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To link to this article: DOI: 10.1016/j.corsci.2016.04.016
URL: http://dx.doi.org/10.1016/j.corsci.2016.04.016

To cite this version: Ciszak, Clément and Popa, Ioana and Brossard, Jean-Michel and Monceau, Daniel and Chevalier, Sébastien NaCl induced corrosion of Ti-6Al-4V alloy at high temperature. (2016) Corrosion Science, vol. 110. pp. 91-104. ISSN 0010-938X

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NaCl induced corrosion of Ti-6Al-4V alloy at high temperature

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A B S T R A C T

This paper presents a study on the Ti-6Al-4V behaviour in presence of NaCl deposit under dry and moist air environments at 560 °C. The results evidence a detrimental effect of the NaCl deposit with a synergistic effect in presence of moist air environment. Treatments under dry and moist air with NaCl deposit for 600 h, lead respectively to weight gains per unit area 5 and 15 times higher than observed under classic oxidation in dry air. Enhancement of the corrosion phenomenon is attributed to the presence of gaseous metal chlorides, leading to the establishment of an active corrosion process.

Keywords:
A. Titanium
A. Alloy
B. SEM
B. X-ray diffraction
C. High temperature corrosion

1. Introduction

Ti-based alloys have been increasingly used in various domains such as chemical industry, medical engineering and aerospace for their high corrosion resistance, bio-compatibility, high specific strength and low-density. Their major asset in high temperature applications is the presence of addition elements, such as aluminium, which enhance their oxidation behaviour. This beneficial effect towards oxidation resistance is mainly due to the formation of alumina Al\textsubscript{2}O\textsubscript{3} in the external part of the rutile layer [1–4].

Nevertheless, oxygen can diffuse deep inside the metallic substrate to form an oxygen dissolution area, which leads to embrittlement of the affected area [3,5,6]. Oxygen dissolution in the Ti matrix explains the gradual increase of the material hardness when approaching the metal-oxide interface. Various studies evaluated the depth of the oxygen dissolution area; they show that the oxygen dissolution is dependent on both temperature and time (for instance, the oxygen dissolution area is 10 \( \mu \)m deep after 70 h at 600 °C [3], but is deeper than 15 \( \mu \)m after only 3 h at 750 °C [6]).

Titanium alloys are sometimes asked to work beyond their original limits, especially in terms of thermal stresses, revealing uncertainties about their behaviour at high temperature (typically above 400 °C for the Ti-6Al-4V alloy). Another parameter that can influence significantly the material behaviour is the gaseous environment, especially the humidity. The material may even, in some cases, come in contact with species such as silica, sea salt or sulphates.

Humidity influences the oxidation mechanisms of the Ti-6Al-4V alloy, as in moist air, the corrosion scale is formed of two layers of rutile TiO\textsubscript{2}, separated by an alumina rich area [7]. These three layers generally contain many defects such as pores and/or whiskers for the outer part [6], which are typical of the water vapour presence [8]. This difference in morphology would therefore suppose the existence of different oxidation mechanisms and oxygen dissolution depth under moist air oxidation. In fact, Pérez [9] showed that after being treated at 700 °C for 150 h, samples of pure titanium exposed to laboratory air and moist air present an oxide layer of 15 \( \mu \)m and 17 \( \mu \)m thick respectively. The oxygen dissolution area thickness enlarges from 30 \( \mu \)m to 40 \( \mu \)m. This larger oxygen dissolution depth could be attributed to the faster diffusion of OH\textsuperscript{−} ions compared to O\textsuperscript{2−} ions (because of their smaller size and lower charge), the hydroxyl ions coming from the adsorption and the dissociation of the water molecules at the oxide/atmosphere interface [10]. Other results are however more contrasted. Motte et al. [7] reported higher weight gain per unit area whereas Champin et al. [11] reported lower weight gain per unit area after treatment in moist air compared with dry air treatment. It is moreover mentioned that the oxidation kinetic is parabolic at the beginning of the oxidation [6,8], tending to a sub-parabolic regime for long duration treatments [8].

