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# The Influence of Temperature and Dissolved Oxygen on the Electrochemical Properties of Three Al-Zn-In Series Sacrificial Anodes

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This paper investigates the corrosion behavior of three commercial anodes under different temperature and dissolved oxygen through a scanning electron microscope and electrochemical test. Our results show uniform corrosion in both environments, and the current efficiency at low temperature and low oxygen are higher than that of saturated oxygen at room temperature. Microtopography reveals that intergranular corrosion is more severe under the condition of low temperature and low oxygen. The polarization curves and electrochemical impedance spectrum tests at the initial stage of erosion illustrate that the activation dissolution resistance is more significant at low temperature and low oxygen. There are varying degrees of passivation tendency, especially Al-Zn-In-Si.

Keywords: Al alloy, temperature, dissolved oxygen, electrochemical property

# **1. INTRODUCTION**

With the depth exploration of offshore and development exceeding 200 meters, Al anodes have gradually become the preferred material for the cathodic protection of marine structures due to its low specific gravity, low cost and high efficiency [1-8]. It is well established that the spontaneous formation of oxide on Al exerts a strong passivation effect; therefore, the activated elements such as Zn, In, Sn, Ga, etc. are added to inhibit the oxidation film on the surface of aluminum anodes [9-11]. Furthermore, Si, Mg, Ti, Mn and other alloying elements are also added to improve corrosion uniformity and current efficiency [12-19].

The deep-sea environment is mainly characterized by low temperature, low oxygen and high pressure, which can significantly reduce the efficiency of sacrificial anodes. Specifically, low temperature retards the oxygen diffusion, lower the solution conductivity and decreases the activity of sacrificial anode [20-21]. Temperature can not only directly affect the corrosion behavior of sacrificial anodes, but it also acts corrosion factors for sacrificial anodes. When the temperature is reduced, it leads to reduce the oxygen diffusion, decrease the conductivity of the water, and compromised the and the activity of sacrificial anodes. Compared with surface seawater, the content of dissolved oxygen in deep-sea water declines sharply. Oxygen content has two main effects on the Al-Zn-In base sacrificial anodes: on the one hand, the production speed of Al<sub>2</sub>O<sub>3</sub> oxide film decreases with the decrease of dissolved oxygen content, which is conducive to the active anodic dissolution; on the other hand, it also leads to the difficulty of "dissolution-redeposition" of In, Zn and other alloying elements, resulting in the decrease of dissolution ability of sacrificial anode [22].

The changes in environmental factors have raised higher requirements for the performance of sacrificial anode. A large number of studies have shown that the performance of conventional sacrificial anodes are compromised in the deep ocean environment, making it challenging to meet the requirements of cathodic protection for deep-sea structures [23-27]. Internationally, the sacrificial anode materials suitable for deep-sea have been developed, and corresponding standards are meet [28-29]. China started work on sacrificial anode materials relatively late, and relevant research work is still in progress [8, 30-34].

The NACE report in 1992 has shown that the static pressure of seawater does not affect the protection performance of the sacrificial anode [35]. Reding and Boyce have tested the performance of Al-Zn-In anode in the simulated seawater at 2Mpa after 165 days and found that the pressure did not affect the anode performance [36]. It has also been reported in the literature [23, 37] that the change of pressure had no significant impact on the electrochemical properties of the anode. Elbeil et al. [38] has reported that the hydrostatic pressure of seawater hardly affected the initial density at 100kPa and 50kPa. Therefore, we only consider the effects of low temperature and low dissolved oxygen on anode properties in the current paper.

At present, most of the documented literature focuses on the application of Al-Zn-In series anodes in a shallow sea environment, and only a few deep-sea studies have been reported. Interestingly, there are few parallel studies on the shallow sea and the deep sea. So in this work, the electrochemical properties of three anodes are compared at low temperature (4°C) with low oxygen (2.0mg/L) and room temperature (about 20°C) with saturated oxygen (7.3mg/L, static solution). This study aims to enrich the application of cathodic protection systems from the shallow sea to the deep sea and visualize the changes in electrochemical performance when temperature and dissolved oxygen go down.

