

PdAgAu alloy with high resistance to corrosion by H₂S

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Abstract

PdAgAu alloy films were prepared on porous stainless steel supports by sequential electroless deposition. Two specific compositions, $\text{Pd}_{83}\text{Ag}_2\text{Au}_{15}$ and $\text{Pd}_{74}\text{Ag}_{14}\text{Au}_{12}$, were studied for their sulfur tolerance. The alloys and a reference Pd foil were exposed to 1000 $\text{H}_2\text{S}/\text{H}_2$ at 623 K for periods of 3 and 30 hours. The microstructure, morphology and bulk composition of both non-exposed and H_2S -exposed samples were characterized by X-ray diffraction (XRD), scanning electron microscopy (SEM) and energy-dispersive X-ray spectroscopy (EDS). XRD and SEM analysis revealed time-dependent growth of a bulk Pd_4S phase on the Pd foil during H_2S exposure. In contrast, the PdAgAu ternary alloys displayed the same FCC structure before and after H_2S exposure. In agreement with the XRD and SEM results, sulfur was not detected in the bulk of either ternary alloy samples by EDS, even after 30 hours of H_2S exposure. X-ray photoelectron spectroscopy (XPS) depth profiles were acquired for both PdAgAu alloys after 3 and 30 hours of exposure to characterize sulfur contamination near their surfaces. Very low S 2p and S 2s XPS signals were observed at the top-surfaces of the PdAgAu alloys, and those signals disappeared before the etch depth reached ~ 10 nm, even for samples exposed to H_2S for 30 hours. The depth profile analyses also revealed silver and gold segregation to the surface of the alloys; preferential location of Au on the alloys surface may be related to their resistance to bulk sulfide formation. In preliminary tests, a PdAgAu alloy membrane displayed higher initial H_2 permeability than a similarly prepared pure Pd sample and, consistent with resistance to bulk sulfide formation, lower permeability loss in H_2S than pure Pd.

Keywords: PdAgAu alloy, sulfur tolerance, ternary alloy, hydrogen separation membrane.

1. Introduction

Hydrogen has received significant attention as an energy carrier because it can be used for electricity generation in fuel cells without emission of CO₂. H₂ production from fossil fuels will continue for many decades to come because of the relatively low cost and high availability of coal and, to an increasing extent, natural gas. Conventional H₂ production from fossil-derived syngas requires multiple stages of purification to completely remove impurities such as CO and H₂S, which can deactivate the platinum electrodes of polymer electrolyte membrane fuel cells.

One strategy for reducing production costs and increasing the yield of ultrapure hydrogen is to use membrane type reactors for reforming and/or water-gas shift. In a single unit operation, membrane reactors deliver high purity hydrogen (by permeation of product hydrogen through reactor walls) and a separate concentrated CO₂ stream that can be sequestered. Pd-based membranes have been extensively studied for the H₂ separation application. The Pd surface is active for dissociation of molecular hydrogen and H-atoms readily diffuse through bulk Pd, enabling high selectivity for hydrogen permeation [1-6]. However, exposure to H₂S, a common minor component of methane and fossil-derived syngas, can compromise membrane permeability by formation of surface or bulk sulfides [7].

Exposure of pure Pd to H₂S is well known to cause formation of a bulk sulfide corrosion product over a wide range of H₂S concentrations (3.5 – 1000 ppm) and exposure temperatures (623 – 1100 K). The corrosion product, Pd₄S, has a hydrogen permeability about an order of magnitude lower than Pd [7, 8]. To improve its tolerance to sulfur, Pd has been alloyed with various minor components [5, 9-15]. PdCu has been extensively studied for its sulfur tolerance over a range of temperatures and H₂S concentrations [16-18]. The potential for bulk corrosion of a PdCu alloy by H₂S exposure depends not only on the alloy composition, but also on the temperature of exposure and the composition of the exposure gas; PdCu alloys will not corrode at conditions at which the sulfide corrosion product is unstable [18]. Thus, at high temperatures

(generally $> 900\text{K}$), PdCu alloys with 30-50% Cu are completely inert to H_2S , even at concentrations of H_2S as high as 1000 ppm [18]. At lower temperatures, which are preferred for H_2 production, surface sulfides are formed, which make the surface inactive for the dissociation step [16-18].

PdAu has also been studied for its sulfur tolerance, but not to the same extent as PdCu [10, 14, 15]. Sulfide formation has been reported in low-Au composition alloys (4-7 %) at mild exposure conditions (673 K and 20 ppm H_2S), while PdAu alloys with higher Au composition ($> 20\%$) have displayed high sulfur tolerance at 623 K, even upon exposure to 66,000 ppm of H_2S [10]. In addition, several authors have reported that the PdAu alloy has a higher permeability than PdCu alloy membranes [19 – 22]. In “clean” environments, PdAg alloys display higher permeability than PdAu, PdCu and even pure Pd. However, in the presence of H_2S , PdAg corrodes as quickly as Pd [5, 10].

With the aim of combining the H_2S tolerance of the PdAu binary with the high permeance of the PdAg binary, we prepared ternary PdAgAu alloys by sequential electroless plating and characterized their resistance to bulk corrosion by H_2S . We exposed a pair of ternary alloys and a reference Pd sample to 1000 ppm of H_2S for periods of 3 and 30 hours at 623 K. We show that, unlike the Pd reference sample, the ternary alloys do not react with H_2S to form bulk sulfide under these conditions. Resistance to bulk corrosion may be linked to co-segregation of Ag and especially Au to the top surface of the alloy.