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http://dx.doi.org/10.1016/j.corsci.2016.04.016
Moreover, as it was described by Grabke et al. for low and high alloyed steels [12] and by Kawahara for a Ni-based alloy [13], corrosion phenomena can be more aggressive if salt crystals condensate on the material. This corrosion phenomenon generally involving solid or liquid mixed environments of NaCl and Na₂SO₄ is called “hot corrosion” [14–16]. Disastrous effects can occur at operating temperatures higher than the melting point of the mixture, what is defined as the type I of hot corrosion [14–16]. The effect appears to be more moderate below this temperature, defining the type II of hot corrosion [14–17], although just as harmful for the material. In his Ph.D. work, Tsaur [17] have also demonstrated that enhanced corrosion phenomenon can occur even if only solid NaCl is present. Very few studies on titanium alloys were done in these particular conditions. They showed parabolic oxidation kinetics 6 times larger than those observed in dry air, for a titanium alloy IMI 685 [18]. The observed corrosion layer is then porous, cracked and non-adherent to the metallic substrate. The oxide scale is rather dense in the inner part and rather porous in the outer part [18]. Both are composed of similar species to those observed after dry air treatment, namely: TiO₂ (rutile [18,19] and anatase [19]) phases optionally containing alloying elements [19], Al₂O₃ in the case of the Ti 624Z alloy [20] and sometimes the bi-component Na₂O, TiO₂ [18]. Concerning the formation of these compounds, the same studies presume an inward diffusion of oxygen [18–20] and Cl⁻ [18,19] anions, concomitant with an outward diffusion of the metallic cations from the substrate [18] (increased by the presence of pores and cracks [19]). These hypotheses are often accompanied by assumptions of reactional mechanisms [19], which are thermodynamically consistent [20], but not demonstrated experimentally. They are systematically based on volatile chloride species, which lead to the formation of TiO₂ rutile and in a particular case, to a mixed oxide Na₂Ti₅O₁₃ as observed by Gurappa [21], accompanied by a release of gaseous Cl₂ (which could justify the presence of blisters and pores). Furthermore, the dissolution phenomenon also appears to be strongly influenced by the presence of salts on the material surface, an oxygen dissolution area of about 500 μm deep has been observed after a treatment at 600 °C for 65 h in presence of molten salts mixture (90% Na₂SO₄, 10% NaCl) [21].

The few studies combining the two previous exposition conditions (moist air and NaCl deposit) report an increase of the weight gain per unit area of the samples (for Ti-6Al-4V [22] and Ti60 [23,24] alloys), up to a factor of 100 for an IMI 685 Ti-based alloy [18]. The morphology of the formed corrosion products is also affected by high porosity and roughness. High weight gains per unit area (compared to those observed after dry air treatment) are justified by an approach of the oxidation mechanisms combining the two previous approaches. Dissociation of water molecules in H⁺ and OH⁻ would take place on defects present on the TiO₂ layer surface [24]. Concerning the reaction between the metallic substrate with NaCl and H₂O, it would result in the formation of titanium oxide TiO₂ [18,22–24], a mixed sodium-titanium oxide Na₂TiO₃ [22–24] and the release of gaseous di-hydrogen [22,23] and hydrochloric acid [22–24].

The objective of the present study is to better understand the influence of the NaCl deposit on the oxidation behaviour at high temperature (500 °C, 560 °C and 625 °C) of a largely used Ti alloy, Ti-6Al-4V, in both dry air and moist air conditions. In this purpose, four environments were considered: laboratory air, moist air, laboratory air with NaCl deposit and moist air with NaCl deposit. Several preliminary tests were performed at the 3 temperatures mentioned above; however this paper focuses on the intermediate temperature of 560 °C.

Corrosion products were analysed by means of X-rays diffraction (XRD) and scanning electron microscopy (SEM) coupled to energy dispersive X-ray (EDS) analysis, while the oxygen disso-

Fig. 1. Weight gain per unit area evolution with time of samples oxidized in laboratory air for 600 h at 500 °C, 560 °C and 625 °C (a), and weight gain per unit area evolution with time square root of same samples (b).

<table>
<thead>
<tr>
<th>Elements</th>
<th>Al</th>
<th>Ti</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>Chemical composition (wt.%)</td>
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<td>90</td>
<td>4</td>
</tr>
<tr>
<td>Chemical composition (at.%)</td>
<td>11</td>
<td>86</td>
<td>3</td>
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</table>

Table 1 Average chemical composition of the as received Ti-6Al-4V alloy, obtained by EDS analysis.
measure approximately 16 mm x 9.0 mm and are 1.0 mm thick. All the specimens were mechanically grinded on SiC papers up to 600 grit [18–20,24], and were then cleaned in ethanol.

NaCl deposits were performed, by spraying in several steps, an aqueous saturated NaCl solution on the samples surfaces which were preheated at 60–80 °C. This method leads to a homogeneous and continuous deposit with a NaCl amount of 3–4 mg cm⁻².

Samples exposed at laboratory air, with and without NaCl deposit, were exposed to laboratory air in Carbolite CWF 1300 muffle furnaces. Air oxidation was performed at three temperatures (500 °C, 560 °C and 625 °C) for eight durations (24, 48, 100, 200, 300, 400, 500 and 600 h) to establish discontinuous weight gain measurements. Exposition to moist air (12 vol. % H₂O(g)), with and without NaCl deposit, was done in an experimental tubular furnace at 560 °C during 600 h. Air enrichment in water vapour is performed by bubbling air in a round bottom flask of distilled water heated at 100 °C, which leads to water vapour saturated air. This mixture passes then through a condenser tube which is heated at 25 °C to set the quantity of water vapour at 12 vol. %, measured with an hygrometer [25]. The steam then passes through the furnace tube with a flow of 8 mL·min⁻¹.

All samples were analysed in two successive steps. Surfaces were first characterized, followed by cross-sections analysis of the same samples. Preparation of cross-sections consisted of 4 steps. At first, a thin layer of Au was sputtered on each sample surface. Then, samples were coated with Cu by an electrolytic method. They were
finally coated in carbon doped phenolic resin and polished up to the mirror finishing.

