Postprint: Formation and Corrosion Resistance of U-Co System Amorphous Alloys

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Abstract

U-rich alloys UxCo100-x (x=50~87.5) in the U-Co binary system were investigated. Master alloy ingots and ribbon samples were prepared by arc melting and melt spinning. The phase composition and stability of the alloy samples were studied by XRD and DSC, and the corrosion resistance of the amorphous alloys was evaluated using potentiodynamic polarization. The results indicate that the composition range for ribbon amorphous formation is 58.5≤x$ 78, $with the optimal composition appearing near U66.7Co33.3. At a heating rate of 20K/min, the crystalli$ $Co amorphous alloys ranges from 534 to 550K, the crystallization enthalpy is 4.8 8.5kJ/mol, and the reduced cryst$ 6 Cl- solution, the corrosion potential of U-Co amorphous alloys is mostly close to -50 mV, their corrosion resistance is significantly superior to that of depleted uranium, and a positive correlation exists between the corrosion resistance and glass-forming ability of U-Co amorphous alloys.

Full Text

U-Co System Amorphous Alloy Formation and Corrosion Resistance Study*

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Abstract
The U-rich alloys UxCo100-x (x=50~87.5) in the U-Co binary system were selected as research subjects. Master alloy ingots and ribbon samples were prepared by arc melting and melt-spinning methods. The phase composition

and stability of the alloy samples were investigated by XRD and DSC, while the corrosion resistance of the amorphous alloys was studied using potentiodynamic polarization methods. The results indicate that the composition range for amorphous phase formation in ribbons is 58.5–78 at% U, with the optimal composition appearing near U66.7Co33.3. At a heating rate of 20 K/min, the crystallization temperature of U-Co amorphous alloys lies between 534–550 K, the crystallization exothermic enthalpy is 4.8–8.5 kJ/mol, and the maximum reduced crystallization temperature (Trx) reaches 0.535. In a 50$×10^{-6}$ Cl^- solution, the corrosion potentials of U-Co amorphous alloys are mostly close to -50 mV, their corrosion resistance is significantly superior to that of depleted uranium, and a positive correlation exists between the corrosion resistance and amorphous forming ability of U-Co alloys.

Keywords amorphous alloy, uranium alloy, rapid solidification, corrosion resistance

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Uranium alloys have important applications in the nuclear field. However, the structure of current uranium alloys is crystalline, and corrosion is one of the key factors limiting their engineering applications [1~3]. Amorphous alloys [4~13] possess characteristics such as single-phase structure, compositional uniformity, and absence of dislocations and grain boundary defects, thus typically exhibiting good corrosion resistance. Developing new uranium-based amorphous alloys with good corrosion resistance promises to expand the applications of uranium alloys.

Uranium alloys have high chemical reactivity, radioactivity, and certain chemical toxicity, making alloy preparation processes complex and difficult. Moreover, phase diagram and thermodynamic data for uranium alloys are very scarce, which restricts systematic research on the formation rules, thermal stability, and properties of uranium-based amorphous alloys. In 1969, Bethune [14] obtained amorphous phases in the U-Si system using irradiation methods. Subsequently, Giessen and Elliott et al. [15~17] prepared and studied amorphous U-M (M=Fe, Mn, Co, Ni, Cr, V, Si, Os, Ir, Pd) systems using the Duwez gun and hammer-anvil methods. Reference [16] reported preliminary results on the thermal stability of U-M amorphous alloys but did not provide specific differential scanning calorimetry (DSC) curves. In particular, the X-ray diffraction (XRD) pattern of U67Fe33 presented in that work contained three diffuse diffraction peaks of varying intensities. However, the XRD pattern of rapidly quenched U85.7Fe14.3 reported in reference [17] showed very sharp first and second strongest diffraction peaks. Comparison of DSC data for three types of samples at this composition —irradiated, melt-quenched plus irradiated, and melt-quenched—revealed that the crystallization exothermic enthalpy of the melt-quenched sample was significantly smaller than the former two. Without microstructural observation evidence, whether the melt-quenched U85.7Fe14.3 sample possessed a purely amorphous structure remains questionable. It should be noted that the melt-quenched samples prepared in reference [16] were all fragmentary with very