2. Experimental

2.1 PdAgAu membrane preparation

Two PdAgAu films were deposited by sequential electroless plating onto porous, $0.1\text{ }\mu\text{m}$ grade (PSSD), stainless steel 316L discs, 1.27 cm diameter x 2 mm thick. Before metal deposition, discs were washed with an alkaline solution consisting of 0.12 M $\text{Na}_3\text{PO}_4 \cdot 12\text{ H}_2\text{O}$,

0.6 M Na_2CO_3 and 1.12 M NaOH using the procedure reported by Ma and coworkers [23], and then annealed in air for 12 hours at 773 K. To avoid intermetallic diffusion between the stainless steel substrate and the PdAgAu alloy components, the supports were modified with alumina by a vacuum assisted dip coating method [24]. Before the electroless plating step, the discs were activated. The sensitizing-activation of the substrate was performed at room temperature, first in an acidic SnCl_2 (1 g/l) solution, and then in an acidic PdCl_2 solution (0.1 g/l). The sensitizing-activation process was repeated six times. Palladium, silver and gold were deposited by sequential electroless deposition using the bath compositions shown in Table 1. For the sample with the lower Ag content, palladium was deposited in two steps of 60 min each, followed by a short Ag deposition of 5 min. In the case of the sample with the higher Ag content, palladium was deposited in two steps of 60 min and 40 min each, followed by Ag deposition for 15 min. For both samples, the Au deposition on top of the Pd–Ag layers was performed for 20 min. After the metal depositions, the samples were rinsed with water and dried at 393 K overnight. The three-metal deposition cycle was repeated for both samples to achieve a target thickness of $\sim 10 \mu\text{m}$.

The ternary samples were annealed at 773 K in H_2 to promote metallic inter-diffusion and alloy formation. The samples were mounted in a tube furnace as shown in Figure 1 and heated from room temperature to 773 K at 0.5 K min^{-1} in flowing nitrogen. The flow was then changed to hydrogen and the samples were held at 773 K in H_2 for 120 h. X-ray diffraction experiments revealed that FCC ternary alloy formation was incomplete at 773 K, so both samples were incrementally annealed at 873 K in H_2 for 48 h.

2.2 PdAgAu membrane H_2S treatment

For the study of the sulfur tolerance of the PdAgAu alloy, two ternary alloys and a $25 \mu\text{m}$ thick Pd foil were exposed to a $\text{H}_2\text{S}/\text{H}_2$ gas mixture. The pure Pd foil was used as a reference

because it is well known that it corrodes upon exposure to H₂S, even at low concentrations [10]. Samples of the ternary alloys and a Pd foil were mounted horizontally in a tube furnace. Then, they were heated from room temperature to 623 K at 1 K min⁻¹ in a flow of high purity argon (40 ml/min). Next, the samples were treated in flowing hydrogen (10 ml/min, high purity) at 623 K for 30 min, followed by a mixture of 1000 ppm H₂S/H₂ (10 ml/min) for 3 hours. Immediately after the H₂S treatment, the reactor feed gas was switched back to Ar (90 ml/min) for 25 minutes to sweep all the H₂S trapped in the reactor volume. After characterization, the 3-hour samples were exposed to H₂S for another 27 hours (for a total of 30 hours of exposure).

2.3 Film characterization

X-ray diffraction. The phase structure of the samples was determined by X-ray Diffraction. The XRD patterns of the films were obtained using a PANalytical X'Pert Pro instrument, having a theta-theta configuration, a Cu X-ray source operated at 45 kV and 40 mA and an X'Celerator detector equipped with a monochromator. Patterns were recorded over a 2 θ range of 20 to 100° at a step size of 0.02°.

Scanning electron microscopy, mapping and energy-dispersive X-ray analysis. The outer surface of the samples was imaged using a Tescan Vega-3 SEM instrument, operated at 20 kV in high-vacuum mode. The instrument was equipped with an Oxford INCA energy dispersive X-ray detection system for composition mapping; compositions were quantified using software provided by the manufacturer.

Elemental depth profiles by X-ray photoelectron spectroscopy (XPS). XPS depth profiles were performed using a ThermoFisher Theta Probe instrument. The instrument has a base working pressure of 1.0 x 10⁻¹⁰ kPa; it is equipped with a monochromatic Al K α X-ray source. The

instrument's hemispherical analyzer was operated in Constant Analyzer Energy (CAE) mode with a pass energy of 200 eV. Depth profiles were performed using argon ion sputtering. A differentially pumped ion gun was operated at 1×10^{-8} kPa, 3 kV and 500 nA, conditions which delivered a sputtering rate of approximately 1 nm min^{-1} . Sputtering was performed in 10 steps of 60 s, followed by 10 steps of 180 s to examine the top $\sim 50 \text{ nm}$ of the sample surfaces. Before sputtering and then at each sputtering step, photoemission spectra for Pd 3d, Pd 3p, O1s, C 1s, Au 4f, S 1s, S 2p core levels were recorded; peak areas were determined by integration employing a Shirley-type background. Sensitivity factors provided by the instrument's manufacturer were used for the quantification of the elements.

3. Results and Discussion

3.1 PdAgAu alloy formation and phase analysis after exposure to H_2S

X-ray diffraction patterns of the Pd foil and the two alloys, before and after exposure to H_2S , are displayed in Figure 2. The XRD patterns of the as-deposited alloy samples (bottom patterns, Figures 2a and 2b) contain diffraction features of a pure Pd phase ($2\theta = 40.12^\circ, 46.62^\circ, 68.12^\circ$) and pure phases of Ag and Au ($2\theta = \sim 38.15^\circ, \sim 44.33^\circ, \sim 64.50^\circ$). After the first annealing cycle performed at 773 K during 120 hours, formation of the FCC ternary alloy phase was incomplete, as illustrated by the continued appearance of small features at $2\theta = 40.12^\circ, 46.62^\circ$ and 68.12° . For both samples, formation of the alloy FCC phase was complete after a second annealing at 873 K for 48 hours.

