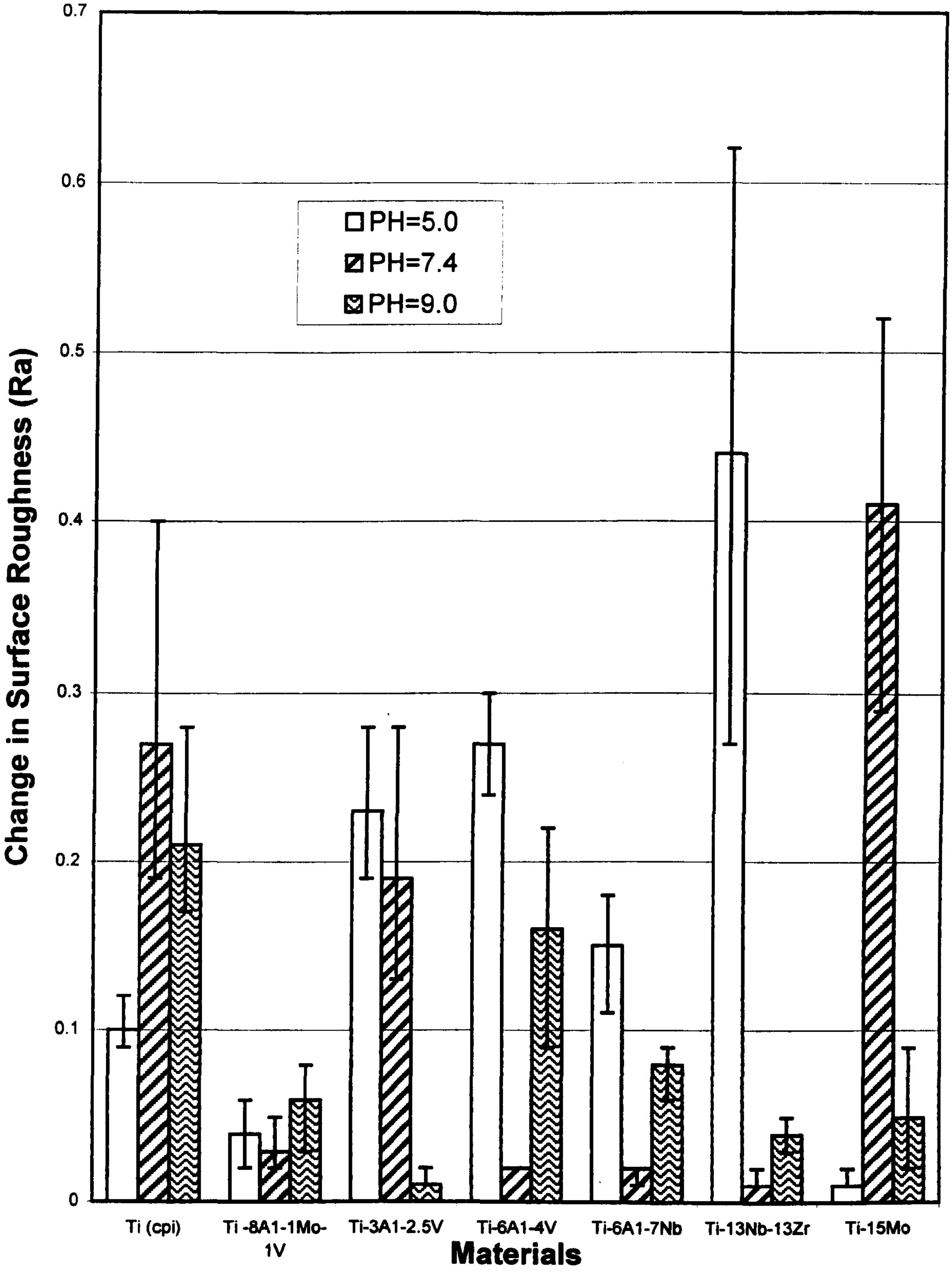


Figure 2.3.5: A histogram showing the mean change in surface hardness following corrosion at varying pH levels

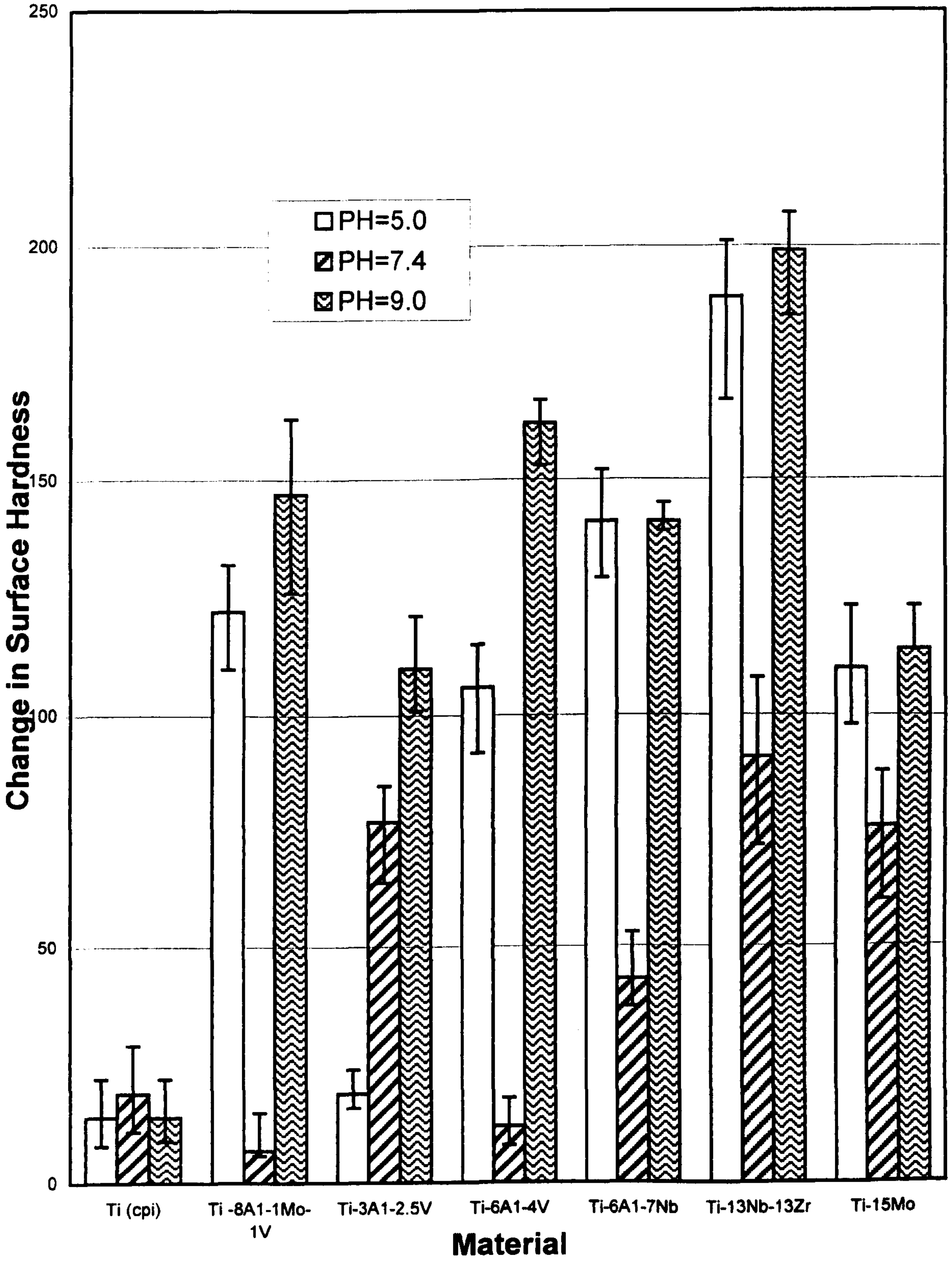
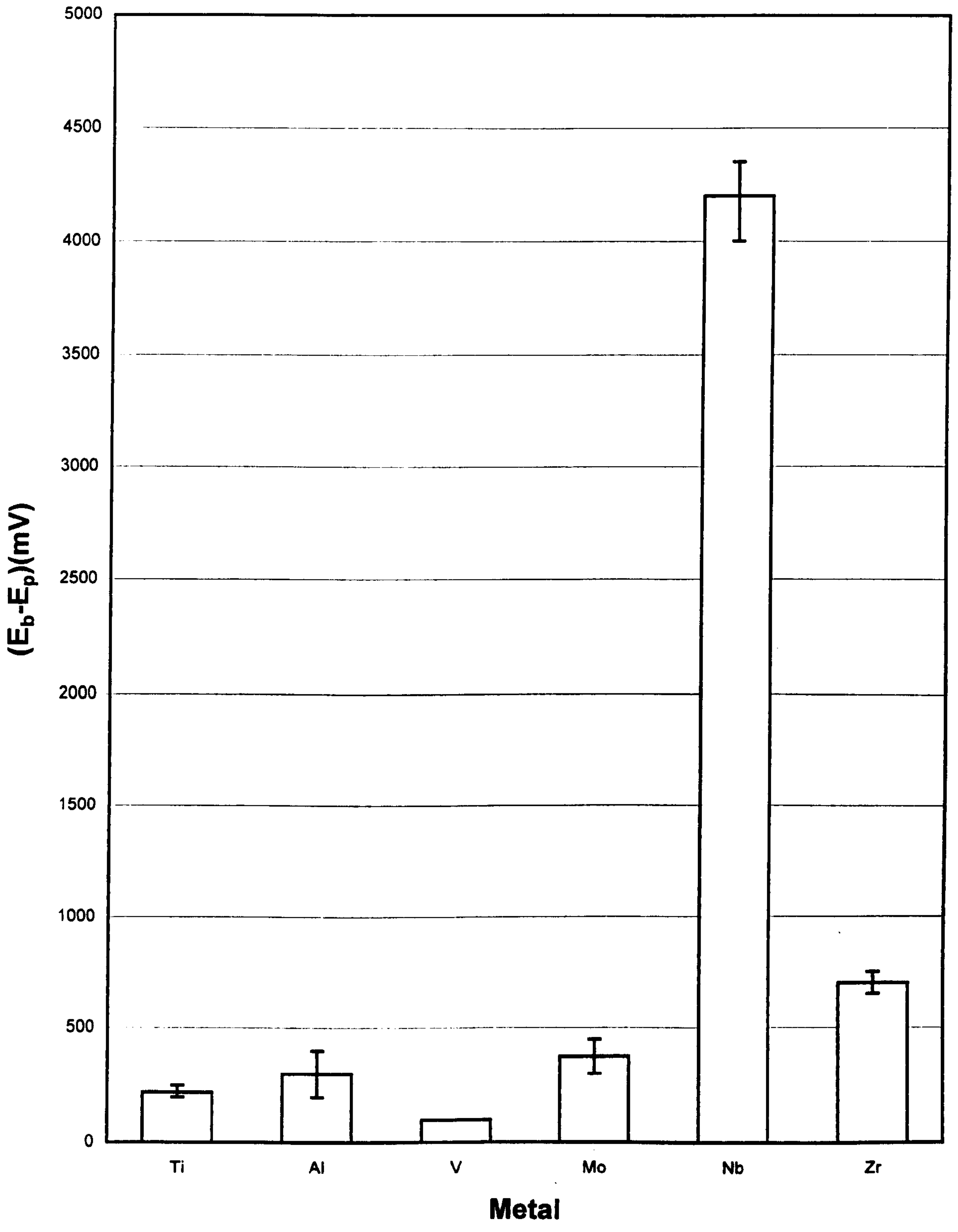


Table 2.3.5 : The pitting corrosion resistance (E_B-E_P) of the metals which constitute the alloys

MATERIAL	E_B - E_P (mV)	E_B - E_P (mV)
Titanium	200	220
	200	
	250	
Aluminium	200	300
	300	
	400	
Vanadium	100	100
	100	
	100	
Molybdenum	300	375
	375	
	450	
Niobium	4000	4200
	4250	
	4350	
Zirconium	650	700
	700	
	750	

Figure 2.3.6: A histogram showing the corrosion resistance (E_B-E_P) of the metals constituting the alloys



2.4 DISCUSSION

There is no doubt, that the results of this study showed Ti (cpi) to have a better corrosion resistance than any of the alloys tested. However, the fact remains that it is an inherently weak material to begin with having a hardness value of less than half that of any of its alloys. Although it does not appear to be a favourable candidate for orthopaedic applications, it continues to fulfil its role well as a suitable biomaterial in applications where corrosion resistance may be of a primary or sole concern.

Of the alloys tested, Ti-13Nb-13Zr, Ti-6Al-7Nb and Ti-15Mo alloys displayed the best corrosion resistance under normal physiological conditions (ie. at pH=7.4), and Ti-13Nb-13Zr alloy even seemed to be most resistant to a slight change in pH in either direction. However, it seemed to develop one drawback, in that it suffered a considerable loss in hardness following corrosion, which may prove to be of paramount importance from the tribological point of view. This material has recently been promoted as a useful alternative to Ti-6Al-4V, as long as a suitable surface treatment (eg. oxygen diffusion hardening) is used. However, the use of surface treatments may be considered as a completely separate issue, and one that is presently surrounded by yet more controversy. During this particular study, an approach was undertaken to investigate more precisely how each material might compare with another as a bulk entity, prone to degradation within a physiologically relevant environment.

An analysis of the alloys tested, suggested that the Ti-6Al-7Nb alloy may be more suitable as an alternative to Ti-6Al-4V. There were two main reasons for this, firstly, its corrosion resistance proved to be as good as Ti-6Al-4V if not slightly better, and secondly, because its resistance to loss of mechanical properties with changes in pH, were equally as good as Ti-6Al-4V.

Overall, as far as 'pitting corrosion resistance' was concerned, it was found that beta phase alloys possessed the best corrosion resistance, but that some alpha phase is necessary to improve the mechanical properties of the alloy. A review of the physical metallurgy in section 1.1 earlier, supports these findings in that although single phase structure alloys should generally produce a better corrosion resistance, it is the alpha-beta mixed phase alloys that display the best combination of strength and ductility [Boyer et al.,1985]. It was also discussed in the same section that the choice of alloying elements plays a vital role in determining the final phase structure produced following heat treatments (ie. annealing procedures used to strengthen the alloys during manufacture) [Donachie, 1988]. Thus, it is not surprising that efforts continue in the search for better alpha/beta phase alloys that also possess a high strength for load-bearing joint replacement materials at the same time as finding better choices of alloying elements.

With regards to the composition of each alloy, it was found that the breakdown potential (E_B) of Ti, Nb, and to a lesser extent Zr, was far greater than Al, V and Mo, in that order as has also been observed by others [Froes and Caplan, 1993, Kovacs and Davidson, 1993, Davidson et al., 1994].

At this stage of the overall study it was necessary to produce an initial set of data regarding the corrosion behaviour of titanium alloys and how slight changes in composition and structure related to such properties. Thus, this particular section of work has confirmed the possible use of a certain group of titanium alloys (Ti-6Al-4V, Ti-6Al-7Nb and Ti-13Nb-13Zr alloys) for use as load-bearing joint replacement materials.

The results confirmed that titanium alloys are very corrosion resistant under normal conditions, but when an accelerated corrosion test is performed, all materials were found to produce breakdown within the ranges used by the testing apparatus. The results obtained regarding slight changes in pH confirmed that the corrosion characteristics of such materials do change and that this can have a significant effect on the breakdown of a particular alloy. It has been shown in previous studies (as discussed in section 1.6) that at a lower pH conditions, Ti-6Al-4V alloy produced a rest potential which decreased with time indicating an increase in active sites explained to be a consequence of thinning or re-modelling of the passive film. However, at a higher pH, the opposite was found with the rest potential value found to increase with time, indicating a decrease in the number of anodic sites [Solar et al., 1979, Ban et al., 1991]. The Pourbaix diagrams for titanium show that the metal is immune to corrosion by gasification (oxygen evolution) across the pH studied. Thus, the corrosion characteristics measured during this study represents only passivation of the oxide layer and any change in pH undertaken during this study does not produce oxygen evolution [Pourbaix, 1984]. These findings suggest that titanium alloys should produce lower breakdown potentials at lower pH values compared with at a higher pH environment. It was also noticed that in lower pH conditions the breakdown potential (E_B) of the alloys was found to be comparatively lower than when testing in an environment at higher pH. However, the results of this study found that the E_B-E_P value was lower at pH=5.0 compared to at pH=9.0. That is to say that the value of the protection potential, E_P, was not influenced by pH.

In aerated PBS solution, the breakdown potential of the alloys was much higher than the reported breakdown potentials in de-aerated solutions [Aragon and Hulbert, 1972, Speck et al., 1980]. One of the problems associated with these conditions is the stability of the aqueous environment at these higher potentials, and therefore, the possible oxygen evolution due to the "degradation" of the electrolyte. We are convinced that we did obtain breakdown of the passive layer on the samples following evidence of pits on the surfaces following the test. Also, electrochemically, the polarization curves demonstrated a hysteresis (ie. a difference between E_B and E_P) which would not have been present if the increase in current density had been due purely to oxygen evolution

CHAPTER 3 : WEAR-ACCELERATED CORROSION, NON-CORROSIVE AND CORROSIVE WEAR OF SEVERAL TITANIUM ALLOYS IN PHOSPHATE BUFFERED SALINE (PBS) SOLUTION

3.1 INTRODUCTION

A simple, low cost sliding-wear apparatus was designed specifically for this study (Figures 3.1.1 and 3.1.2), which allowed to screen the materials in terms of wear within a corrosive environment. Figure 3.1.1 (b) is an image of the specimen disc mounted within the spring supported holder. Figure 3.1.1 (c) is an image of the pin head used in contact with the specimen surface. Wear rates were predicted in terms of changes in surface profilometry [Galante and Rostoker, 1973, Dumbleton, 1978, Stott and Breakell, 1989, Lilley et al., 1992, Poggie et al., 1994, Albert et al., 1994, Streicher et al., 1992, Angelini et al., 1991, Chakraborty et al., 1991, Hedeyat et al., 1992, Cook et al, 1992]. Thus, it was possible to evaluate and to compare the co-joint action of corrosion and wear of titanium and its alloys within a more physiologically relevant environment.

More specifically, it was possible to consider the effect of corrosion on wear by comparing the non-corrosive and corrosive wear of each material. In addition, the effect of wear-accelerated corrosion (by wearing the surface of each specimen prior to ensuing corrosion) was also studied. This evaluation provided evidence of the nature of re-passivation on the surface of each material, and also realizing the effectiveness of oxide coatings in improving the corrosion resistance of such materials.

Figure 3.1.1 : The simple pin-on-disc apparatus used for non-corrosive/corrosive wear testing

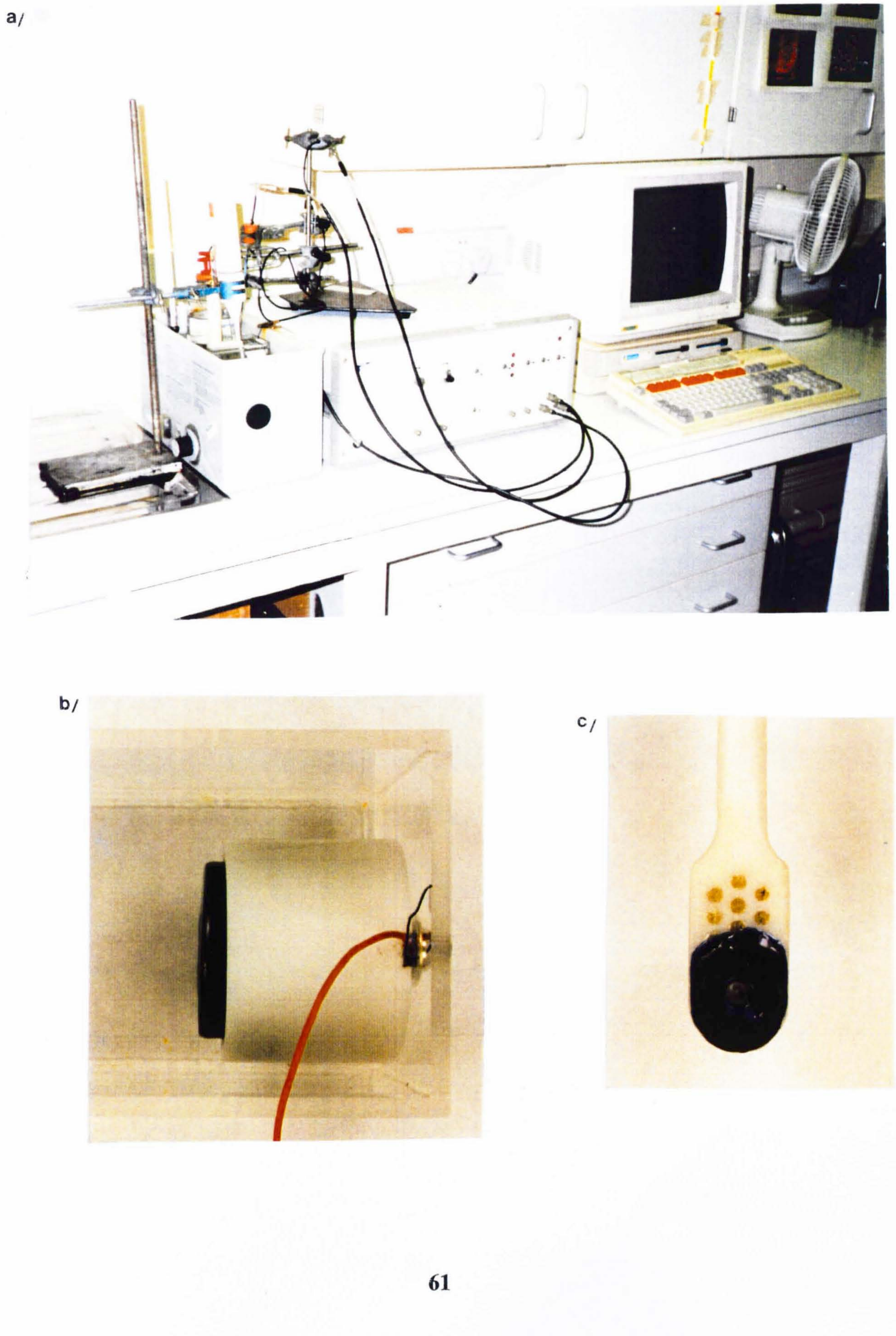
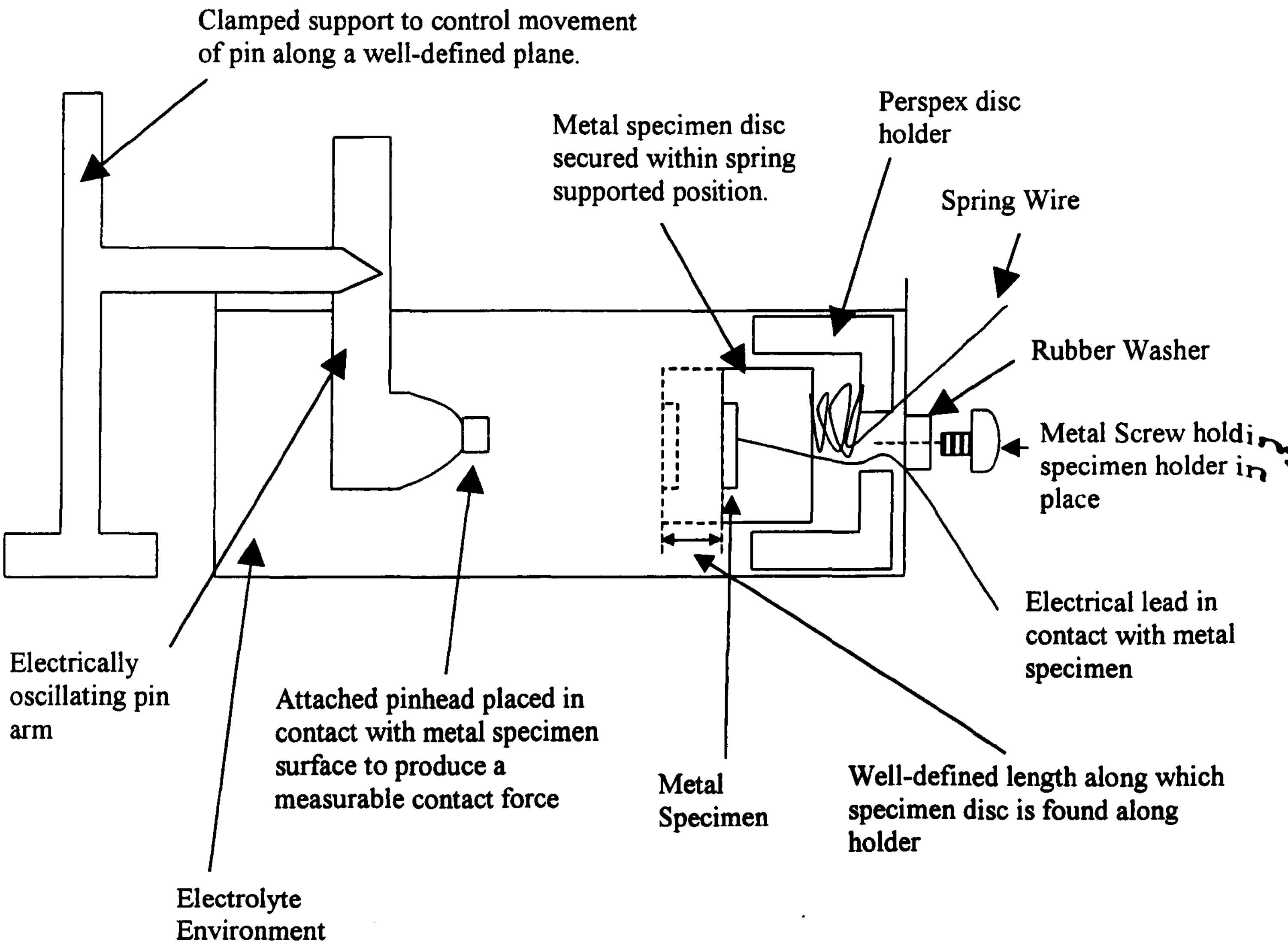


Figure 3.1.2: A schematic layout of the wear testing apparatus used



3.2 MATERIALS AND METHOD

Using the sliding wear apparatus shown (Fig.3.1.1), the specimens were worn continuously for specified periods of time. A constant load of 0.5 N was applied in contact with the specimen surface. The wearing pin was made of alumina and was found to produce a total sliding distance of 250 metres during each corrosion test. Since, the environment was much more aggresive during continuous wear compared to corrosion only, a potential step-up rate of 1000 mV/min was used during these tests as opposed to just 200 mV/min which was used during corrosion testing only, as discussed in Section 2.2 (the electrolyte was maintained at pH = 7.4 throughout). To demonstrate a non-corrosive environment, distilled water was used as the medium replacing PBS as the electrolyte.

Each specimen was measured for changes in surface roughness and hardness, before and after each test. The change in these two properties following non-corrosive/corrosive wear gave a quantitative account of the effect of wear on each material. Once again, triplicate measurements were obtained to maintain a substantial degree of consistency in the data.

To set up an environment allowing measurement of wear-accelerated corrosion, each specimen was subjected to continuous sliding wear, for a period of 5 minutes. Following this, the wear apparatus was removed from the electrolyte cell and the potentiostat was immediately switched on to execute a cyclic polarization curve of the specimen, in a similar way to that described in Section 2.1.

3.3 RESULTS

Table 3.3.1 and figure 3.3.1 show how the pitting corrosion resistance (E_B-E_P) of each material varied with and without the effect of wear before corrosion. With Ti(cpi) and Ti-8Al-1Mo-1V alloy the changes in the values obtained were more significant than with the other alloys. It should be noted that Ti(cpi) is very corrosion resistant to begin with, and although its characteristics have altered significantly following wear-accelerated corrosion, it still produced an E_B-E_P value better than all the alloys. Ti-3Al-2.5V, Ti-6Al-7Nb and Ti-13Nb-13Zr alloys were least affected by wearing of the surface prior to corrosion. Ti-6Al-4V and Ti-15Mo alloys were also affected by wear, but not to the same extent as some of the others. Generally, it might be added that all the alloys tested were very resistant to a wearing effect prior to corrosion. However, the characteristics of pure titanium, did produce a significant reduction in corrosion resistance.

Table 3.3.2 and figure 3.3.2 outline the change in surface roughness for each material in terms of non-corrosive and corrosive wear conditions. In general, it was found that corrosion had an accelerating effect on the wear behaviour of all the materials tested. More specifically, it was noticed that the mixed phase alpha/beta alloys (Ti-6Al-4V and Ti-6Al-7Nb) were more affected by both non-corrosive and corrosive wear than any of the alloys. Other than Ti-15Mo alloy, which was affected the least, the other alloys (ie. Ti-8Al-1Mo-1V, Ti-3Al-2.5V and Ti-13Nb-13Zr) displayed a comparable behaviour.

Table 3.3.3 and figure 3.3.3 outline the reduction in surface hardness for each material following non-corrosive and corrosive wear. Although, it was found that the hardness of all materials was reduced following such degradation, the effect was noted to be more severe in the presence of corrosive conditions compared to non-corrosive conditions. Once again, even within this generalization it was noticed that certain alloys suffered more than others. For example, Ti-6Al-7Nb and Ti-3Al-2.5V alloys followed by Ti-6Al-4V alloy were all superior to the others in resisting a considerable change in hardness when comparing corrosive to non-corrosive wear. However, this group of mixed phase alpha/beta alloys did lose a higher percentage of their initial hardness both following non-corrosive and corrosive wear. Therefore, although they allowed corrosive conditions to affect them least, the fact remains that the phenomenon of wear is more detrimental to this group of alloys.