non-uniform quality, affecting subsequent structural and property characterization. In 1985, Drehman and Poon et al. [18,19] first prepared amorphous ribbons of U-Co, U-Fe, and U-Ni systems using the melt-spinning method and provided XRD patterns for U85.7Co14.3 samples and DSC curves for U72Co28. The XRD pattern of U85.7Co14.3 also contained three diffuse diffraction peaks, while the DSC curve of U72Co28 showed sharp crystallization exothermic peaks in the 275–285 K temperature range. In 1987, Nastasi and Parkin [20] obtained rapidly quenched U6Fe alloys using the hammer-anvil method and studied the effect of amorphous phase content on crystallization kinetics, finding that the crystallization temperature increased with increasing amorphous phase content. In summary, existing research on uranium-based amorphous alloys is very limited, particularly lacking systematic studies on amorphous formation rules, thermal stability, and corrosion resistance for specific systems.

This study selected the U-Co system and systematically investigated the composition rules, thermal stability, and corrosion resistance of amorphous alloys by preparing ribbon samples across a wide composition range of UxCo100-x (x=50.0~87.5, atomic fraction, hereinafter the same) using rapid quenching melt-spinning technology, combined with XRD, DSC, and electrochemical analysis techniques.

1 Experimental Methods

The composition range of alloys in this study was UxCo100-x (x=50~87.5). Based on the characteristics of the U-Co binary phase diagram [21], four compositions—U87.5Co12.5, U85.7Co14.3, U83.3Co16.7, and U78Co22—were selected near the U-rich U6Co phase; six compositions—U58.5Co41.5, U62.5Co37.5, U65Co35, U66.7Co33.3, U69.2Co30.8, and U72Co28—were chosen near the deep eutectic point U62.5Co37.5; and two compositions—U55Co45 and U50Co50—were selected near the UCo phase. The experiment used low-carbon depleted uranium with purity of approximately 99.5% (main impurities being U-C compounds) and Co with purity >99.9%. After 配料 according to the respective compositions, master alloy ingots of about 10 g were prepared using an NMS-II type small metastable alloy preparation furnace under high-purity Ar gas protection (purity >99.999%). To ensure compositional homogeneity, the master ingots were remelted four times. The master alloys were crushed for rapid quenching melt-spinning. The Cu roller (diameter 260 mm, width 50 mm, rotation frequency 0–100 Hz) linear velocity was selected as 50 m/s, Cu roller water cooling flow rate was about 150 mL/s, and the top-blown gas pressure difference was approximately 0.04 MPa. The U-Co amorphous ribbons obtained in this experiment were about 1 mm wide and 20 mm thick, continuous and uniform, straight without burrs, with bright surfaces, and it was found during experiments that ribbon samples containing amorphous phases could be bent 180° without fracturing. [Figure 1: see original paper] shows photographs of the amorphous ribbon samples, including images bent into V and W shapes.

A PANalytical EMPYREAN X-ray diffractometer (Cu Kα) was used to analyze

the phase composition of U-Co alloy master ingots and ribbons, with a diffraction angle 2 scanning range of 20°–100°. A NETZSCH STA-409CD DSC was employed to test the thermal stability of ribbon alloys, with a heating temperature range from room temperature to 1273 K and a heating rate of 20 K/min. Potentiodynamic polarization was used to evaluate the corrosion resistance of U-Co amorphous alloys. The potentiodynamic polarization tests were conducted on a PARSTAT 2263 electrochemical workstation, with amorphous alloys as the working electrode, saturated calomel electrode as the reference electrode, and Pt wire as the auxiliary electrode. The corrosion medium was a 50$×10^{-6}$ Cl^- solution, and the test temperature was room temperature. The potentiodynamic polarization scan rate was 1 mV/s, and the polarization range was -250 mV (relative to open circuit potential) to the pitting potential. If no pitting potential appeared, the test was stopped at 300 mV (relative to open circuit potential).