# 2. EXPERIMENTAL

#### 2.1. Materials and solutions

The sacrificial anodes used in experiments included three kinds of commercial cast anodes, Al-Zn-In-Si, Al-Zn-In-Mg-Ti and Al-Zn-In-Sn-Mg, and were labeled as S, T and M respectively. They were  $\Phi$  20 mm diameter rods and prepared following the Chinese Standard[8]. Their chemical compositions are shown in Table 1.

Anode samples for the current efficiency test with size  $\Phi$  16 mm × 28 mm were machined with a  $\Phi$  3mm × 5mm threaded hole on one end of the lead connected to the wire. Before the experiment, the samples were placed in anhydrous ethanol for ultrasonic cleaning to remove the surface grease and the processed debris in the threaded holes, and then dried in air and weighed by electronic balance. Both ends were coated with silica gel, leaving a working area of 14cm<sup>2</sup> on the surface.

The samples for polarization curves and electrochemical impedance spectra (EIS) measurements were processed into the size  $\phi$  11.3 mm × 15 mm. They had a  $\phi$  1.5 mm perforation at one end for the copper wires, then, sealed with epoxy resin and reserved working area of 1 cm<sup>2</sup>. After grounded with emery paper continuously to grade 2000, they were rinsed with distilled water and anhydrous ethanol.

The middle part of the casting rods was cut transversely and processed into  $\Phi$ 11.3mm × 10mm, then sealed with epoxy resin. After being ground with emery paper sequentially to grade 3000, the samples were polished with 1.5 µm diamond paste and fixed with 0.5% HF solution for 15 seconds. Subsequently, the anode microstructure was observed by scanning electron microscope (SEM-FEI, Quanta 250) and analyzed by energy dispersive X-ray detector (EDX).

The experimental conditions were low temperature (4°C) with low oxygen (2.0mg/L) denoted as condition 1, and at room temperature (about 20°C) with saturated oxygen (7.3mg/L, static solution), denoted as condition 2. Moreover, when the alloy S is in condition 1, it is denoted as S1. Similarly, the rest are labeled as S2, T1, T2, M1 and M2, respectively. The corrosion medium for all tests was simulated seawater with the chemical composition of NaCl 24.53 g/L, MgCl<sub>2</sub> 5.20 g/L, Na<sub>2</sub>SO<sub>4</sub> 4.09 g/L, CaCl<sub>2</sub> 1.16 g/L, KCl 0.695 g/L, NaHCO<sub>3</sub> 0.201 g/L, KBr 0.101 g/L, H<sub>3</sub>BO<sub>3</sub> 0.027 g/L, SrCl<sub>2</sub> 0.025 g/L, and NaF 0.003 g/L.

No.	Zn	In	Sn	Mg	Si	Ti	Others,≤			Δ1
							Si	Fe	Cu	Al
S	5.5-7.0	0.025-0.035			0.10-0.15		0.10	0.15	0.01	Bal.
Т	4.0-7.0	0.020-0.050		0.5-1.5		0.01-0.08	0.10	0.15	0.01	Bal.
М	2.5-4.0	0.020-0.050	0.025-0.075	0.5-1.0			0.10	0.15	0.01	Bal.

Table 1. Chemical composition of the three kinds of commercial anodes (wt.%)

## 2.2. Current efficiency measurements

The current efficiency of the three samples was performed using a three-electrode system. Sample, stainless steel and a saturated calomel electrode (SCE) were applied as the anode, auxiliary cathode and the reference electrode, respectively. The auxiliary cathode was  $\Phi$  110 mm × 130 mm stainless steel tube, which can make the ratio of 1:60 the anode to cathode surface area, using electronic voltmeter acquisition through the sacrificial anode electricity. The working potential of samples was recorded for one or more times daily. The anodic dissolution and adhesion of the corrosion products were observed at the end of the test. Afterwards, the corrosion products were removed with concentrated nitric acid. Then, the actual capacity and current efficiency were calculated according to the following formulas.