The oxide surface and cross-section morphologies were analysed using a JEOL JSM-7600F scanning electron microscope. The chemical composition of the corrosion products was obtained by energy dispersive X-ray spectroscopy (EDS). Phase composition of the oxide scales was determined by X-ray diffraction with an Inel CPS 120 diffractometer using Cu Kα (λ = 0.154 nm) radiation with fixed incidence angle from 1° to 12° in function of corrosion scales thickness. The micro-hardness of each sample was evaluated with a BUEHLER 1600–6100 apparatus using a Vickers diamond head with a charge of 10 g, leading to indentation prints between 5 and 10 μm. Measurement points were done on the whole thickness of the metallic material, each 10 μm near the surface (for the first 100 μm), each 100 μm from 100 μm to 500 μm and then each 500 μm from 500 μm to the last 500 μm in the middle of the sample.

Thermodynamics calculations were performed with the FactSage 6.4™ software in order to propose corrosion mechanisms.

2. Results

2.1. Laboratory air oxidation

The evolution with time of the weight gains per unit area of the samples oxidized under laboratory air at 500 °C, 560 °C and 625 °C for 600 h are presented in Fig. 1a. The oxidation approximately follows a parabolic law rate at the three considered temperatures, according to Fig. 1b. The k_p values obtained from these linearized curves are 7.0 × 10^{-15} g^2 cm^{-4} s^{-2}, 2.0 × 10^{-13} g^2 cm^{-4} s^{-2} and 3.8 × 10^{-13} g^2 cm^{-4} s^{-2} respectively, according to the following formula:

\[
\frac{\Delta m}{S} = k_p t
\]

where \(\Delta m\) is the weight gain per unit area in g cm^{-2}, S is the sample surface area in cm^2, t is the time in s and \(k_p\) is the parabolic constant of oxidation in g cm^{-2} s^{-1}. There are in the literature, only two available values of \(k_p\) for shorter durations of oxidation than 600 h. The first \(1.7 \times 10^{-13} g^2 cm^{-4} s^{-2}\), proposed by Siab et al. [1], was determined after an oxidation treatment of 5.5 h at 600 °C. The second \(3.0 \times 10^{-12} g^2 cm^{-4} s^{-2}\), was proposed by Zhang et al. [26] and determined from an oxidation treatment of 50 h at 600 °C. These values are both higher than the one determined in this study at 625 °C after 600 h oxidation. These differences could be explained by a transient effect, which is known to lead to faster kinetics at the beginning of the oxidation, as it can be noticed on the kinetic curve relative to 625 °C oxidation (Fig. 1a).

Surface observation of the sample treated during 600 h at 560 °C in laboratory air shows a homogeneous aspect composed of fine oxide grains (Fig. 2a). EDS analysis indicate the presence of Ti, O and Al (respectively 23 at.% 68 at.% and 9 at.%). This result coupled with XRD analysis (Fig. 3) indicates mainly the presence of rutile TiO₂ (ICDD n° 00-021-1276), and α-Al₂O₃ transient alumina (ICDD n° 00-056-1186). Cross section samples (Fig. 2b) reveal the presence of a thin (2 μm), dense and adherent oxide layer. EDS elementary maps (Fig. 2c.d.e.f) reveal that aluminium oxide is at the outer part of the corrosion layer (~300 nm), whereas titanium oxide is in the inner part (~1.7 μm).

Inside the metallic substrate, light regions enriched in V can be observed, corresponding to the β-phase present in the Ti-6Al-4V alloy. The dark grains are Al-rich and correspond to the α-phase. Fig. 4 presents the micro-hardness profile versus the distance from the metal/oxide interface. The oxygen EDS profile is equally presented. A very large hardness value of 760HV0.01 is measured at the external part of the substrate close to the metal/oxide interface. Then the hardness value very slowly decreases with the distance from the metal/oxide interface: at around 20 μm, it reaches the average value measured for the Ti-6Al-4V bulk alloy (290HV0.01). In parallel, the O content equally decreases from 44 at.% at the interface to 3 at.% at 10 μm from the interface. It should be noted here that because of O surface pollution, the O content measured by EDS is certainly over-evaluated of several at.%. Therefore the above values should not be taken as quantitative. These evolutions prove the oxygen dissolution in the Ti-6Al-4V metallic substrate resulting in an important evolution of the mechanical properties. Fig. 4 allows evaluating the oxygen dissolution area thickness to around 10 μm.

2.2. Moist air oxidation

The weight gain per unit area is slightly lower after 600 h exposure at 560 °C in moist air (0.415 ± 0.065 mg cm^{-2}) than under laboratory air in the same conditions (0.605 ± 0.07 mg cm^{-2}).

