Dealloying is a common corrosion process that can trigger stress corrosion cracking and catastrophic failure of engineering alloys in corrosive environments. Over the past decade, dealloying has also emerged as a new approach for the fabrication of nanoporous (NP) metals ranging from noble metals such as Au and Pt to reactive elements including Al and Mg for a variety of novel functional and mechanical properties. With dealloying, a full-dense solid metal can be transformed into a new material with open, nanoscale porous structure, simply by mixing it with a less-noble (or more-reactive) element to form a solid solution or compound and then removing the less-noble element by selective dissolution or dealloying (chemically or electrochemically). The occurrence of dealloying is sensitive to the initial composition: dealloying proceeds only if the concentration of less-noble elements in the parent alloy exceeds a critical value, i.e., the partition limit, which is approximately 60 at.% Ag for Au-Ag alloys. The partition limit is associated with a competition between surface passivation (of more-noble elements) and nanoporosity evolution. Below partition limit, the dealloying is prevented by a passive layer of more-noble elements on surface formed at initial stage of corrosion. The competition between passivation and nanoporosity evolution has been discussed in relation to the surface diffusion of more-noble elements and the site percolation of less-noble elements in the bulk.

Similar to a passive layer of more-noble elements discussed above, an oxide passive layer may also influence the kinetics and even shift the partition limit of dealloying. Oxide passive layer is known to improve the corrosion resistance of some metals and engineering alloys. Most notably in stainless steels, addition of Cr (>13 at.%) in Fe can lead to a dense passive layer of Cr oxide that prevents further corrosion and significantly improves the corrosion resistance of the alloy in certain environments. Oxide passivation in Fe-Cr system has also been interpreted in terms of a site percolation mode – a concept that has been successfully used to understand the partition limit of dealloying. Therefore, formation of an oxide passive layer (of either less-noble or more-noble elements) may retard dealloying and increase partition limit. In this study we studied the dealloying behavior of Cu-Rh alloys. Unlike Au-Ag alloys that do not involve any oxide passivation throughout dealloying, the Rh in Cu-Rh alloys is prone to oxidation and surface passivation. One goal of this study is to clarify how the oxide passivation competes with nanoporosity evolution, and how it influences the critical potentials and partition limit for dealloying.

Dealloying of Cu-Rh is also interesting because it generates nanoporous Rh that may act as a catalyst similar to Rh nanoparticles. More interestingly, Rh may form a metastable phase if its size can be shrunk to nanoscale. Bulk Rh exhibits a face-centered-cubic (FCC) crystal structure. A recent study revealed that when the size of Rh particles was shrunk to a few nm, it obtained a hexagonal-close-packed (HCP) crystal structure that may lead to distinct physical and chemical properties compared with its FCC counterpart. Thus the other goal of this study is to clarify whether it is possible to produce a bulk sample of Rh, with nanoporous topological structure and HCP crystal structure, which might facilitate measurements that cannot be conducted on individual nanoparticles.

Our study revealed that during the electrochemical dealloying of Cu-Rh, the dealloying kinetics and the partition limit were influenced by surface oxide passivation. We also found that the dealloyed nanoporous Rh exhibited an FCC structure, which is identical to that of bulk Rh but contradicts the HCP structure reported in Rh nanoparticles.

Experimental

Parent alloys with nominal compositions of Cu$_x$Rh$_{1-x}$ (x = 0.75, 0.77, 0.80, 0.85, 0.91, 0.925, 0.95, 0.98) and Fe$_{0.90}$Rh$_{0.10}$ were prepared by arc-melting of high purity Cu wires (99.999%), Fe wires (99.999%) and Rh sheets (99.95%) under argon atmosphere. The ingots were sealed in a quartz tube with an argon atmosphere (99.999%), homogenized at 950 °C for 50 hours and then quenched in water. The ingots were then rolled to 0.1 mm thickness, cut into alloy sheets with the dimensions of 5 x 5 x 0.1 mm$^3$. These samples were then annealed at 850 °C for 1 hour for recovery. Alloy surfaces were mechanically polished down to 1μm roughness using diamond polishing paste and rinsed using deionized ultra-pure water (18.2 Ωcm/cm; Milli-Q Advantage A10) before electrochemical experiments.