Table 3.3.1: The pitting corrosion resistance (E_B-E_P) of the materials following corrosion and wear-accelerated corrosion

MATERIAL	Corrosion only		Wear-accelerated corrosion	
	$E_B - E_P$ (mV)	$\overline{E_B - E_P}$ (mV)	$E_B - E_P$ (mV)	$\overline{E_B - E_P}$ (mV)
Ti(cpi)	450	600	3000	3150
	650		3200	
	700		3250	
Ti-8Al-1Mo-1V	3150	3300	3850	4000
	3350		3900	
	3400		4250	
Ti-3Al-2.5V	4200	4300	4100	4300
	4300		4350	
	4400		4450	
Ti-6Al-4V	3600	3800	4050	4250
	3900		4200	
	3900		4500	
Ti-6Al-7Nb	3850	4000	3800	4150
	3950		4200	
	4200		4450	
Ti-13Nb-13Zr	3150	3400	3300	3500
	3450		3550	
	3600		3650	
Ti-15Mo	3700	3900	4050	4200
	4000		4200	
	4000		4350	

Figure 3.3.1: Pitting corrosion resistance (E_B-E_P) of the materials following corrosion and wear-accelerated corrosion

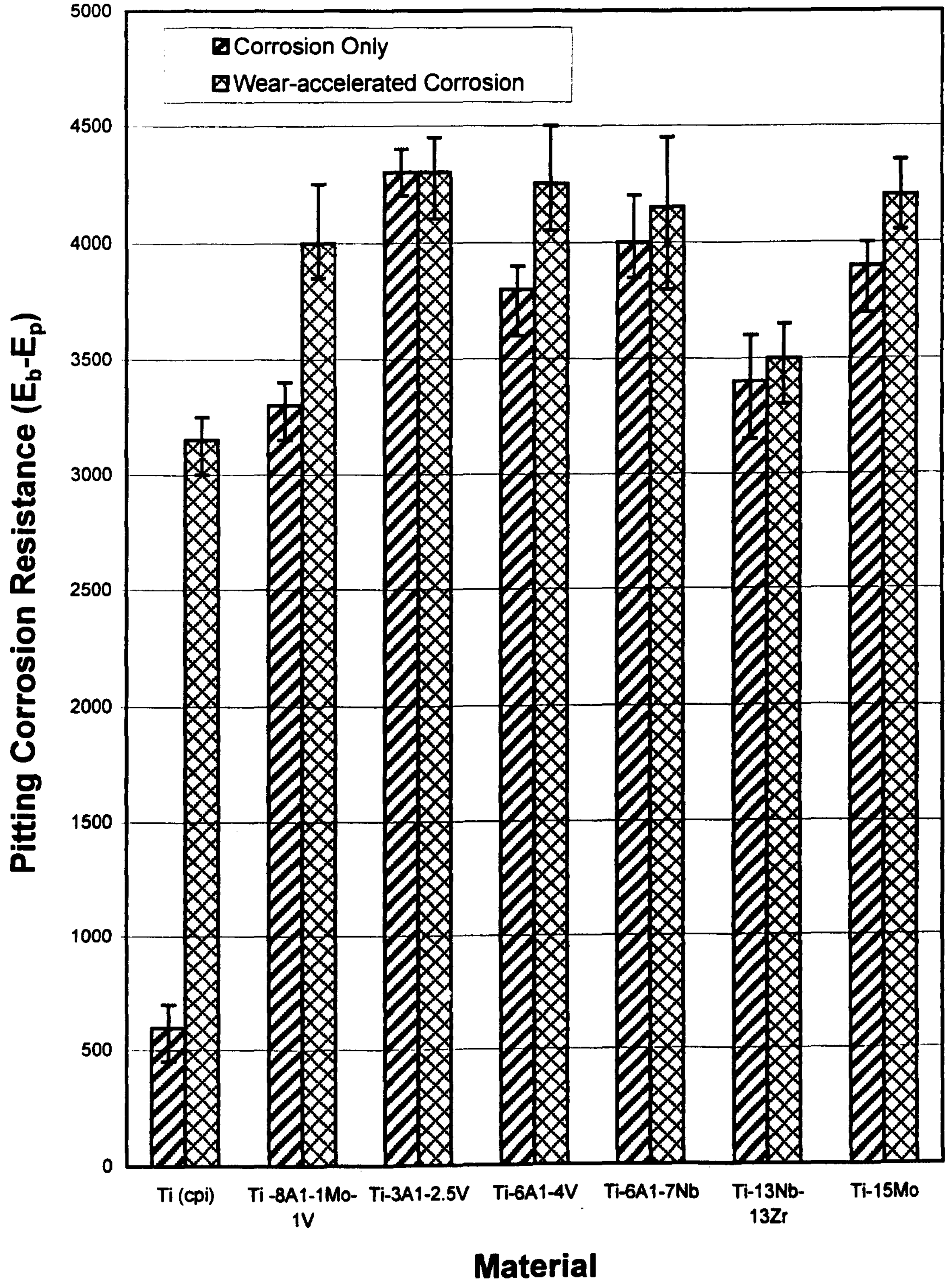


Table 3.3.2: Changes in surface roughness following non-corrosive and corrosive wear

MATERIAL	Initial Surface Roughness (microns)	Change in Surface Roughness (microns)			
		Non-corrosive Wear		Corrosive Wear	
	Ra	Ra	$\overline{Ra}$	Ra	$\overline{Ra}$
Ti (cpi)	0.17	1.38	1.56	2.10	2.58
		1.64		2.39	
		1.66		3.25	
Ti-8A1-1Mo-1V	0.07	1.05	1.16	1.08	1.58
		1.19		1.79	
		1.24		1.87	
Ti-3A1-2.5V	0.10	0.48	0.53	0.62	0.82
		0.55		0.86	
		0.56		0.98	
Ti-6A1-4V	0.02	0.61	0.79	2.01	2.28
		0.77		2.15	
		0.99		2.68	
Ti-6A1-7Nb	0.02	0.80	0.86	1.45	1.68
		0.82		1.79	
		0.96		1.80	
Ti-13Nb-13Zr	0.06	0.98	1.09	1.22	1.36
		1.11		1.34	
		1.18		1.52	
Ti-15Mo	0.14	0.99	1.06	1.00	1.15
		1.08		1.18	
		1.11		1.27	

Figure 3.3.2: A histogram showing the change in surface roughness following non-corrosive and corrosive wear

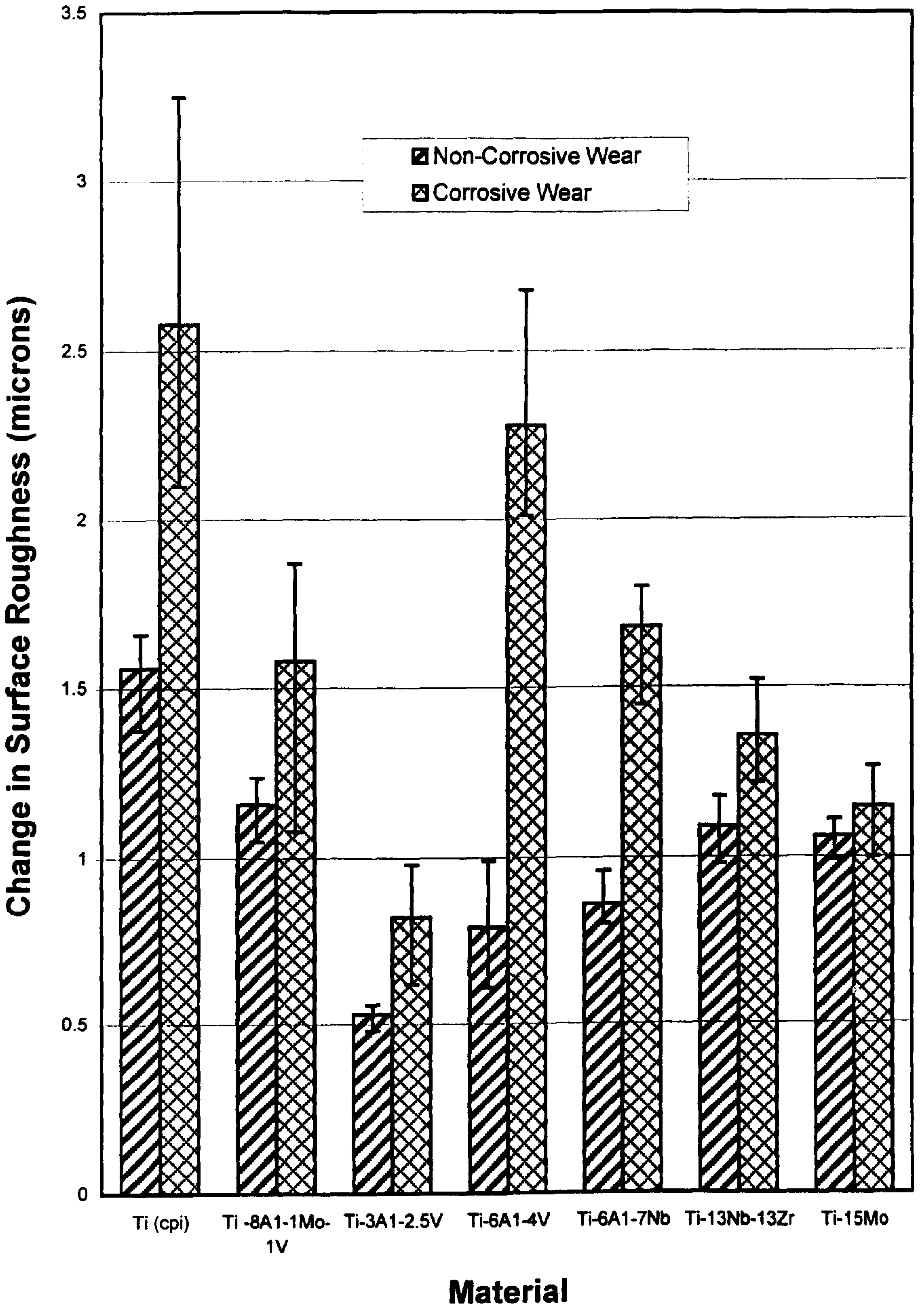
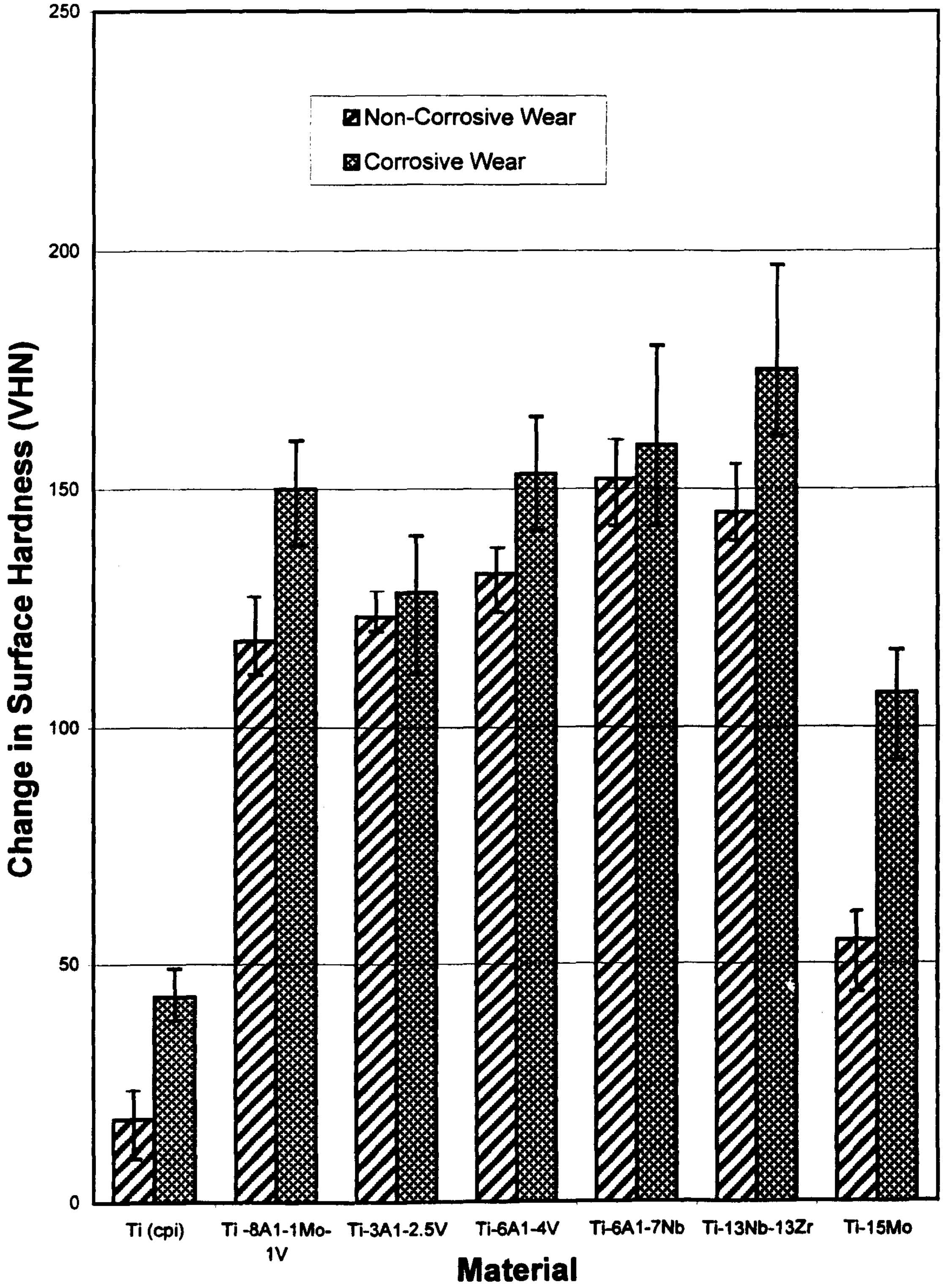


Table 3.3.3 : Changes in surface hardness following non-corrosive and corrosive wear

MATERIAL	Initial Hardness Values	Reduction in Surface Hardness (VHN)			
		Non-corrosive Wear		Corrosive Wear	
	VHN	VHN	$\overline{\text{VHN}}$	VHN	$\overline{\text{VHN}}$
Ti (cpi)	120	9	17 (14%)	38	43 (36%)
		19		44	
		23		49	
Ti-8Al-1Mo-1V	332	111	118 (36%)	138	150 (45%)
		116		152	
		127		160	
Ti-3Al-2.5V	311	120	123 (40%)	111	128 (41%)
		121		133	
		128		140	
Ti-6Al-4V	322	124	132 (41%)	141	153 (48%)
		135		153	
		137		165	
Ti-6Al-7Nb	349	142	152 (44%)	142	159 (46%)
		154		155	
		160		180	
Ti-13Nb-13Zr	353	139	145 (41%)	161	175 (50%)
		141		179	
		155		197	
Ti-15Mo	271	44	55 (20%)	93	107 (40%)
		60		112	
		61		116	

Figure 3.3.3: A histogram showing the change in hardness following non-corrosive and corrosive wear



3.4 DISCUSSION

Wear-accelerated corrosion

The behaviour of wear-accelerated corrosion is supported by a theoretical model which suggests that the wearing mechanism removes the protective oxide film on the surface of the material, thus, leaving the surface unprotected. As a result, when corrosion occurs, it does so at a much faster rate [Jaffee and Burte, 1973, Solar et al., 1979, Baboian and Haynes, 1981]. This explanation suggests that once the oxide layer has been removed, the corrosion resistance of the material will depend upon its ability to grow a "new" protective layer using the OH^- ions in the surrounding environment. Thus, depending on the density and coherency of this complex hydroxide film and its subsequent response to the exchange of metal and chloride ions at the surface of the metal, each material will respond differently in order to protect itself from the on-set of corrosion.

The results obtained during this study, indicate that the resistance to wear-accelerated corrosion may be linked to the physical metallurgy of the titanium alloy itself. More specifically, that those alloys with a mixed alpha/beta phase structure respond better to wear-accelerated corrosion compared to single phase alpha or beta alloys. This finding supports the already growing argument based on initial results outlined in chapter 2, in that, the consideration of 'novel' titanium alloys for use as load-bearing joint materials must be confined to those with a mixed alpha/beta phase structure. It seems that the results, so far, present the opportunity of further testing of alloys within this group.

Non-corrosive wear

When interpreting the results obtained during the study, it is reasonable to suggest that the processes of three-body wear will be contributing to the changes in surface roughness data produced, but, at the same time other factors may also be influencing the overall results. For instance, while the relatively hard alumina pin is positioned to slide along the metal surface to produce continuous wear, the idea of heat generation and heat transfer across the metal surface has not been considered. If the metal surface heats up at the point of continuous contact with the pin, this may alter the removal rate of metal particles from the surface from one material to the next. Thus, one particular alloy may produce more abrasive particles than another which may not as easily be detaching particles into solution. In any case, since the results of this study provide a comparative screening method to allow selection of certain alloys for further consideration, then it is important to be aware of such limitations of the device used as opposed to invalidate the results.

An analysis of the surface hardness values following non-corrosive wear suggests that the mixed alpha/beta phase alloys (ie. Ti-3Al-2.5V, Ti-6Al-4V, Ti-6Al-7Nb and to a lesser extent, Ti-13Nb-13Zr alloy) lost a considerable amount of their hardness following wear degradation compared to the single phase alloys (ie. Ti-8Al-1Mo-1V and Ti-15Mo alloys) and even Ti(cpi) which was relatively softer to begin with.

Corrosive wear

The results of surface roughness within a corrosive wear environment provides evidence for a greater effect on wear scar depth for all materials when corrosion occurs alongside wear. These results reiterate what has also been found by other workers, previously [Chakraborty et al., 1991, Hong and Pyun, 1991, Meyer et al., 1992]. It was noticed that the mixed phase alpha-beta alloys (Ti-6Al-4V and Ti-6Al-7Nb) were more affected by corrosive wear than any of the other alloys. An explanation has been proposed to explain this finding. These two alloys are quite hard initially (Table 2.3.1), and do not lose their hardness as a result of corrosion (Table 2.3.4) in relation to the other alloys which do seem to suffer considerably more. Thus, once they have worn they may be acting as hard abrasives, themselves, giving rise to the effect of three-body abrasive wear. As a consequence of this effect, they would most definitely produce greater wear scars, and thus, a rougher surface. On the other hand, an explanation for Ti(cpi) and Ti-15Mo alloy producing a comparatively lesser change in wear scar depth might be due to these materials being "softer" to start off with (Table 2.3.1). Thus, the particles that detach themselves from these alloys may not be contributing markedly towards three-body wear. This might produce a less prominent wear scar (eventhough the material may have inevitably lost more material as a result of the wear process). This explanation may also provide a valid reasoning for the other materials tested, since they are also not initially as hard as Ti-6Al-4V and Ti-6Al-7Nb alloys.

The changes in surface hardness produced within this environment, once again, showed that the mixed alpha/beta phase alloys suffered considerably more compared to the single phase alloys. However, since this group of materials are relatively harder to begin with, the end result indicates that even during corrosive wear, the alpha/beta alloys produce more favourable results as candidates for joint replacement materials.

It was following this initial study that it was decided to continue to look at variations within the biological environment *in-vitro* and investigate the corrosion resistance of Ti-6Al-4V, Ti-6Al-7Nb and Ti-13Nb-13Zr alloys. Especially, since these two latter materials had recently been proposed as superior alternatives to Ti-6Al-4V alloy which has now been used as an implantable material for several decades. It was decided that an in-depth study of corrosion and wear would be undertaken. The aim of the continuing study would focus on how variations in the biological composition (ie. protein molecules) of the corrosive environment influences the overall corrosion and wear characteristics of these three alloys.

General Comments

The aim of this study was to compare the similarities and differences of titanium alloys under corrosion and wear conditions. In this respect, of all the materials tested, the mixed phase alpha-beta alloys (Ti-6Al-4V and Ti-6Al-7Nb) displayed the best combination of both corrosion resistance and mechanical strength, which would be an important factor, from the point of view of an articulating joint prosthesis. Thus, it is evident that if better alloys are to be found for such applications in the future, then primarily, they must have both alpha and beta phases present in their microstructure, possibly, more beta than alpha to improve corrosion resistance, as has been shown. Also, as far as alloying composition is concerned, although it is preferable that metals such as aluminium be avoided, it seems that this metal continues to be a useful alpha stabilizer with essential strengthening properties to the alloy. However, niobium should continue to be a promising metal in successfully replacing vanadium, as a beta stabilizer, without any significant compromises and potential benefits with respect to less toxicity.

CHAPTER 4 : THE IN-VITRO CORROSION BEHAVIOUR OF Ti-6Al-4V, Ti-6Al-7Nb AND Ti-13Nb-13Zr ALLOYS IN PHOSPHATE BUFFERED SALINE (PBS) SOLUTION WITH AND WITHOUT THE ADDITION OF BOVINE ALBUMIN AND FOETAL CALF SERUM

4.1 INTRODUCTION

In the present study, two alloys (ie. Ti-6Al-7Nb and Ti-13Nb-13Zr) have been investigated in terms of their relative corrosion resistance compared to that of the Ti-6Al-4V alloy. In-vitro testing has made it possible to evaluate these materials within well defined, but varying environments, in order to establish a better scientific understanding of the passivation, re-passivation and breakdown characteristics of the alloys during accelerated corrosion. Using such methods, it was also possible to study the effect of protein adsorption and the presence of proteins as biological macromolecules in the corrosive environment.

It is well known that proteins affect the corrosion behaviour of some metals, and that their presence can either inhibit or accelerate the corrosion phenomena. They are known to behave differently with different metals, since their role in a corrosive environment is governed by many factors such as the surface chemistry of the metal, protein adsorption characteristics, interaction of protein molecules with other ions present in the electrolyte solution to produce organic complexes, and the transport of anionic and cationic charges around and away from the local environment [Cahn et al., 1992, Sousa 1993, Pourbaix 1984, Ban et al., 1991, Clark and Williams, 1982, Williams D.F., 1990 (b), Hanawa 1991, Williams D.F., 1986, Williams R.L., 1986, Ellingsen 1991, Healy and Ducheyne, 1992, Williams R.L. 1988, Dion 1993].

By slightly changing the variables within the environment it has been possible to investigate the corrosion behaviour of these alloys under different conditions. For instance, changes in pH were investigated (ie. at 5, 7.4 and 9) and changes in the concentration of albumin additions to phosphate buffered saline (PBS) solution were also studied (ie. 0.1mg/ml, 1mg/ml and 10mg/ml). In addition, corrosion studies were also undertaken in PBS with the addition of foetal calf serum (10%) in order to simulate more closely the internal biological milieu (in-vivo), but in a controlled manner. Testing in this environment made it possible to test the accuracy of more conventional in-vitro testing using PBS alone as is often reported to evaluate "new" materials. To supplement these studies, surface roughness and hardness measurements were undertaken following corrosion, within the different environments, to investigate the effect of corrosion on the mechanical integrity of each alloy [Williams and Cunningham, 1979, Brandes 1983, Kasemo 1983].

4.2 MATERIALS AND METHOD

The titanium alloys were obtained from different sources: The Ti-6Al-4V alloy from IMI Titanium Ltd., England; the Ti-6Al-7Nb alloy from Sulzer Medical Technology Ltd., Switzerland; and the Ti-13Nb-13Zr from Smith and Nephew Richards inc., U.S.A.;
All three materials were received in the form of 1.0 cm long rods of between 1 and 1.5 cm diameter. After suitable mounting (in a similar way to that described in section 2.2), the specimens were prepared for electrochemical testing by initially polishing down to a 1200 grit specification, followed by finer polishing using 6, 1 and 1/4 micron diamond pastes, respectively, and a final wash in acetone. Great care was taken with the cleaning and polishing regime to ensure that the initial surfaces were reproducible. Surface roughness measurements were undertaken for each specimen before and after corrosion using a 'Taylor-Hobson Surtronic 3' profilometer. This method involved contact with the surface using a diamond stylus measuring the surface roughness to within 0.01 microns. Hardness measurements were also undertaken using a Vickers hardness indentation machine to obtain values of the macrohardness of the surface of each alloy both prior to and following corrosion. Surface roughness and hardness measurements were undertaken before and after each corrosion test. Measurements were undertaken on three specimens in all cases to ensure reproducibility and for the hardness measurements an average of four or five indents was calculated for each surface.

Cyclic polarization curves were obtained using a potential range of between 0 and 5000 mV. The step-up rate used was 200 mV/min starting at the rest potential (a value obtained after the specimens were left under open-circuit conditions for 30 minutes). The electrolyte used was a phosphate buffered saline (PBS) solution to which additions of albumin and foetal calf serum were made when required. The electrolyte cell was maintained at 37°C throughout the tests using a suitable water bath. All tests were carried out under aerated conditions. The electrolyte cell was made of glass and had a capacity of 200 ml. The specimen acted as a 'working' electrode (anode) and was placed in the cell facing a platinum 'auxiliary' electrode (cathode). A saturated calomel electrode was used as the 'reference' electrode.

4.3 RESULTS

The corrosion resistance of the alloys is defined in terms of the difference in the value of the breakdown and protection potentials, E_B and E_P. The breakdown potential is normally that value observed when a large increase in current due to oxygen evolution and pit formation. The protection potential is normally defined as the potential at which the current density on the negative sweep from the maximum anodic potential equals the current density in the passive region (ie. the point at which the hysteresis loop is complete, or, rest potential is reached).

Figure 4.3.1 and 4.3.2 present the corrosion resistance data for the three alloys in different environments. From figure 4.3.1 the corrosion resistance of the alloys (ie. E_B-E_P values) at three different pH levels (ie. 5, 7.4 and 9) in PBS solution can be evaluated. Generally, it was noticed that as the pH level increased, the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb decreased, whereas that of Ti-13Nb-13Zr was seen to increase. It was also noticed that this alloy was most resistant to change in pH in any direction. When albumin, at a concentration of 1 mg/ml, was added the effect of pH on the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb is reduced whereas for Ti-13Nb-13Zr the addition of albumin caused a decrease in corrosion resistance as the pH increased.

From figure 4.3.2 the corrosion resistance of the alloys in an environment containing progressively increasing amounts of albumin can be evaluated. It was observed that by adding albumin (0.1 mg/ml) to PBS, the corrosion resistance increased for Ti-6Al-4V and Ti-6Al-7Nb, whereas, with Ti-13Nb-13Zr it was found to decrease. Further addition of albumin to a concentration of 1 mg/ml reduced the corrosion resistance of Ti-6Al-4V and Ti-6al-7Nb but increased the corrosion resistance of Ti-13Nb-13Zr. The addition of yet more albumin (ie. from 1 mg/ml to 10 mg/ml) caused little change in the corrosion resistance of all three alloys, although, the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb slightly decreased, whereas, that of Ti-13Nb-13Zr slightly increased. The Ti-6Al-4V alloy was more corrosion resistant in all the albumin solutions than in PBS, whereas the Ti-6Al-7Nb and Ti-13Nb-13Zr alloys were generally less corrosion resistant in protein solutions than in PBS. The addition of foetal calf serum produced a similar effect with each alloy as had the presence of higher concentrations of albumin.

Figure 4.3.3 gives an account of the change in surface roughness of the alloy specimens in PBS alone and in that containing varying concentrations of albumin, following corrosion. It was found that the surface roughness increased in the presence of more albumin in the environment, suggesting that the composition of the oxide could be changing as a result of pitting corrosion and that this change could be due to both protein adsorption to the surface and precipitation of corrosion products in the form of metal/protein/oxide complexes. The figure also displays the surface roughness (Ra) of each material following corrosion in PBS with the addition of whole serum. The surface roughness values were found to be influenced to a similar extent to those exposed to low albumin concentrations. Microscopic examinations of the alloy surfaces showed evidence of pits following polarization to the breakdown potential.

Figure 4.3.4 outlines the hardness values of the three alloys following corrosion in PBS alone and in PBS with albumin at varying pH levels and demonstrates that corrosion of all the alloys in all the environments reduced the hardness. In PBS the reduction in hardness was least at pH=7.4 for all three alloys. Following corrosion in albumin for Ti-6Al-4V the hardness was reduced more at pH 7.4 than at pH 5 or 9. For Ti-6Al-7Nb the hardness was least affected at pH 5 and for Ti-13Nb-13Zr the hardness was least affected at pH 7.4.

Figure 4.3.5 shows the influence of protein concentration on the hardness values following corrosion. It demonstrates that the addition of protein reduces the hardness of all surfaces with Ti-13Nb-13Zr being least affected and Ti-6Al-4V the most.

Figure 4.3.1 : Pitting corrosion resistance (E_B-E_P) of alloys in PBS solutions at varying pH levels

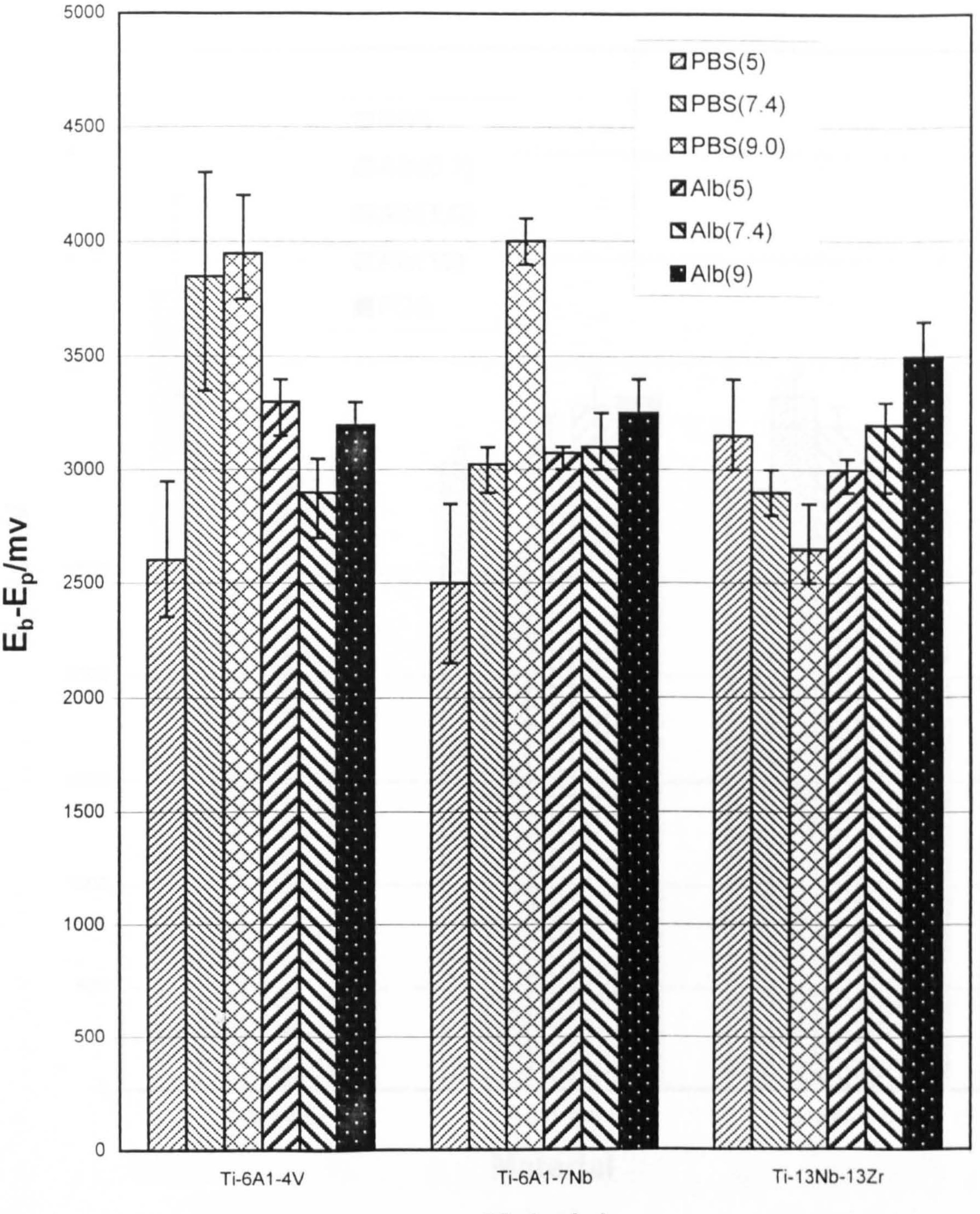


Figure 4.3.2 : Pitting corrosion resistance (E_B-E_P) of alloys in PBS with and without the addition of varying concentrations of Albumin at pH=7.4

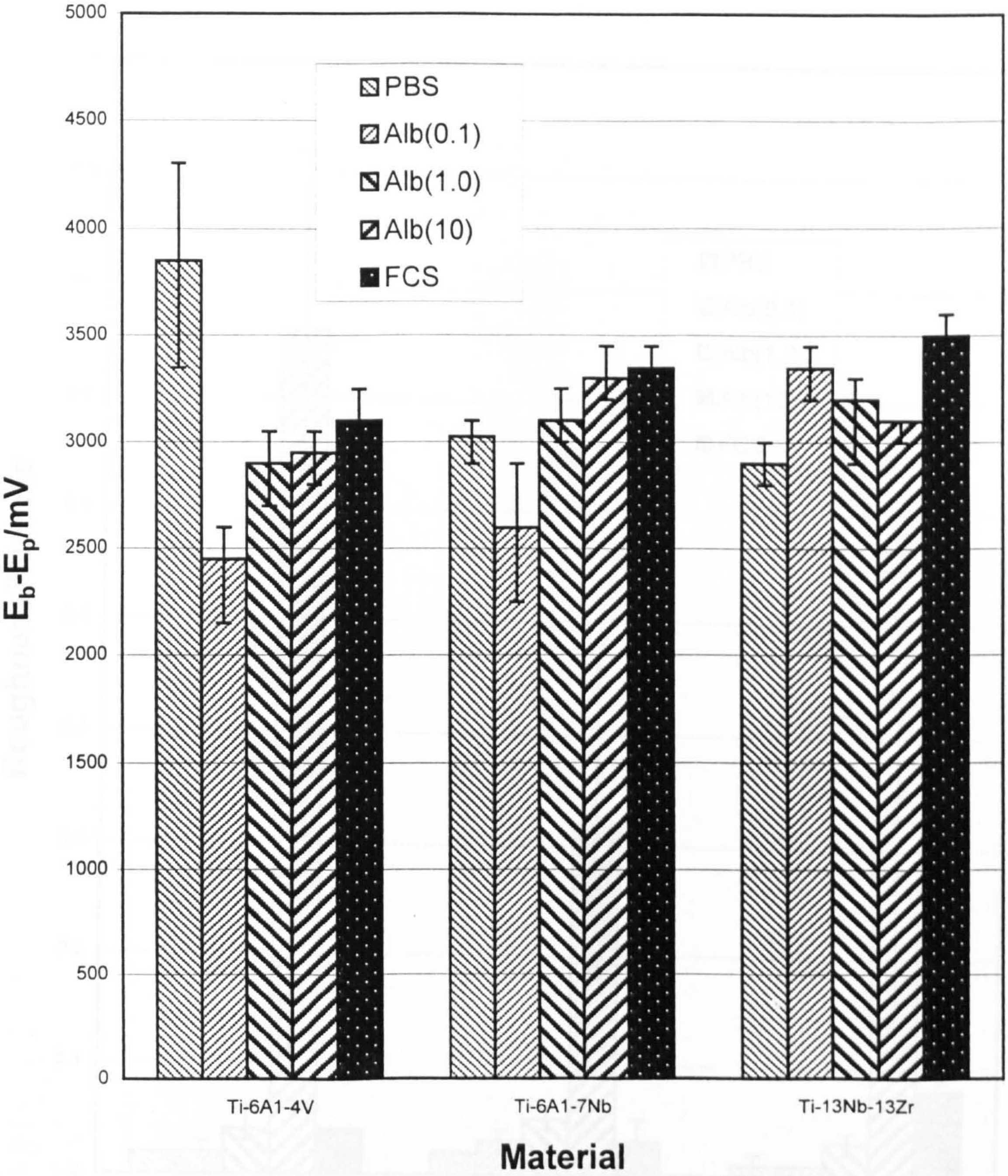


Figure 4.3.3 : Change in surface roughness of alloys following corrosion in PBS only and in PBS with and without the addition of varying concentrations of Albumin at pH=7.4 and foetal calf serum (10%)