The surface morphology of the sample exposed to wet air is very similar to that observed after the laboratory air treatment (Fig. 5a). The oxide layer is homogeneous and composed of fine grains. EDS analysis performed on this surface indicate the presence of Ti, O and Al (respectively 24 at.% 66 at.% and 10 at.%). Cross-section observations reveal that the scale is thin (2 μm), dense and adherent to the metallic substrate (Fig. 5b). The previous EDS analysis coupled with EDS elementary mapping (Fig. 5c–f) and XRD (Fig. 6) indicates that the oxide scale is composed of an outer alumina layer (~600 nm) (ICDD n° 01-080-0855) and an inner rutile layer (~1.4 μm) thick (ICDD n° 00-021-1276).

Fig. 7 presents the micro-hardness profile and the O content as a function of the distance from the metal/oxide interface. Results are similar to those observed for the oxide in laboratory air. The same value of hardness of 760HV0.01 is measured at the external part of the substrate. This hardness value very slowly decreases with the distance from the metal/oxide interface: at 20 μm, it is stabilised to a value close to the average value measured for raw Ti-6Al-4V. In parallel, the O content also decreases from 28 at.% to 3 at.% at 10 μm from the interface. In consequence, the oxygen dissolution area thickness can be evaluated to around 10 μm. Here again,
as explained before, the O content measured by EDS suffers from surface pollution and is certainly over evaluated of several at.%.

2.3. NaCl induced corrosion in air

NaCl coated sample exposed for 600 h at 560 °C to laboratory air show a weight gain per unit area 5 times higher than the sample exposed to laboratory air in the same conditions (Fig. 8).

Surface observations of the sample (Fig. 9) reveal a heterogeneous aspect with blisters and cracked corrosion products. EDS analysis indicate the presence of Ti, O and Na (respectively 22 at.%, 64 at.% and 14 at.%). Based on this average chemical composition, XRD analysis (Fig. 10) evidence the presence of rutile, TiO$_2$ (ICDD n° 01-070-7347), as the main phase and two types of sodium-titanium mixed oxides, Na$_4$Ti$_5$O$_{12}$ (ICDD n° 00-052-1814) and Na$_x$TiO$_2$ (ICDD n° 00-022-1404). In addition, XRD analyses performed at different times of oxidation (24, 48, 100, 200, 300, 400, 500 and 600 h) reveal an evolution of the mixed Na-Ti oxide phases ratio, the Na$_x$TiO$_2$ phase becoming preponderant at long times of oxidation.

Cross sections indicate the presence of a thick (35 μm), porous and non-adherent oxide layer on the material surface. This morphology is in agreement with the work of Dumas and St. John [18] on the NaCl coated IMI 685 alloy oxidized at 600 °C during 24 h. Presence of large cracks between metallic substrate and oxide layer can also be noticed (Fig. 11a). EDS elementary X-ray maps (Fig. 11b–h)
show that the oxide scale is formed of four different layers. The outer part is a mixed sodium-titanium oxide which, in agreement to XRD surface and EDS analyses, should correspond to Na$_2$TiO$_3$ type (with x close to 0.23, according to the XRD pattern) and Na$_4$Ti$_5$O$_{12}$ phases. Just below this layer, there is a very thin aluminium rich oxide, which was not detected by XRD, certainly because of its small amount. The two inner layers are titanium oxides with different Ti contents (the Ti content seems to be higher in the closest part to the metal/oxide interface). Taking into account the XRD results, the outer part of this titanium oxide layer should correspond to TiO$_2$. However, it is very difficult to identify whether the inner part corresponds to a different phase, as its position with respect to the sample surface makes it difficult to be reached by the X-rays. However, according to EDS quantitative analysis (41 at.% of Ti, 51 at.% of O, 5 at.% of Al, and 3 at.% of V), this phase should correspond to TiO$_2$ with a small proportion of addition elements. Presence of a small amount of Cl can also be noticed in this Ti layer that is the closest to the metal/oxide interface. This interface is the less adherent and location for initiation of corrosion product spallation. The outer part of the metallic substrate presents an Al rich layer that was not observed during the previous tests without NaCl deposit.

Fig. 12 presents the micro-hardness profile and the O content versus the distance from the metal/oxide interface. The oxygen content is constant and close to 0 at.%, showing the absence of any oxygen dissolution area. Even if the micro-hardness profile stays constant on the whole thickness of the sample, it nevertheless shows a small decrease at each side of the metallic substrate, corresponding to the first 5 µm from the metal/oxide interface, probably related to the effect of approach of the porous area.

2.4. NaCl induced corrosion in moist air

The sample exposed 600 h at 560 °C to moist air with a NaCl deposit shows a weight gain per unit area 15 times higher than the uncoated samples exposed in laboratory and moist air. Surface observations (Fig. 13) of the sample indicate a heterogeneous aspect with large cavities. EDS analysis indicates the presence of Ti, O and Na (respectively 25 at.%, 65 at.% and 10 at.%). This indicative chemical composition coupled with XRD (Fig. 14) analysis indicates the presence of several kinds of titanium oxide: rutile TiO$_2$ (ICDD n° 01-070-7347) as the main phase, but also Ti$_2$O$_3$ (ICDD n° 00-043-1033) and TiO (ICDD n° 00-012-0754). Two types of sodium-titanium mixed oxides, Na$_4$Ti$_5$O$_{12}$ (ICDD n° 00-052-1814) and Na$_2$Ti$_3$O$_7$ (ICDD n° 00-037-0345), are also identified.