Anodic polarization and other electrochemical experiments were performed using a three-electrode cell under potentiostatic control (AUTOLAB PGSTAT302N). The electrolyte, 1 M H$_2$SO$_4$, was prepared from ultrapure water and concentrated H$_2$SO$_4$ (98% concentration, Guarantee Reagent, Sinopharm Chemical Reagent Co., Ltd). The reference electrode is a saturated calomel electrode (SCE). We used a coated gold wire as counter electrode for anodic polarization and a coated copper wire for dealloying and passivation treatments. The cross-sections of dealloyed sheets were milled with a Leica EM TIC 3X Ion Beam Milling System and then characterized by

E-mail: hjjin@imr.ac.cn
**Results**

**Anodic polarization curves of Cu-Rh alloys.**—Figure 1a shows the polarization curves of Cu$_{0.80}$Rh$_{0.20}$ alloys with x varying between 0 and 1. These curves were obtained by measuring current density during anodic potential sweeping at a constant rate of 1 mV/s in 1 M H$_2$SO$_4$. The polarization starts from a rather low potential at which during anodic potential sweeping at a constant rate of 1 mV/s in 1 M H$_2$SO$_4$, the polarization curves of Cu$_{0.80}$Rh$_{0.20}$ alloys with current density plotted on log-scale.

When x is further decreased to 0.80, the active dissolution or active dealloying vanishes on polarization curve, because the active dealloying potential surpasses the passivation potential. The critical potential for transpassive dealloying also increases with decreasing x or increasing Rh content. As x decreases to 0.77, the polarization curve becomes similar to that of pure Rh: the rapid increase of current density at around 1.45V is accompanied by oxygen evolution process. Although the shiny polished sample surface becomes dull after polarization indicating that corrosion did occur to this sample, the rate of dealloying is extremely slow even at high potentials. Dealing of this sample at 1.50 V for 20 hours leads to a corrosion depth less than 1 μm, according to the mass loss during polarization.

Samples with more Rh are stable against corrosion. A Cu$_{0.75}$Rh$_{0.25}$ sample polarized at 1.80 V for 20 hours remained shiny and its mass did not change at all. This suggests that the composition threshold for dealloying of Cu-Rh under this circumstance is approximately 75 at.%Cu, which is much higher than the parting limit of Au-Ag (~60 at.%Ag), Cu-Zn (~20 at.%Zn) and Cu-Al (~16 at.%Al).42

We emphasize that the dealloying actually occurs in the passive region when the Rh content is not so high, even though the dealloying rate in this region is orders of magnitude lower than that in active and transpassive dealloying regions for a certain alloy composition. More details of passive region are shown in Fig. 1b, where the current density of the above-mentioned polarization curves are plotted on a log-scale. It shows that the passive current density decreases dramatically with increasing Rh content. With x decreasing from 0.925 to 0.85, or Rh content increasing from 7.5 to 15 at.%, the passive current density decreases by two orders of magnitude, and eventually diminishes as the Rh content increases further.

**Critical potentials and parting limit.**—For alloys such as Au-Ag that do not involve passivation, there is only one major characteristic potential for dealloying, i.e., a critical dealloying potential below which the dealloying does not occur and above which the dealloying proceeds to form a nanoporous structure. The corrosion of Cu-Rh alloys involves more critical potentials, as can be identified from their polarization curves (see Figure 2a). The critical potentials, $E_{c}$, $E_{p}$, $E_{t}$, denote the active dealloying potential, the primary passivation potential, and the transpassive dealloying potential, respectively. The $E_{c}$ and $E_{t}$ are determined as intersection point of the tangent extension line in transition region as shown in Figure 2a. For simplicity, $E_{p}$ is determined as a potential at the peak of current density between active dealloying and passivation. The characteristic potentials derived from Fig. 1a for each composition are displayed in Fig. 2b as a function of Cu content, x. This type of diagram was first introduced by Geng and Sieradzki43 for dealloying of Au-Ag (and Li-Sn). As will be presented below, the Cu-Rh shows a different diagram compared with Au-Ag.

For the dealloying of Cu-Rh, the equilibrium Cu$^{2+}$/Cu electrode potential for the dissolution of Cu from the flat surface of Cu-Rh alloy, $E_{c}$, is also plotted for comparison. In case of an ideal solid solution alloy, $E_{c}$ is determined by

$$E_{c} = E_{0} - \frac{RT}{nF} \ln \left( \frac{x (1-x)}{x-1} \right)$$

where $E_{0}$ (=0.101 V) is the equilibrium electrode potential of pure Cu, R is the gas constant, T is the absolute temperature, n is the number of electrons transferred in reaction, and F is the Faraday constant. The active dealloying potential of Cu-Rh alloys, $E_{c}$, is higher than $E_{c}$, and also increases with decreasing x. The primary passivation potential $E_{p}$