Several investigations of PdAg and PdAu alloy formation have been reported in the literature and provide context for our results [14, 25 - 27]. Complete formation of a $\text{Pd}_{86}\text{Ag}_{24}\text{FCC}$ alloy has been observed upon annealing at 773 K in H_2 for 110 hours [26]. $\text{Pd}_{93}\text{Au}_7$ alloy formation, in contrast, is a slower process requiring more than 500 hours in H_2 at 773 K [27]. The difference reflects the low solid-phase mobility of Au (Tammann temperature =

669 K) relative to Ag (Tammann temperature = 618 K). The aggressive anneal conditions that our PdAgAu alloys require for complete alloy formation suggest that the Au component dominates the alloying process.

After alloy formation, both ternary samples and a Pd foil were treated in 1000 ppm $\text{H}_2\text{S}/\text{H}_2$ at 623 K for 3, and then 30 hours. The H_2S -exposed samples were characterized by XRD in the $30^\circ - 80^\circ$ range to study the phases formed during the treatment. The Pd foil shows clear evidence of bulk sulfide growth upon H_2S exposure. A Pd_4S phase starts to form at 3 hours and grows with increasing exposure time. The XRD pattern of the Pd foil exposed for 30 hours (Figure 2c) reveals an almost complete Pd_4S phase in the XRD analysis depth. In contrast, PdAgAu alloys show the same FCC phase before and after 30 hours of exposure, as shown in Figure 2a and b, with no evidence of formation of a crystalline sulfide phase.

Table 2 is a summary of results published in the literature involving the effect of H_2S exposure to Pd, PdAg and PdAu alloys at several temperatures. Several authors have used XRD to identify bulk sulfide phases in Pd and PdAg alloys upon exposure to relatively low H_2S concentrations [5, 10, 11, 17 – 19, 28]. Consistent with the results of our Pd reference sample, O'Brien et al. [16] reported complete conversion of a Pd foil to Pd_4S upon exposure to 1000 ppm H_2S at 623 K for 6 hours. Mundschau et al. also observed the Pd_4S XRD pattern for a Pd foil treated with 20 ppm H_2S at 593 K for 115 hours [5]. These authors also reported formation of mixed $\text{Ag}_5\text{Pd}_{10}\text{S}_5$ sulfide phase upon exposure of a $\text{Pd}_{75}\text{Ag}_{25}$ alloy membrane to 10 ppm H_2S at 593 K for 65 hours.

While not studied as thoroughly as Cu, an Au minor component has also been reported to improve Pd's resistance to H_2S corrosion. Ma and co-workers [14] did not observe sulfide formation upon exposure of a $\text{Pd}_{94.5}\text{Au}_{5.5}$ alloy to 54.8 ppm H_2S in H_2 at 673 K for 24 hours. McKinley reported that a $\text{Pd}_{74}\text{Au}_{26}$ alloy did not tarnish (visual inspection) upon exposure to H_2S (in H_2) concentrations as high as 6.6% at 623 K [10]. In a counter-example, formation of Pd-

sulfide phases, detected by XRD, was observed in low-Au content (6-7 at%) binary and ternary alloys that were exposed to 20 ppm H₂S in a complex background of H₂, H₂O, CO₂, and CO at 673 K [15, 29]; these components might have an influence on the sulfide formation. Our ternary membranes did not react to form bulk sulfides at exposure conditions more severe than those that cause corrosion of PdAg alloys, providing clear evidence of Au's role in imparting sulfur tolerance to the PdAg binary.

3.2 Pd and PdAgAu ternary alloys morphology characterization

The Pd foil and the ternary samples were characterized by SEM both before and after exposure to H₂S. Figure 3 displays micrographs for the three non-exposed samples, and the same samples after 3, and then 30 hours of exposure to 1000 ppm H₂S. After 3 hours of exposure, the Pd foil micrograph (Figure 3b) reveals the appearance of discrete surface aggregates, evidence of partial conversion of the foil to Pd₄S. After 30 hours of treatment (Figure 3c) the micrograph shows that the Pd₄S layer has grown to cover the entire surface. The H₂S-exposed Pd foils were also characterized by X-ray mapping, as shown in Figure 4. The sulfur maps confirm the presence of S in the aggregates formed at 3 hours, and demonstrate complete sulfur coverage after 30 hours. Bulk composition characterization of the 30-hour sample by EDS revealed Pd₄S stoichiometry, confirming that the reference foil had been completely converted to the sulfide.

In contrast to the Pd foil, neither PdAgAu ternary sample exhibited a morphology change upon H₂S exposure (Figures 2d-i). The micrograph of the Pd₇₄Ag₁₄Au₁₂ sample reveals pinhole pores on its surface, which is likely related to the relatively high Ag content of that sample; the dendritic morphology of the silver deposits has been reported by many authors [30, 31]. EDS-mapping of the alloys after 3 and 30 hours of exposure (not shown) did not reveal the presence of S. Thus, in agreement with XRD results, SEM and mapping analyses of the H₂S-exposed PdAgAu alloys did not uncover evidence of bulk sulfide formation upon aggressive treatment

with H₂S.

3.3 XPS depth profile analysis after 3 hours of H₂S exposure

To assess local sulfur contamination at the surfaces of the materials, which may be undetectable by EDS and XRD, XPS composition depth profiles were performed on the Pd foil and the ternary alloys after 3 hours of H₂S exposure. The atomic compositions of Pd, Ag, Au and S as a function of etching time/depth are plotted in Figure 5. The Pd foil has a significant sulfur content (S > 10 at.%) throughout the ~ 40 nm profiling experiment, as shown in Figure 5a. In contrast, the depth profiles of both alloys display only very weak S signals at the top-surface (etch time = 0), which disappear before the etch depth reaches ~ 10 nm. At the same depth, the relative Pd, Ag and Au contents of the surface region reach steady state values that are the same as bulk compositions measured by EDS. Our sulfur profiles are similar to those reported by O'Brien et al. [16], who used XPS depth profiling to study the surface compositions of Pd and Pd₄₇Cu₅₃ foils after exposure to 1000 ppm H₂S at 623 K for 6 hours. These authors observed constant sulfur content of 20 at % (i.e., Pd₄S stoichiometry) throughout the depth profile of their Pd sample. S appeared only at the very top-surface of their Pd₄₇Cu₅₃ alloy.