Cross section observations (Fig. 15a) evidence the presence of a very thick (90 µm), porous and non-adherent oxide layer on the material surface. The presence of a large crack inside the corrosion products layer can also be observed. EDS elementary X-ray maps (Fig. 15b–h) reveal that the oxide layer is stratified into four parts. The outer part is thin (≈ 5 µm) and porous and corresponds to mixed sodium-titanium oxide, which is probably a mixture of Na$_4$Ti$_5$O$_{12}$ and Na$_2$Ti$_3$O$_7$ according to XRD results. The three layers below are titanium oxide with an increasing Ti content. EDS quantifications done in several regions of these different layers suggest that they correspond to TiO$_2$ (32 at.% for Ti and 68 at.% for O), Ti$_2$O$_3$ (41 at.% for Ti and 59 at.% for O) and TiO (48 at.% for Ti and 52 at.% for O), respectively. A small amount of Cl is present inside the
inner titanium oxide layer near the interface with the metal. The oxide layer adherence at this interface is very poor and can lead to spallation. As in the case of the NaCl coated sample exposed to air, at the external part of the metallic substrate, an Al-rich layer was observed, corresponding to a Ti depletion.

Fig. 11. SEM micrograph of a sample of Ti-6Al-4V oxidized in laboratory air with a NaCl deposit at 560 °C for 600 h, (a) cross-section and corresponding X-ray maps of Na (b), Cl (c), Cu (d), V (e), O (f), Al (g) and Ti (h) elements.

Fig. 10. Diffraction pattern of the XRD analysis performed on a sample of Ti-6Al-4V oxidized in laboratory air with a NaCl deposit at 560 °C for 600 h.
of temperature (850 °C) and humidity (P(H₂O) ≈ 220 Pa). In parallel, Pérez et al. [9] showed that in the same temperature range (700 °C), the oxide layers are equivalent after exposures of pure Ti in dry and moist air, but using a dewpoint of 8 °C which corresponds to a very low concentration in water (≈ 1 vol.%). The same oxidation kinetics are equally observed. However, the oxygen diffusion thickness is slightly higher under moist air than in dry air. These observations were related to different oxidant species in moist air. The adsorption of water molecules on the rutile surface leads to their dissociation in H⁺ and OH⁻ ions [27]. Due to its smaller size compared to O²⁻, OH⁻ is supposed to diffuse more rapidly [10], resulting in faster oxidation kinetics (including the formation of the oxide scale and the formation of the oxygen dissolution area) with increasing the water vapour pressure. The differences between the results of the present study and those of the literature could be due to the temperature, which is lower in the present study. It could also be due to the alloying elements, which are known to influence the behaviour of the materials under oxidative conditions. Zeller et al. [27] previously showed that increasing the Al-content of the Ti-alloys reduced the difference between the water vapour containing atmosphere and dry air. For Ti-Al alloys able to form protective alumina scales, no significant effect of the water vapour on the oxidation kinetics was reported [27]. The influence of water vapour strongly depends on the volume contents of TiO₂ and Al₂O₃ in the oxide scale.

The NaCl deposit was shown to be extremely harmful for the material behaviour in laboratory air; the weight gain per unit area being 5 times higher than for the NaCl free sample, while the oxide thickness was more than 15 times higher. This increase and slightly different evolution of weight gains (despite the presence of some divergent values) (Fig. 8) suppose the existence of a different oxidation mechanism. Moreover, presence of pores, blisters, cracks and the poor adherence of the oxide scale to the substrate are complementary evidences of the detrimental behaviour of Ti-6Al-4V in active corrosion conditions. The few studies previously done on the NaCl induced oxidation of the Ti alloys did not allow to clearly identify the reaction mechanism, mainly because of the lack of thermodynamic data needed to confirm the formation of such complex corrosion layers. Xiong et al. [24] suggested, in the case of the Ti60 alloy, that pores and cracks could be due to the rapid inward diffusion of corrosive species through the channels (porosity and cracks) in the corrosion products up to the metal/oxide interface. This accelerated corrosion was explained by the authors through the reaction of titanium dioxide with NaCl to form a mixed titanium-sodium oxide and a volatile chloride [23,24,28], supported by some thermodynamic calculations. However, the mechanism could not be completely demonstrated because of the lack of some thermodynamic data and positive value of standard free enthalpy for the initiating reaction.

Knowing that NaCl(s) melts at 800 °C and vaporizes at 1515 °C [29,30], it is obvious that NaCl remains essentially solid at the corrosion test temperature 560 °C (positive ΔG° for the sublimation of NaCl(s) (Table 2)).

In previous studies related to active oxidation beneath molten salts deposit, it has been demonstrated that if there is no pre-existing chlorine in the gas phase, the chlorine source must be
the molten chlorides [31,32]. In these studies case, chlorine can be released by the oxidation of the molten chloride. However, in our case the sodium chloride is not melted and cannot be oxidized at the given test temperature according to predominance diagram Fig. 19. Thus another source of chlorine should exist and the following hypothesis is suggested.

