Corrosion studies on Fe-based amorphous alloys in simulated PEM fuel cell environment

J. Jayaraj, Y.C. Kim, K.B. Kim, H.K. Seok, E. Fleury*

Division of Materials Science and Engineering, Korea Institute of Science and Technology (KIST), P.O. Box 131, Cheongryang, Seoul 130-650, South Korea

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Abstract

In this paper, we introduce a novel idea of using metallic amorphous material as a bipolar plate for polymer electrolyte membrane fuel cell (PEMFC). The major requirement of the metallic bipolar plate material is their high corrosion resistance. Owing to its chemical homogeneity, and absence of defects (like grain boundaries, dislocations, etc.), amorphous materials exhibit a combination of attractive properties for bipolar plates such as high corrosion resistance and high strength (about 2 GPa). The corrosion properties of Fe_{48}Cr_{15}Mo_{14}Y_{2}C_{15}B_{6} and of a newly developed Fe_{50}Cr_{18}Mo_{8}Al_{2}Y_{2}C_{14}B_{6} bulk metallic amorphous alloys have been investigated under conditions that simulate the fuel cell environment and are compared with those of a stainless steel. Hydrogen gas and pressurized air were bubbled into a 1 M H_{2}SO_{4} solution at 75°C solution, throughout the experiment to simulate the respective anodic and cathodic PEMFC environment. In comparison to the stainless steel, the Fe_{50}Cr_{18}Mo_{8}Al_{2}Y_{2}C_{14}B_{6} amorphous alloy displays significantly higher corrosion resistance.

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1. Introduction

Global warming, pollution, and depletion of oil encourage the development of fuel cell technology as an alternative source of energy. Among the diverse types of fuel cells, polymer electrolyte membrane fuel cells (PEMFCs) are compact, lightweight and can operate at low temperature. In PEMFC, the membrane electrode assembly (MEA) is composed by the anode and cathode separated by the polymer membrane. To ensure a sufficient power, a large number of MEA are stacked, separated by multifunctional bipolar plates. The working principle of the PEMFC is explained elsewhere [1]. In PEMFC, these plates are exposed to severe working conditions combining current, corrosion and stress (shock, vibration, etc.). Thus, materials for bipolar plates are required to possess high strength, good corrosion resistance, and electrical conductivity, in addition to being impermeable to gases. Current bipolar plates used in demonstration are made of carbon, however, due to their low toughness, bulky graphite bipolar plates are unavoidable. For commercialization, cheap and thin metallic bipolar plates should be developed. Intensive researches are thus carried on to develop alternative materials for bipolar plates like graphite composite [2], stainless steel metallic bipolar plates [3], surface modification by thermal nitriding [4], and TiN coating of stainless steel [5], thin PVD coatings of YZU001 like diamond film deposited on either aluminum or stainless steel bipolar plates [6], or titanium coatings on bipolar plates [7]. But it is also known that the loss of electrons of metallic bipolar plates results in dissolution of metallic ions which may poison the MEA. Moreover in the cathode side, the growth of metallic oxide might eventually occurs due to recovery reaction by gain of these electrons, which would lead to the increase of the surface electrical resistance, and results in a reduction of the cell performance.

Recent breakthroughs in materials engineering have generated complex Zr, Ti, Cu, Fe, Ni, La, Pd, and Mg-based alloys that retain a glassy state in bulk form (bulk metallic
glasses or BMGs) via conventional ingot casting. When the conventional alloys are cooled from the liquid melt, they rapidly crystallize just under the melting point. But rather than forming a single crystal, most metals are polycrystalline, with grains of varying shapes and size. In contrast, a glass has such slow crystal nucleation and growth kinetics that the liquid can be under-cooled far below its melting point. Because of the high viscosity of the under-cooled liquid, the atoms cannot rearrange to form crystal, but rather retain a random arrangement to form an amorphous phase (defined as a random atom packing with no long-range order). After being essentially a scientific curiosity, amorphous alloys found applications due to there combination of different properties such as high strength, high hardness, superplasticity in the super cooled liquid region, soft magnetism, high wear resistance, and high corrosion resistance [8].