Current efficiency,  $\eta = Q / Q_0$ 

Actual current capacity, Q = C / W

where  $\eta$  is the anode current efficiency, Q the actual current capacity,  $Q_0$  is the current theoretical capacity, C is the total quantity of charge passed through the electrochemical circuit, and W is the total weight loss. At least three parallel experiments were conducted for each alloy to ensure the reproducibility of the data.

The macro photos before and after the rust removal were captured with a digital camera, and the micromorphology of the corrosion surface under various experimental conditions was observed with SEM.

#### 2.3. Electrochemical measurements

The electrochemical tests were performed with three electrodes systems by the PARSTAT-3F electrochemical test system. A platinum sheet with large-area was used as a counter electrode, SCE served as a reference electrode and the sample as a working electrode. The samples were placed in the test solution, and the time from the end of polishing to the placement in the solution was 10s for reducing the difference of the surface state of each sample. When the open circuit potential (OCP) became stable after 4000 s, the electrochemical impedance spectroscopy (EIS) measurements were calculated with a frequency domain from 100 kHz to 0.01 Hz accompanied by an excitation voltage of 10 mV. We used ZsimpWin software to fit the equivalent circuit diagram of the impedance spectrum. Then, the polarization curves were tested by sweeping the potential at a scan rate of 0.5 mV/s. The scan scope was set from -0.25 to 0.4 V(vs OCP). The electrochemical experiments were repeated several times to ensure the reproducibility.

# **3. RESULTS AND DISCUSSION**

# 3.1. Microstructure

Fig. 1, Fig. 2 and Fig. 3 show the microstructure and element composition of Al-Zn-In-Si, Al-Zn-In-Mg-Ti and Al-Zn-In-Sn-Mg, respectively. The microstructure of the three anodes contains  $\alpha$ -Al

matrix (the grey regions), precipitates at grain boundaries (the bright areas) and defects (the dark areas).

Fig. 1 shows the extensive erosion range of Al-Zn-In-Si grain boundary, visible intermittent chains and spherical phases in crystal boundary, which are rich in the Zn-In precipitated phase analyzed by EDX analysis. A wide segregation band of activated elements on grain boundary is beneficial to restrain corrosion expansion in depth. The size of grains ranging from 50 to 200µm with an irregular shape, and dendrite traces can be observed. As can be seen from the EDX, Si has not been detected in the grains due to its low content. The activated elements, including Zn and In are mainly concentrated at the grain boundary, indicating that corrosion will first occur at the grain boundary. This is consistent with the conclusions of previous literature [39-42].It is inferred from the above discussion that wider grain boundaries and irregular grains lead to uniform corrosion with greater pits formed by the shedding of larger grains.



Figure 1. SEM photographs and EDX images of Al-Zn-In-Si.

Fig. 2 presents that the grain size of Al-Zn-In-Mg-Ti, which is mostly within the range of 100-200 μm with a few uneven distributed huge grains. There are many pits in grain boundaries and grains, which are mainly casting defects formed during solidification and a small number of corrosion pits [39, 43-45]. As shown in EDX analysis, the content of each added element at the grain boundary is slightly higher than inside the grain; that is, the distribution of activated elements is more uniform. Besides, after much observation, we found that some small particles in a few holes and on the corroded surface are detected as rich in precipitates. This agrees with previous conclusions [33]. It is derived from the

above analyses that the uniform distribution of activation elements, casting defects and erosion pits with attachments can generate uniform corrosion accompanied by a lot of small pits.