![Figure 1](https://example.com/f1.png)

Figure 1. a) Anodic polarization curves of Cu$_{1-x}$Rh$_{x}$ alloys in 1 M H$_2$SO$_4$, with a sweeping rate of 1 mV/s. Note that the high current densities at potentials above 1.45 V are mostly contributed by oxygen evolution, instead of metal dissolution. b) Anodic polarization curves of Cu$_{1-x}$Rh$_{x}$ alloys with current density plotted on log-scale.
Figure 2. a) A typical anodic polarization curve of Cu-Rh alloy, which includes active dealloying, passivation and transpassive dealloying regions. The critical potentials for active dealloying ($E_a$), primary passivation ($E_p$), and transpassive dealloying ($E_t$) are defined in this figure. b) A potential-composition diagram for dealloying of Cu$_x$Rh$_{1-x}$ at ambient temperature in 1M H$_2$SO$_4$, which are derived from anodic polarization curves shown in Fig. 1. The “oxide passivation” region is absent in a similar diagram for Au-Ag.$^{6,43}$

changes very little with $x$, while the transpassive potential $E_t$ increases rapidly with decreasing $x$.

The diagram shown in Fig. 2b can be divided into four regions: i) at the bottom of the diagram, below $E_e$, the etching of Cu atoms does not happen at all; ii) between $E_a$ and $E_p$, the Cu atoms on surface may be etched but the surface is quickly passivated with Rh (metallic state), and in consequence the dealloying could not continue; iii) between $E_a$ and $E_p$, the dealloying occurs and can propagate rapidly into the entire sample, leading to the formation of nanoporous structure; iv) between $E_p$ and $E_t$, the sample surface is covered with a passivation film of oxide.

As mentioned earlier, the dealloying does occur slowly in the high-$x$ part of oxide passivation region. As shown in Fig. 1b, the passive current density – an indication of dealloying rate in this region – is potential-insensitive but decreases dramatically with decreasing Cu content $x$ in the parent alloy. As also shown in this diagram (Fig. 2b), in the oxide passivation region (between $E_e$ and $E_a$), when the Cu content is between 0.85 < $x$ < 0.925, the dealloying is slow and the dealloying rate decreases with decreasing $x$; when the Cu content $x$ is decreased to below 0.85, the dealloying is completely stopped by the formation of an oxide passive layer.

This diagram may be used to predict the critical potentials and the shape of the polarization curve by drawing a vertical line at the corresponding $x$. It also clearly shows a competition between active dealloying and oxide passivation in the entire composition range. For samples with $x > 0.95$, the active dealloying prevails against the oxide passivation because the Rh content in parent alloy is too low. For the samples with Cu content below $x = 0.95$, the oxide passivation becomes dominant. Figure 2b shows that $E_a$ increases while $E_p$ remains almost unchanged, and the gap between $E_a$ and $E_p$ shrinks with decreasing $x$ and eventually vanishes at around $x = 0.85$. The transpassive potential $E_t$ also increases with decreasing $x$, and rises steeply at around $x = 0.75$. For the samples with $x \leq 0.75$, dealloying did not occur at all potentials. From a phenomenological point of view, this threshold composition can be defined as a parting limit for the dealloying of Cu-Rh alloy under this circumstance. A major difference between the diagrams of Cu-Rh and Au-Ag is that the oxidation passivation occupies a large region for Cu-Rh but is absent for Au-Ag.$^{6,43}$ Furthermore, the parting limit of Cu-Rh is defined by the formation and stability of oxide passive layer, unlike the Au-Ag whose parting limit is controlled by the passivation of more-noble element (Au).

Figure 3. Potentiostatic dealloying of Cu$_{0.91}$Rh$_{0.09}$. a) Variation of current density with time at various potentials. Note that the current densities at 0.17 V and 0.19 V are negative. b) Variation of corroded thickness (for 5 hours) with potential. The anodic polarization curve of this alloy (1 mV/s) is also plotted for comparison.
In order to determine the “true” critical active dealloying potentials, we conducted a series of SEM examinations (see Fig. 5) on the surface structure of Cu0.91Rh0.09 polarized at different potentials for more than 5 hours. We observed NP structure on surface of samples treated at and above 0.19 V, but not on those treated at and below 0.17 V. This indicates that the true (active) critical dealloying potential for this composition is ~ 0.19 V, which is ~80 mV below the data determined by extrapolation of anodic polarization data (Ey = 0.27 V for Cu0.91Rh0.09). A similar discrepancy was also observed for dealloying of Au-Ag.50 It should be pointed out that the dealloying below Ey is extremely slow. Thus the data of Ey plotted in Fig. 2b, although they are slightly higher than the “true” values, are still meaningful and important in a practical sense.