To the authors’ knowledge, there is no report on the application of metallic amorphous material for bipolar plates for PEMFC. Bulk Fe-based glassy alloys with a thickness of a few millimeter range were recently produced in the Fe–(Al, Ga)–(P, C, B), Fe–(Cr, Mo, Co)–Ga–(P, C, B), Fe–Si–B–(Nb–Zr), and (Fe, Co, Ni)–Zr–B alloy systems [9–12]. The addition of a small amount of Y was found to promote the glass forming ability from millimeter scale to about centimeter scale fully amorphous rods [13,14]. Investigations on Fe-based amorphous alloy containing Cr and other transition metals have also demonstrated higher corrosion resistance than their crystalline counterparts [15,16]. Moreover, the super plasticity behavior of the bulk metallic glasses in the super-cooled liquid region [8,17] opens the window to process the required flow channels in the bipolar plate by metal processing technique like stamping. In general, the electrical resistivity of the amorphous alloy is higher than that of its crystalline state because of concentration of conduction electrons in amorphous state. However, the Fe-based amorphous alloy exhibits an electrical resistivity of the order of 100–130 \( \mu \Omega \) cm at room temperature [18] which is just about values for the stainless steel. Considering all the requirements for bipolar plate and cost effectiveness, Fe-based amorphous alloy could be a candidate material for bipolar plate in PEMFC. However, so far corrosion of amorphous alloy has been investigated only in \( \text{H}_2\text{SO}_4 \) or HCl environment only. Thus this paper intends to understand the corrosion behavior of a Fe-based amorphous alloy, developed by our group, in the simulated PEMFC environment and to compare with the Fe_{48}Cr_{18}Mo_{14}Y_{2}C_{15}B_{6} amorphous alloy, one of the highest glass forming alloy reported till date [13] and the commercially available stainless steel (SS316L) alloy.

2. Experimental procedure

Multicomponent alloys with the composition of Fe\(_{48}\)Cr\(_{18}\)Mo\(_{14}\)Y\(_{2}\)C\(_{15}\)B\(_{6}\) and Fe\(_{50}\)Cr\(_{18}\)Mo\(_{6}\)Al\(_{2}\)Y\(_{2}\)C\(_{14}\)B\(_{6}\) (at.%) were prepared by arc melting of high purity metal, Fe (99.9%), Cr (99.9%), Mo (99.5%), Al (99.9%), Y (99.5%), C (99.9%), and B (99.9%), under an argon atmosphere. Plates of dimension 1.2×5×50 mm\(^2\) were prepared by copper mold casting. Micro structural characterization of the cast rod and plate was performed by X-ray diffraction (XRD) with monochromatic Cu K\(_{\alpha}\) radiation. Differential scanning calorimetry (DSC) and differential thermal analysis (DTA) traces were obtained at a heating rate of 20°C/min to examine the thermal stability associated with the glass transition, the extent of the super-cooled liquid region, and the melting temperature of the alloy. The corrosion behavior of the bulk glassy alloy was studied by electrochemical measurements in a simulated aggressive PEMFC environment; all the experiments were conducted in 1 M \( \text{H}_2\text{SO}_4 \)+2 ppm F\(^-\) at 75°C [19,20]. The hydrogen gas and the pressurized air were bubbled to the solution, throughout the experiments to simulate the respective anodic and cathodic PEMFC environments. Before the corrosion tests, the amorphous plates and SS316L were polished until 1200 grit SiC paper, ultrasonically cleaned in acetone, washed in distilled water, and dried in air to promote reproducible results. The electrochemical studies were studied by a conventional three electrode systems. The amorphous or stainless steel plate, Ag/AgCl and platinum, were used as working electrode, reference and auxiliary electrodes, respectively. The open circuit potential (OCP) was monitored for 5 min then the potentiodynamic polarization tests were performed at a scanning rate of 50 mV/min from −0.5 to 1.2 V. Potentiostatic experiments for about 5 h in PEMFC’s operating condition were performed for the newly developed amorphous alloy. For the PEMFC anodic environment the solution was bubbled with hydrogen under an operating potential in application of −0.1 V (equivalent to −0.1 V\(_{\text{SHE}}\)). For the PEMFC cathodic environment the solution was bubbled with pressurized air with an applied potential of 0.6 V (equivalent to −0.8 V\(_{\text{SHE}}\)) [20]. After the potentiostatic studies, the solution was sampled and the concentration of the dissolved species were analyzed by inductively coupled plasma atomic absorption spectrometry (ICP-AA), which give a thorough measure of the corrosion rate. In order to understand the origin of corrosion resistance, the surface of the glassy specimens before and after potentiostatic tests were examined by X-ray photoelectron spectroscopy (XPS) using a physical electronics (PHI) 5800 spectrometer with Al K\(_{\alpha}\) excitation (\( h\nu = 1486.6 \) eV). From these spectra the composition of the passive film and the underlying alloy surface were quantitatively determined. The thickness of the passive layers was determined from the depth profile obtained from argon ion sputtering at a rate of 3.5 nm/min calibrated to SiO\(_2\) standard layers.
3. Results and discussion