Metals composition profiles (Pd, Ag and Au only; S not included in the calculation) of the PdAgAu alloys samples after 30 hours of H₂S exposure are displayed in Figure 6. Clearly, Ag and Au co-segregate to the top surfaces (etch time = 0) of both alloys. The extent of Au segregation to the surface of Pd₈₃Ag₂Au₁₅ (top-surface composition: Pd₇₃Ag₇Au₂₀; surface/bulk Au ratio = 20/15 = 1.33) is similar to the surface of Pd₇₄Ag₁₄Au₁₂ sample (top-surface composition: Pd₆₆Ag₁₉Au₁₅; surface/bulk Au ratio = 15/12 = 1.25). Au segregation to the surface of PdAu alloys has been experimentally observed by a number of researchers [32-35]. Swartzfager et al. [32] reported a surface Au content of 52 % (at.) for a Pd₈₀Au₂₀ alloy prepared by melting high-purity Au and Pd in a vacuum induction furnace and annealed at 873 K, while

Piccolo et al [33] reported gold concentrations of 75% and 85% in the top layers after annealing the (1 1 1) and (1 1 0) surfaces of a Pd₇₀Au₃₀ alloy above 723 K. Similar surface enrichment, attributed to surface segregation of Au, was observed by Goodman and co-workers [34] using 1:1 Pd-Au / Mo (110) alloy annealed at 800 K. Li et al [35] studied the surface segregation of a Au to the surface of a Au/Pd (1 1 1) alloy by LEED and LEIS. Au and Ag enrichment at the surfaces of alloys with Pd is consistent with the pure-component surface energies: both Ag (1.2-1.4 J/m²) and Au (1.63 J/m²) have lower surface energies than Pd (2.05 J/m²) [36, 37]. However, further experiments would be necessary to corroborate whether the Au-Ag segregation is caused after H₂S or H₂ exposure.

As described earlier, alloying Pd with Au often improves its resistance to bulk corrosion by H₂S; the work we present here shows that Au also imparts corrosion resistance to the PdAg binary. Au preferentially segregates to the surfaces of both the PdAu binary and the PdAgAu ternary alloys; high surface concentrations of Au are likely linked to the lack of reactivity of the alloy surfaces toward H₂S. The high H₂S tolerance of the PdAgAu alloys is consistent with the Au enrichment at the surface and the pure-component standard free energies of sulfides formation at 623 K: Au₂S (- 8,901 J/mol), Ag₂S (- 63,427 J/mol) and Pd₄S (- 132,005 J/mol) [38-40].

3.4 PdAgAu alloy permeation properties: preliminary results

Using an apparatus and procedure that have been described previously [41], the permeability of a 14 µm thick Pd₈₃Ag₂Au₁₅/0.1µmPSS disc (modified with ZrO₂) ternary alloy membrane was compared to that of a similarly prepared pure Pd membrane. At T = 673 K and ΔP = 50 kPa, the alloy membrane displayed pure H₂ permeability of 1.3 x 10⁻⁸ mol m⁻¹ s⁻¹ Pa^{-0.5}, slightly higher than that of the pure Pd sample, 1.2 x 10⁻⁸ mol m⁻¹ s⁻¹ Pa^{-0.5}. Both membranes had H₂/N₂ ideal selectivity >10,000. After 24 hours of exposure to 100 ppm H₂S/H₂ at 673 K, the

alloy membrane retained 32% of its initial permeability, while the pure Pd membrane retained only 15%. These preliminary results provide a link between improved permeability performance in the presence of H₂S and the resistance to bulk sulfide formation imparted by the minor alloy components.

4. Conclusions

Pd₈₃Ag₂Au₁₅ and Pd₇₄Ag₁₄Au₁₂ alloys, synthesized by sequential electroless plating, and a Pd foil reference were treated with 1000 ppm H₂S in H₂ at 623 K for 3 and 30 hours. As expected, the Pd foil reacted with H₂S to form a bulk Pd₄S corrosion product within the first 3 hours of H₂S exposure. In contrast, neither PdAgAu alloy displayed evidence of bulk sulfide development upon post-exposure examination by XRD, EDS, SEM and X-ray mapping. Local sulfur contamination at the surfaces of the ternary alloys and Pd foil was studied by XPS depth profiling. Consistent with the results of bulk characterization, depth profiles of the H₂S-exposed Pd reference samples show significant concentrations of S (~10 at% in the three-hour sample) throughout the 40 nm etch depth. In contrast, the PdAgAu alloys' sulfur profiles displayed only small S signals at the top-surface (etch time = 0), which disappeared before the etch depth reached ~ 10 nm, even after 30 hours of H₂S exposure. The depth profile experiments also revealed significant co-segregation of Ag and Au to the H₂S-exposed alloy surfaces. Preferential location of Au at the alloy surface is likely related to its resistance to bulk sulfidation by H₂S. In a preliminary permeation test, a Pd₈₃Ag₂Au₁₅ alloy delivered H₂ fluxes comparable to those of pure Pd, and a residual permeability of 32 % in 100 ppm H₂S/H₂ mixed gas experiments at 673 K after 24 hours of exposure.

Our work illustrates the resistance of PdAgAu alloys to bulk corrosion by H₂S. It suggests that fabrication of a ternary alloy membrane based on the high-permeability PdAg binary, but with an Au component specifically for H₂S tolerance, could be an effective strategy

for improving the membrane performance in extreme environments.

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Figure Captions

Figure 1: Scheme of the H₂S treatment system.

Figure 2: X-ray diffraction patterns of the Pd film before and after H₂S exposure (a) and both PdAgAu ternary alloys as prepared, after annealing and after H₂S exposure (b, c). Ternary samples annealing conditions: 1st cycle; 120 hours at 773 K in H₂, 2nd cycle; 48 hours at 873 K in H₂. H₂S treatment conditions: T= 623 K, [H₂S]= 1000 ppm, time= 3 and 30 hours.