**Compositional characterization of oxide passive layer.**—The increase of critical potential for transpassive dealloying with decreasing x indicates that the stability and thus the composition of the passivation layer may change with x (of CuRh1-x), and that the trend will be more significant for Au-Ag.50 However, such current transient was not observed in Cu-Rh alloys, probably because their open circuit potentials (OCP) are too close to the anodic dealloying potentials – the cathodic current density decreases with decreasing potential and changes sign at ~ 0.19 V.

![Figure 4](image-url) SEM images of cross section of Cu0.91Rh0.09 alloys after dealloying at a) 0.31 V, b) 0.36 V, c) 0.61 V, d) 0.71 V for 5 hours, respectively.

![Figure 5](image-url) SEM images of surface structure of Cu0.91Rh0.09 after dealloying at a) 0.17 V, b) 0.19 V, and c) 0.21 V for 5 hours, respectively. All samples were annealed at 500°C for 30 minutes after dealloying and before SEM experiments, in order to coarsen the NP structure so that it can be easily characterized by SEM. Scale bars: 500 nm.
reaches ∼77 at.%, which is sufficient to prevent further corrosion according to our results.

**Crystal structure of NP Rh.**—We examined the crystal structure of NP Rh prepared by dealloying Cu$_{0.8}$Rh$_{0.2}$ films (200 μm thickness) at 1.2 V for 30 hours, in the transpassive dealloying region. The residual Cu content in as-dealloyed NP Rh is approximately 10 at.%. NP samples were milled into powders for XRD experiments. Thus the XRD data was obtained from the entire dealloyed sample, rather than sample surface layer. Figure 8a shows the XRD patterns of an as-dealloyed NP Rh and the ones with further electrochemical treatments. By comparing these data with that of bulk FCC Rh, we found that the as-dealloyed NP Rh is composed of FCC Rh, and a considerable amount of oxide according to the two broad peaks at around 34° and 57°. From these data we are unable to identify the crystal structure of oxides. Nevertheless, HCP structure was absent in the as-dealloyed NP samples.

The Rh oxide in as-dealloyed NP Rh can be electrochemically reduced by polarizing at a low potential, such as −0.10 V in this study. The oxides on ligaments were so thick that at this potential, it took more than 10 hours to reach a completely reduced surface state. The XRD pattern of the reduced sample is displayed in Fig. 8a, which shows the FCC structure. The reduced sample was then re-oxidized by polarizing at 1.2V for 20 minutes so that the Rh at ligament surface was oxidized or absorbed with oxygen. XRD pattern reveals that the FCC structure was retained after re-oxidation.

We also prepared NP Rh sample by dealloying Fe$_{0.90}$Rh$_{0.10}$ alloy at a constant potential of 0.00 V, which leads to a NP Rh sample with a clean surface state and small structure size (∼4 nm). The XRD patterns displayed in Fig. 8b reveal that the Fe$_{0.90}$Rh$_{0.10}$ parent alloy exhibits a body-centered cubic (BCC) structure, while the dealloyed NP Rh...
Figure 8. XRD patterns of NP Rh samples fabricated by dealloying a) Cu$_{0.80}$Rh$_{0.20}$ at 1.20 V and b) Fe$_{0.90}$Rh$_{0.10}$ at 0.00 V. The XRD patterns of bulk Rh and parent alloys are plotted for comparison. Samples dealloyed from Cu-Rh are severely oxidized, which were then reduced and re-oxidized for testing. The NP Rh dealloyed from Fe-Rh has a clean surface state.

The microstructure of three types of NP Rh samples prepared by dealloying Cu-Rh alloy mentioned above were also examined by TEM, as displayed in Fig. 9. TEM images show that the ligament size of as-dealloyed sample is approximately 4 nm, which did not coarsen during subsequent electrochemical reduction and re-oxidation. The selected area electron diffraction (SAED) pattern shows an amorphous-like structure and the trace of FCC structure in as-dealloyed sample, and the FCC structure in reduced and re-oxidized samples. These observations are consistent with XRD results.