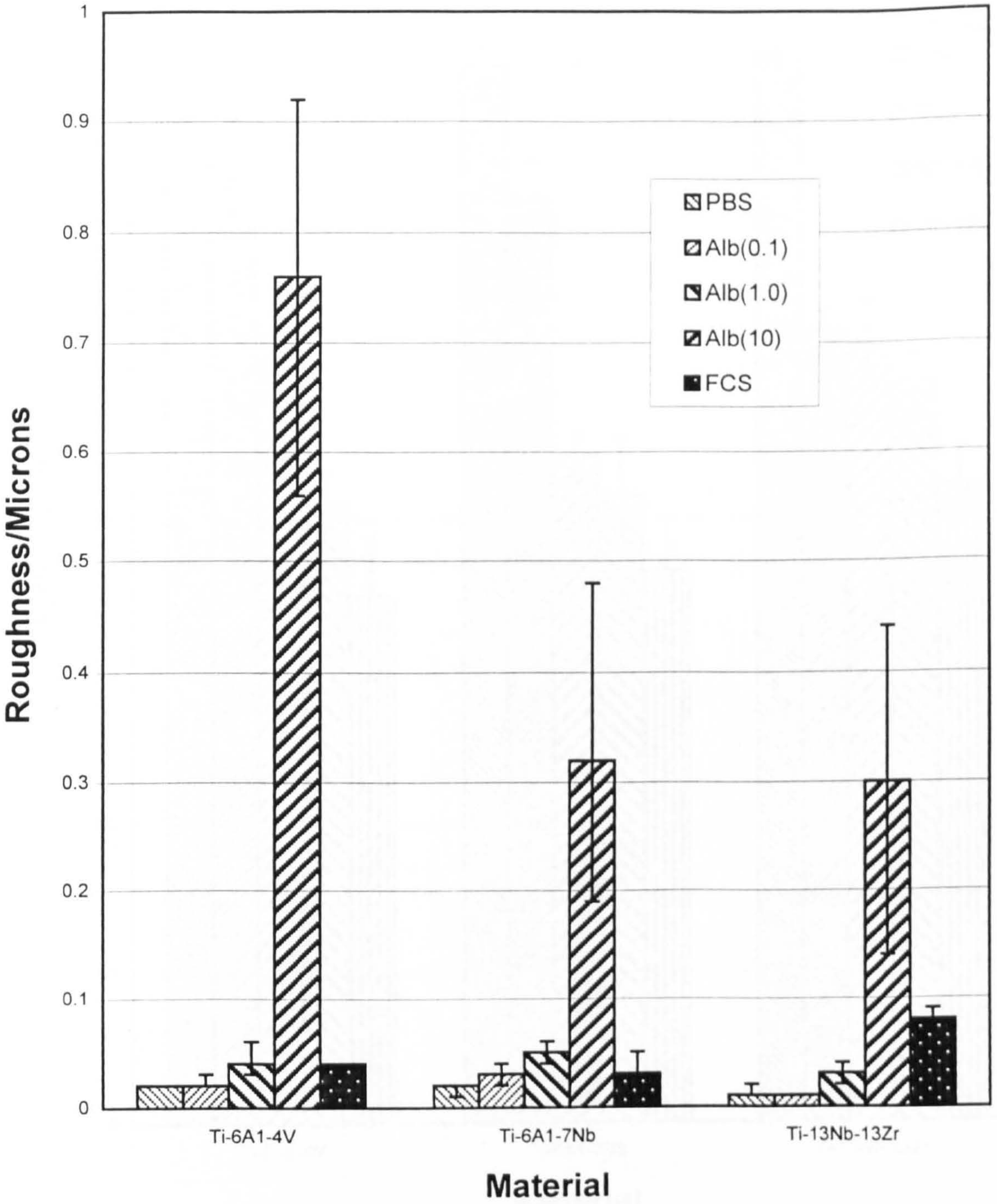


Figure 4.3.4 : Vickers hardness values (VHN) of alloys before and after corrosion in PBS only and in PBS with the addition of varying concentrations of Albumin at pH=7.4

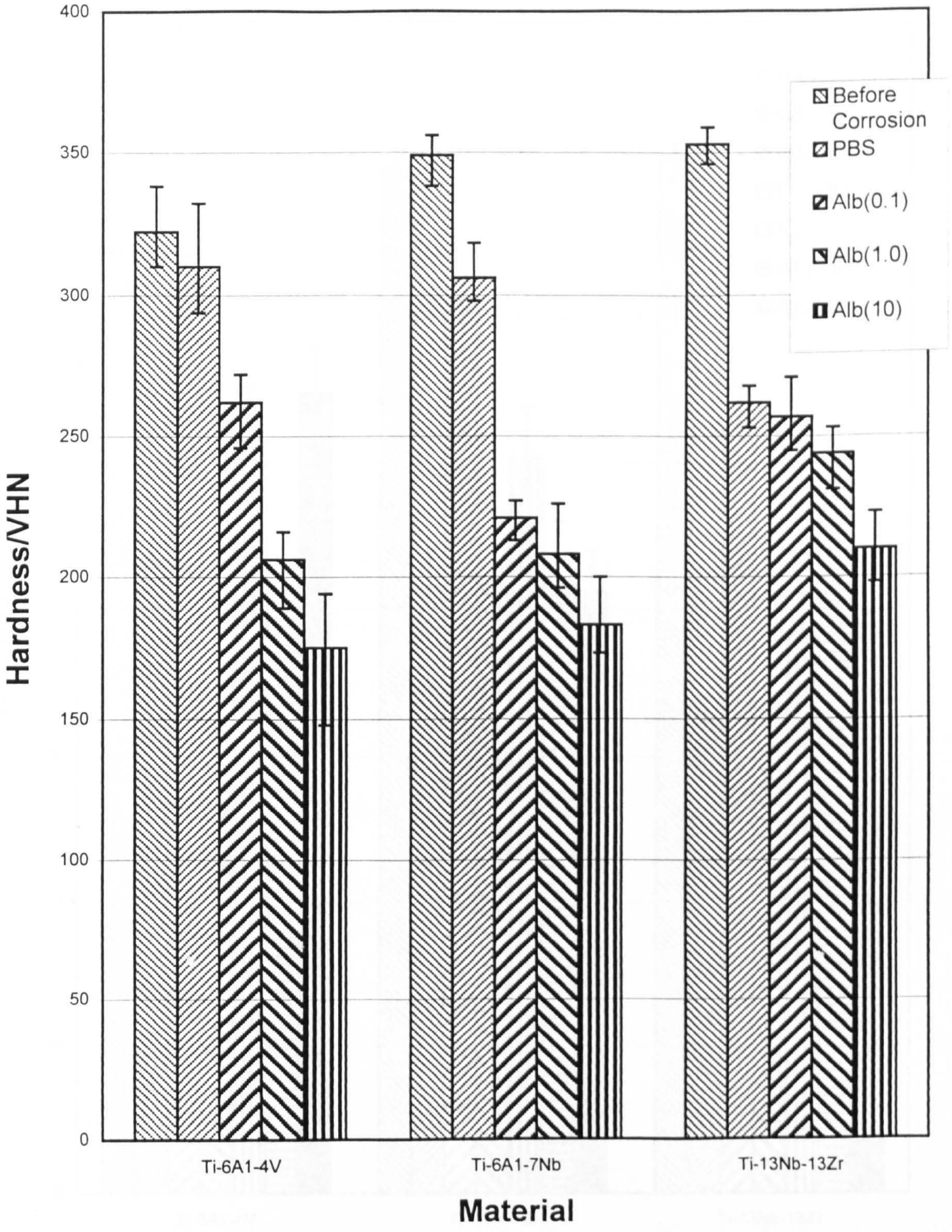
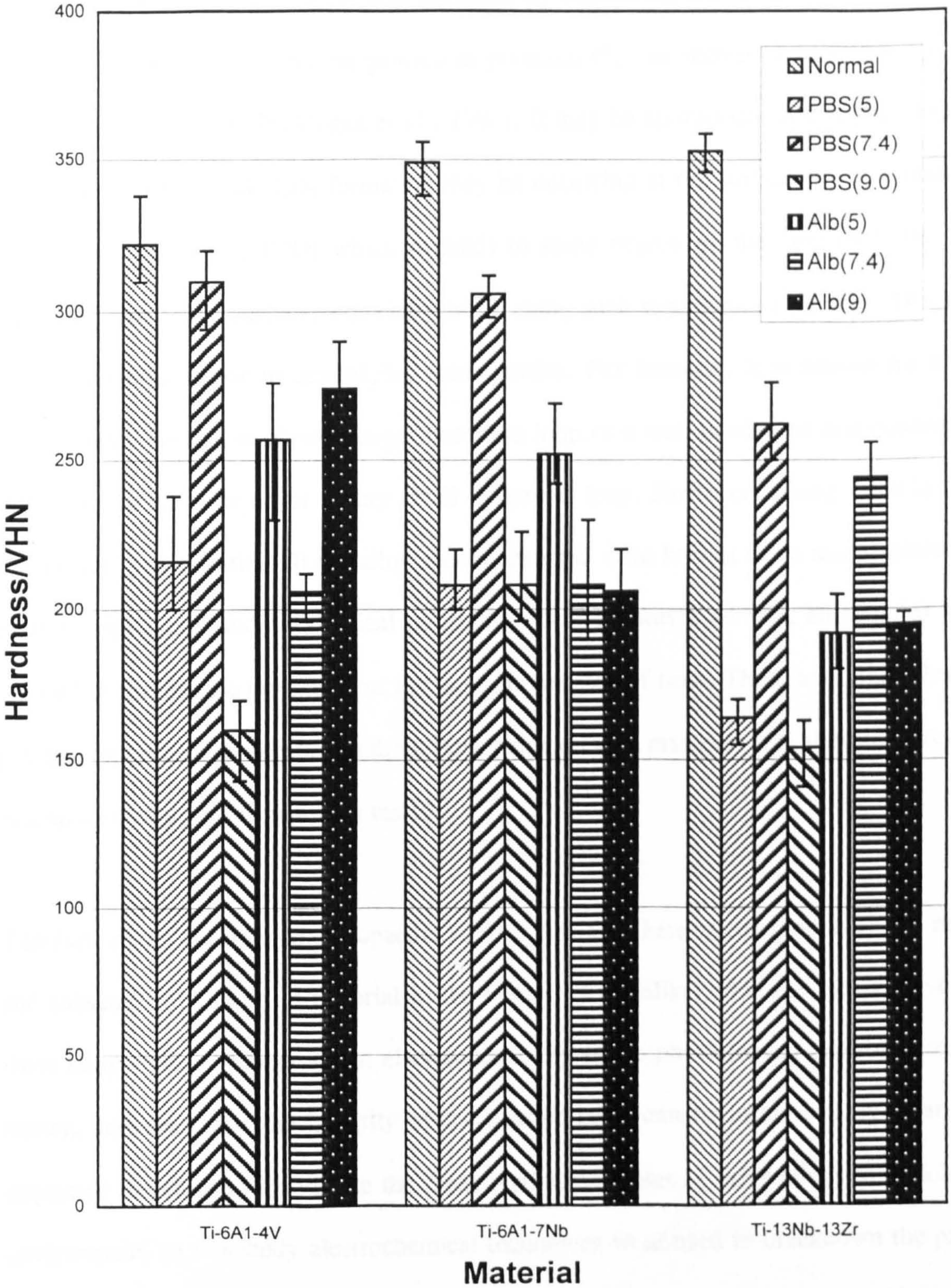


Figure 4.3.5 : Vickers hardness values (VHN) of alloys following corrosion in PBS containing Albumin (conc. 1 mg/ml) at varying pH levels



4.4 DISCUSSION

Susceptibility to pitting corrosion was analyzed by measuring the difference between the breakdown potential (E_B) and the protection potential (E_P) as shown in figures 4.3.1 and 4.3.2 [Cahoon et al., 1975, Cogan et al., 1994]. It may be appropriate at this point to take into account that O_2 and H_2O_2 formation may be occurring at the surface of these titanium alloys [Tengvall et al., 1989] which depends to some degree on the "pre-life", the pre-oxidation state of the surface, especially when using such results quantitatively. This may create a limiting factor in quantifying these results. For instance, it is known for a less passive initial surface to show a large hysteresis loop or a well passivated and conditioned initial surface to show no or a very small hysteresis loop. However, taking these limiting factors into consideration all the alloy specimens used were known to be manufactured for medical application and an identical polishing procedure was used with all samples being exposed to the outside environment for the same lengths of time. These and other cleaning preparation procedures meant that the surface oxide on such materials was allowed to develop in a similar way for all materials tested.

Titanium alloys are used in orthopaedic applications and there is concern about the release and subsequent build up of material in the tissues. It is unlikely that the passive layers on these alloys will be broken down electrochemically in the physiological environment. It is known, however, that the integrity of the passive layer can be influenced by wear. It is important therefore to investigate the repassivation processes of these alloys in a biological environment. In this study electrochemical techniques were used to breakdown the passive layer so as to investigate the repassivation and the interaction of proteins with this process.

To understand comprehensively how corrosion takes place in terms of charge transfer at the metal/electrolyte interface, it is necessary to consider an accepted model for this charge distribution in terms of a charge double-layer structure (see figure 1.6.2 earlier). In this case, a double-layer is present due to hydroxylated ions at the metal surface and charge transfer can occur within the layer depending on whether the metal oxide is stable (ie. acting as a cathode) or whether it is passivating upto and beyond removal of its protective oxide layer (ie. as an anode).

In the absence of proteins when a metal oxide surface breaks down (at breakdown potential, E_B), re-passivation begins to occur almost spontaneously, and as a result of this the underlying metal, which acts as an anode, will simply attract anions in the form of OH^- and Cl^- from the environment and due to re-passivation it will eventually reach what is known as its protection potential (E_P). Thus, a material's ability to re-passivate quickly following its breakdown is a measure of its pitting corrosion resistance.

The influence of pH on the corrosion resistance may be explained by the fact that at a higher pH more OH^- ions are present, and as a result of this more anions are going to be attracted to the metal surface during corrosion, accelerating the corrosion of Ti-6Al-4V and to a lesser extent, Ti-6Al-7Nb. As far as Ti-13Nb-13Zr is concerned, however, the opposite effect may be due to less metal dissolution, since niobium and zirconium are known to be less soluble than aluminium and vanadium therefore the OH^- will increase the oxidation of the titanium and accelerate passivation producing a better corrosion resistance [Kovacs and Davidson, 1993, Bordji et al., 1996]. Furthermore, it is known that Ti-13Nb-13Zr has a more inert oxide layer which seems not to breakdown as easily, and even so once it does it seems to repassivate more quickly even under normal physiological conditions (ie. at pH=7.4) as the data obtained during this study suggests [Davidson et al., 1994]. So, if there are more OH^- ions available at higher pH (ie. at pH=9), then it will most likely repassivate at an even faster rate.

With the more complicated situation in which proteins are introduced into the environment, the corrosion resistance of Ti-13Nb-13Zr was seen to decrease and that of Ti-6Al-4V and to a lesser extent of Ti-6Al-7Nb were seen to increase.

There are six ways in which this might occur [Ban et al., 1991, Clark and Williams, 1982, Williams D.F., 1986, Williams R.L., 1986, Cogan et al., 1994, Browne and Gregson, 1994]:-

1. The biological molecules could upset the equilibrium of the corrosion reactions by consuming one or other of the products of the anodic or cathodic reactions (that is proteins could bind to metal ions and transport them away from the implant surface). This would upset the charge double-layer and allow further dissolution of the metal to occur.
2. The stability of the oxide layer depends on the electrode potential and the pH of the solution. Proteins often participate in electron-carrying roles and thus affect the electrode potential, and bacteria can alter the pH of the local environment through the generation of acidic metabolic products (It is well known that when an implant is inserted, the pH shifts to acidic values of about 5.2 at the injured sites and returns to normal in a few days. This is because the local pH can be affected by inflammation or infection, [Ban et al., 1991]).
3. The stability of the oxide layer also depends upon the availability of oxygen. Adsorption

of proteins onto the surface of materials could limit the diffusion of oxygen to certain regions of the surface. This could cause preferential corrosion of the oxygen deficient regions and lead to the breakdown of the passive layer.

4. Adsorbed protein could act as a catalyst, oxidizing the metal with internal disulphate groups, and form a metal-protein complex with disulphide bonds being subsequently reformed by oxidation with oxygen. Repeated activity of this nature is most likely to lead to cross-linking and precipitation of the protein on the surface.

5. Protein adsorption onto the surface may result in the formation of a protein layer that inhibits corrosion by binding to sites prone to anodic dissolution resulting in a reduced dissolution rate.

6. The cathodic reaction often results in the formation of hydrogen and the build up of hydrogen tends to inhibit the cathodic reaction and restrict the corrosion process. However, biological species in the vicinity of the implant, in particular bacteria, are capable of using up the hydrogen and thus allowing the corrosion to proceed.

It seems that from the results obtained, one or more of these processes will most likely be affecting the outcome of the corrosion behaviour observed for the three titanium alloys studied.

With regards to increasing concentrations of albumin in the corrosive environment, it is known for such proteins to be only sparingly soluble and that once saturation has been reached in terms of corrosion, there should be no further influence of additional protein on corrosion [Williams D.F., 1990 (b), Szycher 1983]. Subsequently, for all three alloys it was found that no considerable difference existed for values of corrosion resistance (E_B-E_P)

between albumin additions of 1 mg/ml and 10 mg/ml, although, there was a notable difference between additions of 0.1 mg/ml and 1 mg/ml. It was also noticed that in the presence of albumin, an increase in pH (from 5 to 7.4 to 9) lowered the corrosion resistance of all three alloys and that a change from a moderate value (pH=7.4) to a higher one (pH=9) resulted in the greatest reduction. The only discrepancy in these results was that Ti-6Al-4V reduced its corrosion resistance with pH increasing or decreasing.

Albumin has an iso-electric point of 4.9, and therefore, at pH of 7.6 it has a charge of -21 mA [Szycher 1983, Fox 1957, Scott 1988]. Thus, as pH increases from 5 to 9, the protein will acquire a greater negative charge. The relevance of this in the corrosion environment is that at higher pH the presence of albumin will play a greater role in either being attracted to the metal surface and/or in transporting metal ions away. At a high pH the presence of the albumin increased the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb in comparison with PBS alone whereas it decreased the corrosion resistance of Ti-13Nb-13Zr. This could be explained by the attraction of protein to the surface and thus the exclusion of OH^- therefore decreasing the dissolution of aluminium and vanadium or the passivation of titanium for the Ti-13Nb-13Zr alloy. Since the corrosion resistance of Ti-6Al-4V also decreases at a lower pH when albumin should not be expected to be attracted to the surface (since it is near to its isoelectric point) then it may be possible that the protein is acting as a metal ion-carrier in this case.

The results obtained from surface roughness measurements (Figure 4.3.6) suggest that the presence of protein on the surface in the form of metal/protein and metal/protein/hydroxylated compounds as precipitates (ie. corrosion products) may have led to a change in the overall composition of the surface, altogether. The change in surface roughness was found to be greater following corrosion in the presence of protein than without it suggesting that not only have pits formed in the surface of the alloys as a result of corrosion, but that interactions of the protein molecules with the oxide layer has led to a change in the oxide composition, also. Observation using an optical microscope confirmed the presence of such pits.

The hardness measurements were useful in defining the extent to which the mechanical integrity of the alloys had been affected, following corrosion (Figures 4.3.7 and 4.3.8). Although in PBS alone, the hardness of Ti-13Nb-13Zr was seen to decrease to a greater extent than the other two alloys, in the presence of albumin it proved to be better in retaining its hardness compared to Ti-6Al-4V and Ti-6Al-7Nb for which the hardness decreased considerably. This phenomenon may be due to the formation of metal/protein/hydroxylated compounds in the form of precipitates at the surface slightly modifying the surface structure, especially, since the evidence showed that the surface of all three alloys softened considerably in the presence of albumin (Figure 4.3.8). This also means that although Ti-13Nb-13Zr had suffered the most in terms of corrosion with albumin present, its hardness was not altered to a large extent. This has some bearing on the argument presented earlier regarding the role of albumin adsorption at the surface of the three alloys, influencing their corrosion behaviour in different ways dependant on compositional change. With regards to this, it seems that a change in the oxide composition has occurred due to either corrosion products or the presence of proteins, thus changing the hardness of each material.

The work using additions of foetal calf serum was performed to try to evaluate the accuracy of simple in-vitro testing, using PBS solution and a single protein solution, allowing a comparison with similar testing in solutions containing serum which might reflect the in-vivo application more closely [Brown and Merritt, 1980]. The results showed that in the presence of whole serum (10%) the corrosion resistance of Ti-6Al-4V improved (in agreement with other workers [Boyer and Gall, 1985, Brown and Merritt, 1980]) and the corrosion resistance of Ti-13Nb-13Zr and Ti-6Al-7Nb reduced.

It was found that a similar pattern to that in the presence of albumin existed in these tests, only this time the effect was reduced. The reason for this may partly be due to the fact that albumin makes up between 55-62 % of the proteins present in whole serum and so its influence will still be considerable when considering the role of biological molecular interactions with proteins and their influence on corrosion as discussed earlier [Scott and Eagleson, 1988]. This argument is strengthened by the fact that albumin has strong adhesive properties and in the presence of other proteins it competes successfully during the adsorption process, although, it is known that when proteins compete for the same surface their overall adsorption capacity is reduced. This effect is brought about due to the overall change in charge as a result of the accumulation of multiple charges on different proteins.

From the surface hardness data obtained (Figure 4.3.8) for the alloys following corrosion in this environment it is apparent that, although, there is a reduction in hardness for all three alloys, Ti-6Al-7Nb was most resistant to change retaining 73% of its initial hardness, whereas, both Ti-6Al-4V and Ti-13Nb-13Zr were affected considerably more retaining only 66% and 57%, respectively.

CHAPTER 5 : WEAR-ACCELERATED CORROSION, NON-CORROSIVE AND CORROSIVE WEAR OF Ti-6Al-4V, Ti-6Al-7Nb AND Ti-13Nb-13Zr ALLOYS PHOSPHATE BUFFERED SALINE (PBS) SOLUTION WITH AND WITHOUT THE ADDITION OF BOVINE ALBUMIN AND FOETAL CALF SERUM

5.1 INTRODUCTION

When considering materials for application as orthopaedic prostheses, it is not only important to consider properties such as corrosion resistance, but also those of wear behaviour. Thus, when evaluating 'new' materials it seems reasonable to study these two phenomena, both on their own and in a co-joint manner, especially since the in-vivo situation predicts that both corrosion and wear may be a cause for concern, either acting separately along different time scales or together at any one instant [Buchanan et al., 1987, Angelini et al., 1991, Hong and Pyun, 1991, Chakraborty et al., 1991, Hedeyat et al., 1992, Cook et al., M^{c}Donald and Bloebaum, 1995, Dumbleton 1978, Buchanan et al., 1983, Thomsen et al., 1995, Kumar et al., 1990].

An investigation of how corrosion affects the wear behaviour, and also how wear may accelerate the corrosion of three titanium alloys, Ti-6Al-4V, Ti-6Al-7Nb and Ti-13Nb-13Zr. The environment within which the material operates is very important and phosphate buffered saline (PBS) with and without addition of bovine albumin (conc. 1 mg/ml) and foetal calf serum (10%) was used [Buchanan et al., 1987, Kumar et al., 1990]. By maintaining a somewhat more complicated, but controlled environment, it was possible to monitor and to analyze the reaction of each of the three titanium alloys in terms of their relative surface phenomena, especially the interaction of protein molecules, such as albumin, with each alloy under differing conditions of degradation [Williams D.F., 1986, Curtis and Forrester 1984, Kothari et al., 1995, Scott and Eagleson, 1988, Williams R.L. et al., 1990, Sousa and Barbosa, 1993, Silva et al., 1990, Clark and Williams, 1982].

When considering the action of wear in an environment containing protein molecules, it is known for such biological molecules to provide a lubricating effect towards the wear behaviour [Kumar et al., 1990]. Furthermore, in the presence of corrosion, if such protein molecules are attracted due to a larger driving force than for simple adsorption, then it may follow that such materials will be more effectively protected by the presence of protein molecules, and thus produce less severe wear scars due to enhanced lubrication.

It is important to investigate how the hardness of the materials changes as a result of non-corrosive and corrosive wear when trying to speculate what might happen to these materials following degradation. For instance, excessive loss in hardness value of the material may result in either an increase or a decrease in wear depending on the wear mechanism involved. More precisely, if simple two body wear is present, it is possible to find that more of a softer material will be removed. On the other hand, a relatively harder material may act as a source of harder abrasive particles to assist three-body abrasive wear [Williams and Cunningham, 1979, Brandes 1993]. This study investigated the changes in surface hardness, the depth of the wear scar, the weight changes and the corrosion resistance of three titanium alloys under conditions of corrosion and wear.

5.2 MATERIALS AND METHOD

Three alloys were tested, of which Ti-6Al-4V is presently used as an implantable material. The samples of this alloy were obtained from IMI (UK) Ltd.. The other two alloys were developed as alternatives to Ti-6Al-4V, and were obtained from two independent sources, Ti-6Al-7Nb from Sulzer Medical Technology Ltd., Switzerland, and Ti-13Nb-13Zr from Smith and Nephew Richards inc., U.S.A. as mentioned in section 4.2.

All materials were received in the annealed condition suitable for implantation as a medical device. The specimens were prepared for corrosion and wear testing, in a similar way, producing a fine polish using a 1/4 micron diamond paste followed by a thorough cleaning in acetone. Surface roughness measurements were undertaken for each specimen before and after corrosion using a surface profilometer (a 'Taylor-Hobson Surtronic 3' profilometer was used). Hardness measurements were also performed before and after each test using a Vickers hardness indentation machine [Williams and Cunningham, 1979, Lilley et al., 1992, Streicher et al., 1992, Poggie et al., 1994].

The pitting corrosion resistance of the alloys was evaluated quantitatively by measuring the difference between the breakdown and protection potentials, E_B and E_P respectively, as discussed in section 4.2, earlier.

<u>Wear-accelerated corrosion</u>

Part of the surface of each specimen was continuously abraded, using a simple pin-on-disc type apparatus, for 5 minutes prior to corrosion testing. Cyclic polarisation data were obtained in the form of potential vs. current density curves between a range of 0 - 5000 mV using a 'AUTOSTAT' computer controlled potentiostat. The potential was increased at a rate of 200mV/min, starting with the rest potential (E_R). Electrode potentials were measured against a saturated calomel reference electrode. Similar results were also obtained for corrosion without the prior wear degradation, and the results from these types of test were compared. Three different corrosive environments (i.e. PBS only, PBS + albumin (1mg/ml) and PBS + whole serum (10%)) were used as the electrolyte.

The surface hardness was measured for each material before and after degradation using a Vickers hardness indentation machine. The hardness measurement was made of various locations within the central region of the wear tracks which was clear of corrosion debris. An average value was obtained using at least five individual measurements for each specimen.

Non-corrosive and Corrosive Wear

For non-corrosive wear, the specimens were worn continuously for a period of 10 minutes and measurements of their weight change and surface roughness, to evaluate the depth of the wear scar, were undertaken before and after each test. To simulate a non-corrosive environment, distilled water was used as the non-corrosive medium with and without additions of albumin and whole serum. For corrosive wear, a similar procedure was used, only this time using PBS as the corrosive medium with and without additions of albumin (1 mg/ml) and whole serum (10%). All data were consolidated in order to analyze the effect of corrosion in enhancing the wear behaviour of these alloys.

5.3 RESULTS

Wear-accelerated corrosion

Figure 5.3.1 shows the pitting corrosion resistance of the alloys, within the three different environments, following corrosion and wear-accelerated corrosion. Following wear-accelerated corrosion in PBS, it was noted that the corrosion behaviour of Ti-6Al-4V and Ti-13Nb-13Zr was accelerated by the wear process by a similar extent, whereas, that of Ti-6Al-7Nb was influenced to a much greater extent (i.e. the value of E_B-E_P was seen to increase by twice as much as for the other two alloys). However, it was still evident that Ti-13Nb-13Zr alloy was the most corrosion resistant alloy of the three followed by Ti-6Al-7Nb and Ti-6Al-4V respectively, during both corrosion and wear-accelerated corrosion testing.

In the presence of albumin the corrosion resistance of Ti-6Al-4V is significantly improved whereas that of Ti-6Al-7Nb was basically unchanged and that of Ti-13Nb-13Zr was reduced. Following the wear process however, a large decrease in the corrosion resistance of Ti-6Al-4V was observed, although it was more corrosion resistant than without albumin, whereas the corrosion resistance of Ti-6Al-7Nb was affected to a far less extent such that the presence of albumin appeared to protect the Ti-6Al-7Nb surface from corrosion following the wear process. The corrosion behaviour of Ti-13Nb-13Zr following the wear process was not influenced by the presence of albumin. In the presence of albumin the corrosion resistance of Ti-6Al-7Nb and Ti-13Nb-13Zr were not significantly different either with or without the wear process and they were both more corrosion resistant than Ti-6Al-4V.

In the presence of FCS the corrosion resistance of all three alloys was slightly lower than it was in albumin solution without the wear process. Following the wear process the corrosion resistance of each alloy was reduced further and to a greater extent than in the albumin solution. The corrosion resistance of Ti-6Al-4V and Ti-13Nb-13Zr were lower following wear in FCS than following wear in PBS and that of Ti-6Al-7Nb was very similar to that following wear in PBS. That is all three alloys showed their lowest corrosion resistance following wear in FCS.

Figure 5.3.2 displays the Vickers hardness values of all three alloys, following corrosion and wear-accelerated corrosion (the hardness values of the three alloys before testing were Ti-6Al-4V = 322, Ti-6Al-7Nb = 349 and Ti-13Nb-13Zr = 353). Corrosion of the three alloys in PBS caused a slight decrease in all their hardness readings. Following the wear process, however, a significantly larger decrease in the hardness was observed for all three alloys. In the presence of albumin a much larger decrease in hardness was observed than in PBS alone under pure corrosion conditions. Under wear-accelerated corrosion conditions no significant further change in hardness was measured. It follows that the alloy surfaces had harder surfaces under wear-accelerated conditions in albumin than in the presence of PBS alone.

Corrosion in FCS produced similar hardness measurements on all three alloys to those following corrosion in albumin. Slightly larger changes were observed following wear-accelerated corrosion in FCS than in albumin with the hardness of the Ti-6Al-4V surface increasing and those of the Ti-6Al-7Nb and Ti-13Nb-13Zr surfaces decreasing.

Non-corrosive and Corrosive wear

Figure 5.3.3 summarises the weight change of the specimens, before and after non-corrosive and corrosive wear. This data showed that the only condition under which a loss of weight occurred was during wear in distilled water. There was a small weight gain following wear in a corrosive environment of PBS alone. Following the addition of albumin there was a weight gain under both non-corrosive and corrosive wear conditions. This gain being slightly larger for Ti-6Al-4V and Ti-13Nb-13Zr under corrosive than non-corrosive wear conditions. In the presence of FCS there was a further slight increase in weight. This again being slightly larger under corrosive than non-corrosive wear conditions.

Figure 5.3.4 presents the surface roughness data evaluating the depth of the wear scars produced under each condition. It is generally observed that in the presence or absence of proteins the wear scar was less deep following non-corrosive wear than following corrosive wear for all materials. It is also demonstrated that the addition of protein reduced the depth of the wear scar for all materials under both non-corrosive and corrosive wear conditions. Following corrosive wear it is generally observed that the wear scars were less deep for Ti-6Al-7Nb and Ti-13Nb-13Zr than for Ti-6Al-4V. Under non-corrosive wear this trend was not so obvious.