2 Results and Discussion

2.1 Phase Composition

2.1.1 Phase Composition of Master Alloy Samples [Figure 2: see original paper] presents the XRD patterns of UxCo100-x (x=50~87.5) master alloys. As shown, alloys with x=50, 55, 58.5, 62.5, 65, 66.7, 69.2, 72, 78, and 83.3 are all two-phase mixtures consisting of the body-centered tetragonal (bct) U6Co phase and the bcc UCo phase. These two phases have many atoms per unit cell, with complex atomic occupancy in the U6Co phase. lists the measured lattice parameters of each phase in different samples. The U6Co phase has a=1.0266–1.0483 nm, c=0.5123–0.5276 nm, while the UCo phase has a=0.6317–0.6428 nm, which are close to the lattice parameters reported in literature [22] for U6Co phase (a=1.03600 nm, c=0.52100 nm) and UCo phase (a=0.63557 nm). As can be seen from [Figure 2: see original paper] and , with increasing U content, the diffraction peak positions and lattice parameters of U6Co and UCo phases show no systematic variation trend. This indicates that in the slowly solidified master alloy ingots, the actual compositions of both U6Co and UCo phases deviate from their strict stoichiometric ratios. Furthermore, the relative intensities of diffraction peaks for these alloy phases are not completely consistent with standard powder patterns, reflecting that their grains possess a certain degree of crystallographic preferred orientation. For example, in the as-cast alloy with x=62.5, the (200) crystal plane shows strong preferred orientation. For alloys with x=85.7 and 87.5, the diffraction peaks of the UCo phase basically disappear in the XRD patterns, leaving mainly the U6Co phase, whose lattice parameters are also given in . In summary, the phase composition of U-Co master alloys formed by slow cooling in Cu crucibles is basically consistent with that indicated by the thermodynamic equilibrium phase diagram at each composition, but the U6Co and UCo phases exhibit certain compositional metastability.

2.1.2 Phase Composition of Ribbon Samples [Figure 3: see original

paper] shows the XRD patterns of rapidly quenched UxCo100-x (x=50~87.5) ribbons. Near-single-phase UCo structures formed at x=50 and 55 compositions, while amorphous phases, U6Co, and UCo coexisted in x=58.5 and 62.5 alloys, with amorphous diffuse peaks located between 35°-45°, indicating poor amorphous forming ability for the eutectic composition U62.5Co37.5. It should be noted that the UCo phase shows strongest diffraction from the (220) crystal plane in alloy ribbons with x=50 and 55, whereas only (200) diffraction peaks appear in ribbons with x=58.5 and 62.5, showing obvious preferred orientation. In samples with x=65, 66.7, 69.2, 72, and 78, the amorphous phase is dominant, coexisting with small amounts of UCo phase. Based on the symmetry and smoothness of the main diffuse peak of the amorphous phase in the XRD patterns, it can be preliminarily judged that alloy ribbons at x=65, 66.7, and 69.2 have higher amorphous content and greater amorphous forming ability. By calculating the peak areas of UCo and amorphous phases, the UCo phase contents in these three alloys can be semi-quantitatively determined as 9.5%, 4.8%, and 14.5%, respectively, further indicating that U66.7Co33.3 alloy has greater amorphous forming ability. In contrast, ribbon alloys at x=83.3, 85.7, and 87.5 contain basically no amorphous phase, forming supersaturated solid solution phases isostructural with orthorhombic α-U, with lattice parameters: for x=83.3, a=0.2860 nm, b=0.5840 nm, c=0.5010 nm; for x=85.7, a=0.2858 nm, b=0.5860 nm, c=0.4960 nm; for x=87.5, a=0.2856 nm, b=0.5832 nm, c=0.4961 nm. These lattice parameters are similar to those of pure α-U [23] (a=0.2854 nm, b=0.5870 nm, c=0.4955 nm). Thus, rapid quenching at x=83.3, 85.7, and 87.5 compositions causes significant compositional metastability. For U85.7Co14.3 alloy, Drehman and Poon [18] reported its ability to form amorphous phases and presented XRD patterns (Mo Kα, λ=0.07093 nm) showing three diffuse peaks, with the first and second diffuse peaks located near 2 =17.5° and 27.5°, respectively, exhibiting relatively high diffraction intensities. When converted to Cu Kα radiation (λ=0.15406 nm), the amorphous diffuse peaks are located around 2 =37° and 62° (the diffraction peak near 2 =28° is attributed to the UCo phase), which is close to the amorphous diffraction peak positions for alloys with x=65, 66.7, 69.2, 72, and 78 in this experiment ([Figure 3: see original paper]). Since the melt-spinning rate used in reference [18] was between 50-75 m/s, to further compare experimental results, this work also prepared U85.7Co14.3 alloy ribbons at a melt-spinning rate of 75 m/s, but no amorphous phase was obtained, with the α-U solid solution phase being the main product.