Figure 3: Effect of time of exposure on the morphology of Pd foil (a, b and c), Pd₈₃Ag₂Au₁₅ (d, e and f) and Pd₇₄Ag₁₄Au₁₂ (g, h and i). H₂S treatment conditions: T= 623 K, [H₂S]= 1000 ppm, time= 3 and 30 hours.

Figure 4: SEM surface mapping images of the Pd foil after 3 and 30 hours of exposure. H₂S treatment conditions: T= 623 K, [H₂S] = 1000 ppm, time= 3 and 30 hours. Color: S - Pink, Pd - red.

Figure 5: X-ray photoelectron spectroscopy depth profiles of the Pd foil (a), Pd₈₃Ag₂Au₁₅ (b) and Pd₇₄Ag₁₄Au₁₂ (c) samples after 3 hours of H₂S exposure. Elemental concentrations shown for Pd (black), Ag (blue), Au (green) and S (red). Analysis conditions: 1st 10 sputtering cycles of 60 seconds each. 2nd 10 sputtering cycles of 180 seconds each. H₂S treatment conditions: T = 623 K, [H₂S]= 1000 ppm, time= 3 hours.

Figure 6: X-ray photoelectron spectroscopy depth profiles of the Pd₈₃Ag₂Au₁₅ (a) and Pd₇₄Ag₁₄Au₁₂ (b) samples after 30 hours of H₂S exposure. Sulfur was not included in the Pd/Ag/Au composition. Analysis conditions: 1st 10 cycles of 60 seconds each. 2nd 10 cycles of

180 seconds each. H₂S treatment conditions: T= 623 K, [H₂S]= 1000 ppm, time= 30 hours.

Table 1. Chemical composition of Pd, Ag and Au electroless plating solutions, activation and plating conditions.

	Activation / pH modifier	Plating bath		
		Pd	Ag	Au
SnCl ₂ (g/l)	1	-	-	-
PdCl ₂ (g/l)	0.10	-	-	-
HCl (M)	1	-	-	-
Hydrazine (mM)		10	10	-
28-30% NH ₄ OH (M)	-	9.8	9.8	-
Na ₂ EDTA (g/l)	-	180	180	-
PdCl ₂ (mM)	-	20.3	-	-
AgNO ₃ (mM)	-	-	10	-
Na ₂ S ₂ O ₃ . 5 H ₂ O (mM)	-	-	-	149.9
Na ₂ SO ₃ (mM)	-	-	-	169.9
C ₆ H ₈ O ₆ (mM)	-	-	-	339.9
AuCl ₃ . HCl. 4 H ₂ O	-	-	-	6.9
pH	-	11.5	11.5	11.0
Temperature (°C)	-	50	50	30

Table 2. Summary of literature results involving the effect of H₂S exposure to Pd, PdAg and PdAu alloys at several temperatures.

Reference	Membrane composition (Mol %)	Temperature (K)/ Time of exposure (h)	Feed composition	Experiment result
McKinley [10]	Pd	623 / 144	4.5 ppm H ₂ S in H ₂	Dull and etched surface.
Mundschau [5]	Pd	593 / 120	20 ppm H ₂ S/60% H ₂ -He	Pd ₄ S formation.
O'Brien [16]	Pd	623 / 6	1000 ppm H ₂ S/10% He-H ₂	Pd ₄ S formation.
This work	Pd	623 / 3	1000 ppm H ₂ S in H ₂	Pd ₄ S formation.
Mundschau [5]	Pd ₇₅ Ag ₂₅	593 / 65	10 ppm H ₂ S/80% H ₂ -He	Pd ₄ S and Ag ₅ Pd ₁₀ S ₅ formation.
McKinley [10]	Pd ₇₃ Ag ₂₇	623 / 48	3.5 ppm H ₂ S in H ₂	Dull and etched surface.
McKinley [10]	Pd ₇₄ Au ₂₆	623 / 144	4.3 ppm H ₂ S in H ₂	Retained original luster.
McKinley [10]	Pd ₇₄ Au ₂₆	623 / 168	20.6 ppm H ₂ S in H ₂	Retained original luster.
McKinley [10]	Pd ₇₄ Au ₂₆	623 / 6	66000 ppm H ₂ S in H ₂	Retained original luster.
Gade [15]	Pd ₉₄ Au ₆ (Cold-worked)	673 / 80	20 ppm H ₂ S/50% H ₂ / 29% H ₂ O/19% CO ₂ /1% CO	Pd ₄ S and Pd _{2.8} S formation.
Ma [14]	Pd _{94.5} Au _{5.5}	673 / 24	54.8 ppm H ₂ S in H ₂	No bulk sulfide formation.
Coulter [29]	Pd ₈₆ Au ₇ Pt ₇	673 / 100	20 ppmv H ₂ S/50% H ₂ / 29% H ₂ O/19% CO ₂ /1% CO	Pd ₄ S formation.
This work	Pd ₇₄ Ag ₁₄ Au ₁₂	623 / 30	1000 ppm H ₂ S in H ₂	No bulk sulfide formation.
This work	Pd ₈₃ Ag ₂ Au ₁₅	623 / 30	1000 ppm H ₂ S in H ₂	No bulk sulfide formation.