Figure 5.3.1 : The pitting corrosion resistance (E_B-E_P) of the alloys following corrosion and wear-accelerated corrosion in PBS with and without additions of albumin and foetal calf serum

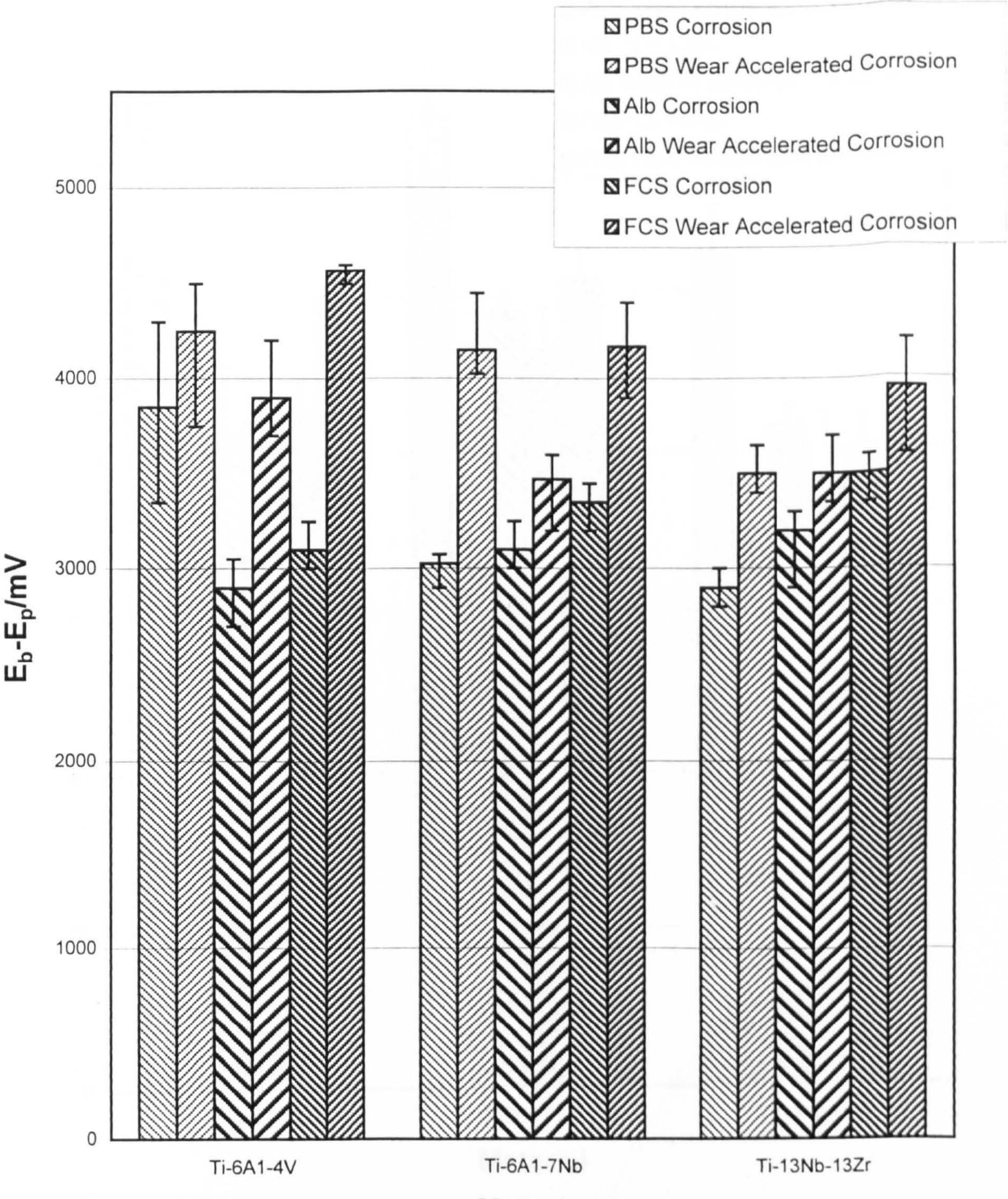


Figure 5.3.2 : The surface hardness values (VHN) of the alloys following corrosion and wear-accelerated corrosion in different environments

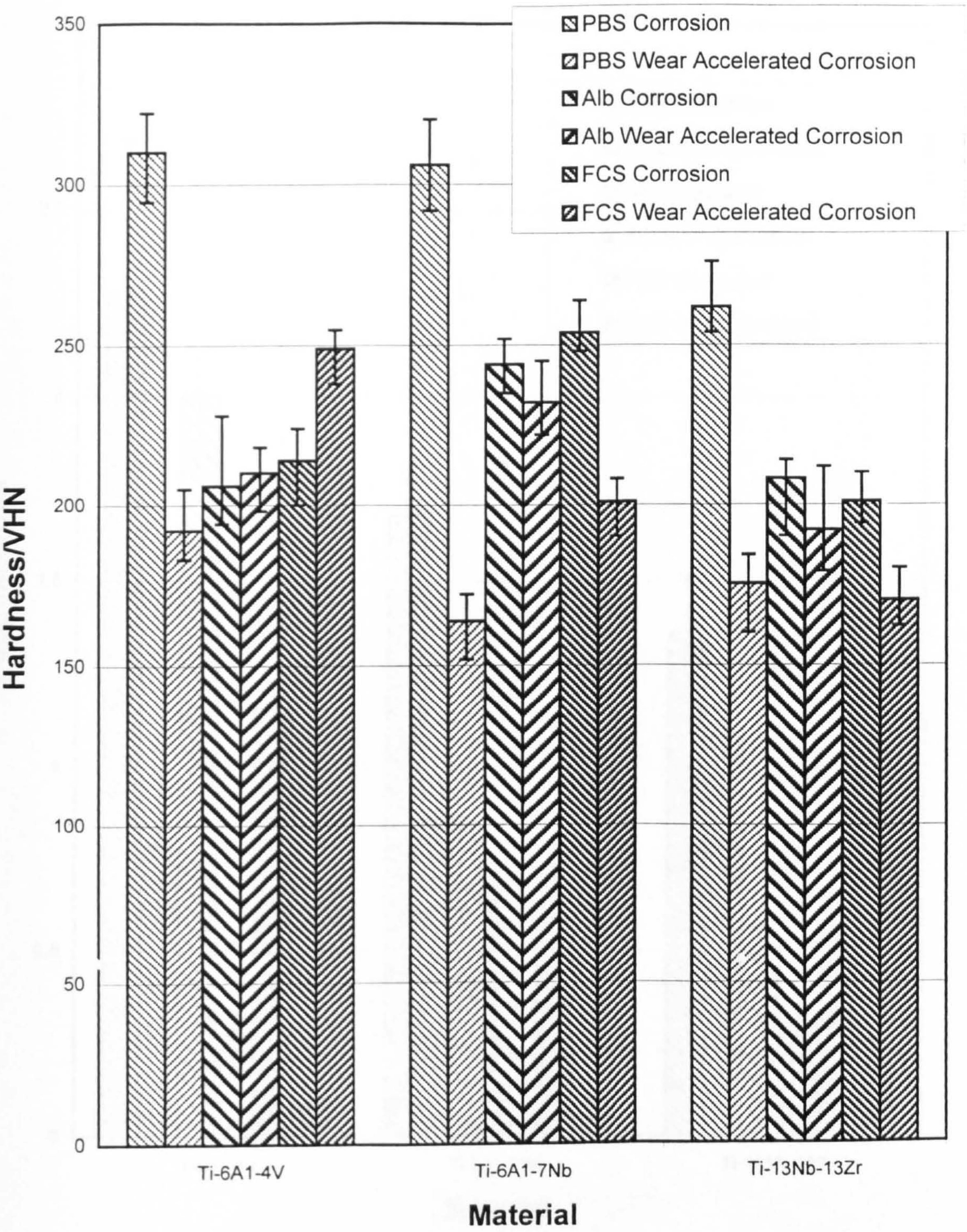


Figure 5.3.3: The weight change of the alloy specimens following non-corrosive and corrosive wear in different environments

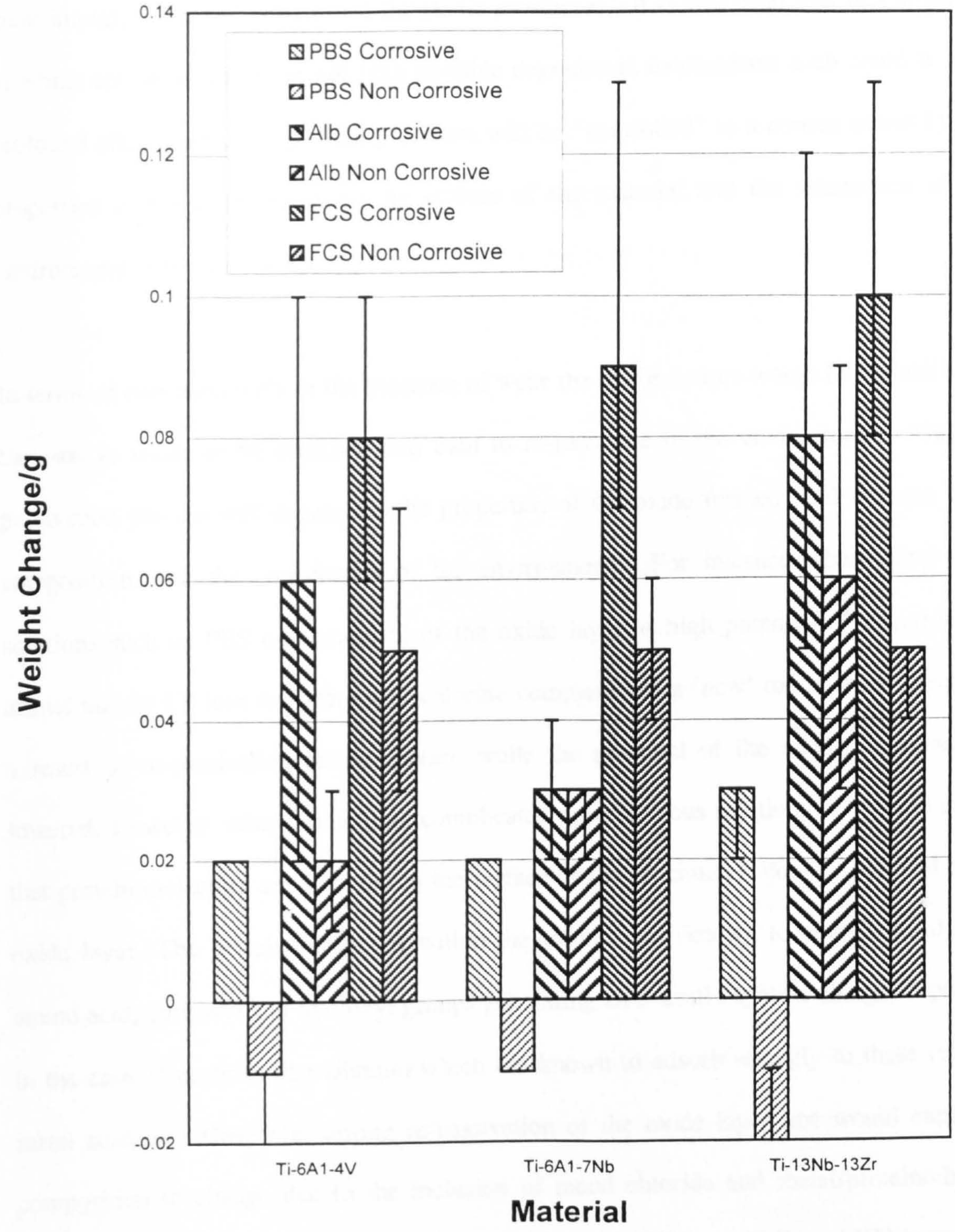
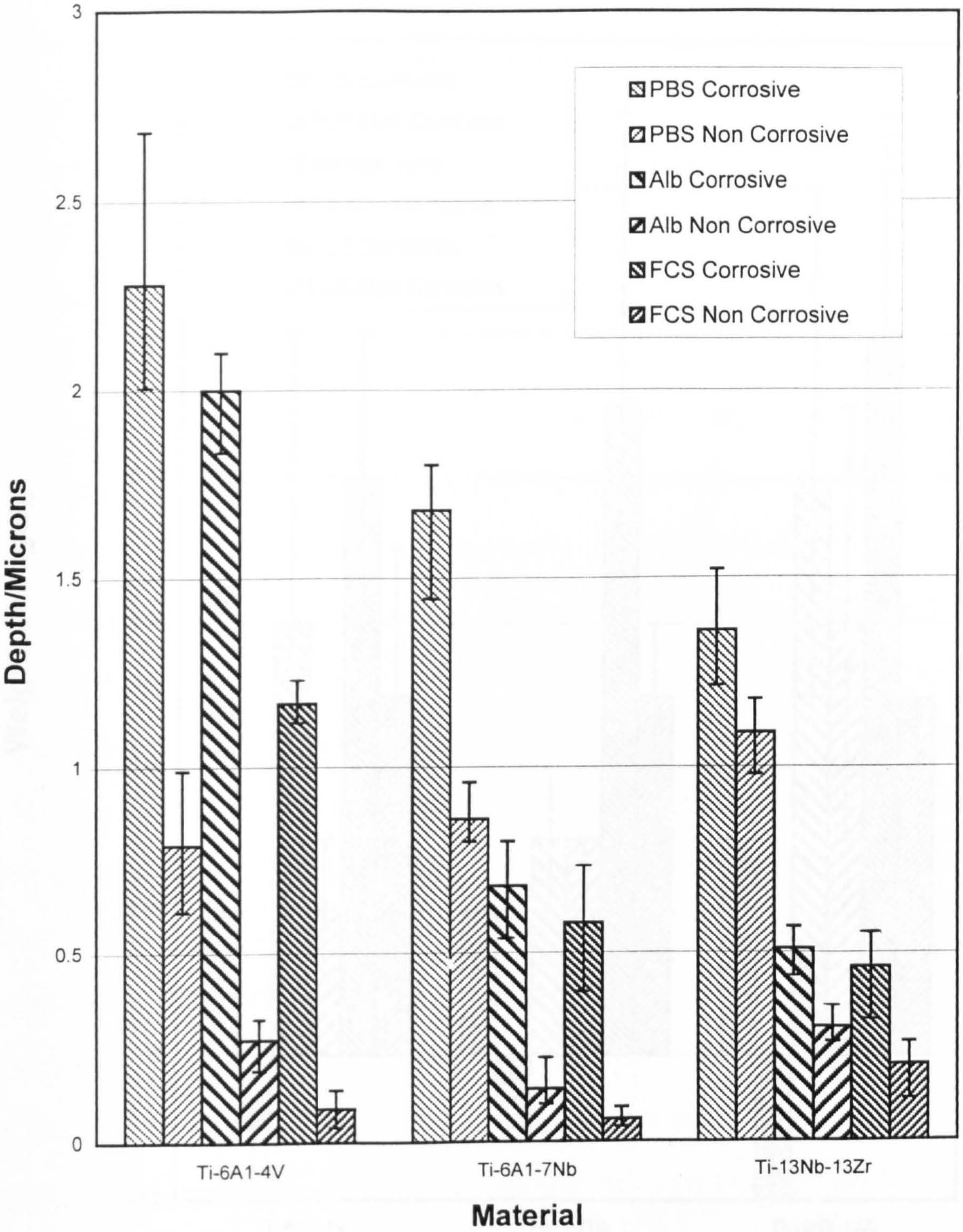


Figure 5.3.4 : The change in surface roughness (Ra) of the alloys following non-corrosive and corrosive wear in different environments



5.4 DISCUSSION

The co -joint degradation processes of corrosion and wear of metal surfaces is clearly of great importance in the design of orthopaedic prostheses. It is also clear that in a situation in which corrosion and wear are both possible degradation mechanisms each could have a profound effect on the other. Both processes will be "controlled" to a certain extent by the properties of the oxide layer on the surface of the material and the interaction of the environment with that surface.

In terms of corrosion without the presence of wear the major factors would be the ability of the passive oxide to be breached and then to re-passivate in the environment. This re-passivation process will depend on the properties of the oxide formed, and thus the alloy composition, and the constituents of the environment. For instance when using ionic solutions such as PBS only removal of the oxide layer at high potentials (pitting) would attract mainly Cl^- ions to form metal chloride complexes of a 'new' oxide layer formed as a result of re-passivation of the surface while the potential of the anode (specimen) is lowered. However, when using more complicated proteinaceous solutions one would expect that protein molecules are attracted to the surface of the specimen following removal of the oxide layer. The protein molecules within the solution are known to contain disulphide, amino acid, carboxyl and hydroxyl groups generating an overall negative charge, especially in the case of proteins like albumin which are known to adsorb strongly to these types of metal surfaces. Therefore, during re-passivation of the oxide layer one would expect its composition to change due to the inclusion of metal chloride and metal/protein/chloride complexes [Ellingsen 1991, Healy and Ducheyne, 1992, Williams D.F. and Williams R.L., 1988, Brown and Merritt, 1990]. Thus, the re-passivated oxide may have a different composition to the original oxide and this may be reflected in the hardness of that surface.

In terms of wear without corrosion the amount of wear or the susceptibility of a surface to wear will be "controlled" to a certain extent by the hardness of the surface oxide. The depth of the wear scar, however, will depend on the wear mechanism that is occurring. If the main mechanism is wear of a softer material by a harder material in abrasive wear then one would expect the softest material to wear to the greatest depth. If, however, 3-body wear could be a significant wear mechanism then the wear of a harder material could produce harder wear particles which could cause a greater amount of wear if trapped between the surfaces.

During pure wear, weight changes can be related to loss of material worn away or adsorption of material from the environment. During corrosion processes weight changes can be associated with material lost via dissolution or build up of material due to repassivation incorporating material from the environment. Therefore, under conditions of corrosion and wear the final weight change will be a complex phenomena taking all these possible processes into account.

This study aimed to investigate the co -joint action of corrosion and wear of three possible orthopaedic titanium alloys. From previous studies certain information is already known about these alloys. Ti-6Al-4V and Ti-6Al-7Nb are both alpha-beta titanium alloys with the Al component stabilising the alpha phase and thus maintaining the improved mechanical properties over pure titanium and V and Nb stabilising the beta phase to maintain the corrosion resistance. Nb has been put forward as a replacement for V owing to possible toxic effects of V [Blumenthal and Cosma, 1989, Wapner 1991, Maurer et al., 1994]. Ti-13Nb-13Zr is a near-beta titanium alloy which possesses a higher corrosion resistance to the other two alloys but still has enough alpha phase to provide the necessary mechanical properties. The increased corrosion resistance is reasoned to be due to the fact that Nb and Zr are less soluble than Al and V and that the passive oxide layer on the surface of the alloy is more inert since it consists of a dense rutile-like structure, when formed in air at room temperature, providing greater protection to the underlying alloy [Kovacs and Davidson, 1993 (a)].

Under conditions of corrosion alone in PBS it can be clearly seen that Ti-13Nb-13Zr has the highest corrosion resistance as has been reported previously [Kovacs and Davidson, 1993 (b)]. Following the addition of albumin the corrosion resistance of Ti-6Al-4V is higher than both Ti-6Al-7Nb and Ti-13Nb-13Zr and in fact albumin increases the corrosion resistance of Ti-6Al-4V and decreases the corrosion resistance of the other two alloys. Following the addition of FCS there was a decrease in the corrosion resistance of all three alloys. This demonstrates the interaction of the protein with the corrosion reactions of these three alloys and suggests a different interaction of the protein with each alloy. It is well known that proteins influence the corrosion behaviour of some metals and that their presence can either inhibit or accelerate the corrosion phenomena. The role of proteins in a corrosive environment is governed by many factors such as the surface chemistry of the metal, protein adsorption characteristics, interaction of protein molecules with other ions present in the electrolyte to produce organic complexes and the transport of anionic and cationic charges around and away from the local environment [Sousa and Barbosa, 1993, Silva et al., 1990, Clark and Williams, 1982].

When the specimens were subjected to wear immediately prior to corrosion testing it was demonstrated, as would be expected, that in PBS the corrosion resistance was reduced. This is presumably due to the removal of the surface oxide by the wear process which is not so effectively re-passivated. Changes in the surface oxide formed during re-passivation are reflected by changes observed in the hardness measurements on the materials. It is shown for all three alloys that the as-prepared specimens have the hardest surface oxides with Ti-13Nb-13Zr being the hardest and Ti-6Al-4V the lowest. Following corrosion in PBS the hardness of all the surfaces are reduced indicating a change in surface oxide structure following corrosion. Ti-13Nb-13Zr is reduced to the greatest extent and Ti-6Al-4V only minimally. Following wear-accelerated corrosion in PBS the hardness values for all three alloys were reduced even further suggesting that the wear process prior to corrosion testing had a significant influence on the surface oxide formed during re-passivation.

The presence of proteins in the environment also caused a reduction in the hardness both in pure corrosion and generally to a slightly greater extent following wear-accelerated corrosion, although this was not so for Ti-6Al-4V in FCS. It might be possible that protein from the environment was interacting in some way with the surface oxide formed during re-passivation, and it may be by this route that the corrosion resistance was reduced. In some cases it is known for protein molecules attached to the metal surface to cause preferential pitting depleting the underlying area of oxygen.

We then turned the phenomena the other way round, i.e. we investigated the influence of corrosion on the wear process. Wear was evaluated by measuring weight change and depth of wear scar in either distilled water (to represent a non-corrosive environment) or in PBS (to represent a corrosive environment). In both solutions measurements were also taken following the addition of proteins. It needs to be taken into account that only a small portion of the surface was subjected to wear but weight change measurements represent weight change of the whole specimen.

The only condition under which a loss of weight was observed was following wear in distilled water which must be due to the removal of material by the wear process. This resulted in a wear scar of 0.79um for Ti-6Al-4V, 0.86um for Ti-6Al-7Nb and 1.09um for Ti-13Nb-13Zr such that the greatest weight loss corresponded to the deepest wear scar. Although Ti-13Nb-13Zr suffered the greatest degradation via wear under these conditions we also know that it had the hardest surface oxide prior to testing. This could suggest that 3-body wear was playing a role in the wear process. Following the addition of albumin there was an increase in weight for all three alloys, this being greatest for Ti-13Nb-13Zr and least for Ti-6Al-4V. The wear scar for all three alloys under these conditions were very small. This could suggest two things: the first is that the weight gain could be due to adsorption of protein to the entire specimen surfaces and the second that the adsorbed protein could have a lubricating effect thus reducing the wear. This process was increased further by the addition of FCS rather than albumin except for the weight gain for Ti-13Nb-13Zr.

When wear occurs in a corrosive environment in all cases there is an increase in weight. The weight gain increased as albumin and further when FCS was added and increased more for Ti-13Nb-13Zr than for Ti-6Al-7Nb which was more than for Ti-6Al-4V. In PBS alone the weight gain was presumably due to the increase in oxide thickness on the total specimen surface which outweighed the loss of material at the wear scar since it was also demonstrated that a large increase in the depth of wear was measured. The weight gain increased in the order Ti-6Al-4V = Ti-6Al-7Nb < Ti-13Nb-13Zr whereas the wear scar depth decreased Ti-6Al-4V > Ti-6Al-7Nb > Ti-13Nb-13Zr, which is the opposite trend to that obtained using distilled water suggesting that the ionic molecules within PBS solution and their interaction with the surface during oxide removal seems to affect the resulting wear effect. This suggests that corrosion has less of an accelerating effect on the wear of Ti-13Nb-13Zr than on Ti-6Al-7Nb or Ti-6Al-4V.

The addition of protein further increased the weight gain of all three alloys and decreased the wear scar depths. This could be due to protein adsorption occurring or the protein becomes incorporated with the wear or corrosion products adhered to the surface and thus lubricated the wear process. Also as before, the wear scar on Ti-13Nb-13Zr is the lowest and that of Ti-6Al-4V the highest under each condition and similarly the weight gain the highest on Ti-13Nb-13Zr and the lowest on Ti-6Al-4V. In other words, a greater amount of protein was incorporated into the surface oxide or adsorbed to the surface of Ti-13Nb-13Zr than the other alloys and this resulted in a shallower wear scar.

CHAPTER 6 : THE EFFECT OF SURFACE ROUGHNESS AND FOETAL CALF SERUM ON THE CORROSION BEHAVIOUR OF Ti-6Al-4V, Ti-6Al-7Nb AND Ti-13Nb-13Zr ALLOYS

6.1 INTRODUCTION

Under normal physiological conditions, it is well known for titanium and its alloys to be corrosion resistant as has been discussed in chapter 1 [Donachie 1988, Jaffee and Promisel, 1970]. However, realizing that passive dissolution may still be occurring, it is necessary to consider the consequences of particulate material or even dissolved ions of the constituent components of the alloy present within the tissue surrounding the implant. Thus, metal ions and metal-protein complexes could still cause tissue damage, and therefore, passive corrosion rates are still a significant factor as far as the corrosion behaviour of these materials is concerned [Browne and Gregson, 1994, Ban et al., 1991, Brown and Savoury, 1983, Williams R.L., 1986, Healy and Ducheyne, 1992, Clark and Williams, 1982]. Also, as a result of the implant surface becoming worn during failure, a rough surface may be present influencing the corresponding corrosion behaviour [Cogan et al., 1994, Solar et al., 1979, Healy and Ducheyne, 1992, Jacobs et al., 1991].

Two alloys (Ti-6Al-7Nb and Ti-13Nb-13Zr) were studied alongside Ti-6Al-4V alloy to analyze how altering the surface roughness during preparation for corrosion testing might influence the stability of the materials under open circuit conditions, as well as during accelerated corrosion testing. A proteinaceous solution was also used to investigate how a change of surface roughness might affect the influence of serum proteins on the dissolution and corrosion behaviour of these alloys.

It has previously been shown, using polarization studies, that the surface roughness affects the corrosion resistance of titanium [Jaffee and Burte, 1973]. Considerable differences have also been noted for the value of breakdown potential, E_B, partly due to variations in surface roughness [Suttow et al., 1976]. Smoother surfaces have been recommended for implant use, since a finer initial polish may shorten the time necessary for the specimen to reach a steady potential (ie. the rest potential). It has been shown that Ti (cpi) is least affected by changes in surface preparation, whereas, the Ti-6Al-4V alloy was seen to become progressively more electropositive when the potential is measured following the mechanical polishing with 600 grit paper, and 0.05 micron diamond paste, respectively [Solar et al., 1979, Jaffee and Burte, 1973]. Rest potentials were found to be dependent on the thickness of the passive film, becoming more positive as the film becomes thicker. The effective area can be thought of as a measure of the number of anodic areas. Thus, the rest potential, E_R, becomes more positive as the film becomes thicker and the number of anodic areas decreases (ie. a smoother surface). It follows that as the degree of polishing is varied, the effective area decreases as the size of the polishing compound decreased reducing the number of anodic sites [Solar et al., 1979].

During implantation, titanium releases corrosion products into the surrounding tissue even though it is covered by a thermodynamically stable oxide film. Incorporation of elements from the extracellular fluid (P, Ca and S) into the oxide has been observed as a function of implantation time. Moreover, changes in the stoichiometry, composition, and thickness of the oxide have been associated with the release of titanium corrosion products in-vitro. It was found that the dissolution rate decreased as the oxide thickness increased. The dissolution reaction depended on at least two distinct chemical reactions: hydrolysis of the oxide and exchange reactions between the Ti-OH surface sites and the adsorbed P-containing species [Solar et al., 1979].

It has recently been shown for proteins to have a beneficial effect on the corrosion resistance of titanium in the absence of film breakdown. However, when attack takes place these species are known to have an adverse effect [Healy and Ducheyne 1992]. During adsorption and desorption studies it has been shown that more albumin is adsorbed onto TiO_2 than Ti. In the same study, it was found that low levels of albumin were adsorbed onto Ti-6Al-4V and Ti, and that this amount remained constant with time. In contrast, low levels were initially adsorbed onto V and TiO_2 surfaces, but the amount increased throughout the time period of the experiment [Sousa and Barbosa, 1993]. In a separate study, it was shown that electropolished titanium samples bound significantly larger amounts of both albumin and fibrinogen than their coarser mechanically polished counterparts [Walivaara et al., 1994].

As well as understanding that changes in surface roughness may influence the corrosion behaviour within the physiological environment, it is also important to bear in mind that implant materials having a higher surface roughness may produce greater frictional effects between components. With reference to this phenomenon it has been found that the surface roughness of titanium plates was higher than for stainless steel plates explaining the higher amount of metal found in tissue adjacent to titanium implants compared to stainless steel ones [Walivaara et al., 1994].

More recently, it has been shown during studies involving surface treatments on Ti-6Al-4V and Ti-5Al-2.5Fe alloys that although alloys with the same surface treatment led to a similar response with fibroblasts and osteoblasts, rougher and irregular surfaces (ie. those produced by plasma diffusion treatment or plasma assisted chemical vapour deposition) were found to be more cytocompatible. This finding actually favoured the use of rougher surfaces in view of the enhanced biocompatibility response of the implant [Torgerson and Gjerdet, 1994]. Other workers have also shown the advantages of rougher surfaces to improve the biocompatibility of the implant. It was found that smooth electropolished implants were surrounded by less bone than machined implants with similar oxide thicknesses and those that were anodically oxidized with thicker oxides [Larsson et al., 1996]. A rougher type anodized titanium surface showed a significantly thinner soft tissue capsule compared to an electropolished one [Ungersbock and Pohler, 1994]. Smooth or finely grooved titanium surfaces appear to be optimal in supporting the attachment of human gingival fibroblasts. On the other hand, sand-blasted titanium surfaces, on which cells lack focal contact were not able to support fibroblast adhesion [Hormia and Kononen, 1994]. Yet another study has shown that grit-blasted titanium provides an excellent surface for bone-implant integration [Goldberg et al., 1995].

The present study investigated how the corrosion resistance of such titanium alloys may vary as a consequence of varying the surface roughness. When considering the in-vivo situation, the surface of such materials is known to undergo wear following failure of the prosthesis, producing a rough surface exposed to the biological environment. Under such circumstances, a material that would produce a small variation in corrosion resistance may prove to be a better substitute material. On the other hand, a material whose corrosion behaviour is known to accelerate significantly as a consequence of a rough surface, may be avoided. This argument has particular relevance when considering that a rougher surface oxide layer might more easily produce deeper pits and crevices and that such irregularities in the surface may further accelerate any following corrosion/wear. This may also result in a greater amount of particulate material present in the surrounding tissue presenting increasing problems of toxicity.

6.2 MATERIALS AND METHOD

The specimen discs were mounted in preparation for polishing (as described in section 4.2). All samples were initially polished down using different grades of grit paper, 180, 300, 600, and 1200 respectively. Two grades of surface roughness were investigated during this study. Firstly, at a coarse specification using 180 grit emery paper and secondly by continuing to polish the specimens finely down to 1/4 micron diamond paste specification. A surface profilometer ('Taylor-Hobson Surtronic 3' profilometer) was used which involved a direct contact method with the surface in the form of a diamond stylus which measured the surface roughness profile of each specimen to within 0.01 micron. Triplicate measurements were undertaken in all cases to ensure reproducibility. An insulating varnish was applied around the edges of each specimen surface to leave a well defined area (ie. an area of square dimensions, 7mm x 7mm) exposed to the electrolyte environment. A constant area of 0.49 cm^2 was used throughout the study for all specimens.

During the rest potential studies, each specimen was exposed to the electrolyte environment under open circuit conditions, for a period of up to 180 minutes. During this time, potential measurements were undertaken every 10 minutes until a steady potential change of approx. 2mV/min was observed. At this time, the potential was noted and defined as the rest potential (E_R) [Thompson et al., 1979, Meyer and Reclaru, 1995]. For each material in each environment, this procedure was repeated three times and from the data obtained, an average value for E_R was accepted.

Having attained rest potential, the potentiostat was switched on to undergo an accelerated corrosion test between 0 and 5000 mV. Cyclic polarization curves were obtained for each material for the two surface roughness specifications within each of the two biological environments. To simulate such biological environments, a phosphate buffered saline (PBS) solution was prepared to which additions of foetal calf serum (10%) were added when required. In all cases, the electrolyte cell was maintained at 37°C throughout the tests using a suitable water bath. All testing was carried out under aerated conditions. The electrolyte cell was made of perspex and a volume of 200 ml was used for the electrolyte. The specimen was used as the 'working' electrode (anode) and was placed in the cell facing a platinum 'auxiliary' electrode (cathode). A saturated calomel electrode was used as the 'reference' electrode.

6.3 RESULTS

Figure 6.3.1 consists of a histogram of the mean values obtained from the data in table 6.3.1 as well the standard deviation in values noted from one sample to the next. It was noted for Ti-13Nb-13Zr alloy to produce a slightly rougher surface finish with respect to the other two alloys following both smooth and coarse polishing.

Figure 6.3.2 shows a histogram displaying the values of rest potential (E_R) of the alloys obtained from table 6.3.2. Figure 6.3.3 uses a histogram to outline the corresponding times taken to reach the rest potential (obtained from the data in table 6.3.3), with respect to each environment and surface preparation tested. In an environment of PBS alone, it was found that the time taken to attain E_R was reduced in the same way for each alloy when considering a change in surface roughness. Generally, it was noted that a longer time was required for E_R to be attained for specimens following a coarse polishing method compared to those following a fine polish (ie. approx. 1.5 times longer). The addition of serum on specimens following a coarse polishing procedure (180 grit) indicated that the time taken was reduced by 1/3 of that when considering PBS alone. However, when using a smooth surface roughness only Ti-13Nb-13Zr was affected with E_R now being attained in just 1/3 the time of that in PBS alone, whereas, for the other two alloys the time taken to reach E_R was more or less the same.

When analyzing the values of rest potential it was noticed that, in PBS alone, the E_R of all three alloys became more electropositive when changing from 180 grit to 1/4 micron polish as found by other workers when testing similar materials [Solar et al., 1979, Jaffee and

Burte, 1973]. At 1/4 micron polish specification, the addition of serum proteins made the E_R value of Ti-6Al-4V more electropositive, whereas, that of Ti-6Al-7Nb and Ti-13Nb-13Zr more electronegative in that order of increasing effect. When considering the addition of serum proteins at 180 grit specification, it was found for the E_R value of Ti-6Al-4V to become more electronegative, whereas, that of Ti-6Al-7Nb and Ti-13Nb-13Zr became increasingly more electropositive in that order.

Figure 6.3.4 and table 6.3.4 outline the pitting corrosion resistance (E_B-E_P) of the alloys at the two different surface roughness specifications within the two different environments. In PBS alone, when using a 1/4 micron polish, the corrosion resistance (E_B-E_P) is such that Ti-6Al-7Nb > Ti-13Nb-13Zr > Ti-6Al-4V. When using a 180 grit polish, the corrosion resistance of Ti-13Nb-13Zr > Ti-6Al-7Nb > Ti-6Al-4V. Ti-13Nb-13Zr alloy is least affected by a change in surface roughness followed by Ti-6Al-4V and Ti-6Al-7Nb in that order. At 1/4 micron specification, the presence of serum proteins, reduces the corrosion resistance of Ti-6Al-4V and has relatively no effect on both Ti-6Al-7Nb and Ti-13Nb-13Zr alloys. At 180 grit specification, the effect of serum proteins, improved the corrosion resistance of Ti-6Al-4V, has relatively no effect on Ti-6Al-7Nb and slightly reduced that of Ti-13Nb-13Zr alloy.