The above experimental results indicate that under slow cooling solidification conditions, near-single-phase U6Co formed near x=85.7 and 87.5 compositions, while compositional metastable U6Co and UCo phases were obtained at other compositions. Under melt-spinning conditions at 50 m/s, significant compositional metastability occurred at high U contents (x$ $83.3), forming supersaturated α-U phases; at low U contents (x=50-55), UCo phases formed; and amorphous phases could form in the range of x=58.5-78, with x=62.5 alloy showing poor amorphous forming ability. The amorphous forming ability of the U-Co binary system is asymmetric around the eutectic point, with greater amorphous

forming ability at high U content compositions near the right side of the eutectic point in the phase diagram, roughly in the composition range of x=66.7–69.2. The determined amorphous composition range for the U-Co binary system is similar to that reported by Giessen et al. [15] (U60Co40–U80Co20) and also falls within the composition range of U40Co60–U90Co10 reported by Drehman and Poon [18]. Meanwhile, Giessen et al. [15] pointed out that in U-Co amorphous alloys, small amounts of U6Co and another major unknown phase coexist with the amorphous phase, whereas in this experiment the coexisting phases were U6Co (in U58.5Co41.5 and U62.5Co37.5) and UCo phases, with the latter being dominant.

2.2 Thermal Stability

[Figure 4: see original paper]a shows the DSC curves of UxCo100-x (x=62.5~78) ribbons at a heating rate of 20 K/min. No obvious glass transition temperature (Tg) features were observed. The x=78 alloy sample exhibits typical single-peak crystallization behavior, while the others show double-peak crystallization. lists the initial dynamic crystallization temperature parameters (Tx) for these alloys, which range from 534–550 K. With increasing U content, Tx first increases then decreases, reaching a maximum value of 550 K at x=66.7. Giessen et al. [15] measured the thermal stability of amorphous samples between U60Co40–U80Co20 at a heating rate of 40 K/min, finding that their crystallization process contained 2–3 crystallization peaks with Tx in the 545–555 K range. Drehman and Poon [18] reported Tx values of 509–534 K for amorphous alloys between U42Co58–U90Co10 at a heating rate of 10 K/min. For the same sample, higher heating rates yield higher measured dynamic crystallization temperatures, and clearly the Tx data from this experiment and literature reports reflect this trend. On the other hand, the crystallization exothermic enthalpies (ΔHc) of these ribbon samples range from 4.8–8.5 kJ/mol, while Drehman and Poon [18] reported ΔHc values of 4–9 kJ/mol for amorphous alloys in the U42Co58–U90Co10 composition range, and Elliott et al. [17] reported ΔHc of 4.2 kJ/mol for U85.7Fe14.3 amorphous alloy. At the same heating rate (20 K/min), bulk amorphous alloys Zr55Cu30Al10Ni5 and Ca65Mg15Zn20 have ΔHc values of 4.2 kJ/mol [24] and 1.7 kJ/mol [25], respectively, while Zr41.2Ti13.8Cu12.5Ni10.0Be22.5 amorphous alloy and Pd43Ni10Cu27P20 amorphous alloy have ΔHc values of 0.34 kJ/mol [26] and 0.3 kJ/mol [27], respectively. Clearly, the exothermic heat of these conventional amorphous alloys is significantly lower than that of U-Co amorphous alloys, which exhibit stronger thermal effects during crystallization.