3.1. Microstructural characterization

The X-ray diffraction pattern of the Fe-based amorphous alloys shown in Fig. 1 is characterized by a broad peak with the diffraction angle \(2\theta\) in the range of 40–50°, demonstrating the amorphous nature of alloy in contrast to crystalline materials, which exhibit several sharp peaks at position defined by the Bragg’s law of diffraction. Fig. 2 shows the DSC trace obtained from 1 mm diameter Fe\(_{50}\)Cr\(_{18}\)Mo\(_8\)Al\(_2\)C\(_{14}\)B\(_6\)Y\(_2\) rod during continuous heating at rate of 0.33 K/s, in which arrows indicate the glass transition temperature, \(T_g\), and crystallization temperature, \(T_x\). For this alloy composition as well as for Fe\(_{48}\)Cr\(_{15}\)Mo\(_{14}\)C\(_{15}\)B\(_6\)Y\(_2\), two exothermic peaks due to the crystallization of the glassy phase were found to follow the glass transition reaction. Table 1 indicated that variation of the Cr and Mo contents and the addition of 2 at.% Al do not affect significantly the thermal properties of the amorphous phase. However, modification of the alloy composition altered the glass forming ability, and in alloy containing 18 at.% Cr the maximum specimen diameter resulting in fully amorphous structure was only 5 mm as compared to the 9 mm diameter rod for the Fe\(_{48}\)Cr\(_{15}\)Mo\(_{14}\)C\(_{15}\)B\(_6\)Y\(_2\) alloy [13].

3.2. Potentiodynamic behavior of Fe-based amorphous alloy and stainless steel

3.2.1. PEMFC anodic environment

The polarization curves for the Fe-based amorphous materials and SS316L, in 1 M H\(_2\)SO\(_4\)+2 ppm F\(^-\) at 75 °C with hydrogen bubbling is shown in Fig. 3. The corrosion resistance of the Fe-based amorphous and crystalline alloys depends on the concentration of the major passivating elements, here chromium, as explained elsewhere for the different corrosive environments [15,16,21]. In this study, the same perspective was used to explain the passivating behavior of the alloys in the simulated PEMFC environment. The amorphous alloys with higher chromium content (18 at.%) shows passive behavior, indicative of an excellent corrosion resistance. However, a distinct passive region could not be noticed for the amorphous alloy with low chromium content (about 15 at.%) in the hydrogen bubbled environment, as shown in Fig. 3. With high chromium content, Fe-based amorphous alloys passivated spontaneously with a current density of the order of 0.75 mA/cm\(^2\) at the approximate operating potential of the PEMFC anodic environment of about −0.1 V. However, the stainless steel SS316L containing the same amount of chromium (18 at.%), is in active–passive transition region as indicated with the current density of 1.1 mA/cm\(^2\), which means the stability of the passive layer is higher for the amorphous alloy when compared to the stainless steel.

3.2.2. PEMFC cathodic environment

Similarly, the polarization curves for the Fe-based amorphous materials and SS316L in 1 M H\(_2\)SO\(_4\)+2 ppm F\(^-\) at 75 °C with air bubbled conditions are shown in Fig. 4. For this condition, a low passivating current of the order of 78.4 \(\mu\)A/cm\(^2\) was recorded for the high chromium Fe-based alloys and even a distinct passive region but with a higher passivating current density of the order of 0.91 mA/cm\(^2\) was observed for the low chromium content Fe-based amorphous alloy. In comparison to the hydrogen bubbled environment, the higher corrosion resistance of the alloys with air bubbling reflects the useful effect of the air (i.e. oxygen) in the formation of the passive layer. Moreover, the corrosion potential of the amorphous alloy becomes nobler with the value of 0.062 V during air bubbling as compared to −0.268 V for hydrogen bubbling, for the high chromium Fe-based amorphous alloy, and 0.097 V from −0.201 V for low chromium Fe-based amorphous, indicating again that the air bubbled environment is less corrosive than hydrogen.
bubbled. In contrast an almost identical corrosion potential of about $K_{0.241}$ V was observed for SS316L stainless steel both in hydrogen and in air bubbled environment. A lowest passivating current, of the order of 0.19 mA/cm$^2$ was recorded for the high chromium content Fe based amorphous (Fe$_{50}$Cr$_{18}$Mo$_8$Al$_2$Y$_2$C$_{14}$B$_6$) alloy at the PEMFC cathodic operating potential of about 0.6 V (as shown in Fig. 2) when compared to the two other alloys studied in this work indicating a better choice for PEMFC bipolar plate applications.