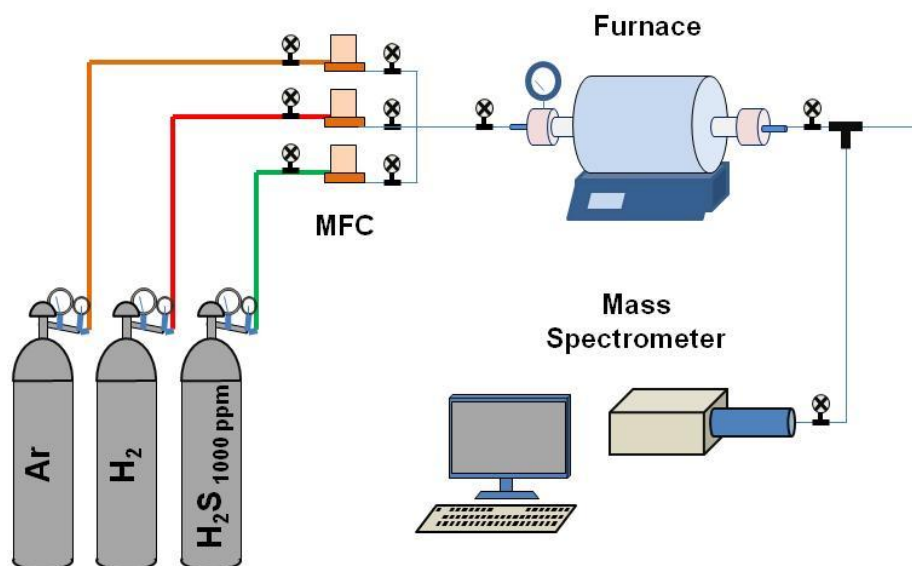


Figure 1

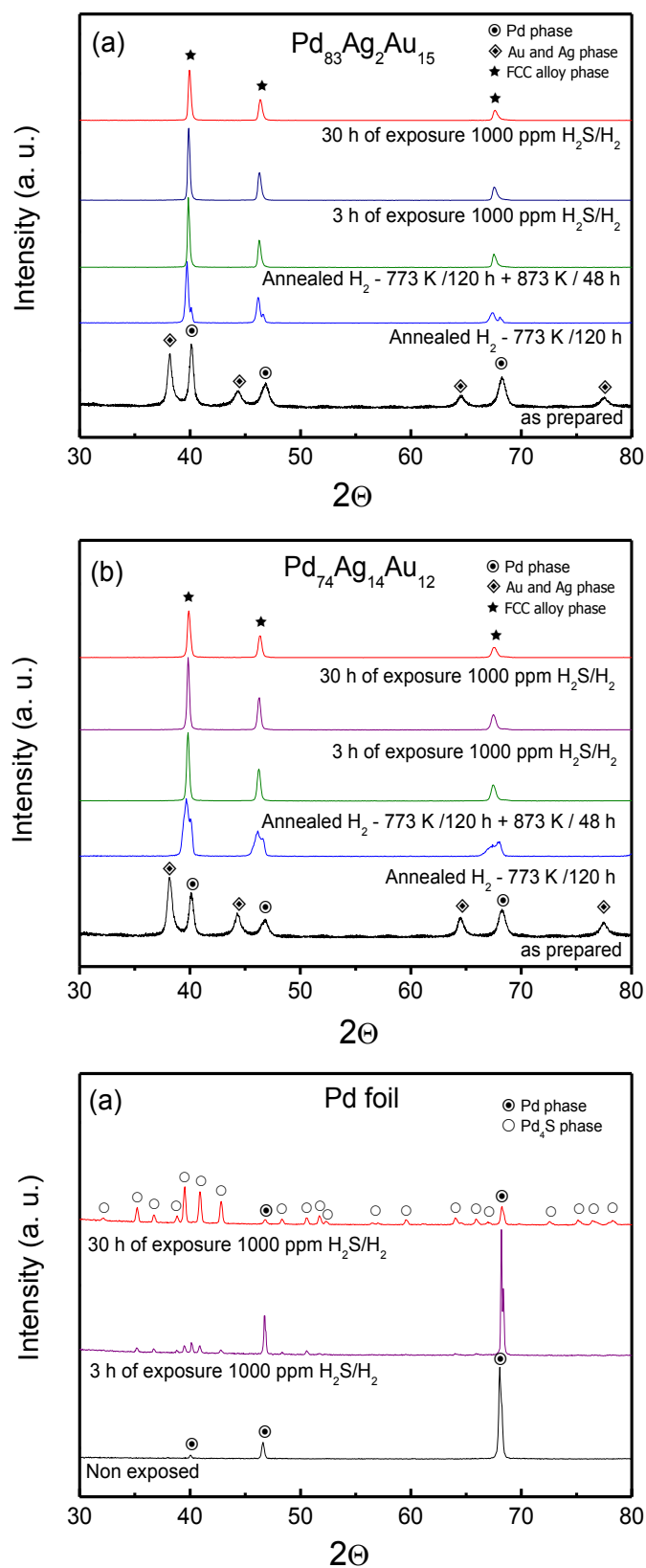


Figure 2