Fig. 3 indicates that the grain size of Al-Zn-In-Sn-Mg is around 100 µm, and the overall structure is relatively uniform. Besides, the anode displays a small fraction traces of dendrite microstructure [1]. Like Al-Zn-In-Mg-Ti, there are also many pits on the eroded surface. According to EDX analysis, the content of each element at the grain boundary is higher than inside the grain. Moreover, the activated elements Zn, In and Sn accumulate in large quantities at the grain boundary. However, the content of In and Sn inside the grain is negligible to be detected. This is basically consistent with previous findings [46]; except that the Sn content of the sample used in this experiment is not detected in the grain because it is less than 0.1% (wt.). Moreover, after a large number of tests with EDX in the pits (Fig. 2 and Fig. 3), we found that the content of activated elements decreases significantly, which can inhibit the development of the pits in-depth, and conducive to the horizontal corrosion expansion. It is deduced from the above statements that the same size grains, aggregation of activated elements at grain boundaries and many pits containing less active elements conduce to uniform corrosion with minor surface undulation.



Figure 2. SEM photographs and EDX images of Al-Zn-In-Mg-Ti.



Figure 3. SEM photograph and EDX analysis of Al-Zn-In-Sn-Mg.

# 3.2. Current efficiency test

Fig. 4 shows the macroscopic corrosion morphology of the three commodity anodes in both environments. The a and b are the corrosion morphologies before and after the removal of products. We found that the three samples have almost no product coverage in both environments. The corrosion surface of Al-Zn-In-Si at low temperature and low dissolved oxygen displays little fluctuation. However, in saturated oxygen at room temperature, the surface undulation increases, and the shape of many corrosion pits changes from circular to elongated. This phenomenon may be due to the interconnection of some small corrosion pits, and the activated elements are more likely to deposit at both ends of the connected corrosion pits during redeposition. Al-Zn-In-Mg-Ti anode has a small number of bright spots of different sizes on the corrosion surface in the low-temperature (4°C) and low-dissolved oxygen (2.0mg/L) environment. The bright areas cover very little on the whole, and a large number of small corrosion pits are distributed on the surface. When the condition becomes saturated oxygen at room temperature, there are still a few bright spots, and the number of small corrosion pits is significantly reduced, with minor surface fluctuation. The corrosion morphology of Al-Zn-In-Sn-Mg in the two environments is the same, which presents uniform corrosion, accompanied by a few bright spots and little fluctuation at room temperature under saturated oxygen. The above description reveals that the corrosion morphologies of the three anodes are consistent with the microstructure analyses. In addition, the previous description of the morphology was basically relaying on whether it was flat, uniform or not, and whether there were pits or not [39,44]. Here, the morphology was described in more detail, and the possible reason for formation was analyzed, which is very rare. However, the corrosion described above is generally uniform, which is consistent with the previous research [47-54].



**Figure 4.** Constant potential corrosion morphology of Al-Zn-In-Sn-Mg in both cases: a, before removing corrosion products; b, after removal of corrosion products.

Table 2 lists the open-circuit potentials (OCP), closed-circuit potentials (CCP), actual capacitance and current efficiency of three commodity anodes in two different environments. The current efficiencies at low temperature and low oxygen are higher than that at room temperature and saturated oxygen.

No.	OCP (V vs SCE)	CCP (V vs SCE)	Capacity $(A \cdot h \cdot kg^{-1})$	Efficiency (%)
S1	-1.120	-1.083~-1.100	2481.2	87.24
S2	-1.124	-1.096~-1.104	2452.4	86.23
T1	-1.134	$-1.076 \sim -1.094$	2558.8	88.22
T2	-1.138	$-1.082 \sim -1.087$	2344.2	84.43
M1	-1.132	-1.093~-1.111	2581.1	90.54
M2	-1.141	-1.110~-1.104	2452.4	86.03

 Table 2.
 Electrochemical parameters of three commodity anodes.



Figure 5. The microcorrosion morphology of three anodes in two simulated seawater environments.

Among them, the current efficiency of Al-Zn-In-Sn-Mg is the highest at low temperature and low oxygen, reaching more than 90%, followed by Al-Zn-In-Mg-Ti and Al-Zn-In-Si. However, the value difference of current efficiency, only one percentage point is the smallest when the Al-Zn-In-Si alloy was placed in the two different environments. Meanwhile, the other two have a difference of four percentage points; however, the overall difference is not significant. Simultaneously, as shown in Fig. 4, there is a slight difference in corrosion morphology between the two different environments. However, it does not violate the principle that the current efficiency is consistent with the uniformity of the surface corrosion morphology. Overall, they are uniformly corroded [33,43,54].