**Discussion**

**Compositional effect on critical potentials.** — As displayed in Fig. 2h, for Cu-Rh alloys (and other alloys that involve oxide passivation in corrosion), there are two potential regions that are suitable for rapid dealloying: the active dealloying in low potential region and the transpassive dealloying in high potential region. These regions are bounded by the critical potentials that vary with alloy composition. The lower boundary of transpassive dealloying, i.e., $E_a$, increases with decreasing $x$, because the surface passive layer formed on lower-$x$ alloys contains more Rh and is more stable against electrochemical “break-up”. On the other hand, the active dealloying potential gap shrinks, as $E_a$ increases while $E_p$ remains almost constant with decreasing $x$. The increase of $E_a$ with decreasing $x$ is similar to the variation of critical dealloying potential with Ag content in Au-Ag alloys, which has been studied previously.\(^3\) The composition independence of $E_p$ may result from the fact that, in this case, the oxide passivation is solely controlled by the oxidation of Rh. Cyclic voltammogram curves displayed in Fig. 10 indeed show that for both flat Rh and NP Rh, the OH adsorption or oxidation occurs above $\sim$0.40 V, which is identical to the $E_p$ determined for Cu-Rh alloys. The active dealloying potential window closes as the active dealloying potential surpasses the oxidation potential of Rh, while the Rh content in parent alloy exceeds a critical value (between 15 and 20 at.%).

Parting limit: intrinsic or extrinsic?.— Our study shows that the dealloying cannot proceed when the Cu content in Cu-Rh alloy is below 75 at.% - it can be defined as a parting limit of this alloy under this circumstance, which is much larger than that of the Au-Ag alloys. As we have also demonstrated in this study, the surface passivation of Rh oxide plays an important role in corrosion and may also have shifted the parting limit of this alloy. The formation and stability of oxide may be sensitive to the corrosion environment, such as the pH value of the electrolyte. It is thus possible that in a different electrolyte the dealloying of Cu-Rh alloys may show a different parting limit. The parting limit identified in this study for Cu-Rh may be an environment-sensitive extrinsic property, rather than an intrinsic parting limit that is solely controlled by the atomic structure of parent alloy and the passivation of more-noble element (Rh, not the oxide). We also anticipate that the corrosion of alloys containing passive noble metals (such as Pt) may show a similar behavior and a
similar high extrinsic parting limit. Their intrinsic parting limit may be obtained by dealloying in a non-aqueous solution where the oxides do not form, or in an environment where the oxides are unstable.

**Oxide passivation vs dealloying.**—The polarization curves of Cu-Rh alloys (see Fig. 1a) are similar to those of Fe-Cr alloys, which contain active, passive and transpassive corrosion regions, even though Rh oxides are more stable than Cr oxide and Rh is less reactive than Cr. For Fe-Cr alloys, a stable passive layer develops and the material becomes “stainless” as the Cr content exceeds 13 at.%, in some acidic solutions. Figure 1b shows that for Cu-Rh, the passive current density decreases dramatically to almost zero when x decreases from 0.90 to 0.85, which indicates that the threshold composition for passivity of Cu-Rh sits between 10–15 at.% Rh, which is close to that of Fe-Cr. The passivity of Fe-Cr has been interpreted by percolation models: the one proposed by Sieradzki et al. considers the percolation of third-neighboring Cr atoms in the underlying material where the separation between Cr atoms can be bridged by adsorbed oxygen anions; the other suggested by McCafferty considers the connectivity of oxide network on sample surface. These models may also apply to the passivity of Cu-Rh alloys.

It is noted that the extrinsic parting limit for Cu-Rh identified in this study is not defined as a threshold composition where the passive layer develops, but a critical Cu content below which the passive layer no longer “breaks down” at even high potentials. Such a “parting limit” does not exist for the Fe-Cr system, because even pure Cr can eventually be dissolved at sufficiently high potentials. From this point of view, the (extrinsic) parting limit for Cu-Rh is associated with the stability of the oxide passive layer. But how its value is related to the parent alloy composition is still an unsolved issue. Is it connected with the percolating Rh in the parent alloy? Or is it simply associated with the enrichment of Rh in surface layer and an accelerated growth kinetics of surface oxide layer? What is the role of metallic Rh in surface passivation? A more-detailed characterization of passive layer may provide new clues to these issues. Further studies are also required to clarify how the intrinsic and extrinsic parting limits are related.