It is well known that for any liquid compound, equilibrium between liquid and gaseous states is established, leading to the creation of a vapour pressure of this compound (mainly dependent of the temperature). This phenomenon, at a lesser extent, is also true for solids [33,34]. Considering constant equilibrium between NaCl$_{(l)}$ and NaCl$_{(g)}$ (Eq. (2)), it appears that NaCl vapour pressure at 560 °C is around $3.12 \times 10^{-7}$ bar, according to Table 2, Fig. 17 and
Fig. 16. Micro-hardness (○) and relative oxygen content (●) (obtained by EDS) profiles versus distance from the metal/oxide interface Ti-6Al-4V oxidized in moist air with a NaCl deposit at 560 °C for 600 h.

Fig. 17. Vapor pressure evolutions of major chloride compounds formed by the reaction of NaCl with Ti-6Al-4V in air, and vapour pressure of NaCl between 460 °C and 760 °C (determined by FactSage 6.4/FactPS – FTSalt – thermodynamic calculation using reaction module for 1 mol of species).

Fig. 18. SEM surface micrograph of the metal/oxide interface of Ti-6Al-4V oxidized in laboratory air with a NaCl deposit at 560 °C for 600 h, presenting some NaCl grains with rounded edges.

![Table 3](image)

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<th>No.</th>
<th>Equilibrium equation</th>
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<td>(5)</td>
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<td>(6)</td>
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<td>(7)</td>
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<td>(8)</td>
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<td>(9)</td>
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<td>(10)</td>
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</tbody>
</table>

which may attest of their progressive gas releasing (to maintain the solid ↔ gas equilibrium at the given temperature). Secondly, this phenomenon has also been verified by exposing an uncoated sample to gaseous environment including a crucible containing NaCl upstream. The characterisations of this sample revealed the presence of Na on the sample surface, which suppose a transport of the NaCl through the gas flow to the sample surface, accrediting the thesis of a gaseous state for NaCl.

Moreover, calculations of standard free enthalpies provide positive values for reactions involving solid NaCl (Eq. (3) in Table 3), whereas negative values are obtained with gaseous NaCl (Eq. (4)).

4NaCl(s) + O₂(g) + 2TiO₂(g) ↔ 2Na₂TiO₃(s) + 2Cl₂(g)  
4NaCl(g) + O₂(g) + 2TiO₂(g) ↔ 2Na₂TiO₃(s) + 2Cl₂(g)

In consequence, initiating reaction between NaCl, O₂, and TiO₂ should involve gaseous NaCl instead of solid NaCl. According to thermodynamic predictions, these considerations are valid for reaction involving Na-Ti oxides, which are referenced in databases as (Na₂O)(TiO₂)₃ (equivalent to Na₂Ti₃O₇), (Na₂O)₂TiO₂½ (equivalent to Na₂Ti₃O₇) and Na₂TiO₃. But, SEM-EDS and XRD analyses
evidenced the presence of Na₄Ti₅O₁₂ and Na₂TiO₃ (Eq. (5)) for which no thermodynamic data are available.

$$4NaCl \text{(g)} + O_2 \text{(g)} + 5TiO_2 \text{(s)} \leftrightarrow Na_4Ti_5O_{12(s)} + 2Cl_2 \text{(g)} \quad (5)$$

The assumption of the formation of Cl₂(g) is confirmed by thermodynamic predictions of gaseous compounds present after the reaction 4 (Table 3, Fig. 19). In the experimental conditions of the present study, Cl₂(g) appears to be the first of the major gaseous species that are formed during the reaction mentioned above. Moreover, according to the predominance diagram of species (Fig. 20), at 560 °C in high P(O₂) and low P(Cl₂) area (which would correspond to the oxide/atmosphere interface in the present case), NaCl should be destabilised, leading to the formation of Na-Ti oxide in presence of TiO₂. This Na-Ti oxide would be (Na₂O)(TiO₂)₆, which would correspond to Na₂Ti₃O₁₃. As a consequence, the presence of both Na₄Ti₅O₁₂ and Na₂TiO₃ intermediate compounds (Fig. 21) suggests that the system would not be in the equilibrium state, which could lead to the formation of such compounds. The evolution of the Na₂Ti₅O₁₂/Na₂TiO₃ ratio with time, according to XRD analyses, could be another evidence of this non-equilibrium state.

However, it is rather difficult to state where this reaction takes place. In the present case, a majority of NaCl remains at the top of the oxide layer, while a few grains are present inside the oxide scale and at the oxide/substrate interface, probably stabilised by the low P(O₂) and the high P(Cl₂), according to the predominance diagram of species (Fig. 20). This presence of NaCl grains at the top of the oxide layer suggests that the reaction of Eq. (4) would take place at the NaCl/native TiO₂ interface.