Fig. 5 exhibits the microscopic morphology after the efficiency tests. It is well established that the corrosion degree of grains boundary is positively correlated with temperature. However, intergranular corrosion of Al-Zn-In-Si and Al-Zn-In-Sn-Mg is significantly more severe at low temperature and low oxygen. The possible reason for this phenomenon is that, in the room temperature and saturated oxygen environment, the activation reaction is more intense, resulting in shedding of grains. Moreover, the surface grains still fall off during the process of rust removal. Therefore, the intergranular corrosion on the surface appears to be relatively lighter after a large number of grains are exfoliated due to severe intergranular corrosion. Here, we make a different interpretation from the previous literature [33]. The micromorphology of Al-Zn-In-Mg-Ti differs slightly in the two different environments, which may be due to the difference in the content of activated elements in the grain boundary and interior of grain. At this point, the influence of the environment on activation dissolution is weakened.

### 3.3. Potentiodynamic polarization curves measurement

The polarization curves of the three anodes in the two different environments are shown in Fig.6, and the corrosion parameters obtained from the polarization curves are listed in Table 3. The corrosion potential of the three anodes is lower at low temperature and low oxygen than that at room temperature and saturated oxygen. The possible reason is that the formation rate of surface oxide film at the initial stage is slow, and the area of the contact surface between the substrate and the solution is large under low temperature and low oxygen. However, low temperature also inhibits the speed of activation reaction, leading to a small corrosion current. With the increase of  $E_{corr}$  value, the corrosion current gradually rises. The slope of the anode curve is relatively large, indicating that the anode has a relatively excellent passivation tendency, especially the clear passivation of Al-Zn-In-Si. Nonetheless, when the  $E_{corr}$  value of Al-Zn-In-Si exceeds  $E_{pit}$  value(-1.004V), the current starts to increase linearly, and hydrogen evolution is observed. These indicate that the oxide film ruptures, forming stable pitting corrosion and corrosive pitting begins to grow. The above description manifests that low temperature and low oxygen environment have a more significant influence on the initial activation of Al-Zn-In-Si alloy.

Under the condition of saturated oxygen at room temperature, due to the influence of temperature and saturated oxygen, the oxidation film is continuously generated. Then, the contact surface between the matrix and solution becomes smaller, resulting in a decrease of activation points.

As a result, the corrosion potential of the three anodes is significantly increased compared with under low temperature and low oxygen condition.



**Figure 6.** Polarization curves of three commodities after 4000s in simulated seawater, 1: 4°C, 2.0mg/L; 2: 20°C, 7.3mg/L. (a: Al-Zn-In-Si, b: Al-Zn-In-Mg-Ti, c: Al-Zn-In-Sn-Mg)

Table 3. Corrosion parameters of the samples in simulated seawater solution.

Samplas	$E_{corr}$	Icorr
Samples	(V vs. SCE)	$(\mu A \cdot cm^{-2})$
S1	-1.090	0.113
<b>S</b> 2	-1.051	0.884
T1	-1.203	0.221
T2	-1.143	0.275
M1	-1.233	0.130
M2	-1.123	0.388

At this condition, a large number of micro electrolytic cells are formed between the active points on the surface and the oxide film, and the increase of the oxide film results in the small anode and a larger cathode. Therefore, the activation reaction is more intense than that under low temperature and low oxygen, and the corrosion current ( $I_{corr}$ ) is more massive. From the above statements, it is clear that the activation dissolution of the anode is more difficult under low temperature and low dissolved oxygen than that under saturated oxygen at room temperature. Meanwhile, at room temperature and saturated oxygen condition, the decrease in anode curve slope and the increase of the corrosion current density indicate that the self-corrosion is gradually increasing, which reduce the working efficiency of the anode. The above conclusions under the two conditions are is similar to the results of Wang et al. 's study at different temperatures [55], indicating that temperature has a great effect on  $E_{corr}$ .