**Crystal structure of Rh at nanoscale.**—Some materials undergo a phase transformation when their crystal sizes are shrunk to nanometer scale. Such a “size effect” may originate from a “surface effect”. At nanoscale, the contributions from surface (e.g., excess free energy, surface stress) to overall free energy may become dominant in determining their crystal structure. Our study shows that dealloyed NP Rh exhibits an FCC structure just like bulk Rh. It appears to contradict Huang’s observation of an HCP crystal structure in Rh nanoparticles. But it should be noted that NP Rh and Rh nanoparticles are different in many aspects: i) the ligament diameter of our NP Rh is ~4 nm, which is larger than the particle size (~2 nm) reported by Huang; ii) inevitably our NP Rh samples contain residual Cu (>5 at.%), while Huang’s nanoparticle are pure; iii) topologically, NP Rh may resemble a network of nanowires rather than an assembly of nanoparticles, and the mean surface curvature of nanoporous structure is also lower than that of similar-sized nanoparticles; iv) NP Rh contains defects such as dislocations and grain boundaries, while nanoparticles can easily be defect-free. Further studies are required to clarify whether and how these factors, particularly the characteristic size and the residual Cu, affect the crystal structure of NP Rh.

**Conclusions**

Electrochemical dealloying of Cu-Rh alloys was systematically investigated in 1 M H$_2$SO$_4$ aqueous solution. The corrosion of Cu-Rh under this circumstance is similar to that of Fe-Cr: polarization curves are characterized by active, passive, transpassive corrosion (or dealloying) regions. Unlike Fe-Cr alloys that eventually dissolve at high potentials, the Cu-Rh becomes “stainless” at even very high potentials when the Cu content is decreased to below 75 at.% Cu, which represents a “parting limit” of this alloy. This value is much higher than that of Au-Ag and other alloy systems, where the oxide passivation does not occur. The high parting limit is attributed to a passive layer of Rh oxide that hinders a dealloying. Therefore, the parting limit identified in this study is in fact an extrinsic property. The intrinsic parting limit may be obtained in future by dealloying under other environments where the passivity can be suppressed or avoided. NP Rh samples fabricated by dealloying exhibit an FCC structure, instead of an HCP structure reported recently for Rh nanoparticles. This inconsistency may result from the differences in their characteristic size, impurity, surface geometry and other defects.

**Acknowledgments**

Supported by the National Key R&D Program of China (grant No. 2017YFA0204401) and National Natural Science Foundation of China (grant No. 51571206). The authors thank R. Yu (Tsinghua Univ.) and B.(Bo) Zhang for stimulating discussions, and C.L. Dai, X.L. Wang, B. (Bin) Zhang, Y. Li’s Group for the assistance in experimental work.

**ORCID**

Feng Liu https://orcid.org/0000-0003-3935-4028
Hai-Jun Jin https://orcid.org/0000-0002-9445-2651

**References**

1. K. Sieradzki and R. C. Newman, “Stress-Corrosion Cracking,” *Journal of Physics and Chemistry of Solids*, 48, 1101 (1987).
2. K. Sieradzki and R. C. Newman, “Brittle Behavior of Ductile Metals During Stress-Corrosion Cracking,” *Philosophical Magazine a-Physics of Condensed Matter Structure Defects and Mechanical Properties*, 51, 95 (1985).
3. J. Erlebacher, M. J. Aziz, A. Karma, N. Dimitrov, and K. Sieradzki, “Evolution of nanoporosity in dealloying,” *Nature*, 410, 450 (2001).
4. J. McCue, E. Benn, B. Gaskey, and J. Erlebacher, Dealloying and Dealloyed Materials, in Annual Review of Materials Research, Vol 46, edited by D. R. Clarke (Annual Reviews, Palo Alto, 2016), pp. 263.
5. J. Weissmüller, R. C. Newman, H.-J. Jin, A. M. Kodge, and J. W. Kysar, “Nano-porous Metals by Alloy Corrosion, Formation and Mechanical Properties,” *MRS Bulletin*, 34, 577 (2009).
6. J. Weissmüller and K. Sieradzki, “Dealloyed nanoporous materials with interface-controlled behavior,” *MRS Bulletin*, 43, 14 (2018).
7. Q. Chen and K. Sieradzki, “Mechanisms and Morphology Evolution in Dealloying,” *Journal of the Electrochemical Society*, 160, C226 (2013).
8. A. J. Forty and P. Durkin, “A micro-morphological study of the dissolution of silver-gold alloys in nitric-acid,” *Philosophical Magazine a-Physics of Condensed Matter Structure Defects and Mechanical Properties*, 42, 295 (1980).