According to the hypothesis that NaCl(g) acts as precursor in the Cl₂(g) releasing mechanism (Eq. (4)), it is supposed that only Cl₂(g) takes part in the next mechanism steps. Thus, the as-released Cl₂(g) can be then partially released in the atmosphere, but some might migrate down to the metal/oxide interface, according to its activity gradient, and react with metallic Ti in order to form a Ti chloride following the here below reaction (Eq. (6) in Table 3) with a high negative value of standard free enthalpy.

$$Ti(s) + 2Cl_2(g) \leftrightarrow TiCl_4(g) \quad (6)$$

This reaction of Cl₂(g) with Ti(s) at the metal/oxide interface will lead to a continuous evaporation of TiCl₄(g), supporting an increased corrosion kinetics. The existence of this mechanism is experimentally supported by the residual presence of Cl in the part of the oxide layer closest to the metal/oxide interface (probably resulting from titanium chloride condensation during cooling and partially dissolved by water used for grinding). The rapid Ti consumption is proven by an Al enrichment at the metallic substrate surface and by the presence of a porous metal area below the substrate surface, in agreement with Xiong et al. results [24]. In parallel, the low adherence of corrosion products to the substrate also supports an active oxidation mechanism involving metallic chloride volatilisation.

According to the diagram of species predominance established as a function of O₂ and Cl₂ partial pressure (Fig. 20), TiCl₄(g) is con-
considered to be the major compound formed by the reaction between Ti(s) and Cl(g) (Eq. (6)) due to its stabilisation under high partial pressure of Cl₂ and low partial pressure of O₂, which is consistent with the expected atmosphere at the metal/oxide interface. Moreover, this titanium tetra-chloride appears to have the highest pressure vapour (Fig. 17) and in consequence, to be the most stable species of titanium chlorines in these experimental conditions.

The last step of the mechanism would involve the reaction of TiCl₄(g) with the inward diffusing oxygen leading to the formation of titanium dioxide and gaseous chlorine (Eq. (7)):

\[ \text{TiCl}_4(g) + O_2(g) \rightarrow \text{TiO}_2(s) + 2\text{Cl}_2(g) \] (7)

In this way, gaseous Cl₂ is produced again and can be used to react with Ti as in reaction 4. The last two reactions will take place cyclically and keep the mechanism of active oxidation up.

Same reactional sequences could be applied to Al and V (with the formation of AlCl₃(g) [24] and VC₁₂(g)), however neither aluminium-sodium oxide nor vanadium-sodium oxide were put in evidence in our case. These reactions were probably not favoured thermodynamically and/or their kinetic was not dominant.

The three suggested steps of the mechanism of Ti corrosion in presence of NaCl (Eqs. (4), (6) and (7)) are summarised in Fig. 22.

Concerning the absence of an oxygen diffusion area, this finding also attests of the preponderance of the corrosion kinetics over the inward oxygen diffusion kinetics into the metal. The metal surface recession due to active corrosion is faster than the inward diffusion of oxygen, and then no O-enriched metal layer is found.

The detrimental effect of the NaCl deposit is increased in presence of water vapour, as it is demonstrated by the very high mass gain per unit area, 15 times higher than in laboratory air. The corrosion products layer is even thicker (more than 2.5 times) than the one formed in laboratory air with NaCl deposit only. The similarity of the compounds obtained with NaCl deposit in both laboratory and moist airs, residual presence of Cl on the part of the oxide layer closest to the metal/oxide interface, Al enrichment at the metallic substrate surface and absence of an oxygen diffusion area suppose that the corrosion mechanisms are equivalent. Xiong et al. [24] mentioned that the presence of water vapour could lead to the formation of two other harmful compounds [23]: HCl and H₂. Water molecules can react with titanium dioxide, gaseous NaCl (coming from the solid NaCl deposit) and oxygen to form a mixed Na-Ti oxide and gaseous hydrochloric acid (Eq. (9) and (10)). This assumption is supported by the results of thermodynamic calculation presented in Fig. 23, which shows that HCl can be formed in higher quantity than Cl₂. As in laboratory air exposition, the thermodynamic equilibrium reaction involving solid NaCl (Eq. (8)) has a positive standard free enthalpy, while the reaction involving gaseous NaCl (Eq. (9)) has a negative standard free enthalpy (see Table 3).

\[ 4\text{NaCl}(s) + 2\text{H}_2\text{O}(g) + \text{TiO}_2(s) \rightarrow 2\text{Na}_2\text{TiO}_3(s) + 4\text{HCl}(g) \] (8)

\[ 4\text{NaCl}(g) + 2\text{H}_2\text{O}(g) + \text{TiO}_2(g) \rightarrow 2\text{Na}_2\text{TiO}_3(g) + 4\text{HCl}(g) \] (9)

As Na₄Ti₅O₁₂ phase is equally present after moist air exposition, a corresponding reaction can be written (Eq. (10)), but for the reasons explained above, its standard free enthalpy cannot be calculated.

\[ \text{NaCl}(g) + 2\text{H}_2\text{O}(g) + 5\text{TiO}_2(g) \rightarrow \text{Na}_4\text{Ti}_5\text{O}_{12}(s) + 4\text{HCl}(g) \] (10)