Figure 6.3.5 and table 6.3.5 display the values of breakdown potential, E_B, for the alloys in simple saline and proteinaceous solutions with both a smooth (1/4 micron) and rough surface (180 grit) finish. When evaluating the effect of serum proteins at a 1/4 micron polish specification, it was found that the E_B values decreased for Ti-6Al-4V, stayed more or less the same with Ti-6Al-7Nb and increased with Ti-13Nb-13Zr. At 180 grit specification, The

E_B value increased for Ti-6Al-4V, stayed the same once again for Ti-6Al-7Nb and increased once again for Ti-13Nb-13Zr. It was found that in PBS alone, the value of E_B is lowered following corrosion at the rougher surface specification compared to a smooth one, with Ti-6Al-7Nb having the lowest value followed by Ti-13Nb-13Zr and Ti-6Al-4V in that order. In a medium containing serum proteins the E_B value of Ti-6Al-7Nb is once again lowered, considerably, with that of Ti-13Nb-13Zr and Ti-6Al-4V showing an increase in their respective E_B values using a rougher surface specification compared to that with a smooth surface finish.

Table 6.3.1: The surface roughness profiles (Ra) of the alloys at a varying surface roughness

Alloy	**¼ Micron Polish (Microns)**		**180 Grit Polish (Microns)**	
	X	**$\bar{X}$**	**X**	**$\bar{X}$**
Ti-6Al-4V	0.03	0.027	0.27	0.28
	0.03		0.28	
	0.02		0.29	
Ti-6Al-7Nb	0.04	0.037	0.26	0.253
	0.04		0.25	
	0.03		0.25	
Ti-13Nb-13Zr	0.07	0.073	0.31	0.327
	0.08		0.33	
	0.07		0.34	

Figure 6.3.1 : A histogram showing the mean surface roughness profile values (Ra) of all three alloys at a varying surface roughness

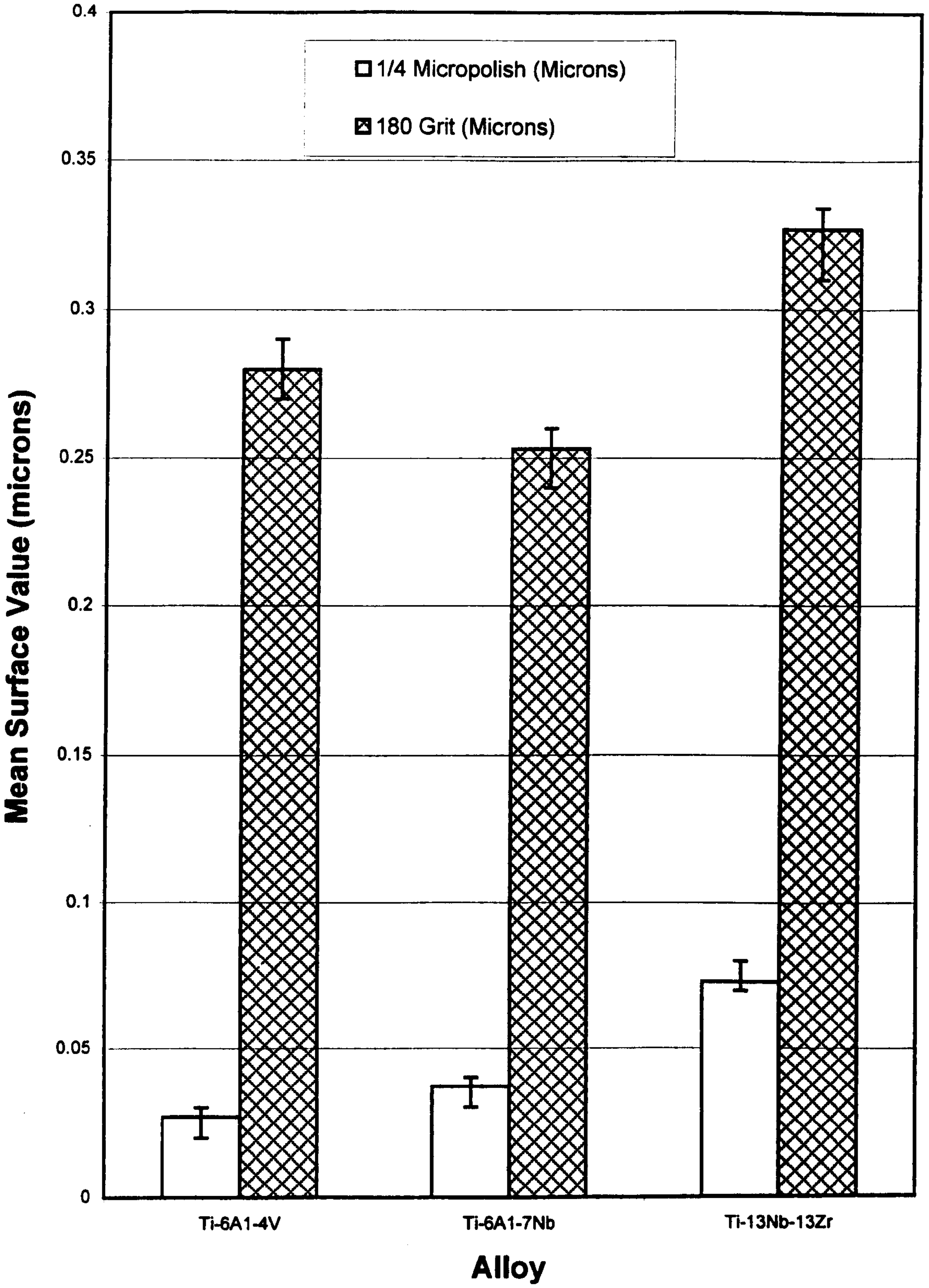


Table 6.3.2: The rest potential (mV) of the alloys at varying surface roughness

Alloy	**PBS (¼)**		**PBS (180 grit)**		**PBS + Serum (¼)**		**PBS + Serum (180 grit)**	
	X	**$\bar{X}$**	**X**	**$\bar{X}$**	**X**	**$\bar{X}$**	**X**	**$\bar{X}$**
Ti-6Al-4V	- 249	-242	-317	-309	-122	-163	-478	-465
	- 280		-276		-187		-441	
	- 198		-333		-180		-476	
Ti-6Al-7Nb	- 97	- 93	-186	-181	-226	-259	- 98	-94
	- 110		-151		-281		-93	
	- 72		-205		-270		-91	
Ti-13Nb-13Zr	- 191	-156	-295	-268	-333	-358	-80	82
	- 131		-231		-351		-81	
	- 147		-277		-390		-86	

Figure 6.3.2 : Mean rest potential (mV) values of alloys at a varying surface roughness

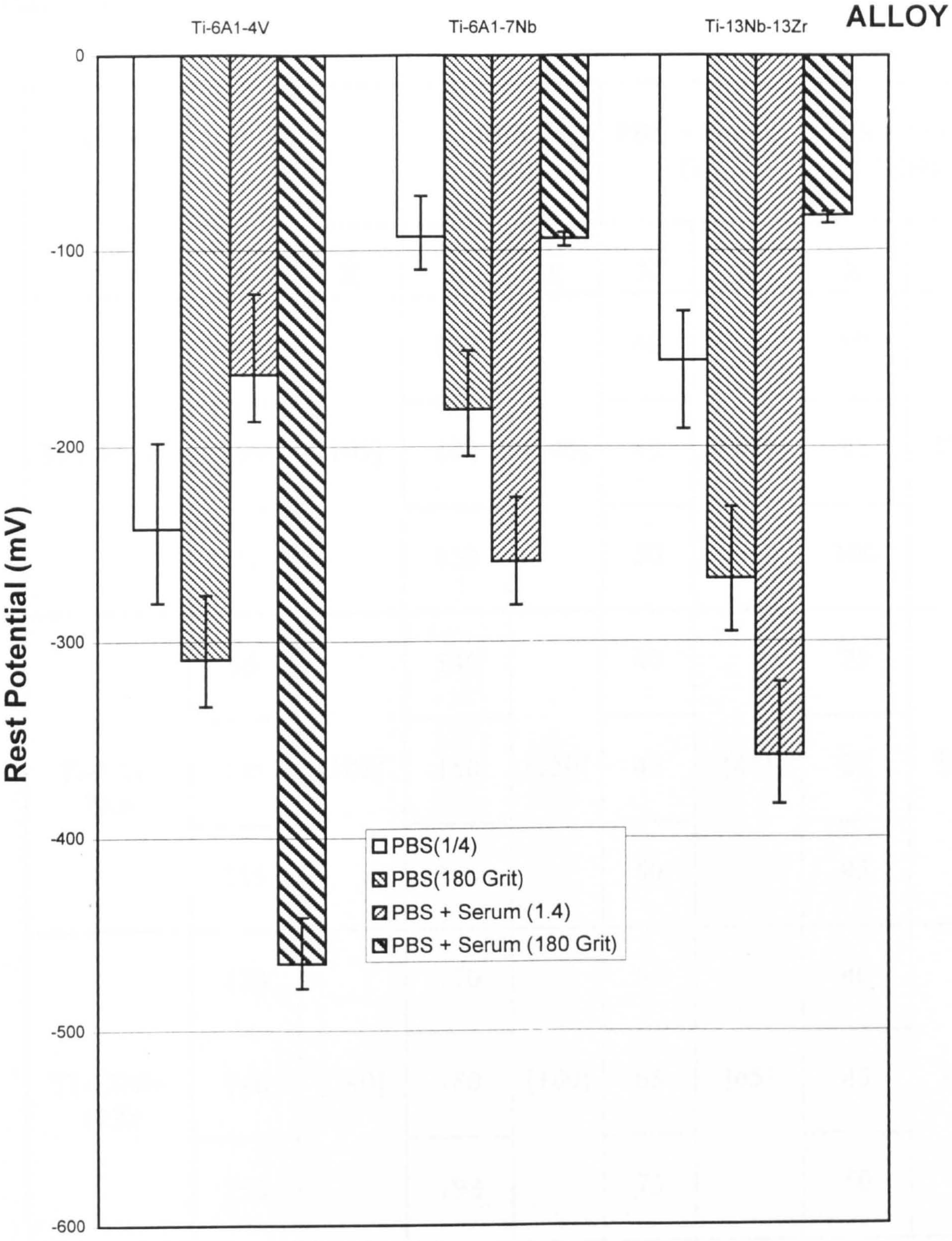


Table 6.3.3: The times taken (mins) to reach rest potential for all three alloys at varying surface roughness

Alloy	PBS (¼)		PBS (180)		PBS + Serum (¼)		PBS + Serum (180)	
	X	$\bar{X}$	X	$\bar{X}$	X	$\bar{X}$	X	$\bar{X}$
Ti-6A1-4V	95	[105]	130	[140]	40	[45]	80	[90]
	105		140		45		90	
	115		150		50		100	
Ti-6A1-7Nb	95	[105]	140	[150]	40	[45]	75	[85]
	105		150		45		85	
	115		160		50		95	
Ti-13Nb-13Zr	130	[140]	170	[180]	55	[65]	40	[45]
	140		180		65		45	
	150		198		75		50	

Figure 6.3.3 : A histogram showing mean times taken (mins) to reach rest potential for alloys at varying surface roughness

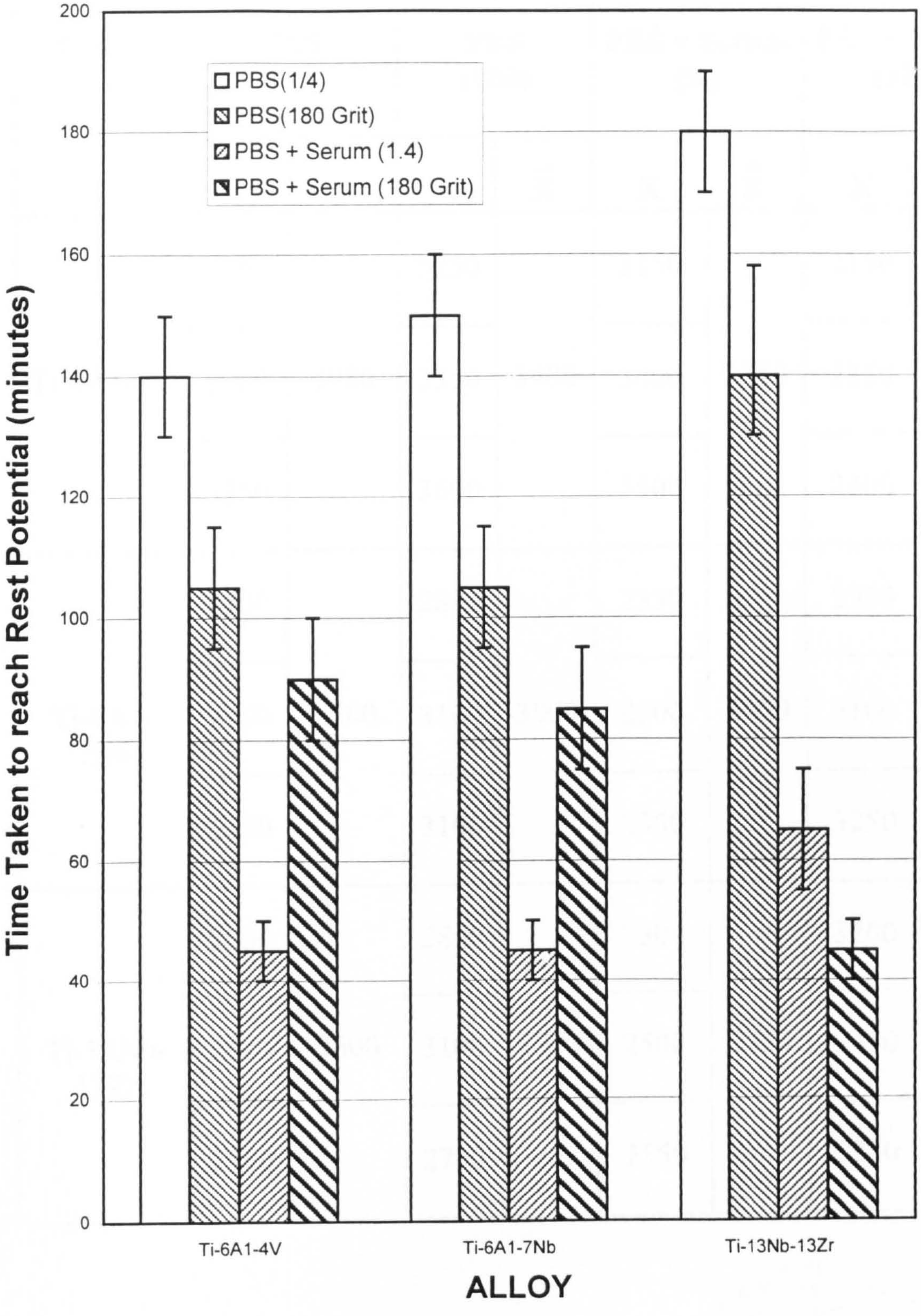


Table 6.3.4: The pitting corrosion resistance (E_B-E_P) of the alloys at a varying surface roughness within varying biological environments

Material	**PBS (¼)**		**PBS (180)**		**PBS + Serum (¼)**		**PBS + Serum (180)**	
	X	**$\bar{X}$**	**X**	**$\bar{X}$**	**X**	**$\bar{X}$**	**X**	**$\bar{X}$**
Ti-6A1-4V	2700	2950	3250	3400	3150	3350	2150	2300
	2900		3350		3400		2350	
	3250		3600		3500		2400	
Ti-6A1-7Nb	2100	2200	2800	3000	2350	2400	2950	3100
	2350		3100		2500		3100	
	2150		3100		2350		3250	
Ti-13Nb-13Zr	2500	2600	2850	2900	2300	2450	3200	3300
	2600		3100		2500		3450	
	2700		2750		2550		3250	

Figure 6.3.4: A histogram showing the pitting corrosion resistance (E_B-E_P) of the alloys at varying surface roughness within varying biological environments

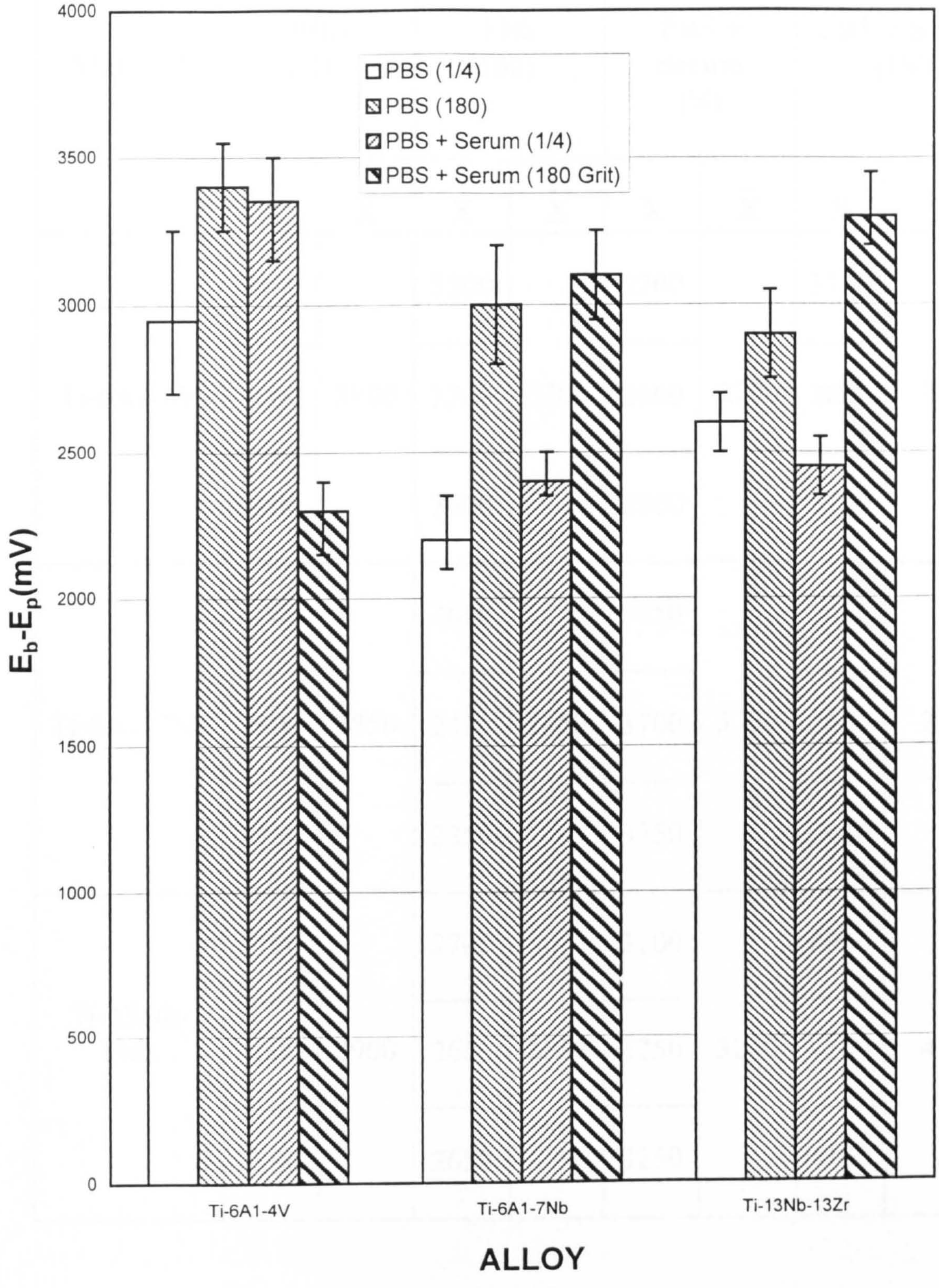
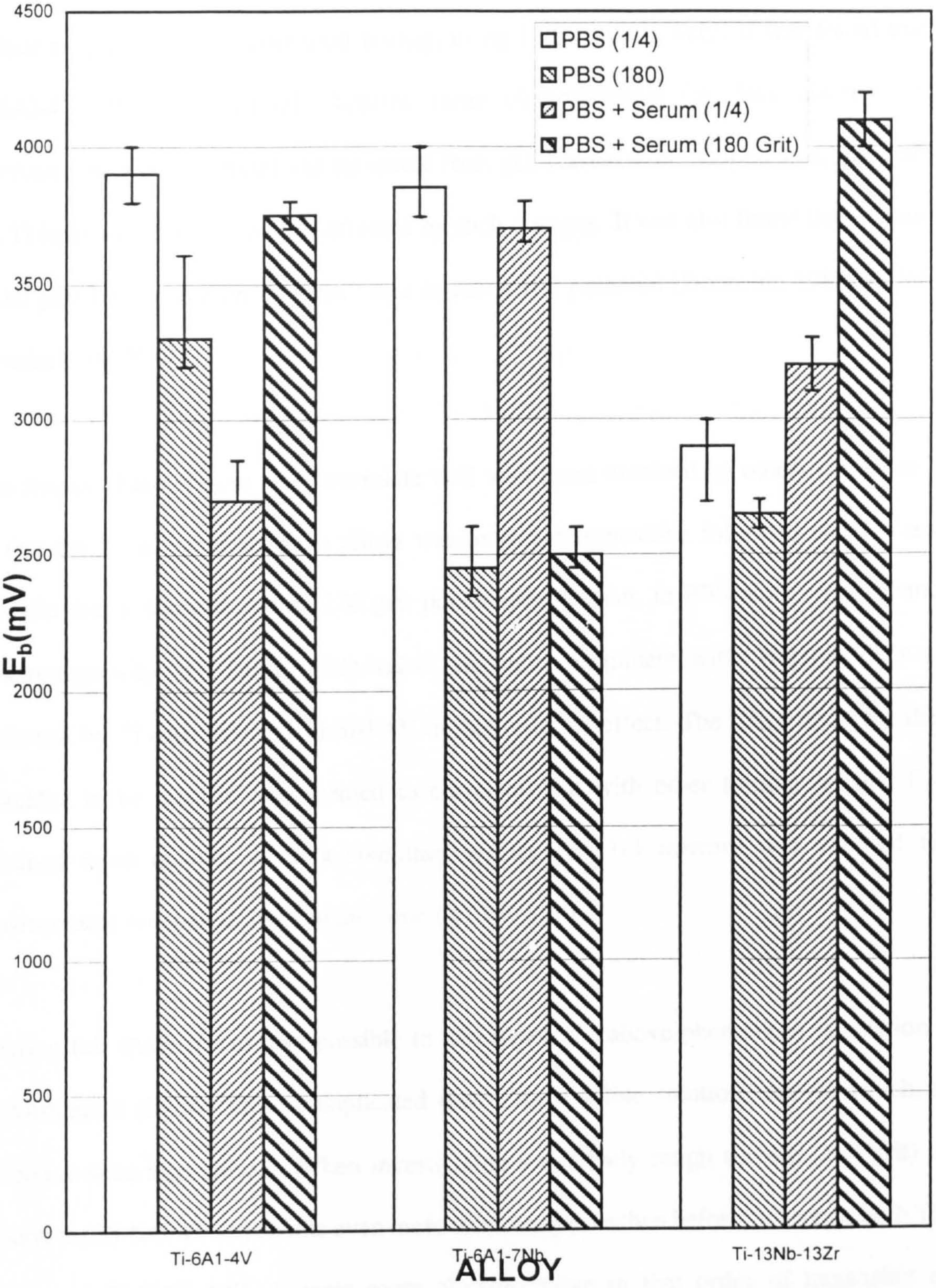


Table 6.3.5: The breakdown potential values (E_B) at a varying surface roughness

Material	PBS (¼)		PBS (180)		PBS + Serum (¼)		PBS + Serum (180)	
	X	$\bar{X}$	X	$\bar{X}$	X	$\bar{X}$	X	$\bar{X}$
Ti-6A1-4V	3800	3900	3200	3300	2700	2700	3700	3750
	4000		3200		2800		3800	
	3900		3600		2900		3750	
Ti-6A1-7Nb	4000	3850	2600	2450	3650	3700	2450	2500
	3800		2400		3700		2600	
	3750		2350		3750		2450	
Ti-13Nb-13Zr	2700	2900	2700	2650	3100	3200	4000	4100
	3000		2600		3250		4200	
	3000		2650		3250		4100	

Figure 6.3.5: A histogram showing the variations in E_B values due to a change in surface roughness



6.4 DISCUSSION

Other workers have performed similar studies with Ti(cpi) and Ti-6Al-4V alloy varying the surface roughness before corrosion testing, using PBS solution only. It was found that the Ti-6Al-4V alloy progressively became more electropositive (ie. less reactive to the environment) as the potential was measured from grit-blasted to electropolished, respectively, but Ti (cpi) was found to be less affected by such changes. It was also found that a smoother initial polish may shorten the time taken to reach rest potential [Donachie 1988, Jaffee and Promisel, 1970].

The results of the present study correlate well with those obtained by others earlier in PBS, in that the E_R value of all three alloys was more electropositive following the 1/4 micron specification compared to the 180 grit polish specification, in PBS alone. This change in electronegativity to a more noble value was most prominent with Ti-13Nb-13Zr alloy, followed by Ti-6Al-7Nb and Ti-6Al-4V in that order of effect. The times taken for the rest potential to be reached also seemed to correlate well with other findings in that E_R was attained more quickly when a smoother surface (ie. 1/4 micron) was exposed to the environment compared to a rougher one (ie. 180 grit).

During this study it was also possible to investigate the above phenomena in relation to an environment slightly more complicated than simple saline solution, one in which serum proteins were also present. When investigating a relatively rough surface (180 grit) it was found for Ti-6Al-4V to become even more electronegative than before, whereas, both Ti-6Al-7Nb and Ti-13Nb-13Zr became more electropositive in that order of increasing effect.

Finally when correlating the differences noticed between 1/4 micron and 180 grit in an environment containing proteins it was noticed that Ti-6Al-4V produced a more electropositive potential on a smoother surface as was the case in PBS alone, only to a much greater extent. Ti-6Al-7Nb and Ti-13Nb-13Zr alloys were found to become more electronegative in that order.

Although, there are numerous factors which will be governing both the time taken to reach rest potential as well as the direction of change and final value, there are certain trends which have become apparent from this study. Generally it seems that the presence of proteins slows down the effect of passive dissolution, probably due to adsorption. There is also a definite correlation between the influence of these proteins at the metal oxide surface and how this varies depending on the surface roughness of any one particular alloy. In this respect, the results suggest that Ti-13Nb-13Zr alloy takes slightly longer to reach E_R at 180 grit, whereas, at 1/4 micron it reaches a stable potential relatively quickly compared to the other two alloys. Both Ti-6Al-4V and Ti-6Al-7Nb reach E_R in a shorter time to Ti-13Nb-13Zr at 180 grit specification (same time as Ti-13Nb-13Zr at 1/4 micron), however, at 1/4 micron they both take twice as long as Ti-13Nb-13Zr to reach E_R.

In an environment containing PBS only, when considering a change from a smooth (1/4 micron) to a rough (180 grit) surface roughness, the corrosion resistance of all three alloys was reduced. On the other hand, in an environment containing PBS + serum, a change from a 1/4 micron to 180 grit showed that the corrosion resistance of Ti-6Al-4V improved, whereas, that of Ti-6Al-7Nb and Ti-13Nb-13Zr reduced in that order of effect. These results suggest that a smooth surface finish would produce a more favourable response in terms of corrosion resistance, for Ti-6Al-7Nb and Ti-13Nb-13Zr alloys both in the absence and in the presence of proteins. This suggests that the role of proteins at the surface of such materials may be different during corrosion simply as a consequence of the surface being rougher. Generally, it was found that more noble values of E_R produce a better corrosion resistance (ie. a smaller value for E_B-E_P). Therefore, it might be suggested that rest potential values be used to predict the corrosion resistance of such materials in well defined, but varying environments.

It was noticed from the results of this study that the breakdown potential, E_B, was relatively high in both a non-protein and protein environment, for Ti-13Nb-13Zr alloy, at a rough surface specification. At a glance this may actually be a promising sign knowing that the oxide layer is more dense and therefore does not as easily breakdown [Kovacs and Davidson, 1993]. However, when considering the results obtained about pitting corrosion resistance, a greater difference in the values of E_B and E_P make it possibly more prone to pitting corrosion in such an environment. An analysis of the breakdown potential values (E_B) revealed that Ti-6Al-7Nb alloy remained quite stable when considering the addition of proteins into the environment using a 1/4 micron polish specification. However, its E_B value fell significantly as a result of a change in surface roughness following a 180 grit polish specification, with and without the presence of proteins. The E_B value of Ti-13Nb-13Zr alloy improved in a protein environment, with the greatest increase noticed as the surface became rougher. Finally, Ti-6Al-4V alloy seemed to produce a relatively higher value of E_B compared to the other two alloys, when considering a PBS environment, and even in the presence of proteins when using a 1/4 micron polish.

General Comments

When undertaking an analysis of surface roughness factors, it was intended to become aware of how simple preparation or handling methods may influence the overall breakdown characteristics of these materials. It is often documented in clinical trials of implants, how retrieved prostheses appear abraded with coarse surface finishes during optical examinations. It has been shown during this study that the corrosion behaviour of these alloys may vary considerably if surfaces are rougher. It has also been shown during this study that the overall breakdown characteristics of the material depends upon the rest potential (E_R). If this value is found to be less noble during implantation due to changes in surface roughness, then this may enhance the on-set of corrosion at a lower breakdown potential (E_B). To summarize, it is a valid point to suggest that more scientific evidence be established regarding surface properties, and that alloys be designed or manufactured which only show a slight variation in corrosion and wear behaviour as result of surface roughness change.

CHAPTER 7 : THE INFLUENCE OF HUMAN MACROPHAGE CELLS ON THE CORROSION BEHAVIOUR OF Ti-6Al-4V, Ti-6Al-7Nb AND Ti-13Nb-13Zr ALLOYS IN PBS SOLUTION CONTAINING FOETAL CALF SERUM (10%)

7.1 INTRODUCTION

It has been suggested on numerous occasions that *in-vitro* testing of materials to understand how metals and alloys behave during accelerated corrosion tests should be carried out using proteinaceous solutions. It has been argued that materials behave differently when exposed to an environment containing such biological macromolecules compared to that when using simple saline solutions alone. This argument is justified by the fact that proteins can either inhibit or accelerate the corrosion behaviour of metals in a number of ways. It also seems more appropriate to undergo testing in an environment which mimics the *in-vivo* situation more closely.

Titanium and its alloy, Ti-6Al-4V, have now been used quite extensively as biomaterials for many years. Although, under normal physiological conditions it is known for such materials to remain resistant to corrosion, it has been noted on numerous occasions that these materials do breakdown under the influence of accelerating factors such as wear. Moreover, the passive dissolution of the constituent metal ions have been found in the form of corrosion products within the tissue surrounding the implant.

It is for this reason and the controversy surrounding the use of aluminium and vanadium as potentially toxic alloying elements that "new" alloys have been shown to cause less adverse biological response, to be more corrosion resistant and in some cases have a lower stiffness and greater tensile strength.

During this study it was possible to test two such alloys alongside Ti-6Al-4V alloy (ie. Ti-6Al-7Nb and Ti-13Nb-13Zr) and directly compare the corrosion behaviour of all three alloys. In a previous study, the effect of the presence of serum proteins such as albumin was also studied. To continue to try to replicate the *in-vivo* situation in a controlled manner and to try to understand the exact mechanisms leading to metal dissolution and breakdown of such materials, an electrolyte containing serum proteins was used with and without the presence of non-activated human macrophage cells. Not only would this compare the alloys in a more closely simulated *in-vivo* environment, but, also by monitoring the change in their rest potential it might be possible to understand how macrophage cells affect the stability of these materials in such an environment. Also, if during accelerated corrosion testing the cells are interfering with the proteinaceous environment they may be influencing the corrosion resistance of the alloys.

It has been noted on occasions that whenever titanium alloy particles have been found in tissue surrounding the implant, either, in the form of corrosion or wear particles that an inflammatory response occurs involving the presence of phagocytic cells such as macrophages. Sometimes in a more severe inflammatory response macrophages have been known to be present in the form of multi-nuclear giant cells when they fuse together to try to enhance their phagocytic activity [Williams D.F., 1986, Williams R.L., 1986, Maloney et al., 1996, Goodman et al., 1994, Howie et al., 1993, Williams D.F., 1981].

During recent studies in-vitro, Ti-6Al-4V and Ti-6Al-7Nb alloys were compared to detect differences in the cellular response to wear particles. It was found that Ti-6Al-4V alloy particles stimulated the release of PGE_2, IL-1, TNF (alpha and beta) and IL-6 from human peripheral blood monocytes and rodent peritoneal macrophages to a greater extent than for Ti (cpi) and Ti-6Al-7Nb alloy [Haynes et al., 1993, Shanbhag et al., 1994, Rogers et al. 1997].

Macrophage cells are known to become activated releasing enzymes which may interact with the other biological species in the environment, leading to reactions with corrosion and wear products. The interaction of these components, may be the cause of black staining of tissue surrounding a titanium implant [Scales 1991, Korovessis and Repanti, 1994]. More precisely, macrophages play several major roles in the inflammatory response. They may assist in controlling acute inflammation by their release of monokines such as interleukin-1. Also, macrophages affect any inflammatory response by the regulation of both T- and B-lymphocytes, and fibroblasts.

The next few paragraphs will try to summarize the role of the various enzymes that are released during activation of the macrophage cell.