It should be noted that the corrosion behavior of the alloys studied in PEMFC simulated environment differs by two contexts, the concentration of chromium content for amorphous alloys and the structure of the alloy, i.e. amorphous and crystalline nature. From the results shown in Figs. 3 and 4, a minimum chromium content of about 18 at.% is required in the amorphous alloys for high performance in PEMFC environment. At the same time, with the same chromium content, the amorphous alloy shows higher corrosion resistance than the crystalline stainless steel. The introduction of chemical heterogeneity in the crystalline alloy prevents the formation of the uniform passive film and the grain boundaries act as potential sites for preferential dissolution to occur, resulting in a higher corrosion rate [22].

### 3.3. Potentiostatic polarization of Fe$_{50}$Cr$_{18}$Mo$_8$Al$_2$Y$_2$C$_{14}$B$_6$ amorphous alloy

#### 3.3.1. PEMFC anodic environment

In order to understand the higher corrosion resistance of the newly developed Fe$_{50}$Cr$_{18}$Mo$_8$Al$_2$Y$_2$C$_{14}$B$_6$ amorphous alloy, the stability of this alloy was studied by potentiostatic polarization in simulated PEMFC environment. From Fig. 3 it is noticed that the corrosion potential ($-0.268$ V) of the alloy is close to the $-0.1$ V, the operating potential at the anodic PEMFC bubbled with hydrogen. Since oxidation of hydrogen is the anodic reaction in PEMFC, at this potential the alloy may experience dissolution. When a voltage of $-0.1$ V was applied to the alloy at the anodic condition of PEMFC, as seen in Fig. 5 the current firstly dropped rapidly at the very first stage of polarization and then stabilized. The current drop from 1.1 mA/cm$^2$ at the beginning of the experiment to 2.48 μA/cm$^2$ after 39.25 min of the polarization test is related to the formation of passive layer. As the passive layer is formed on the whole surface, the current required for maintaining the layer is reduced. It should be noted that beyond the 39.25 min, the current changes from positive to negative. The reason for this change is not yet clear, but similar results were reported by Ma et al. [23] for 316L stainless in an acidic solution of pH 1 at 80 °C.

| Alloy composition | $T_g$ (°C) | $T_{x1}$ (°C) | $T_{x2}$ (°C) | $T_m$ (°C) | $\Delta H (J/g)$ | Ref. |
|-------------------|------------|--------------|--------------|------------|-----------------|-----|
| Fe$_{44}$Cr$_{16}$Mo$_{16}$C$_{18}$B$_6$ | 597 | 659 | 712 | 1141 | – | [15] |
| Fe$_{48}$Cr$_{15}$Mo$_{14}$C$_{15}$B$_6$Y$_2$ | 575 | ~612 | ~660 | ~1140 | – | [13] |
| Fe$_{50}$Cr$_{18}$Mo$_8$Al$_2$C$_{14}$B$_6$Y$_2$ | 560 | 606 | 669 | 1148 | –73 | This work |

Fig. 3. Potentiodynamic behavior of Fe$_{50}$Cr$_{18}$Mo$_8$Al$_2$Y$_2$C$_{14}$B$_6$ glassy alloy, Fe$_{48}$Cr$_{15}$Mo$_{14}$C$_{15}$B$_6$Y$_2$ glassy alloy and stainless steel-SS316L in 1 M H$_2$SO$_4$ + 2 ppm F/75 °C/H$_2$ bubbled.
However, the negative current indicates that the passive layer is cathodically protected, which means there is no active dissolution of the amorphous alloy. It could also be realized from Fig. 5 that the current is relatively constant indicating that the passive film is stable under the operating potential of 0.1 V of PEMFC anodic environment.