### 3.4. EIS measurements

The electrochemical impedance spectrum (EIS) plots of the three commodity anodes after immersion for 4000s in simulated seawater solution with different environmental conditions are shown in Fig. 7. Under low temperature and low oxygen conditions, the Nyquist diagrams of the three kinds of anodes are characterized by both high and low frequencies capacitive arcs, which is basically consistent with Qu et al. 's conclusion at low temperatures [56]. The capacitive arcs at high frequency correspond to the initial surface of the sample, while the capacitive arcs at low frequency correspond to the newly formed pit surface. Meanwhile, the low-frequency capacitive arc of Al-Zn-In-Si has the largest capacitive arc radius. The possible reason is that there are two processes of oxidation and activation on the anode surface at the initial stage, and the formation process of the initial oxidation film is dominant, leading to considerable resistance to activation reaction. The number of activation points is small, and the surface still reveals a passivation state, which is consistent with the polarization curve. The reason for the smaller capacitive arcs of Al-Zn-In-Sn-Mg at low frequency is the dissolution-deposition reaction of the active elements in some stable pitting pits.

Under the condition of saturated oxygen at room temperature, the EIS diagrams of the three anodes show high-frequency capacitive loops and low-frequency inductive loops. Compared with the conditions of low temperature and low dissolved oxygen, except Al-Zn-In-Mg-Ti, which has similar capacitive loops at high frequency in the two environments, the capacitive loops of the other two anodes are significantly reduced, indicating that the polarization resistance is decreased. The polarization impedance is mainly controlled by charge transfer resistance and diffusion resistance. The increased temperature accelerates the rate of charge transfer and diffusion, promoting the activation reaction and leading to a significant reduction in polarization resistance. Moreover, inductive retraction occurs at low frequency, suggesting that the anode surface has pitting corrosion due to Cl<sup>-</sup> adsorption and the dissolution of activated elements. Meanwhile, the appearance of a low-frequency inductive arc also illustrates that the electrode process is mainly determined by the electrode potential and another surface state variable that depends on the electrode potential as described by Li et al. [57]. It has been proposed in the previous literatures that the inductive arc is the occurrence mark of pitting corrosion in the pitting model of Al alloy [58-59].



**Figure 7.** Nyquist and Bode plots of Al-Zn-In-Si (a-b), Al-Zn-In-Mg-Ti (c-d) and Al-Zn-In-Sn-Mg (e-f) with different environments, 1: 4°C, 2.0mg/L; 2: 20°C, 7.3mg/L.



Figure 8. Electrical equivalent circuits used for fitting the EIS results

To further analyze the anodized electrode process, the electrical equivalent circuits (EECs) were adopted to fit the EIS data, as shown in Fig. 8. The EECs in Fig. 8a and Fig. 8b used to fit the EIS results under condition 1 and condition 2. Where  $R_s$  is the resistance of the simulated seawater solution. *CPE*<sub>1</sub> represents the electric double-layer capacitance at the interface of the working electrode and the electrolyte.  $R_t$  is charge transfer resistance controlled by the interfacial electrochemical reaction kinetics. *CPE*<sub>2</sub> stands for the capacitance between new surface and solution in pitting pits, and  $R_2$  is the charge transfer resistance in pitting pits. The series combination of  $R_L$  and L characterize the inductive loop at low-frequency band and are the inductance parameter of pitting pits.

The impedance of the constant phase component *CPE* is expressed as:

 $Z_{CPE}=1/((j\omega)^n Y_0)$ 

Where  $\omega$  is the angular frequence,  $Y_0$  and n are the two parameters of the equivalent element. *CPE* is used to describe the physical quantity when the parameter of capacitance deviates, so  $Y_0$  always takes a positive value, like a capacitor. n is the dimensionless exponent.