In response to bacteria, viruses, other pathogens or possibly tumours, cells of the immune system such as macrophages can produce cytokines which are vital to the maintenance of cellular homeostasis. One such cytokine, initially identified as a product of macrophages, is tumour necrosis factor - α (TNF-α). It is actively involved in regulation of the immune system and is a major effector molecule in macrophage-mediated tumour cell toxicity. Since the production of TNF-α in macrophages is not constitutive but requires activation, it seems an appropriate cytokine to study when trying to analyze the activation of macrophages.

Interleukin-1 (IL-1) has been isolated in two forms: IL-1(alpha) and IL-1(beta). Both of these polypeptides have been found in human and other animal species. These two forms represent the products of two genes synthesized as propeptides of 270 and 269 amino acids, respectively. A number of cell types have been shown to produce IL-1, including monocyte and macrophage cell lines, natural killer cells (NK) and neutrophils. Some of the reported IL-1 activities include: the activation of T-cells in the presence of co-stimulation through the T cell receptor; augmentation of IL-2 and interferon(IFN)-induced activation of NK-mediated cytotoxicity; osteoblast and endothelial cell activation; By acting on vascular endothelial cells, IL-1 influences the inflammatory response, and affects the thermoregulatory centre in the hypothalamus leading to the fever response [Kuby, 1994]. The actions of IL-1 on connective tissue cells suggest a central role in recovery of tissue after injury, promoting extracellular matrix turnover and allowing proliferation and migration of cells. IL-1 has also been implicated *in-vivo* in the pathogenesis of a variety of acute and chronic inflammatory diseases such as septic shock, rheumatoid arthritis, and atherosclerosis; much of our view of the physiological role of IL-1 is based on *in-vitro* studies and the effects of injection of IL-1 into animals [Dower et al., 1992, Danirello, 1989].

Another important cytokine with potential involvement is Interleukin-4 (IL-4) which exhibits a multitude of *in-vitro* properties. Some of the reported activities of IL-4 are: the inhibition of release of inflammatory mediators (eg. TNF, IL-1, IL-8, PGE_2) from activated monocytes; a growth co-stimulator for B cells, T cells and mast cells; upregulates numerous activation antigens on B cells and T cells; augments cytolytic T cell differentiation; induces activated B cells to switch to IgE synthesis, and specifically enhances secretion of human IgG_4 and IgG antibodies; [Finkelman et al., 1990, Mosmann and Coffman, 1989, Tepper et al., 1989, Golumbek et al., 1991]. Additionally, IL-4 may act as an autocrine growth factor for antigen-specific T cells in combination with IL-1 and enhance antigen presentation and phagocytosis in macrophages [Zlotnik et al., 1992].

Interleukin-6 (IL-6) is a novel cytokine which is multi-functional acting on B-cells, T-cells, monocytes, haematopoietic stem cells, hepatocytes and nerve cells. Thus, IL-6 may play a central role in infections, inflammation and tissue injury. IL-6 is produced in various tissues and it has been suggested that its deregulated production may be involved in the pathogenesis of auto-immune diseases, lymphoid malignancies, and multiple myeloma. IL-6 is produced by many different types of lymphoid and non-lymphoid cells, either constitutively or in response to various stimuli. The production of IL-6 by normal T-cells is monocyte dependent, thus, it is a useful measure of cytotoxicity when studying an environment containing T-cells [Zembala M. and Asherson G.L., 1989].

Previous studies have reported the presence of numerous T-lymphocytes and no B-lymphocytes in tissues where metallosis was evident. The presence of an abundance of activated macrophages was also reported in these tissues [Lalor and Revell, 1993, Lalor et al., 1991, Santavirta et al., 1991]. These studies suggest that T-lymphocytes are the likely cell for activated macrophages to be interacting with, and that there is some form of immunological reaction occurring in tissues in which metallosis has occurred. However, the fact that this response is due to the debris is still not conclusive. Thus, the presence of T-lymphocytes in the environment alongside macrophages is very closely related to the *in-vivo* situation.

Finally, macrophages phagocytose cellular and molecular debris, and also detoxify and /or sequester toxic materials [Vince et al., 1991 and Williams D.F., 1994]. The presence of macrophages during the inflammatory response in the ways described has also been linked to the idea that such cells play a major role in the process leading to aseptic loosening of femoral implants and bone resorption [Williams D.F., 1994, Pizzoferrato et al., 1991, Kim et al., 1994, Shanbag et al., 1994, Horowitz and Purdon, 1995, Cogan et al., 1994]. Thus, it becomes quite clear why it is an important issue to study their presence within a corrosive environment.

7.2 MATERIALS AND METHODS

Corrosion testing

The titanium alloy specimens were all obtained in the annealed condition. The Ti-6Al-4V alloy was supplied by IMI Titanium Ltd., England; the Ti-6Al-7Nb alloy from Sulzer Medical Technology Ltd., Switzerland; and the Ti-13Nb-13Zr alloy from Smith and Nephew Richards Inc., U.S.A.;

The specimens were received as 1.0 cm long discs of between 1 and 1.5 cm diameter. After suitable mounting, the specimens were prepared for electrochemical testing by initial polishing down to a 1200 grit specification using 180, 300, 600 and 1200 grit paper respectively. Following this finer polishing was undertaken to obtain a smooth 'mirror-like' finish using 6, 1 and 1/4 micron diamond pastes, respectively. Finally, an insulating lacquer ('Alcomit') was applied around the edges of each specimen surface to leave a well defined area (ie. a square of dimensions 7x7 mm) exposed to the environment. A constant area of 0.49 cm^2 was used throughout the study for all specimens.

During rest potential studies, each specimen was exposed to the electrolytic environment under open circuit conditions, for a period of up to 180 minutes. The potential was monitored every 10 minutes until a steady potential change of less than 2 mV/ 10 minutes was observed. The potential was noted as well as the time taken to reach this potential, which was defined as the 'rest potential (E_R)' [Scales 1991, Williams D.F., 1994]. For each material in each in environment, this was repeated three times and from the data an average value was accepted.

Once, rest potential had been attained, the specimen was subjected to an accelerated corrosion test to obtain an anodic cyclic polarization curve using a potential range of between 0 and 5000 mV. The step-up-rate used was 200 mV/min starting at the rest potential. The electrolyte used was a phosphate buffered saline (PBS) solution to which additions of foetal calf serum (10%) and human macrophage cells were made as and when required. The electrolyte cell was maintained at 37°C throughout the tests using a suitable water bath. All tests were carried out under aerated conditions. The electrolyte cell was made of perspex and a volume of 200 ml was used for these experiments. The specimen acted as a 'working' electrode (anode) and was placed in the cell facing a platinum 'auxiliary' electrode (cathode). A saturated calomel electrode was used as the 'reference' electrode.

The mononuclear cells were obtained from human blood serum and were separated from other cell types by virtue of their mass using a ficoll density gradient obtained from Pharmacia, U.K.. When testing the specimens in an environment containing serum and cells, it was important to initiate the experiment promptly following the separation of the cells. The macrophages were known to be in a non-activated state at the beginning of the experiment as determined by flow cytometry analysis and expression of CD-14 and CD-54 which are proven cell markers for these cells. The cells were found to survive for up to 4 hours after which time they began to die. The cells were added carefully, using a pipette, to the surface of the specimen which was placed so as to sit flat at the bottom of the electrolyte cell. The PBS solution containing the serum was then added to the electrolyte cell. The final cell population consisted of macrophages and lymphocytes.

Laser scanning confocal microscopy

A laser scanning confocal microscope was used to observe the cells before and after corrosion had taken place to assess the behaviour of the cells at the material surface. Images were also obtained to analyze any differences in the attachment of cells to each material before corrosion, following the attainment of rest potential and after the accelerated corrosion test. The expression of CD-14 (to distinguish both non-activated and activated cells) and CD-54 (to distinguish activated cells only) as cell surface phenotypic markers was studied using immunofluorescent staining. A contrast in the colours on the photographs helped visualize the proportionate representation of the cells that were activated both as a result of exposure to the environment and following an accelerated corrosion test.

Immunological assays

Interleukin-6 (IL-6), interleukin-1ß (IL-1ß), interleukin-4 (IL-4) and tumour necrosis factor-α (TNF-α) were analyzed using commercially available solid phase sandwich, ELISA (enzyme linked immunosorbant assay) kits obtained from Biosource International Inc., Camerillo, California.

This technique used a 96-well plate that is pre-coated with an antibody to the particular cytokine that is being studied. This antibody binds the cytokine to the plate, after which another antibody is added that recognises different epitopes. This second antibody is coupled to a detection system that yields a colour change proportional to the amount of cytokine bound to the first antibody. Samples including standards of known cytokine content, control specimens, and unknowns, were pipetted into the wells followed by the addition of biotinylated second antibody. During the first incubation, the cytokine antigen binds simultaneously to the immobilized (capture) antibody on a second site. After removal of excess antibody, Streptavidin-Peroxidase (enzyme) was added. This was allowed to bind to the biotinylated antibody to complete the four-member sandwich. After a second incubation and washing to remove all of the unbound enzyme, a substrate solution was added which was acted upon by the bound enzyme to produce colour. The intensity of the coloured product was directly proportional to the concentration of the cytokine present in the original specimen.

7.3 RESULTS

Table 7.3.1 and 7.3.2 display the values of rest potential (E_R) and the times taken in minutes to reach this potential, in the three different biological environments which vary in terms of degree of biological complexity. The data shown represents the mean values of three independent evaluations for each material. Figures 7.3.1 and 7.3.2 gives a pictorial representation of this data. It is noticed that in a simple environment of PBS solution alone, the nobility of the alloys is in the order of Ti-6Al-7Nb, Ti-13Nb-13Zr and Ti-6Al-4V. With reference to the data in Table 7.3.2 it seems that Ti-6Al-4V and Ti-6Al-7Nb reach rest potential in a shorter time than Ti-13Nb-13Zr alloy.

In a slightly more complicated environment in which serum proteins are present, Ti-6Al-4V appears to have a more noble rest potential than Ti-6Al-7Nb and Ti-13Nb-13Zr in that order. Although, the time taken to reach E_R is reduced for all three alloys in this environment, it was noticed that Ti-6Al-4V and Ti-6Al-7Nb attain rest potential in a shorter time than Ti-13Nb-13Zr alloy, as in PBS solution.

However, in an environment containing serum proteins and human macrophage cells there is a contrast in results to the environment with proteins only, with the nobility of Ti-6Al-7Nb and Ti-13Nb-13Zr being much greater than Ti-6Al-4V. As far as the time periods taken to reach E_R are concerned, it seems that all three alloys reach rest potential in more or less the same time as when serum proteins are present without macrophages.

It was noticed that the E_R value became more negative for Ti-6Al-7Nb and Ti-13Nb-13Zr alloys in an environment containing serum and serum + macrophages. These results suggest that the proteins are playing a more prominent role in covering the surface of the alloys within a matrix of protein reducing the rate of dissolution of the metal ions during passivity. In the case of Ti-6Al-4V, the E_R value becomes more electropositive in the presence of proteins, but when macrophages are also present, the E_R value becomes more electronegative again. Although, the time taken to reach E_R is reduced similarly to the other two alloys.

Table 7.3.3 and Figure 7.3.3 display the pitting corrosion resistance (E_B-E_P) in terms of the difference in breakdown (E_B) and protection potential (E_P), in progressively more complicated PBS environments. The results represent mean values in the form of a histogram to show qualitatively the corrosion behaviour of all three alloys in biological environments of varying degrees of complexity. Generally, the smaller the difference in the values of E_B and E_P, the greater the corrosion resistance of the material [Cogan et al., 1994]. It seems that the order of increasing corrosion resistance for the materials tested was Ti-13Nb-13Zr > Ti-6Al-7Nb > Ti-6Al-4V in an environment of PBS alone. When testing in PBS + foetal calf serum (FCS), the order of corrosion resistance appeared to be the other way around , Ti-6Al-4V > Ti-6Al-7Nb > Ti-13Nb-13Zr, and when using an environment containing PBS + FCS + macrophage cells (mac), Ti-6Al-4V = Ti-6Al-7Nb > Ti-13Nb-13Zr.

Figures 7.3.4 to 7.3.7 display the release of cytokines, IL-6, TNF-alpha, IL-1(beta), and IL-4 respectively, from the activated macrophages and T-lymphocytes that were present in the environment both before and after corrosion had occurred. There was a marked release of IL-6 cytokine for Ti-6Al-4V alloy following corrosion as opposed to exposure to the environment without corrosion taking place. Generally, both Ti-6Al-4V and Ti-13Nb-13Zr alloys showed a greater release of this cytokine compared to Ti-6Al-7Nb alloy which was comparatively much less represented.

The release of TNF-α cytokine showed a very similar pattern to that of IL-6 release, but with higher concentrations released in the same pattern as far as material and condition was concerned. Generally it was found for all three alloys to express a low concentration of IL-1ß compared to both IL-6 and TNF-α, and there was a marked increase in cytokine expression following corrosion of Ti-6Al-4V alloy compared to without corrosion. For the other two alloys the amount of IL-1(beta) was not significantly affected by corrosion.

The release of IL-4 confirmed the presence of T-lymphocytes, and the activation of this population, within the cell matrix used for these experiments. The release of this enzyme was greater with Ti-13Nb-13Zr and Ti-6Al-4V alloys compared to Ti-6Al-7Nb alloy which did not show much expression at all. Overall, IL-6, TNF-α and IL-4 showed a greater expression following corrosion compared to without it, as far as both Ti-6Al-4V and Ti-13Nb-13Zr alloys were concerned. It was found for Ti-6Al-7Nb alloy to express a similar amount of release of all four cytokines both before and after corrosion.

Figures 7.3.8 to 7.3.10 show images of the surface of the three alloys before and after corrosion had occurred in an environment containing PBS + FCS + macrophages using a laser confocal microscope. These photographs simply provide a qualitative account in support of the data obtained in Figures 7.3.4 to 7.3.7 of how the macrophage cells are seen to become 'flattened' as opposed to a 'rounded' shape when they have been activated.

Table 7.3.1: The rest potential values (mV) of the alloys in biological environments varying in degrees of complexity

Alloy	Rest Potential (mV)					
	PBS		PBS + FCS		PBS + FCS + Mac	
	X	$\bar{X}$	X	$\bar{X}$	X	$\bar{X}$
Ti-6A1-4V	- 249	- 242	- 122	- 163	- 413	- 376
	-280		187		- 369	
	- 198		180		- 347	
Ti-6A1-7Nb	- 97	- 93	- 226	- 259	- 177	- 190
	- 110		- 281		- 210	
	- 72		- 270		- 183	
Ti-13Nb-13Zr	- 191	- 156	- 333	- 358	- 262	- 214
	- 131		- 351		- 201	
	- 147		- 390		- 179	

Figure 7.3.1: A histogram showing the rest potential values (mV) of the alloys in varying biological environments

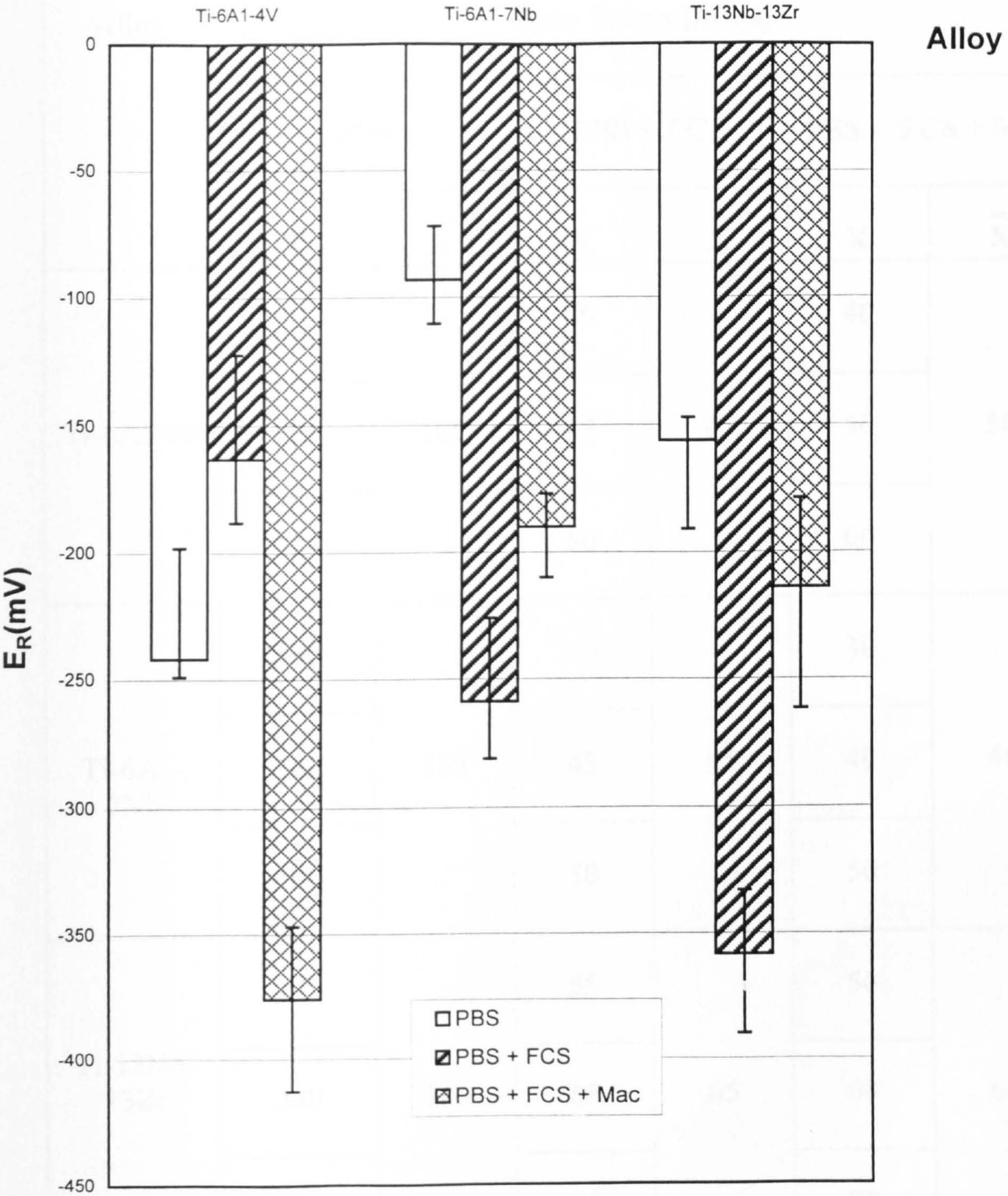


Table 7.3.2: The times taken (mins) to reach rest potential in varying biological environments

Alloy	Time Taken (Mins)					
	PBS		PBS + FCS		PBS + FCS + Mac	
	X	$\bar{X}$	X	$\bar{X}$	X	$\bar{X}$
Ti-6Al-4V	95	105	40	45	40	50
	105		45		50	
	115		50		60	
Ti-6Al-7Nb	130	105	40	45	30	40
	140		45		40	
	150		50		50	
Ti-13Nb-13Zr	130	140	55	65	50	60
	140		65		60	
	150		75		70	

Figure 7.3.2: A histogram showing the mean times taken (mins) to reach rest potential in varying biological environments

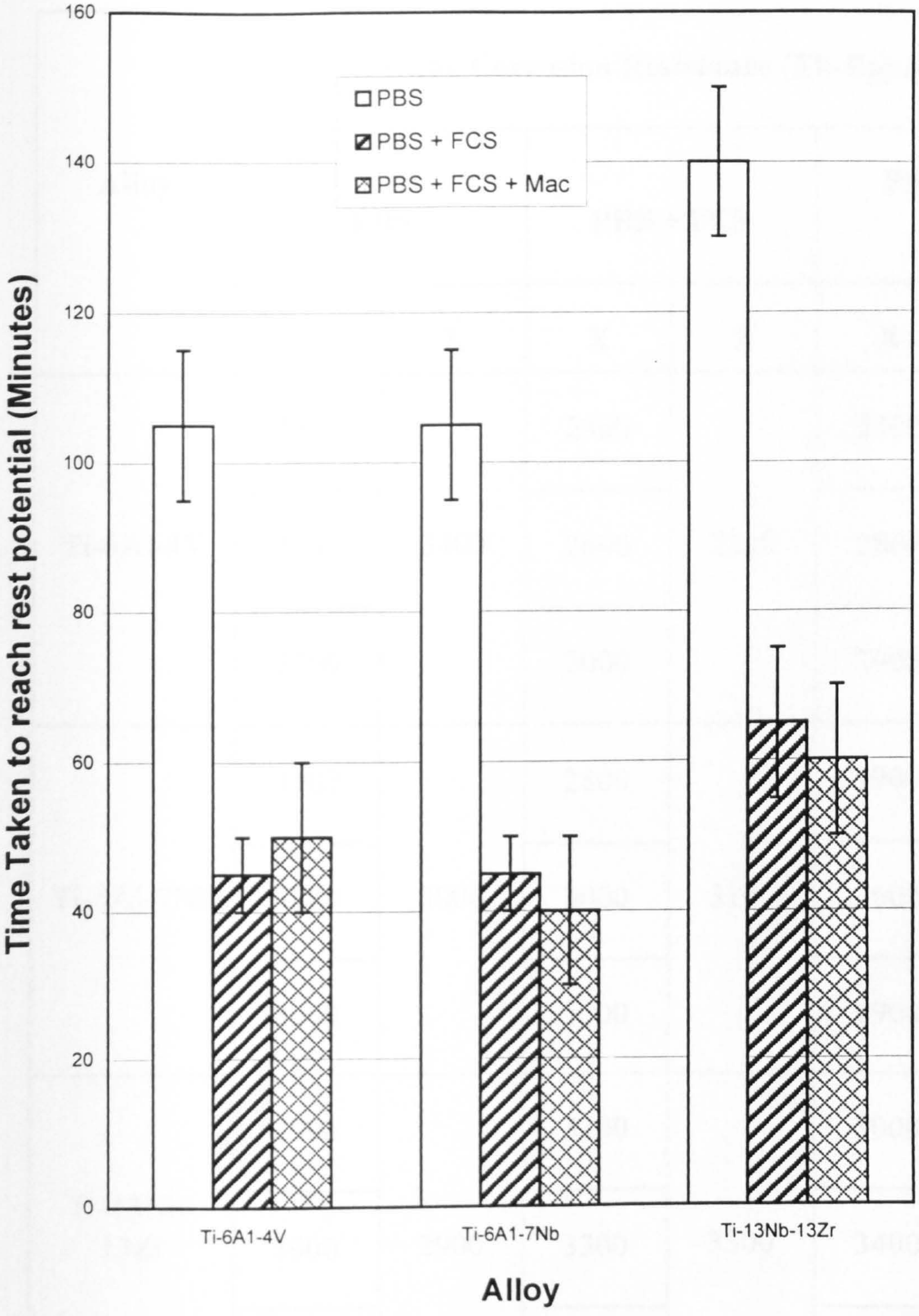


Table 7.3.3: The pitting corrosion resistance (E_B-E_P) of the alloys in varying biological environments

Alloy	**Pitting Corrosion Resistance (Eb-Ep) (MV)**					
	PBS		**PBS + FCS**		**PBS + FCS + Mac**	
	X	$\bar{\mathbf{X}}$	**X**	$\bar{\mathbf{X}}$	**X**	$\bar{\mathbf{X}}$
Ti-6Al-4V	3200	3400	2300	2300	2700	2800
	3500		2600		2800	
	3500		2000		2900	
Ti-6Al-7Nb	3100	3000	2800	3100	2900	2800
	2900		3000		2600	
	3000		3500		2900	
Ti-13Nb-13Zr	2700	2900	3100	3300	3000	3200
	3000		3300		3400	
	3000		3500		3200	

Figure 7.3.3: Histogram of the pitting corrosion resistance (E_B-E_P) of the alloys in varying biological environments

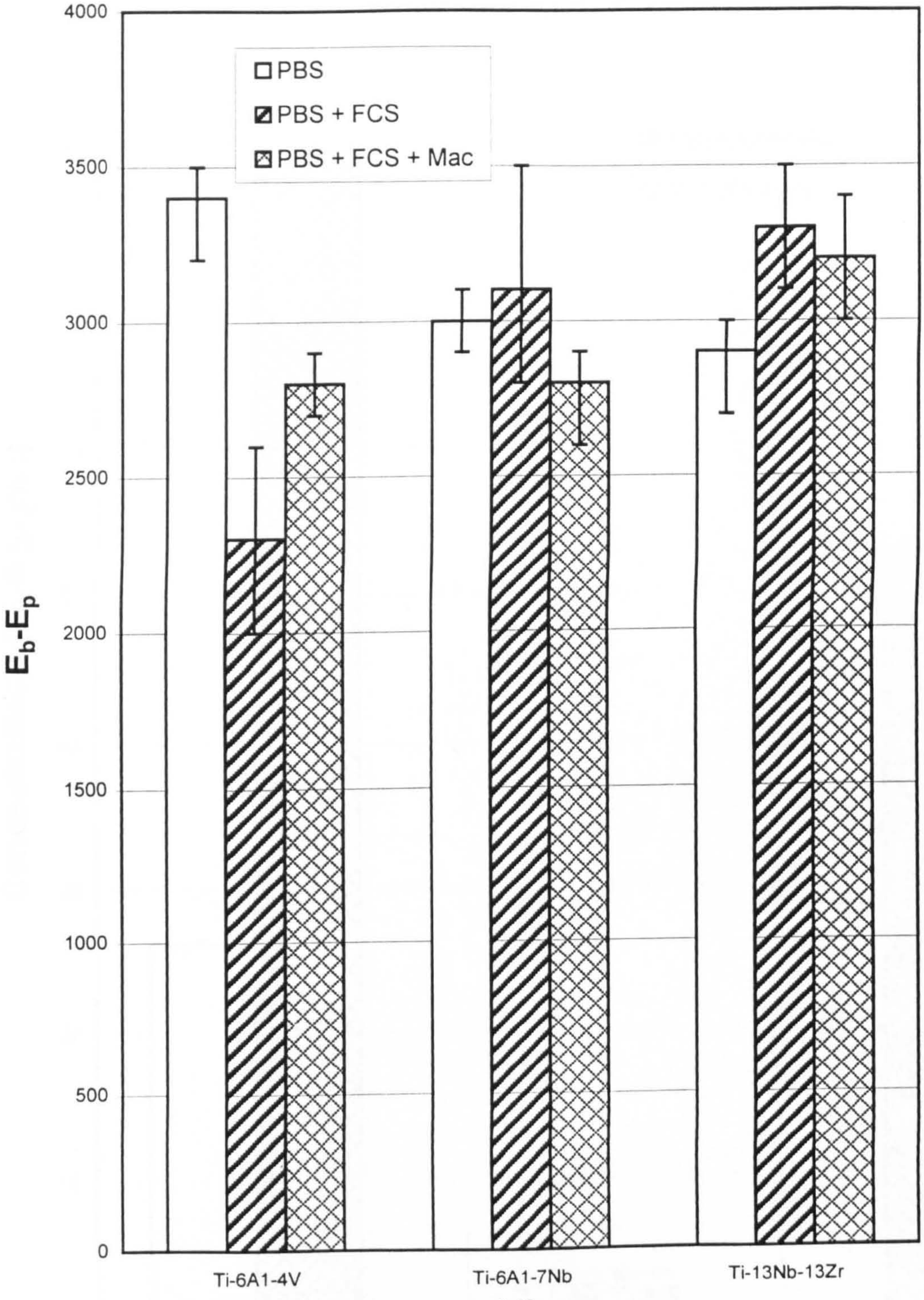


Figure 7.3.4: The release of IL-6 cytokine from activated macrophages before and after corrosion

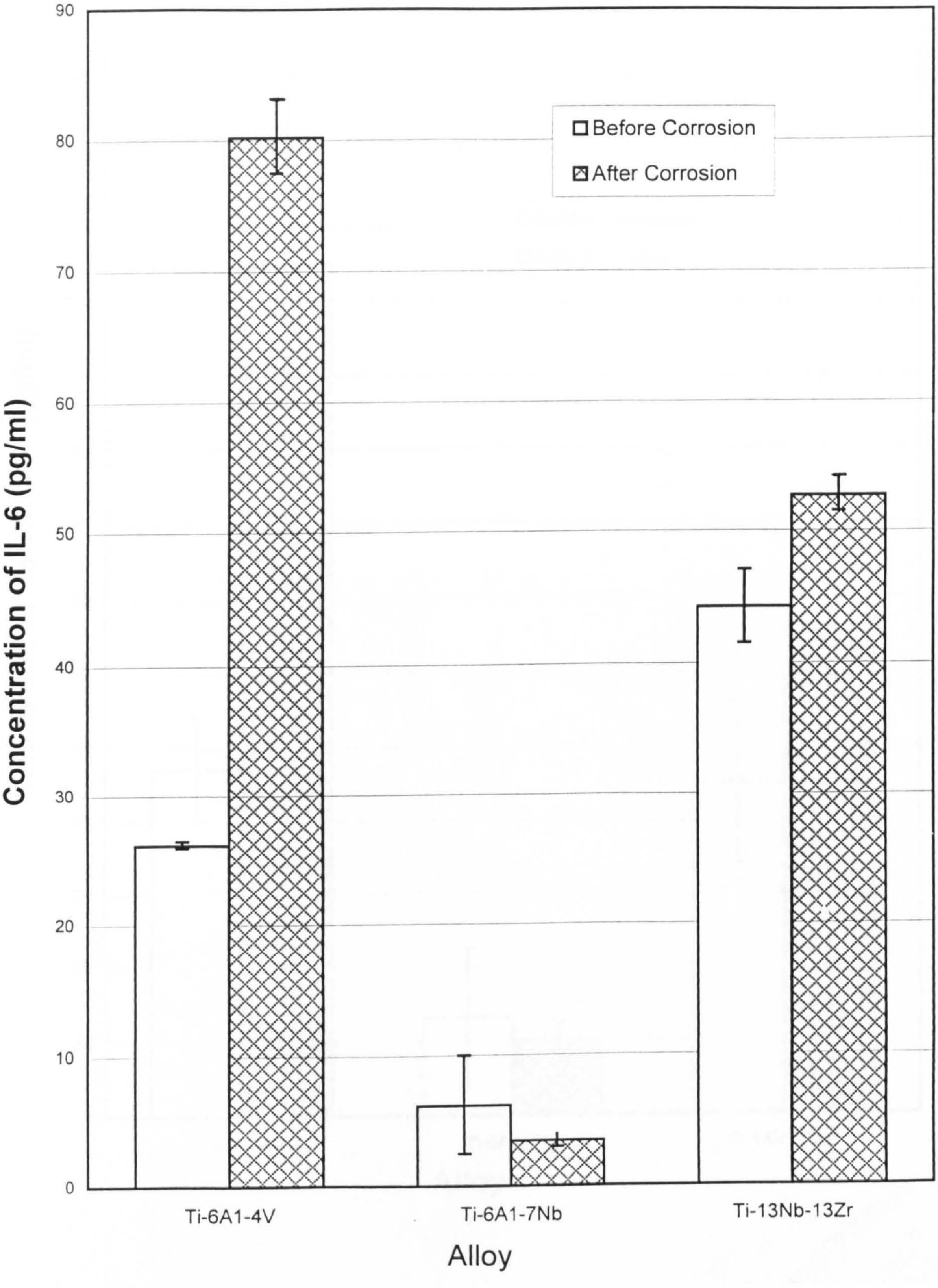


Figure 7.3.5: The release of TNF-α cytokine from activated macrophages before and after corrosion