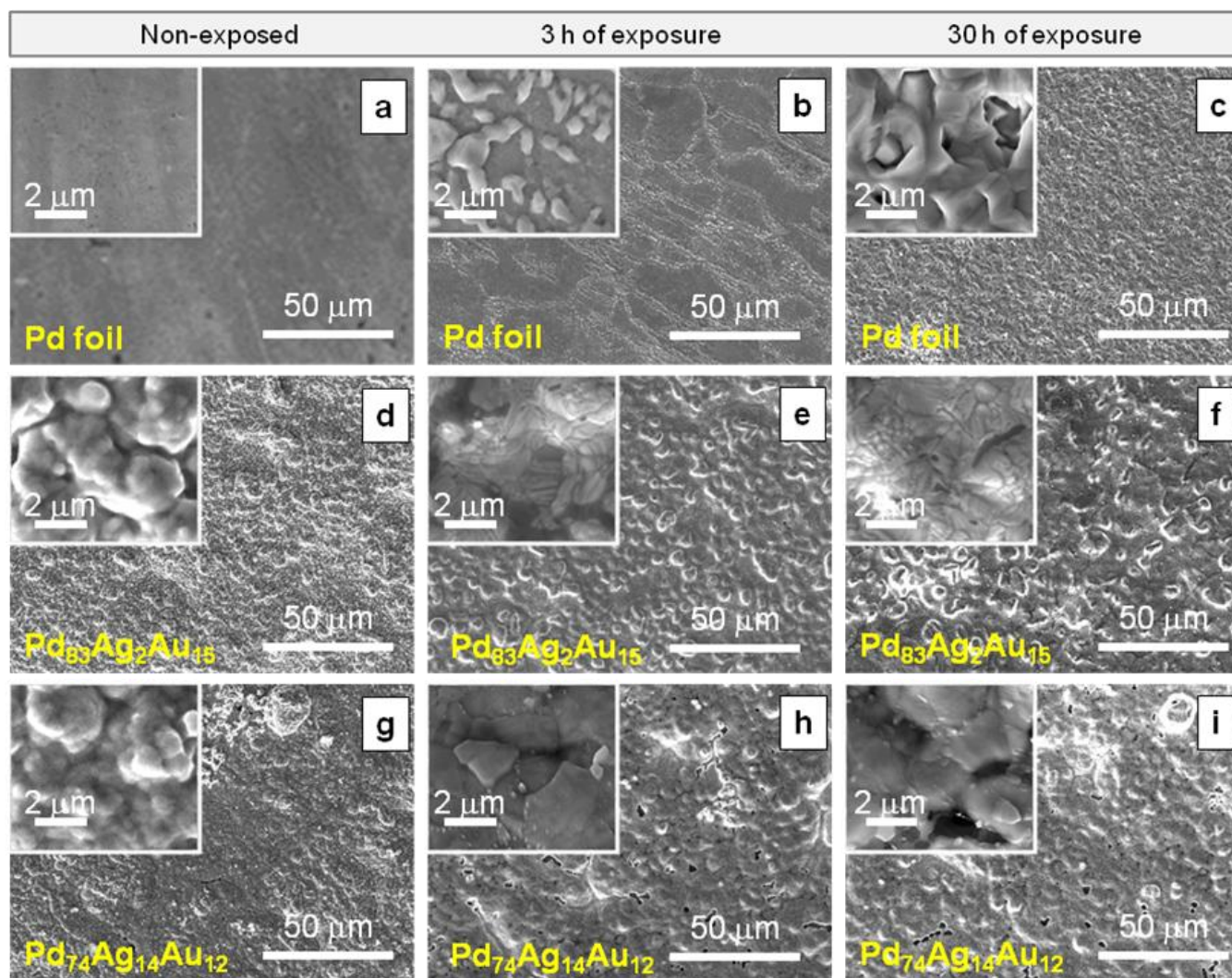


Figure 3

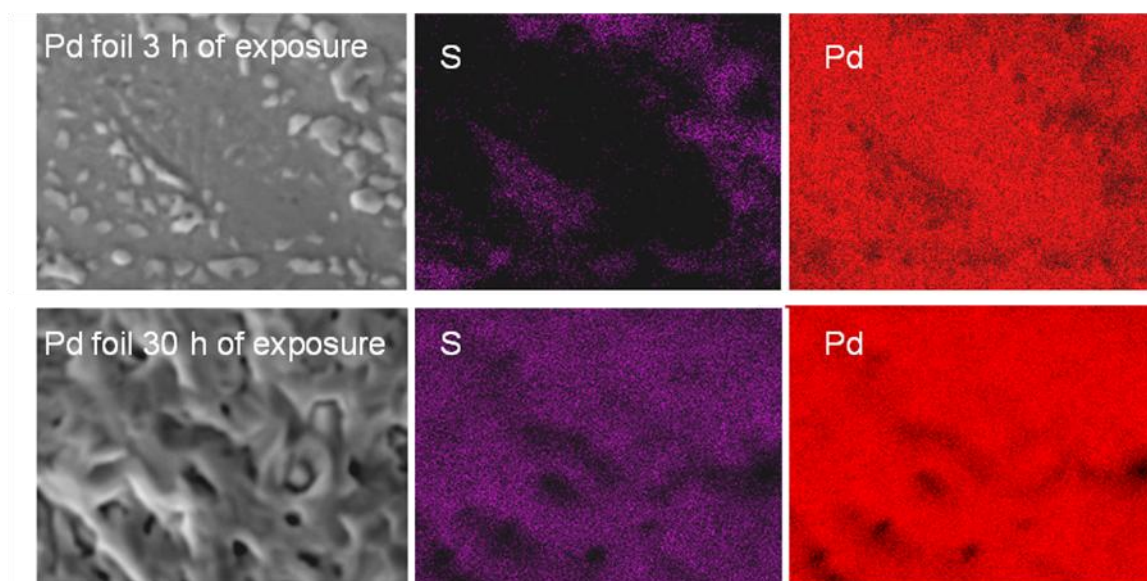


Figure 4