3.3.2. PEMFC cathodic environment

Similarly, for cathodic PEMFC environment bubbled with air under the operating potential of 0.6 V applied to the amorphous alloy, a typical passivation behavior was noticed as shown in Fig. 6. In contrast to the anodic environment, a short time of about 16 min is required for the passive film formation and the current remains stable at an order of 0.12 mA/cm², for a period of 5 h. Moreover, no cathodic current is observed, and this could be attributed to the fact that the applied voltage of 0.6 V is on the middle of the passive layer region. However, the time required to stabilize the passive layer is different for the anodic and cathodic PEMFC environment, which could be explained by the difference in thickness and composition of the passive layers. This is further verified by analyzing the passive layer surface using XPS technique, which will be discussed later. The potentiostatic study of the Fe₅₀Cr₁₈Mo₈Al₈Y₂C₁₄B₆ glassy alloy confirms the stability of the passive layer formed in the both anode and cathode simulated PEMFC environment. Quantitative indication of the corrosion rate can be obtained from these potentiostatic polarization studies and it is discussed in the following chapter.

3.4. ICP-AA analysis of the solution

At the end of the potentiostatic test, the solution was sampled and analyzed for Fe, Cr, Mo, Al, B, and Y by inductively coupled atomic spectroscopy. According to the Faraday’s law, the chemical change occurring at an electrode is proportional to the quantity of electricity passing through the cell and it is given by the following equations [24]:

\[ Q = nFN \]  
\[ Q = it \]

The amount \( N \) of the each chemical species (Fe, Cr, Mo, Al, Y, B) dissolved during the corrosion process measured by ICP-AA technique is given in Table 2 and \( n \), the number of electrons involved per chemical species in the cell during the corrosion process, is assumed to be 3. Substituting these values in the above equations, the amount dissolved can be calculated.
However, if \( Q \) with pressurized air.

Fe50Cr18Mo8Al2Y2C14-

the applied voltage is close to the corrosion potential [25].

of hydrogen or even partial electrochemical reactions when

reactions like evolution of gases, electrochemical oxidation

i

current (calculated from Eq. (2) and it corresponds to the area of the

using Eq. (1). The total charge density (\( Q \))

\( Q \) (given as

values and \( F \), the Faraday’s constant, the charge density

given as \( Q_s \)) was calculated from the solution analysis by

using Eq. (1). The total charge density (\( Q \)) was also

calculated from Eq. (2) and it corresponds to the area of the

current (\( i \)) time (\( t \)) curve shown in Figs. 5 and 6. The
difference in the \( Q \) and \( Q_s \) could be related to non-corrosion

reactions like evolution of gases, electrochemical oxidation of hydrogen or even partial electrochemical reactions when

the applied voltage is close to the corrosion potential [25].

However, if \( Q \) is identical to that of the \( Q_s \) value, it implies

that the current in the potentiostatic test is entirely due to the

corrosion reaction. Consequently, the corrosion product in the

solution, analyzed by ICP-AA technique can give a direct

and reliable measure of corrosion as compared to the

potentiostatic and/or potentiodynamic techniques. From

Table 2, the amount of Cr (passivating element), dissolved

in the hydrogen purged environment is higher than in air

bubbled PEMFC environment indicating that the hydrogen

bubbled environment is very aggressive when compared to

the air bubbled, which is in good agreement with the

potentiodynamic studies. The corrosion current density is a

direct measure of the corrosion resistance offered by the

materials, so the equivalent corrosion current density (\( I_{corr-S} \))

was calculated by averaging the charge density obtained

from solution analysis (\( Q_s \)) over the time and reported

in Table 2. For comparison, the corrosion current density

(\( I_{corr-P} \)) was also estimated from the potentiodynamic curves

shown in Figs. 3 and 4 by applying the Tafel slope method

[26]. Under hydrogen bubbling, the corrosion current density

value from the electrolysis (\( I_{corr-S} \)) is lower and a slightly

higher corrosion rate was observed from the potentiody-

namic test (\( I_{corr-P} \)). Indeed, these potentiodynamic tests were

performed for shorter time, indicating that not enough time

was given for the passive film formation. However, these

corrosion data already provide useful information on the

corrosion performance. For air bubbling, the values of \( I_{corr-S} \)

and \( I_{corr-P} \) are compatible indicating that the passive layer

reaches its stability at a faster rate. By comparing the \( I_{corr-P} \)

values (Table 2), it can be observed that the amorphous alloys

exhibit a low corrosion current density when compared with

stainless steel, thus demonstrating a superior corrosion resistance. This preliminary study indicates that the newly

developed high chromium Fe-based amorphous alloy is a

potential candidate material for bipolar plates in PEMFC.