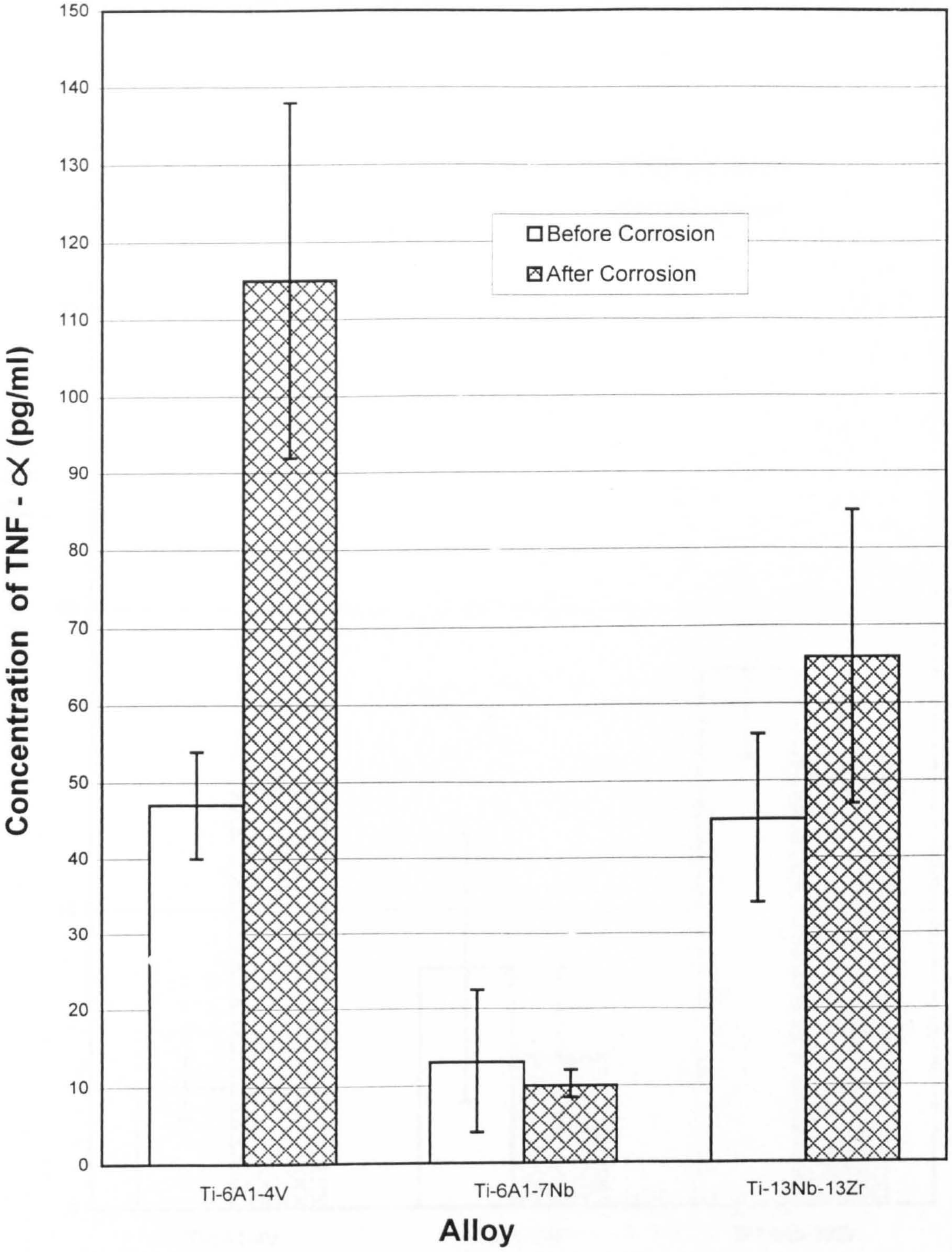


Figure 7.3.6: The release of IL-1ß cytokine from activated macrophages before and after corrosion

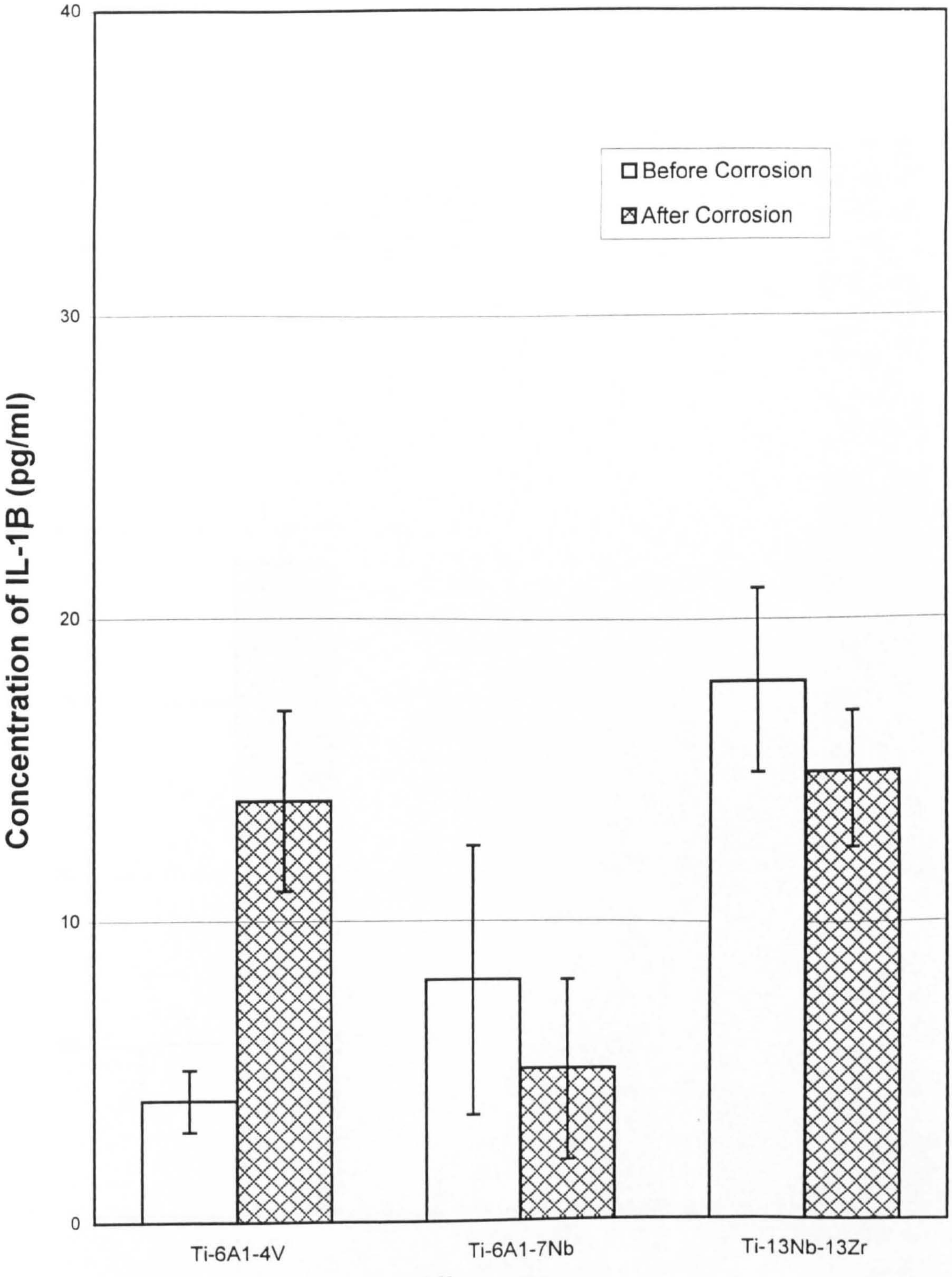


Figure 7.3.7: The release of IL-4 cytokine from activated macrophages before and after corrosion

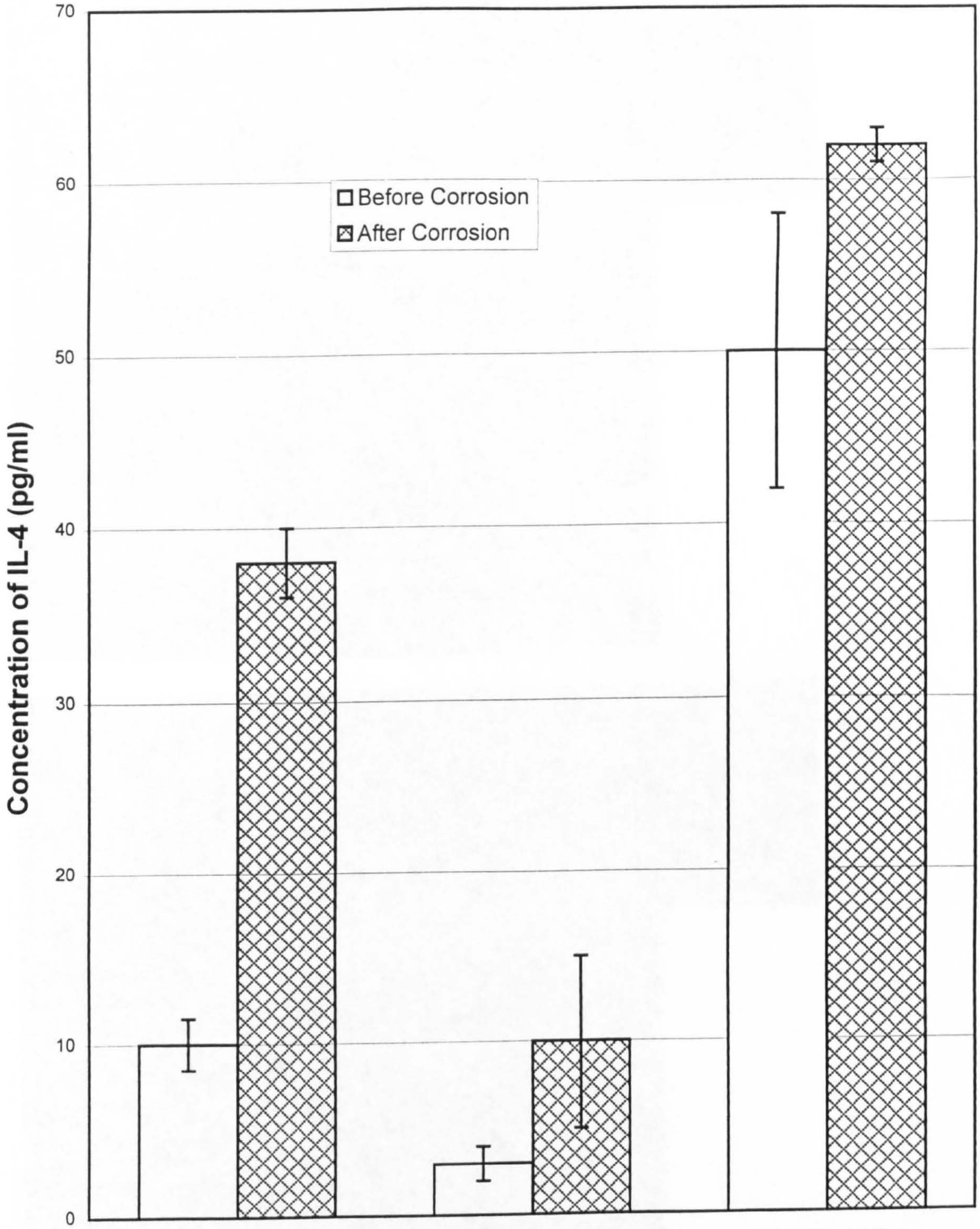


Figure 7.3.8: Laser confocal microscope photographs of Ti-6Al-4V alloy surface before and after corrosion in an environment containing PBS + serum + human macrophage cells

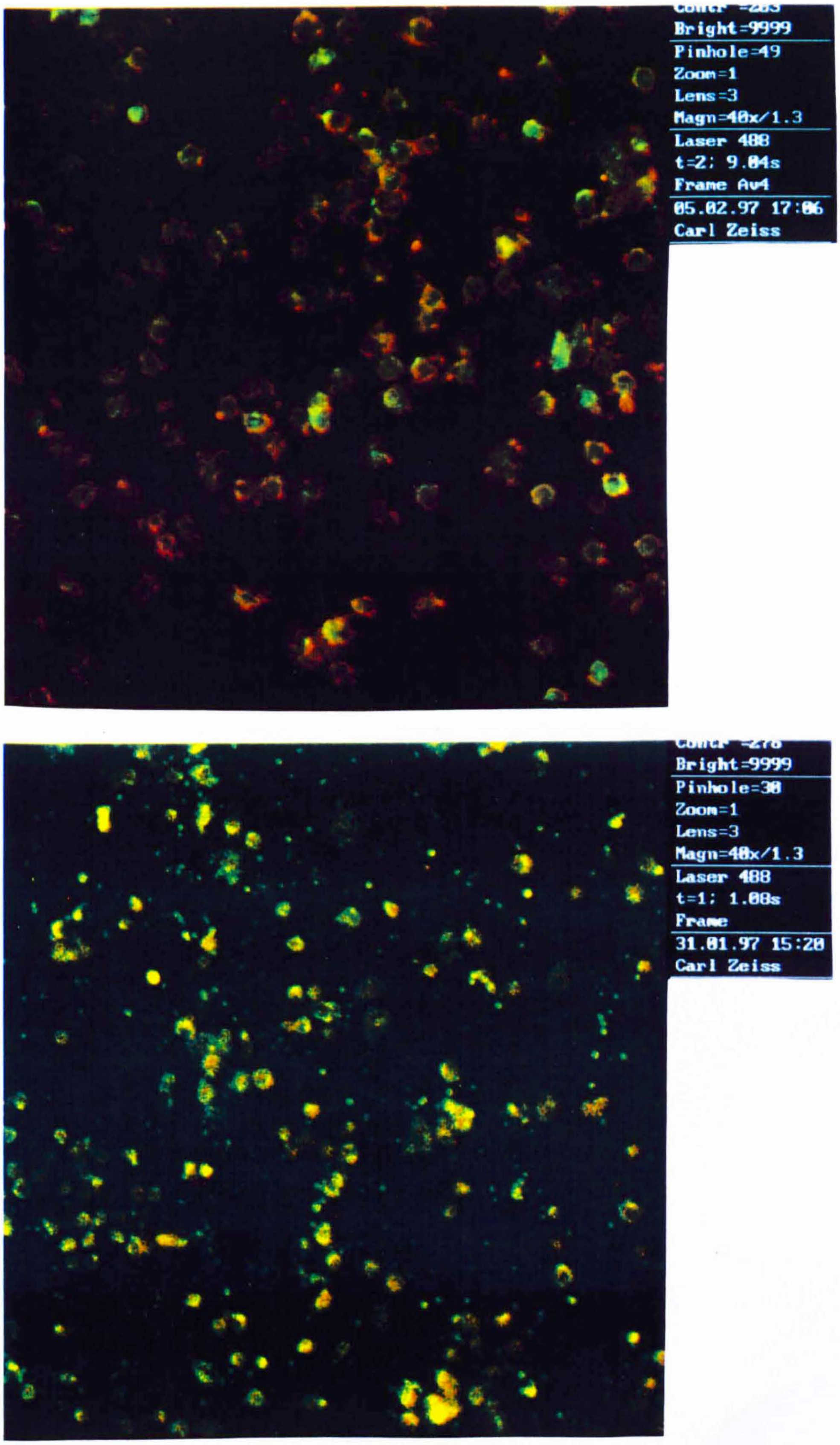


Figure 7.3.9: Laser confocal microscope photographs of Ti-6Al-7Nb surface before and after corrosion in an environment containing PBS + serum + human macrophage cells

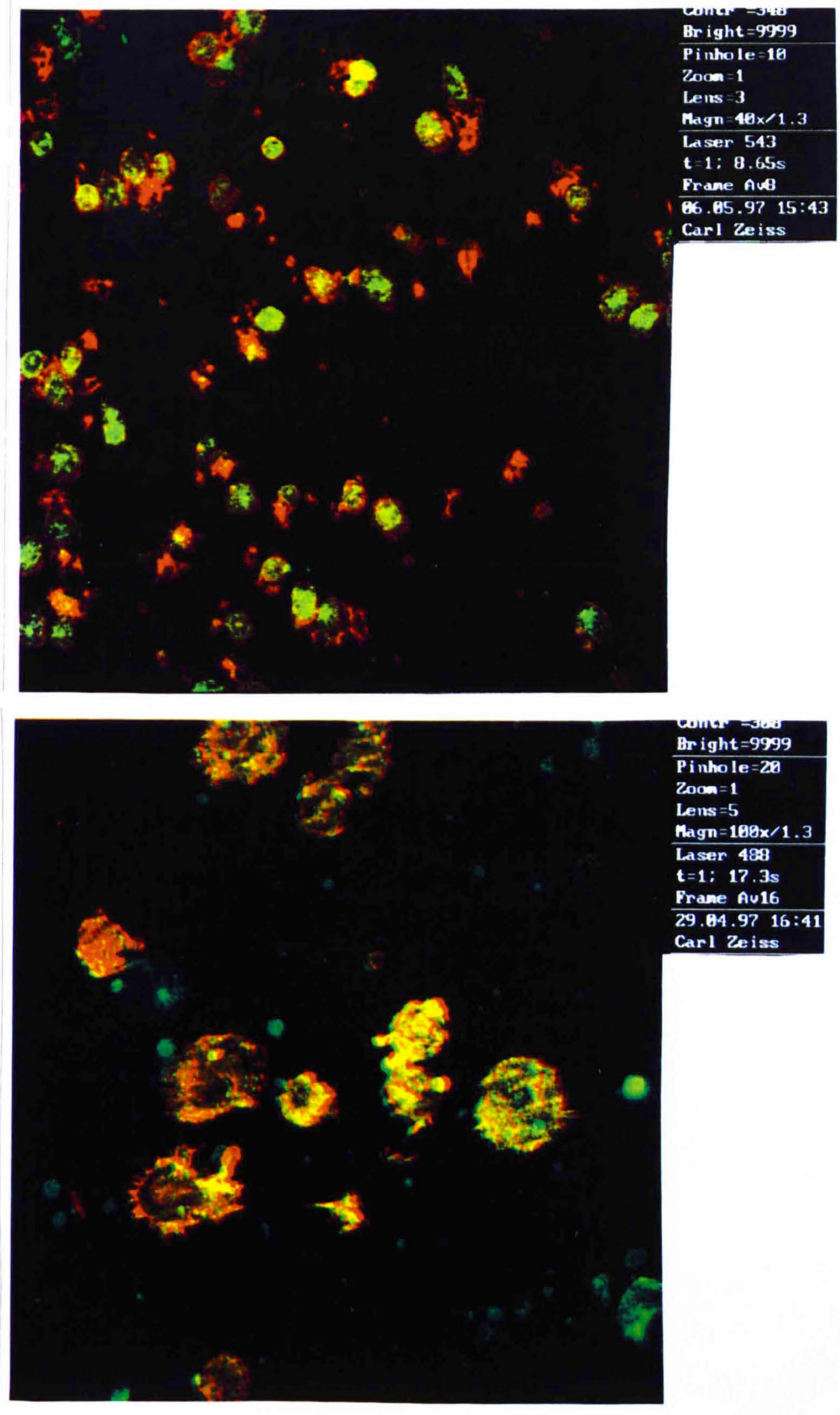
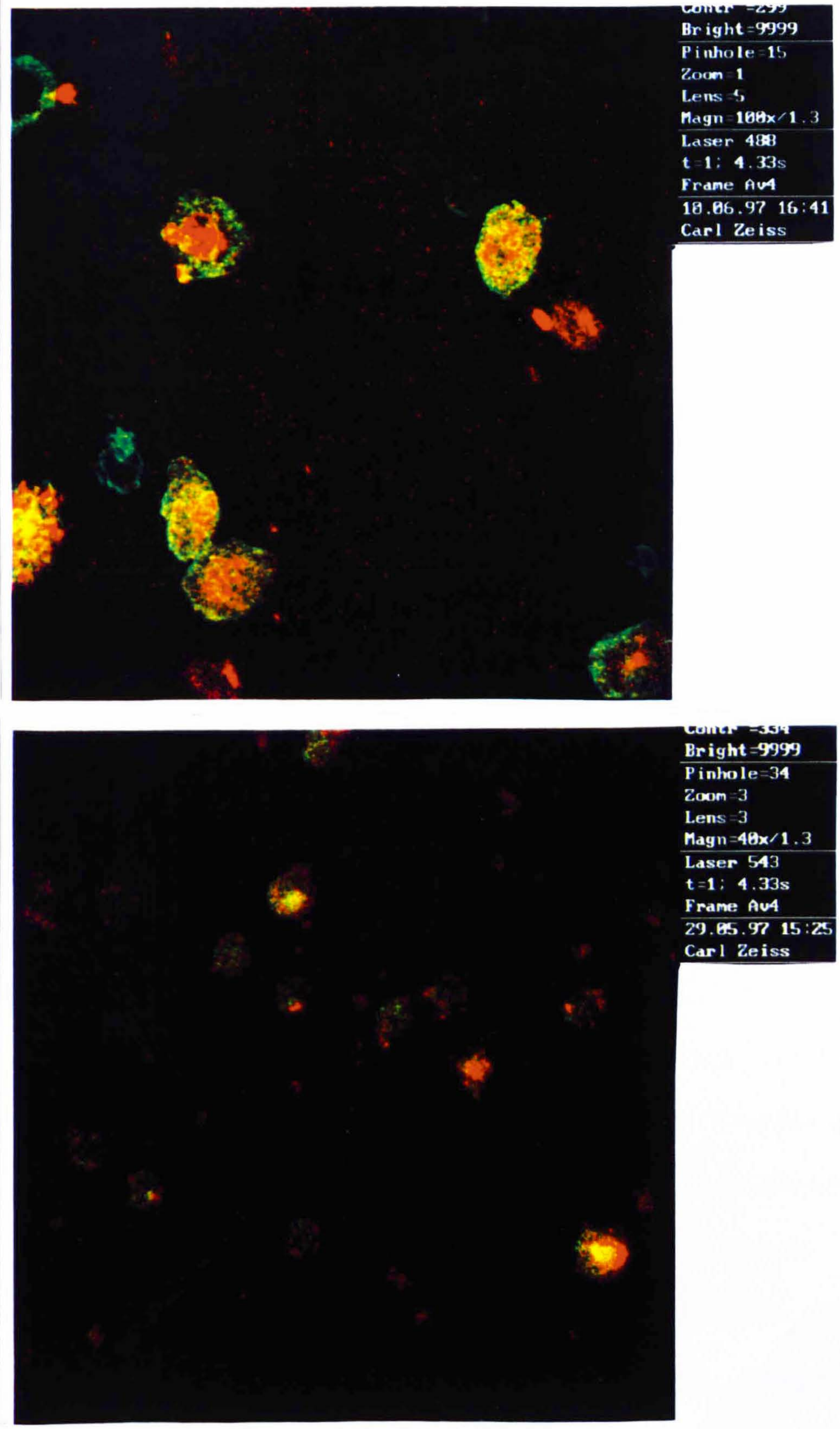


Figure 7.3.10: Laser confocal photographs of Ti-13Nb-13Zr alloy surface before and after corrosion in an environment containing PBS + serum + human macrophage cells



7.4 DISCUSSION

In order to explain the pattern of the results as accurately as possible it is important to understand the mediating factors involved between the cells and serum proteins interacting at the specimen surface, both during passivation and accelerated corrosion testing, in the controlled biological environment investigated.

When considering the interaction of macrophage cells with the serum proteins present in the biological environment it is necessary to outline the charge carrying role of both the protein molecules and the macrophage cells. Also, to understand their relative behaviour in terms of charge carriers at the metal/electrolyte interface. Cationized bovine serum albumin and other cationic proteins are significantly more immunogenic than their less charged counterparts causing more macrophage activity. There is an indication that specific cell-surface binding sites which occur on macrophages may mediate selective uptake of certain proteins with highly charged domains [Dohlman 1991]. However, it has also recently been reported that protein pre-adsorption on polymer surfaces showed lower levels of cytokines suggesting a passivating effect by the protein adsorption phenomenon on macrophage activation [Anderson 1995]. Also, protein deficient related changes in carbohydrate composition of macrophages could lead to modification of their receptor activity and charge related functions [Machaiah 1994]. It was shown that the cell surface sialic acid which contributes to the negative charge of the cells, is a constituent of albumin and may be related to the cell activity in some way, yet unknown [Dohlman 1991]. However, albumin has also been shown to play a role in macrophage activation in its sialic acid free form, even though, it seems that the presence of this sialic acid increases macrophage phagocytic activity.

Generally, it is suggested that the presence of serum proteins plays a role in decreasing the rate of dissolution of metal ions from the surface of the alloy by covering the surface layer. However, for Ti-6Al-4V an adverse effect was noticed in which the E_R value became more negative suggesting an increase in the rate of metal ion dissolution. Previous studies have proposed that the Ti-6Al-4V alloy surface involves a more proliferative dissolution mechanism of Al^{3+} and V^+ ions compared to the other two alloys which contain less soluble metals such as niobium and zirconium. In this study, the macrophage cells seem to be interacting in some way to either become activated releasing certain enzymes or play a role in attracting the positive metal ions which are released from the oxide surface more readily towards relatively more negatively charged sialic acid coated cell surfaces [Dohlman, 1991 and Davidson et al., 1994].

The measurement of cytokine secretion is a sensitive indicator of cellular responses in the presence of biomaterials. Interleukin-1ß (IL-1ß), tumour necrosis factor-α (TNF-α) and interleukin-6 (IL-6), have all been identified in periprosthetic tissue and implicated in the inflammatory response to wear particles and periprosthetic loosening [Andrew et al., 1993 and Chiba et al., 1994].

It has been shown that in response to the release of interleukin-4 (IL-4), stimulated monocytes produce less IL-1ß [Vannier et al., 1992 and Hart et al., 1996], which may explain why in the present study the results obtained are in agreement with this finding. If the cell suspension consisted of both macrophages and T-lymphocytes, then, the expression of IL-4 by the T-lymphocytes may have inhibited the expression of IL-1ß from these cells. Thus, the overall concentration of IL-1ß expressed would be considerably less as confirmed by the results obtained during this study.

Overall, it was observed that the presence of proteins improves the corrosion resistance of Ti-6Al-4V and reduces that of Ti-13Nb-13Zr, with that of Ti-6Al-7Nb remaining unchanged. However, the same environment in the presence of macrophage cells improves the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb and slightly reduces that of Ti-13Nb-13Zr alloy.

Although, it was found for the corrosion resistance (E_B-E_P) to be similar for both Ti-6Al-4V and Ti-6Al-7Nb in an environment containing serum proteins and macrophage cells, it was found for Ti-6Al-7Nb alloy produced less cell activation. Thus, it seems that the ions released by Ti-6Al-7Nb alloy are less toxic than those of Ti-6Al-4V. This suggests that the presence of vanadium ions within the alloy enhances the activation of macrophages during both passivation and corrosion. The same argument may be used when discussing the effects of Ti-13Nb-13Zr alloy on cell activation. In this case the increase in cell activation may be linked to the release of zirconium ions into solution during passivation and corrosion. Having considered this argument, it is equally as important to take into account the relatively active rest potential value (E_R) found for Ti-6Al-4V alloy in the presence of serum proteins and macrophage cells, especially, when considering the passive state of the alloy. Since, if the potential for release of ions from the alloy surface is considerably higher as a result of a matrix layer of negatively charged species (ie. proteins and macrophages), then this may also be contributing to a greater release of ions into solution both during passivation and corrosion.

CHAPTER 8 : GENERAL CONCLUSIONS

1. This study has shown that following Ti(cpi), the beta (Ti-15Mo) or near-beta (Ti-13Nb-13Zr) alloys display the greatest resistance to corrosion. An investigation of the alloying elements has shown that niobium and to a lesser extent zirconium produced the most favourable effects in terms of corrosion resistance, as opposed to molybdenum, vanadium and aluminium, in that order. The phenomenon of corrosive wear (contribution of corrosion to wear) as opposed to non-corrosive wear is more detrimental to the mixed phase alpha-beta alloys (ie. Ti-6Al-4V and Ti-6Al-7Nb), than any of the others. Wear by itself, is more detrimental to the other alloys than to these two. As far as the influence of wear on the corrosion of materials is concerned (ie. wear-accelerated corrosion), it has been shown that Ti-3Al-2.5V, Ti-13Nb-13Zr and to a lesser extent Ti-6Al-7Nb, are least affected. Conversely, Ti, Ti-15Mo, Ti-8Al-1Mo-1V and Ti-6Al-4V are all affected to a greater extent.

2. The corrosion resistance of the three titanium alloys (Ti-6Al-4V, Ti-6Al-7Nb and Ti-13Nb-13Zr) was influenced by the specific environments within which they were tested. The presence of low concentrations of proteins was demonstrated to increase the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb but reduce the corrosion resistance of Ti-13Nb-13Zr. The pH of the environment was also very important. In the absence of proteins an increase in pH decreased the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb but increased the corrosion resistance of Ti-13Nb-13Zr. However, in the presence of proteins the corrosion resistance of Ti-13Nb-13Zr decreased as pH increased. The hardness of the alloy was also influenced by the corrosive environment. Ti-13Nb-13Zr initially had the harder surface, however, following corrosion in PBS it was lower than both Ti-6Al-4V and Ti-6Al-7Nb.

Following corrosion in albumin, the Ti-13Nb-13Zr surface was harder than either Ti-6Al-4V or Ti-6Al-7Nb.

3. The data produced during this study suggests that, without corrosion, the wear of Ti-13Nb-13Zr was greater than that of Ti-6Al-7Nb or Ti-6Al-4V; and that in the presence of protein, the wear of all three alloys was reduced. In the presence of corrosion without wear, Ti-13Nb-13Zr was more corrosion resistant Ti-6Al-7Nb which was more corrosion resistant than Ti-6Al-4V, without protein. However, in the presence of protein, the corrosion resistance of Ti-13Nb-13Zr and Ti-6Al-7Nb is reduced and that of Ti-6Al-4V increased. In the presence of corrosion and wear, the corrosion resistance of Ti-13Nb-13Zr was higher than that of Ti-6Al-7Nb and Ti-6Al-4V, and in the presence of proteins the corrosion resistance of Ti-13Nb-13Zr and Ti-6Al-7Nb are very similar but higher than that of Ti-6Al-4V.The wear of Ti-13Nb-13Zr is lower than that of Ti-6Al-7Nb and Ti-6Al-4V with or without the presence of proteins. Therefore the overall degradation when both corrosion and wear processes are occurring is lowest for Ti-13Nb-13Zr and highest for Ti-6Al-4V and the presence of proteins reduces the degradation of all three alloys.

4. a) The rest potential (E_R) of all three titanium alloys becomes more electropositive in PBS alone when polishing using a 1/4 diamond paste polish compared to 180 grit paper. This effect was most prominent with Ti-13Nb-13Zr followed by Ti-6Al-7Nb and Ti-6Al-4V in that order. In the presence of serum proteins, at 1/4 micron specification, Ti-6Al-4V became more electropositive, whereas, both Ti-6Al-7Nb and Ti-13Nb-13Zr became more electronegative in that order. At 180 grit specification, the opposite was found, with Ti-6Al-4V becoming more electronegative, whereas, Ti-6Al-7Nb and Ti-13Nb-13Zr became more

electropositive.

b) In PBS alone, following a 1/4 micron polish, the corrosion resistance of Ti-6Al-7Nb > Ti-13Nb-13Zr > Ti-6Al-4V. When using a 180 grit polish, the corrosion resistance of Ti-13Nb-13Zr > Ti-6Al-7Nb > Ti-6Al-4V. In this environment, Ti-13Nb-13Zr was least affected by a change in surface preparation followed by Ti-6Al-4V and Ti-6Al-7Nb in that order. At 1/4 micron specification, the presence of proteins, reduces the corrosion resistance of Ti-6Al-4V and has relatively no effect on both Ti-6Al-7Nb and Ti-13Nb-13Zr alloys. At 180 grit specification, the addition of serum proteins, improves the corrosion resistance of Ti-6Al-4V, has relatively no effect on Ti-6Al-7Nb and slightly reduces that of Ti-13Nb-13Zr alloy. Overall when considering an environment containing serum proteins, it was found that a fine polish preparation was more favourable for both Ti-6Al-7Nb and Ti-13Nb-13Zr alloys.

c) In PBS alone, the value of E_B was lower for all three alloys as a result of a change in surface roughness from 1/4 micron to 180 grit specification, in the order, Ti-6Al-7Nb, Ti-6Al-4V and Ti-13Nb-13Zr respectively. However, in the presence of proteins, a change from 1/4 micron to 180 grit specification showed a lowering in E_B value for Ti-6Al-7Nb and an increase for both Ti-6Al-4V and Ti-13Nb-13Zr alloys.

5. a) In an environment that more closely depicts the in-vivo situation, in which human macrophage cells are present, it was found for Ti-6Al-7Nb to have the most noble rest potential (E_R) followed by Ti-13Nb-13Zr and Ti-6Al-4V in that order.

b) The time taken to reach rest potential (E_R) is reduced for all three alloys when exposed to an environment containing serum proteins and/or human macrophage cells. This effect is most prominent for Ti-13Nb-13Zr followed by Ti-6Al-7Nb and Ti-6Al-4V in that order.

c) The presence of serum proteins improves the corrosion resistance of Ti-6Al-4V and reduces that of Ti-13Nb-13Zr, with that of Ti-6Al-7Nb remaining unchanged. However, the same environment in the presence of human macrophage cells improves the corrosion resistance of Ti-6Al-4V and Ti-6Al-7Nb and slightly reduces that of Ti-13Nb-13Zr alloy.

d) There is substantial evidence to suggest that rest potential and in-vitro corrosion studies should whenever possible be carried out using proteinaceous electrolytic solutions including macrophage cells, to assess more accurately the corrosion characteristics of titanium alloys as biomaterials, especially when trying to compare or predict their end use durability and breakdown mechanisms.

e) This study has shown that Ti-6Al-4V alloy and Ti-13Nb-13Zr alloy, to a lesser extent, give rise to an increased amount of macrophage activation following corrosion, Ti-6Al-7Nb alloy caused a comparatively lower response.

PROPOSALS FOR FURTHER WORK

This project has created numerous opportunities for further work following on from this in-depth study into the corrosion and wear behaviour of certain titanium alloys within varying biological solutions. Similar testing might be contemplated on other 'novel' titanium alloys presently available as better alternatives to Ti-6Al-4V alloy (eg. Ti-5Al-15Mo-5Zr alloy and certain Ti-Nb-Zr-Pd-Sn alloys). Surface treatments on titanium alloys might be studied, especially, with some of the latest techniques available to produce diamond-like coatings (DLC) onto titanium surfaces. This might include a review of either one specific surface treatments applied to a number of alloys and/or a study of various treatments presently available to try to assess methods of improving the corrosion and wear behaviour of titanium alloys by slightly altering the surface.