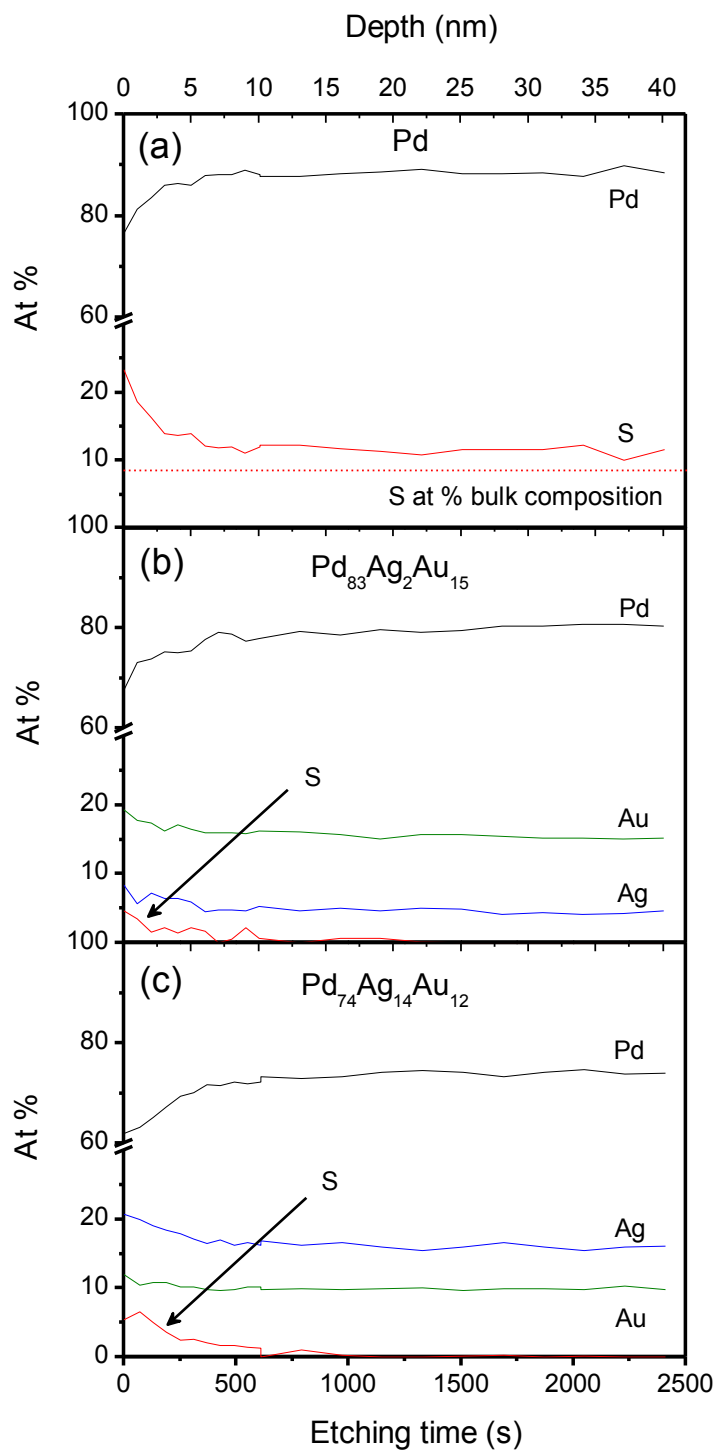


Figure 5

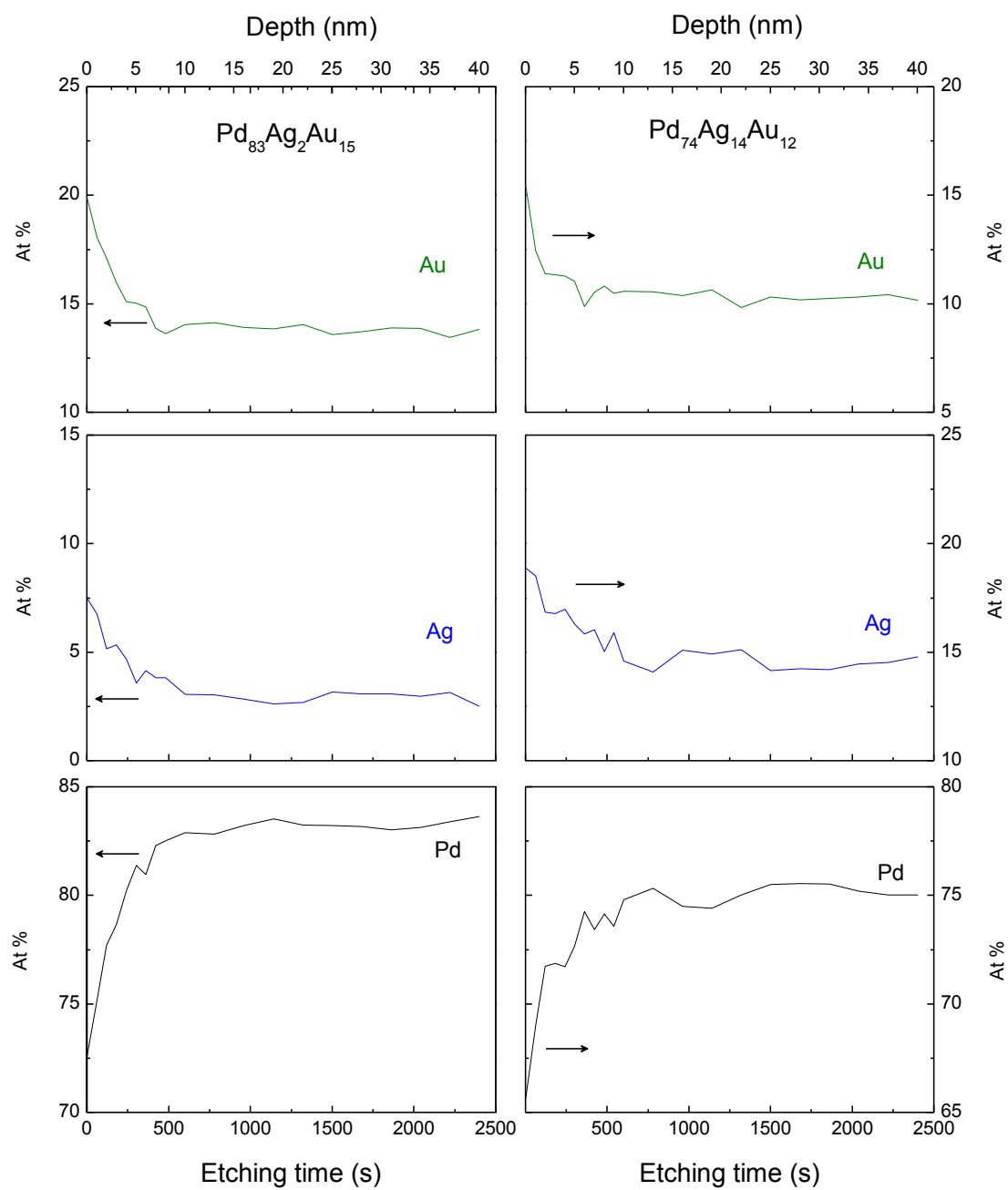


Figure 6