For a capacitive element, *n* is a constant evolved from it due to its heterogeneous effects. *n*=1 corresponds to a pure capacitor, *n*=0.5 Warburg impedance, and *n*=0 a resistance. The fitting parameters of the impedance spectrum are listed in Tables 4 and 5. As shown in Table 4, under low temperature and low oxygen environment,  $n_1$  values are all-around 0.9, indicating that the deviation to ideal capacitance is small. The difference of  $n_2$  value is massive, which is due to the diffusion effect of Al-Zn-In-Si at the low-frequency range to a certain extent, and also denotes that the pitting degree is relatively light. The  $n_2$  value of Al-Zn-In-Mg-Ti alloy is equal to 1, and the *CPE*<sub>2</sub> value is the largest, implying that compared with the other two kinds of alloys, the activation reaction in pitting pits of Al-Zn-In-Mg-Ti alloy is the most intense. The  $R_L$  and L values of Al-Zn-In-Mg-Ti alloy are also the highest under saturated oxygen at room temperature, followed by Al-Zn-In-Sn-Mg alloy, which is consistent with the surface corrosion after the constant current test. The  $\chi^2$  depicts a small error in the fitting data, as shown in Tables 4 and 5, representing a good agreement between the fitted data and the experimental results.

No.	<b>S</b> 1	<b>T</b> 1	M1
$R_s/(\Omega \cdot cm^2)$	9.178	10.71	11.63
$CPE_1 - Y_0 / (\Omega^{-1} \cdot cm^{-2} \cdot s^{-1})$	$2.206 \times 10^{-5}$	$1.462 \times 10^{-5}$	$1.710 \times 10^{-5}$
$n_1$	0.8955	0.9066	0.8728
$R_t/(\Omega \cdot cm^2)$	2535	3455	3904
$CPE_2 - Y_0 / (\Omega^{-1} \cdot cm^{-2} \cdot s^{-1})$	$1.733 \times 10^{-5}$	$9.023 \times 10^{-6}$	$4.755 \times 10^{-6}$
<b>n</b> <sub>2</sub>	0.6450	1.000	0.8656
$R_2/(\Omega \cdot cm^2)$	41840	3542	9972
$\chi^2$	6.643×10 <sup>-4</sup>	3.033×10 <sup>-3</sup>	$1.995 \times 10^{-3}$

**Table 4.** EIS fitting parameters of alloys at low temperatures and low dissolved oxygen after 4000s of immersion

No.	S2	T2	M2
$R_s/(\Omega \cdot cm^2)$	11.77	7.072	6.704
CPE <sub>1</sub> -	$4.487 \times 10^{-5}$	$1.816 \times 10^{-5}$	$1.812 \times 10^{-5}$
$Y_0/(\Omega^{-1} \cdot cm^{-2} \cdot s^{-1})$			
$n_1$	0.8347	0.8834	0.8937
$R_t/(\Omega \cdot cm^2)$	6413	6904	7233
$L/(H \cdot cm^2)$	829.3	85730	6175
$R_L/(\Omega \cdot cm^2)$	7284	70200	12900
$\chi^2$	$1.317 \times 10^{-3}$	$1.398 \times 10^{-3}$	$5.133 \times 10^{-4}$

**Table 5.** EIS fitting parameters of alloys at room temperature and saturated oxygen after 4000s of immersion

## **4. CONCLUSIONS**

The three different commercial anodes exhibit uniform corrosion behaviors in both environments, and their efficiencies at low temperature and low oxygen are higher than that at saturated oxygen and room temperature.

After constant current corrosion, the macromorphological characteristics of anodes is consistent with the microstructure analyses. Micromorphology reveals that intergranular corrosion is more intense at low temperature and low oxygen. The reason is that at room temperature and saturated oxygen, the intergranular corrosion appears lighter due to the grain shedding in the severe intergranular corrosion area.

The polarization curves and EIS measurements at the initial stage of corrosion illustrate that the activation dissolution resistance is more significant at low temperature and low oxygen than at room temperature and saturated oxygen. Under low temperature and low oxygen, Al-Zn-In-Si has the highest degree of passivation.

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