Analogously to the previous case (reaction between Ti-6Al-4V and NaCl in laboratory air), which involved the migration of gaseous chlorine, this volatile hydrochloric acid could then migrate up to the metal/oxide interface and react with metallic titanium to form gaseous titanium chloride and hydrogen (Eq. (11)).

\[ \text{Ti}_4(s) + 4\text{HCl}(g) \rightarrow \text{TiCl}_4(g) + 2\text{H}_2(g) \] (11)

Created hydrogen may then be evacuated out of the reaction area by outward diffusion through pores and cracks or/and diffuse inward to dissolve inside the metallic substrate. Beyond the solubility limit of hydrogen in both α and β phases, hydrogen reacts with titanium to form titanium hydrides. These specific compounds are known to degrade the mechanical properties of Ti and its alloys. In fact, the precipitation of such compounds leads to an embrittlement of the metallic substrate and thus to its more pronounced cracking ability [37–43]. Such cracks occurring near the surface of the metal substrate could then facilitate the inward migration of corrosive gaseous species, contributing to the enhancement of the overall corrosion process.

The titanium chloride can then react with the oxygen and lead to the formation of titanium dioxide and gaseous chlorine, which could also migrate to the metal/oxide interface and keep working the mechanism of active corrosion (Eq. (7)), until the HCl (and by extension the NaCl and H₂O) tank is consumed.

All the steps of the mechanism of active corrosion in NaCl/H₂O environment are summarised in Fig. 24.

In presence of moisture, Deacon equilibrium (Eq. (12)) will also contribute to the mechanism.

\[ 2\text{H}_2\text{O}(g) + 2\text{Cl}_2(g) \rightarrow 4\text{HCl}(g) + \text{O}_2(g) \] (12)
Nevertheless, positive value of standard free enthalpy for Deacon equilibrium provided by thermodynamic predictions in presence of $\text{Cl}_2(g), \text{H}_2\text{O}(g), \text{Cl}_2(g)$ and $\text{HCl}(g)$ at $560 \, ^\circ\text{C}$ indicates that this reaction is favoured in the way of $\text{Cl}_2(g)$ formation. Thus, the major part of $\text{HCl}(g)$ can be converted in $\text{Cl}_2(g)$, which is known to be more reactive than $\text{HCl}(g)$ in the corrosion mechanisms induced by chlorine [32].

The enhancement of the corrosion process is the major evidence of the synergistic effect of NaCl and water vapour on titanium alloys corrosion [23,44]. It can be explained by the conversion of $\text{HCl}(g)$ in $\text{Cl}_2(g)$, via the Deacon equilibrium, which lead to the release of a higher quantity of $\text{Cl}_2(g)$ than in the previous conditions without moisture. This higher quantity of $\text{Cl}_2(g)$ can then promote the reaction 6 (Eq. (6)) and enhance the entire mechanism of active corrosion.

Furthermore, the presence of $\text{H}_2\text{O}$ must be also discussed. Previous works [27,45] showed that the $\text{H}_2\text{O}$ dissociated mainly at the defects on $\text{TiO}_2(\alpha)(110)$ planes into free H atoms on OH* groups. Their dissolution inside the $\text{TiO}_2$ layer might change the defect chemistry and therefore the transport mechanisms. These considerations might also be possible explanation for the synergistic effect of NaCl and $\text{H}_2\text{O}$ vapour. A third hypothesis that can be discussed is the presence of an amount of gaseous NaOH, according to thermodynamic predictions (Fig. 23). Knowing that metal hydroxides are precursors to metal chlorides [46,47], the potential impact of this gaseous NaOH on the synergistic effect could be another explanation of the corrosion enhancement phenomenon.

**Conclusion**

Enhancement of corrosion phenomenon was shown for Ti-6Al-4V alloy with NaCl deposit at high temperature. A synergistic effect was also noticed in presence of water vapour, increasing by a factor of 15 the weight gain per unit area with respect to samples oxidized in dry or moist air at the same temperature. This enhancement is attributed to the presence of gaseous chlorine coming from the NaCl decomposition, leading to the establishment of an active corrosion phenomenon at the metal/oxide interface. The synergistic effect, observed under moist air could be connected to the formation of gaseous HCl or/and NaOH. This large increase in corrosion rate in the presence of NaCl and $\text{H}_2\text{O}$ is of course of first practical importance since planes and helicopters are used in sea areas. Nevertheless, it should be noticed that the enhanced corrosion mechanism is accompanied by a lack of oxygen enriched area under the corrosion product. Therefore, the consequence of these two phenomena, i.e. oxygen embrittlement under pure oxidation conditions and fast active corrosion under $\text{O}_2$–NaCl–$\text{H}_2\text{O}$ conditions, over the mechanical properties of the structural alloys should be evaluated.

**Acknowledgements**

This work was financially supported by the ICB. The authors would also like to thank Michel VILLASI and Stéphane MATHIEU for fruitful discussions and Frédéric HERBST, Nicolas GEOFFROY and Nathalie ROUERGUES for technical support.

**References**