The Fe50Cr18Mo8Al2Y2C14B6 metallic amorphous bipolar

plates for PEMFC are under construction in the authors’

laboratory and will be investigated under real operating

conditions.

3.5. Origin of passivity analyzed by XPS technique

To understand the difference of corrosion resistance of

the Fe50Cr18Mo8Al2Y2C14B6 amorphous alloy under both

PEMFC anodic and cathodic environments the passive films

formed in air before tests and during the potentiostatic

studies were analyzed by XPS technique. The XPS depth

profile of the passive film formed before the test (Fig. 7(a)),
during potentiostatic polarization in simulating PEMFC

anodic (Fig. 7(b)) and cathodic (Fig. 7(c)) environment,

reveals that the air passivated film was mainly composed of

iron rich oxide layer of atomic concentration of about

18 at.% with the inner layer of chromium oxide of about

5 at.%. With the sputtering rate of 3.5 nm/min calibrated to

a SiO2 standard layer, the thicknesses of the passive layers

formed in air, anodic and cathodic PEMFC environments

were found to be about 2.1, 3.8, and 5.6 nm, respectively.

Interestingly, the passive film formed in simulated PEMFC

environment shows a significant difference when compared

Table 2

Summary of the corrosion tests result

| Alloy                | Condition       | Dissolved species by ICP- AA (mg/l) | Charge density (C/cm²) | Corrosion current density (μA/cm²) |
|---------------------|-----------------|------------------------------------|------------------------|-----------------------------------|
| SS316L stainless steel | H2 bubbled      |                                     |                        |                                    |
|                     | Air bubbled     |                                     |                        |                                    |
| Fe50Cr18Mo8Al2Y2C14B6 amorphous alloy | H2 bubbled | 1.09/0.60/0.05/0.23/2.16/0.091/2.72/0.29/0.38/0.25/0.66/0.12/ | 0.179 0.22 | 9.95 197 |
|                     | Air bubbled     |                                     | 0.242 2.07             | 13.44 15 |

Fig. 6. Current density as function of time for Fe50Cr18Mo8Al2Y2C14B6 amorphous alloy at −0.1 V in 1 M H2SO4+2 ppm F− at 75 °C bubbled with pressurized air.
to the air passivated. An outer chromium rich oxide layer was formed with the iron oxide layer, and careful inspection shows that the chromium oxide concentration doubles and quadrupled for the respective PEMFC anodic and cathodic environment when compared to the air passivated film, while the iron oxide concentration decreased by two fold or remains almost constant. Furthermore, from Fig. 7((b) and (c)), we can assume that, the thin Fe-oxide forms first. Since it is porous, oxygen diffuses inward and chromium diffuses outward, thereby forming a Cr2O3 protective oxide layer. It should be noticed that in each case, the molybdenum oxide concentration remains almost constant at about 4 at.%. Thus the XPS analysis indicated that the chromium enriched passive film is the origin of the high corrosion resistance for Fe50Cr18Mo8Al2Y2C14B6 amorphous alloy in PEMFC environment.

4. Conclusion

The corrosion property of two amorphous alloy in 1 M H2SO4 + 2 ppm F- at 75 °C purged either with hydrogen or pressurized air, to simulate the PEMFC environment for bipolar plate applications, are compared with that of stainless steel SS316L. The potentiodynamic studies show that the chromium content plays an important role on the corrosion resistance and that the Fe50Cr18Mo8Al2Y2C14B6 amorphous alloy exhibits a better corrosion resistance when compared to the SS316L of identical Cr content. The potentiostatic studies for Fe50Cr18Mo8Al2Y2C14B6 alloy confirms the stability of the passive layer formed in the PEMFC environment. Analysis of the corrosion products by ICP-AA technique gave a direct measure of the corrosion rate and also indicated the importance of the chromium content, on the corrosion resistance behavior. Since the Fe50Cr18Mo8Al2Y2C14B6 amorphous alloy shows a high corrosion resistance, by the formation of chromium enriched passive layer (confirmed by XPS profile) in the PEMFC environment, from a corrosion resistance standpoint this amorphous alloy is a candidate material for the bipolar plates in PEMFC.

Acknowledgements

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