Further corrosion work might also be a possibility by looking at introducing other biological cells into the environment (eg. fibroblasts, neutrophils etc..). The influence of macrophage cells on other metal surfaces might be investigated, especially using dissimilar metals (eg. titanium and stainless steel). This might include looking at the response of such cells to metal samples constituting titanium alloys to correlate relative degrees of cytocompatibility. Particulate material (ie. metal/alloy particles of varying sizes) could be added to the environment and the corresponding corrosion/wear behaviour investigated to mimic the effect of wear debris already present in the vicinity of a material which then begins to corrode. A more comprehensive corrosion/wear testing apparatus could be designed to be incorporated within the corrosion cell. Perhaps an investigation into improving the stability of such a device to produce more consistency in producing corrosion/wear data might be pursued.

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APPENDIX 2 : PUBLICATION

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In-vitro corrosion and wear of titanium alloys in the biological environment

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Cyclic anodic polarization studies were undertaken for several titanium alloys of varying composition and phase structures. All materials were exposed to an accelerated corrosion test using a potentiostat and their electrochemical behaviour was analysed within a potential range of 0 to 5000 mV. The electrolyte used was a phosphate buffered saline (PBS) solution at pH = 5, 7.4 and 9. The polarization curves obtained represented both the passive and active regions of the materials and these curves were used to compare the resistance to pitting corrosion of each material. The sliding-wear of these materials was studied in both non-corrosive and corrosive environments. A simple pin-on-disc type wear apparatus was designed and built to simulate the co-joint action of corrosion and sliding-wear. Using this apparatus, it was also possible to evaluate the effect of wear-accelerated corrosion, which was also evaluated by wearing the surface of the specimens prior to corrosion. It was evident that the mixed phase alpha–beta alloys (Ti–6Al–4V and Ti–6Al–7Nb) possessed the best combination of both corrosion and wear resistance, although commercially pure titanium and the near-beta (Ti–13Nb–13Zr) and beta (Ti–15Mo) alloys displayed the best corrosion resistant properties. © 1996 Elsevier Science Limited

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Titanium (commercially pure grade) and Ti–6Al–4V alloy have both been used in implantable devices for several decades. Although they have an excellent reputation for corrosion resistance and biocompatibility, considerable controversy has developed over the biocompatibility of the alloy[1–13].

There are two main points to consider in this controversy.

(a) Firstly, although titanium has an excellent corrosion resistance, if titanium is released into the tissue even by the way of passive dissolution or perhaps by another process involving wear, the tissue reaction may vary. This reaction could be anything ranging from a mild response (e.g. a discolouration of the surrounding tissue) to a more severe one (e.g. inflammatory reaction causing pain and even leading to loosening owing to osteolysis)[14–19].

(b) Secondly, if deposits of titanium alloy particles are present in the tissue, then concentrations of aluminium and vanadium (which themselves have considerably different biological properties than titanium) will also be present[1–3,6–7,12,20,21]. It has thus been proposed that better performance in terms of corrosion resistance and biocompatibility would be achieved through the use of alloys of different composition. However, even though several such alloys have been investigated, there is still very little data to support this view[6,22–27].

The physical metallurgy of titanium shows that it undergoes an allotropic transformation at about 885°C, changing from a close-packed hexagonal (c.p.h.) crystal structure (i.e. alpha phase) to a body-centred cubic (b.c.c.) crystal structure (i.e. beta phase). Alloying elements are known to either lower or raise the transformation temperature. Depending on their microstructure, titanium alloys fall into five categories: alpha, near-alpha, alpha–beta, near-beta or beta. Each of these denotes the general type of microstructure present after heat treatment and processing[28–31]. Substitutional alloying elements play an important role in controlling the microstructure and properties of titanium alloys.

Vanadium and niobium are beta amorphous (i.e. have similar phase relations) with b.c.c. titanium[28]. Aluminium has significant solubility in both alpha and beta phases. Zirconium is unique in that it is isomorphous with both the alpha and beta phases of titanium. Almost all titanium alloys contain one or more of these two elements because they are soluble in both phases, and particularly because they improve creep strength in the alpha phase. Also, beta isomorphous elements are preferred since they do not form intermetallic compounds. Generally, it is known

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that commercially pure titanium and single phase alloys demonstrate the best corrosion resistant properties. However, the alpha–beta alloys, when properly heat-treated, display an excellent combination of strength and ductility. They are also stronger than both the alpha or the beta alloys[28,29].

Under normal physiological conditions, titanium and its alloys should not break down, without the presence of other accelerating factors, such as wear. However, there is still some release of material into tissue owing to passive dissolution. Thus, metal ions and metal–protein complexes could cause tissue damage, and therefore, passive corrosion rates are a significant factor as far as the corrosion behaviour of these materials is concerned[11,32–36]. The corrosion characteristics of some of these materials have been studied previously, especially Ti and Ti–6Al–4V; however, these studies have only shown the corrosion behaviour within a limited potential range, usually up to approximately 2000 mV, using a de-aerated electrolyte solution, at which point they have been shown to break down. Also, any novel material has been presented simply as an alternative to the Ti–6Al–4V alloy, with an improvement in corrosion resistance[6,22,37]. In some cases, such alloys have also been shown to display better wear resistance following prior treatment with an appropriate surface treatment[24,25].

In this study, it has been possible to compare the corrosion resistance of a variety of alloys, including two recently developed novel ones (i.e. Ti–6Al–7Nb and Ti–13Nb–13Zr). The materials were investigated using a potential range of 0–5000 mV, and in an electrolyte which was not de-aerated to model the availability of oxygen in the physiological environment. It was also possible to consider the effect of corrosion on wear by comparing the non-corrosive and corrosive wear of each material. To supplement these studies, the effect of wear-accelerated corrosion (by wearing the surface of each specimen prior to corrosion) was studied.

Wear is not known to be the main cause of failure of prosthetic devices in the short term, as are infection, loosening or fracture. However, in the long term it may contribute to reasons for prosthesis removal[38]. Thus, it becomes an important factor to consider the effects of wear debris that may accumulate in the tissues, leading to problems of toxicity and unfavourable inflammatory responses[39–43]. A simple, low cost sliding-wear apparatus was designed specifically for this study (*Figure 1*), to evaluate the materials in a corrosive environment, and thus predict wear rates using surface profilometry[44–54]. Therefore, it was possible to assess the predictability of the degradation of these titanium alloys under more physiologically relevant conditions.

MATERIALS AND METHODS

Corrosion studies

The materials used in this study were obtained from various sources: Ti(cpi), Ti–6Al–4V, Ti–8Al–1Mo–1V and Ti–15Mo from IMI Ltd., UK; Ti–6Al–7Nb from Sulzer Medical Technology Ltd., Switzerland; Ti–13Nb–13Zr and all of the constituent metal samples from Smith and Nephew Richards Inc., USA, and Ti–3Al–2.5V from Aerospace Metals (Bury) Ltd., UK.

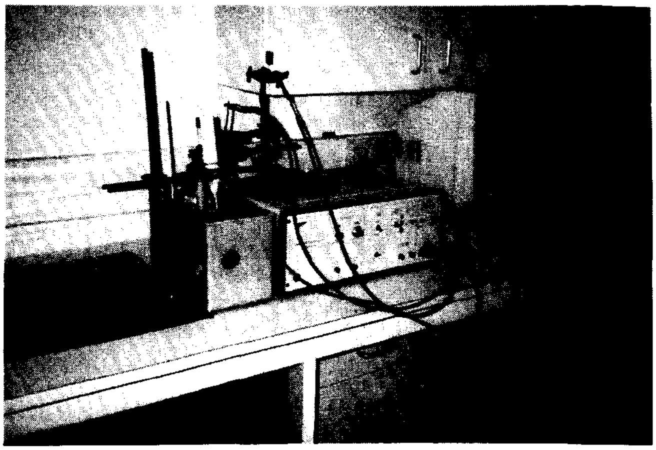

Figure 1 The simple pin-on-disc apparatus used for non-corrosive/corrosive wear testing.

All the materials were received in the annealed condition, in the form of rods, which varied between 1 and 1.5 cm in diameter.

The test specimens were then cut up to form discs of approximately 0.5 cm thickness. A wire lead was attached to the back of each disc, using a small amount of conducting paste. Following this, each specimen was mounted into a thermosetting resin mould, in the form of a 3 cm diameter rod. A large portion of the surface of the specimen was covered with an insulating lacquer varnish, leaving only a given specimen area exposed to the test solution. The specimen area used varied between approximately 0.1 and 0.5 cm^2.

Each specimen was then polished down to a 600 grit specification. Finer polishing was also undertaken to produce a 'mirror-like' finish, using 6, 1 and 1/4 μm diamond paste, respectively. Surface roughness measurements were undertaken for each specimen before and after corrosion using a surface profilometer (a 'Taylor-Hobson Surtronic 3' profilometer was used). Hardness measurements were also undertaken using a Vickers hardness indentation machine. For each specimen, an average value was obtained, using three individual readings.

Finally, the specimen (anode) was placed in solution facing an auxiliary electrode (cathode) which had a surface area four or five times greater than the specimen.

Cyclic polarization curves were obtained using a potential range of between 0 and 5000 mV (an 'AUTOSTAT' computer controlled potentiostat was used (*Figure 1*)). The potential was increased at a rate of 200 mV min^{-1}, starting with the rest potential (E_R). Electrode potentials were measured against a saturated calomel reference electrode. The electrolyte solution used in these tests was a phosphate buffered saline (PBS). The solution was made up by adding pre-packed tablets to distilled water, which produced a 0.1 M buffered solution at a pH of 7.4. The pH could then be altered as required, by adding measurable amounts of hydrochloric acid or sodium hydroxide. The solutions were maintained at 37°C throughout the tests using a water bath.

Corrosive-wear studies

Using the sliding-wear apparatus, mentioned earlier, it was possible to monitor the effects of wear of the materials within a corrosive environment. A constant load of 0.5 N was applied in contact with the specimen surface. The wearing pin was made of alumina and produced a total sliding distance of 250 metres during each corrosion test. Since the environment was much more aggressive during continuous wear compared to corrosion only, a potential step-up rate of $1000\,\mathrm{mV\,min^{-1}}$ was used (the electrolyte was maintained at pH = 7.4 throughout the wear studies). To demonstrate a non-corrosive environment, distilled water was used as the electrolyte solution. Each specimen was measured for changes in surface roughness and hardness, before and after each test. The change in these two properties following corrosive-wear gave a quantitative account of the effect of wear on each material.

In addition to the above, the concept of wear-accelerated corrosion was also investigated. In this case, each specimen was subjected to wear, for a period of 5 min, prior to corrosion testing in a similar way as that described earlier.

RESULTS AND DISCUSSION

Corrosion studies

To study the corrosion characteristics of the materials, potential vs current density curves were obtained (*Figures 2* and *3*). In general, all of the alloys tested showed a breakdown potential (E_B) of between 3000 and 4000 mV, very similar to the breakdown potential of commercially pure titanium, Ti(cpi), which produced breakdown at approximately 4000 mV. As a means of cross-reference, the materials were also tested in de-aerated solution and were found to break down at between 1500 and 2000 mV instead of around 3000 mV, which was comparable to the results obtained by other workers[26, 55–58].

The breakdown potential is known to be observed with a large increase in current at E_B associated with pit formation[59]. Observations using an optical microscope confirmed such pits. In practice, there were no real differences noted in the value of E_B for any one material. However, it has been known for considerable differences in E_B to occur partly owing to variations in the surface roughness[60]. The protection potential (E_p) is normally defined as the potential at which the current density on the negative sweep from the maximum anodic potential equals the current density in the passive region (i.e. the point at which the hysteresis loop is complete, or rest potential is reached)[59, 60–63]. The difference between the breakdown and protection potentials (E_B–E_p) is related to the 'pitting corrosion resistance' of the material (*Tables 1*, *2* and *3*). The lower the value, the more corrosion resistant the material. *Table 1* shows how the corrosion resistance of each material varies with changes in the pH of the environment (i.e. pH = 5 and 9). In addition, *Tables 4* and *5* show how the surface roughness and hardness values varied following corrosion at different levels of pH.

There is no doubt that Ti(cpi) displays a better corrosion resistance than any of the alloys tested, and even its mechanical integrity is similar to that of its alloys, following corrosion. However, the fact remains that it is an inherently weak material to begin with, having a hardness value of less than half that of any of its alloys. Still, although it does not appear to be a favourable candidate for orthopaedic applications, it continues to fulfil its role well as a suitable biomaterial in areas where corrosion resistance may be of sole concern.

In general, it was found that alpha–beta/beta alloys possessed the best corrosion resistance. Of all the alloys tested, Ti–13Nb–13Zr displayed the best corrosion resistance under normal physiological conditions (i.e. at pH = 7.4), and even seemed most resistant to changes in pH. However, it has one major drawback, in that it undergoes a considerable loss in hardness following corrosion, which is of paramount importance from the tribological point of view. This material has recently been promoted as a useful alternative to Ti–6Al–4V, as long as a suitable surface treatment (oxygen diffusion hardening in this case) is used[26]. However, the use of surface treatments is a completely separate issue, and one that is surrounded by yet more controversy. Therefore, the advantages and disadvantages of such treatments will not be discussed in this study.

These results suggest that Ti–6Al–7Nb is a better alternative to Ti–6Al–4V and there are two main reasons for this. Firstly, its corrosion resistance proved to be slightly better, if not as good as Ti–6Al–4V. Secondly, because its resistance to loss of mechanical properties with changes in pH were equally as good as Ti–6Al–4V.

With regards to the composition of each alloy, and thus the corrosion resistance of each constituent metal (*Figure 3* and *Table 6*), Ti, Al, V, Mo and Zr to a lesser extent displayed a strong resistance to both pitting and crevice corrosion. Conversely, Nb appeared to be much more vulnerable as a metal. However, it was also found that the breakdown potential (E_B) of Ti, Nb and Zr to a lesser extent is far greater than for Al, V and Mo, as has also been observed by others[23, 24, 26].

Non-corrosive wear and corrosive wear

Table 5 shows how the wear characteristics of each material change in both non-corrosive and corrosive conditions. In general, it was found that for all materials tested, corrosion is an accelerating factor on the wear behaviour. However, there is a notable variation as to the extent of its influence, between any one material. As far as changes in surface roughness

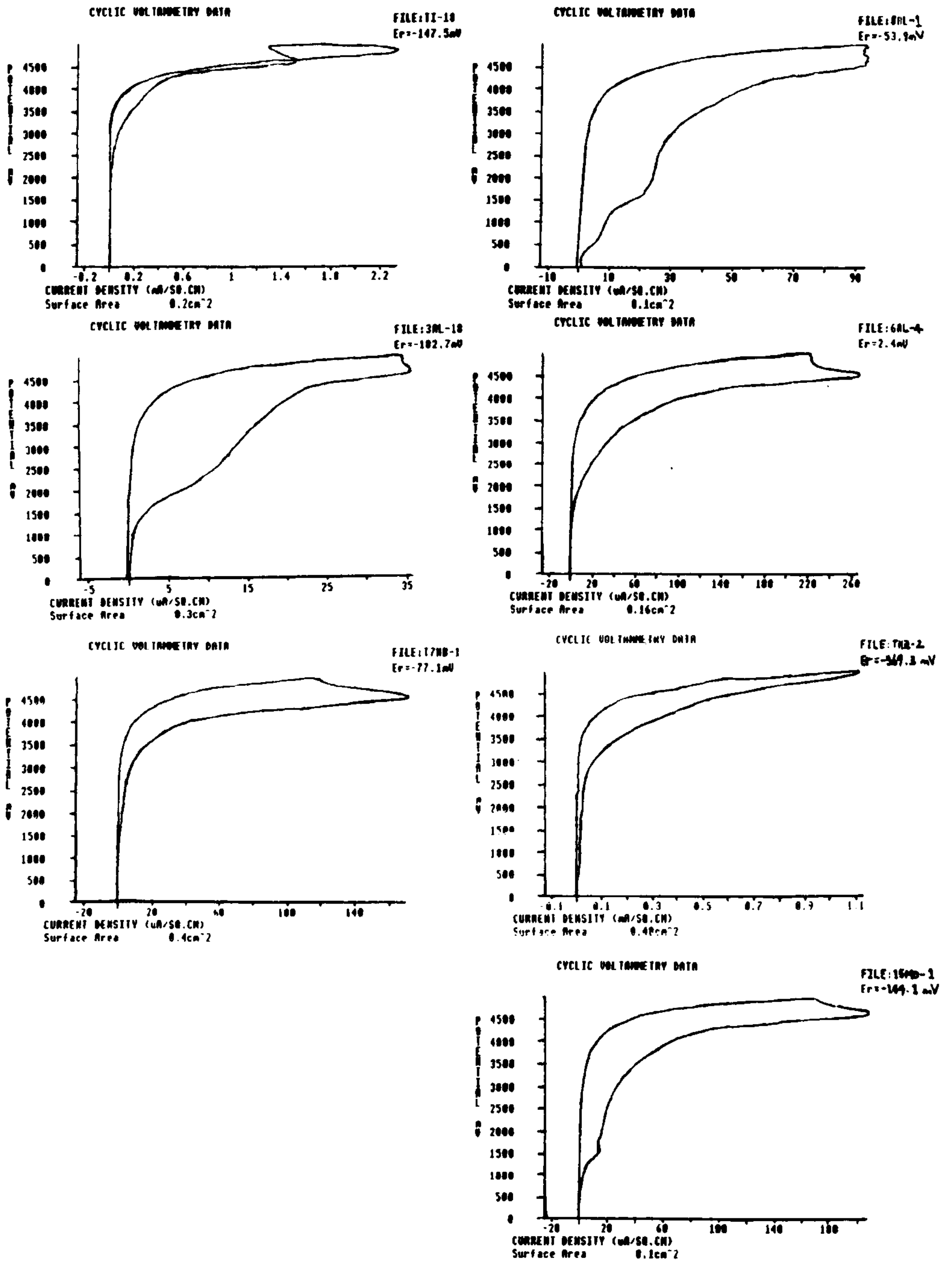

Figure 2 Potential vs current density curves for titanium alloys of varying composition.

are concerned (*Table 7*), the mixed phase alpha–beta alloys (Ti–6Al–4V and Ti–6Al–7Nb) were affected more than the other alloys. The reason for this may be that these alloys are quite hard initially (*Table 8*), and do not lose their hardness as a result of corrosion (*Table 5*). Thus, once they have worn they may be acting as hard abrasives themselves, giving rise to the effect of three-body wear. As a result of this, they would most definitely produce greater wear scars, and thus a rougher surface.

Other then Ti–15Mo, which was affected the least, the rest of the alloys (Ti–8Al–1Mo–1V, Ti–3Al–2.5V and Ti–13Nb–13Zr) displayed a comparable behaviour. An explanation for Ti–15Mo might be that it is a comparatively ‘soft’ material to start off with (*Table 8*), and thus may not contribute markedly towards three-body wear. This would produce the effect of less prominent wear scars on the surface (although it may have lost more material as a result of the wear process). This might also be the case for the other alloys, since they are also not as hard as Ti–6Al–4V and Ti–6Al–7Nb. In terms of changes in surface hardness, Ti–6Al–7Nb and Ti–3Al–2.5V followed by Ti–6Al–4V were superior in resisting any considerable change (*Table 6*). This suggests that corrosion has a more detrimental effect on the mechanical properties of either the alpha or beta alloys and that the mixed phase alpha–beta alloys suffer the least.

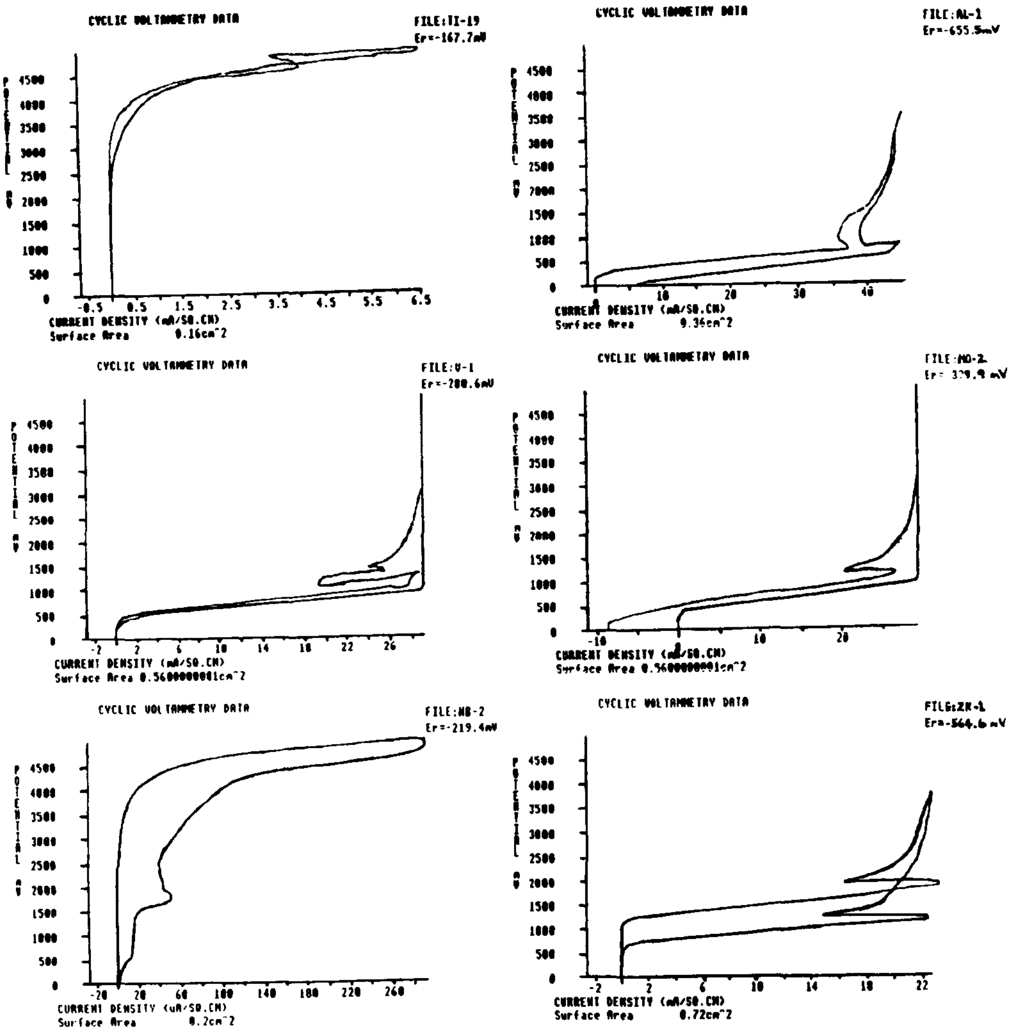

Figure 3 Potential vs current density curves of the different metals constituting the alloys.

Wear-accelerated corrosion

The effect of wearing the specimen prior to corrosion testing was investigated to observe how this action might influence the corrosion characteristics of each material. *Table 2* shows the pitting corrosion resistance (E_B–E_p) of each material with and without the effect of wear before corrosion.

Table 1 Pitting corrosion resistance (E_B–E_p) of the titanium alloys at varying pH levels

Material	Phase	E_B-E_p (mV) at pH = 5.0	$\overline{E_B-E_p}$ (mV) at pH = 5.0	E_B-E_p (mV) at pH = 7.4	$\overline{E_B-E_p}$ (mV) at pH = 7.4	E_B-E_p (mV) at pH = 9.0	$\overline{E_B-E_p}$ (mV) at pH = 9.0
Ti(cpi)	pure metal	200 300 400	300	200 550 600	450	500 700 750	650
Ti–8Al–1Mo–1V	near alpha	1600 2150 2400	2050	3250 3500 3600	3450	4050 4100 4150	4100
Ti–3Al–2.5V	alpha–beta	2900 2975 3200	3025	4200 4300 4400	4300	3350 3650 4100	3700
Ti–6Al–4V	alpha–beta	1900 2700 3200	2600	3350 3900 4300	3850	3850 3900 4100	3950
Ti–6Al–7Nb	alpha–beta	2300 2400 2800	2500	2900 3075 3100	3025	3850 4050 4100	4000
Ti–13Nb–13Zr	near beta	3100 3150 3200	3150	2800 2900 3000	2900	2300 2700 2950	2650
Ti–15Mo	beta	3100 3300 3350	3250	2800 3000 3200	3000	4050 4300 4400	4250

Table 2 A comparison of the pitting corrosion resistance (E_B–E_p) following corrosion and wear-accelerated corrosion

Material	Corrosion only		Wear-accelerated corrosion	
	E_B–E_p (mV)	$\overline{E_B-E_p}$ (mV)	E_B–E_p (mV)	$\overline{E_B-E_p}$ (mV)
Ti(cpi)	450	600	3000	3150
	650		3200	
	700		3250	
Ti–8Al–1Mo–1V	3150	3300	3850	4000
	3350		3900	
	3400		4250	
Ti–3Al–2.5V	4200	4300	4100	4300
	4300		4350	
	4400		4450	
Ti–6Al–4V	3600	3800	4050	4250
	3900		4200	
	3900		4500	
Ti–6Al–7Nb	3850	4000	3800	4150
	3950		4200	
	4200		4450	
Ti–13Nb–13Zr	3150	3400	3300	3500
	3450		3550	
	3600		3650	
Ti–15Mo	3700	3900	4050	4200
	4000		4200	
	4000		4350	

It was found that although Ti(cpi) possessed a much better corrosion resistance than any of the alloys during corrosion, following wear-accelerated corrosion its corrosion resistance was found to be comparable with the alloys. It seems that Ti–3Al–2.5V, Ti–13Nb–13Zr and to a lesser extent Ti–6Al–7Nb were least affected by wear. However, Ti–15Mo and Ti–8Al–1Mo–1V along with Ti and Ti–6Al–4V were all affected. It is also possible to assume that the wear-accelerated corrosion behaviour of all the materials is more severe than for corrosion only. However, there was a slight variation of this effect for each material. This behaviour is supported by a theoretical model, which suggests that the wearing mechanism removes the protective oxide film on the surface of the material, thus leaving the surface unprotected. As a result, when corrosion occurs it does so at a faster rate[55,57,61].

Table 3 The pitting corrosion resistance (E_B–E_p) of the metals constituting the alloys

Material	E_B–E_p (mV)	$\overline{E_B-E_p}$ (mV)
Titanium	200	250
	200	
	250	
Aluminium	200	300
	300	
	400	
Vanadium	100	100
	100	
	100	
Molybdenum	300	375
	375	
	450	
Niobium	4000	4200
	4250	
	4350	
Zirconium	650	700
	700	
	750	

General comments

The aim of this study was to compare the similarities and differences of titanium alloys under corrosion and wear conditions. In this respect, of all the materials tested, the mixed phase alpha–beta alloys (Ti–6Al–4V and Ti–6Al–7Nb) displayed the best combination of both corrosion resistance and mechanical strength, which would be an important factor from the point of view of an articulating joint prosthesis.

Thus, it is evident that if better alloys are to be found for such applications in the future, then primarily, they must have both alpha and beta phases present in their microstructure, possibly more beta than alpha to improve corrosion resistance, as has been shown. Also, as far as alloying composition is concerned, although it is preferable that metals such as aluminium be avoided, it seems that this metal continues to be a useful alpha stabilizer with essential strengthening properties to the alloy. However, niobium should continue to be a promising metal in successfully replacing vanadium as a beta stabilizer, without any significant compromises, and with potential benefits with respect to less toxicity.

CONCLUSIONS

This study has shown that following Ti(cpi), the beta (Ti–15Mo) or near-beta (Ti–13Nb–13Zr) alloys display the greatest resistance to corrosion. An investigation of the alloying elements has shown that

Table 4 Changes in surface roughness following corrosion at varying pH levels

Material	Change in surface roughness (μm)					
	At pH = 5.0		At pH = 7.4		At pH = 9.0	
	Ra	$\overline{Ra}$	Ra	$\overline{Ra}$	Ra	$\overline{Ra}$
Ti(cpi)	0.09	0.10	0.19	0.27	0.17	0.21
	0.09		0.22		0.18	
	0.12		0.40		0.28	
Ti–8Al–1Mo–1V	0.02	0.04	0.02	0.03	0.03	0.06
	0.04		0.02		0.07	
	0.06		0.05		0.08	
Ti–3Al–2.5V	0.19	0.23	0.13	0.19	0.01	0.01
	0.22		0.16		0.01	
	0.28		0.28		0.02	
Ti–6Al–4V	0.24	0.27	0.02	0.02	0.09	0.16
	0.27		0.02		0.17	
	0.30		0.02		0.22	
Ti–6Al–7Nb	0.11	0.15	0.01	0.02	0.06	0.08
	0.16		0.02		0.09	
	0.18		0.02		0.09	
Ti–13Nb–13Zr	0.27	0.44	0.01	0.01	0.03	0.04
	0.43		0.01		0.04	
	0.62		0.02		0.05	
Ti–15Mo	0.01	0.01	0.29	0.41	0.02	0.05
	0.01		0.42		0.04	
	0.02		0.52		0.09	

niobium and to a lesser extent zirconium produced the most favourable effects in terms of corrosion resistance, as opposed to molybdenum, vanadium and aluminium, in that order. The phenomenon of corrosive wear (contribution of corrosion to wear) as opposed to non-corrosive wear is more detrimental to the mixed phase alpha–beta alloys (i.e. Ti–6Al–4V and Ti–6Al–7Nb) than any of the others. Wear by itself is more detrimental to the other alloys than to these two. As far as the influence of wear on the corrosion of materials is concerned (i.e. wear-accelerated corrosion), it has been shown that Ti–3Al–2.5V, Ti–13Nb–13Zr and to a lesser extent Ti–6Al–7Nb are least affected. Conversely, Ti, Ti–15Mo, Ti–8Al–1Mo–1V and Ti–6Al–4V are all affected to a greater extent.

Table 5 Changes in surface hardness following corrosion at varying pH levels

Material	Change in surface hardness (VHN)					
	At pH = 5.0		At pH = 7.4		At pH = 9.0	
	VHN	$\overline{VHN}$	VHN	$\overline{VHN}$	VHN	$\overline{VHN}$
Ti(cpi)	8	14	11	19	9	14
	12		17		11	
	22		29		22	
Ti–8Al–1Mo–1V	110	122	6	7	126	147
	124		12		152	
	132		15		163	
Ti–3Al–2.5V	16	19	64	77	101	110
	17		82		108	
	24		85		121	
Ti–6Al–4V	92	106	8	12	153	162
	111		10		166	
	115		18		167	
Ti–6Al–7Nb	129	141	37	43	139	141
	142		39		139	
	152		53		145	
Ti–13Nb–13Zr	167	189	72	91	185	199
	199		93		205	
	201		108		207	
Ti–15Mo	98	110	60	76	100	114
	109		80		119	
	123		88		123	

Table 6 Changes in surface hardness following non-corrosive and corrosive wear

Material	Reduction in surface hardness (VHN)			
	Non-corrosive wear		Corrosive wear	
	VHN	$\overline{\text{VHN}}$	VHN	$\overline{\text{VHN}}$
Ti(cpi)	9	17	38	43
	19		44	
	23		49	
Ti–8Al–1Mo–1V	111	118	138	150
	116		152	
	127		160	
Ti–3Al–2.5V	120	123	111	128
	121		133	
	128		140	
Ti–6Al–4V	124	132	141	153
	135		153	
	137		165	
Ti–6Al–7Nb	142	152	142	159
	154		155	
	160		180	
Ti–13Nb–13Zr	139	145	161	175
	141		179	
	155		197	
Ti–15Mo	44	55	93	107
	60		112	
	61		116	

Table 8 Initial surface roughness and hardness values of alloys

Material	Surface roughness (μm)		Surface hardness (VHN)	
	Ra	$\overline{\text{Ra}}$	VHN	$\overline{\text{VHN}}$
Ti(cpi)	0.15	0.17	110	120
	0.16		118	
	0.19		132	
Ti–8Al–1Mo–1V	0.06	0.07	320	332
	0.07		336	
	0.07		340	
Ti–3Al–2.5V	0.10	0.10	300	311
	0.10		307	
	0.20		326	
Ti–6Al–4V	0.01	0.02	310	322
	0.02		318	
	0.03		338	
Ti–6Al–7Nb	0.01	0.02	338	349
	0.02		353	
	0.02		356	
Ti–13Nb–13Zr	0.05	0.06	346	353
	0.06		354	
	0.07		359	
Ti–15Mo	0.10	0.14	255	271
	0.16		266	
	0.16		292	

Table 7 Changes in surface roughness following non-corrosive and corrosive wear

Material	Change in surface roughness (μm)			
	Non-corrosive wear		Corrosive wear	
	Ra	$\overline{\text{Ra}}$	Ra	$\overline{\text{Ra}}$
Ti(cpi)	1.38	1.56	2.10	2.58
	1.64		2.39	
	1.66		3.25	
Ti–8Al–1Mo–1V	1.05	1.16	1.08	1.58
	1.19		1.79	
	1.24		1.87	
Ti–3Al–2.5V	0.48	0.53	0.62	0.82
	0.55		0.86	
	0.56		0.98	
Ti–6Al–4V	0.61	0.79	2.01	2.28
	0.77		2.15	
	0.99		2.68	
Ti–6Al–7Nb	0.80	0.86	1.45	1.68
	0.82		1.79	
	0.96		1.80	
Ti–13Nb–13Zr	0.98	1.09	1.22	1.36
	1.11		1.34	
	1.18		1.52	
Ti–15Mo	0.99	1.06	1.00	1.15
	1.08		1.18	
	1.11		1.27	

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