From the melting curves shown in [Figure 4: see original paper]b, three alloys (x=62.5, 72, and 78) show double-peak melting behavior, but the second melting peak is very weak. In contrast, x=65, 66.7, and 69.2 alloys show near-single-peak melting, meaning that during heating and crystallization of these ribbons, the supercooled liquids of these three alloys deviating from the equilibrium eutectic composition formed eutectic-like structures. On the other hand, it also reflects that when more UCo or U6Co crystalline phases exist in amorphous ribbons,

the melting behavior becomes more complex. lists the melting temperature (Tm) and liquidus temperature (TL) parameters for these alloys. Their Tm values are basically consistent, located near 1010 K, close to the corresponding eutectic temperature of 1007 K in the equilibrium phase diagram. The TL values of amorphous samples with x=65, 66.7, and 69.2 are similar, ranging from 1026–1028 K, TL for x=62.5 is 1057 K, while TL for x=78 composition is 1099 K, approaching the equilibrium melting temperature of U6Co (1099 K).

To evaluate the amorphous forming ability of alloys, Turnbull [28] suggested using the reduced glass temperature Trg=Tg/TL as a criterion. Since no Tg was observed in the DSC curves of U-Co amorphous alloys, this work adopted the reduced crystallization temperature Trx=Tx/TL as the evaluation parameter for amorphous forming ability, following the treatment in references [15,18]. The Trx values for the x=62.5–78 composition range are 0.486–0.535, first increasing then decreasing with increasing U content. The lowest value corresponds to the x=78 alloy, while the highest value corresponds to the x=66.7 alloy (). Combined with XRD results, it can be preliminarily concluded that the x=66.7 alloy in the U-Co system has the strongest amorphous forming ability. In the report by Giessen et al. [15], the Tx/Tm values for alloys between U60Co40–U80Co20 ranged from 0.54–0.55, with no obvious trend. In contrast, Drehman and Poon [18] reported Tx/TL values of 0.38–0.51 for alloys between U42Co58–U90Co10, first increasing then decreasing with U content, with the maximum value corresponding to U72Co28 alloy. Significant differences exist between literature reports and this experimental results.

2.3 Corrosion Resistance

[Figure 5: see original paper] shows the potentiodynamic polarization curves of UxCo100-x (x=62.5~78) amorphous alloys and depleted uranium (DU) in a 50$×10^{-6}$ Cl^- solution. Except for the x=78 alloy whose corrosion potential (Ecorr) is below -100 mV, the Ecorr values of other alloys are all close to -50 mV, significantly higher than that of DU, and much higher than that of Al-plated uranium metal (below -500 mV) [2]. Thus, U-Co amorphous alloys possess strong corrosion resistance. The Ecorr and pitting potential (Epit) of three U-Co alloys (x=62.5, 65, and 69.2) are relatively close, and all exhibit "pseudo-passivation" characteristics, indicating that protective oxide films formed on the alloy surfaces. Although the x=72 alloy has relatively high Ecorr, it shows pitting characteristics when the polarization potential is slightly higher than Ecorr. The Ecorr of x=78 alloy is significantly lower than other alloys, and the anodic polarization part only shows active dissolution characteristics. From the potentiodynamic polarization curves, among these U-Co amorphous alloys, x=62.5, 65, and 69.2 alloys have relatively good corrosion resistance, x=72 alloy is second, and x=78 alloy is the worst.

Combined with the previous thermal stability data, there exists a positive correlation between the corrosion resistance and amorphous forming ability of U-Co amorphous alloy series, which also implies that the presence of crystalline phases

affects the corrosion resistance of amorphous alloys.

Discussion

The formation of U-Co amorphous alloys in this experiment is discussed from three aspects: thermodynamic metastable equilibrium, crystallization kinetics, and impurity effects. First, the amorphous composition characteristics of the U-Co system deviate from the eutectic principle proposed by Cohen and Turnbull [29]. Chen [30] summarized the amorphous composition characteristics of several typical binary systems including Pd-Si, Zr-Ni, Ca-Al, Zr-Cu, and Hf-Be in an early review, finding that compositions with high amorphous forming ability in these systems all deviate from the eutectic points in equilibrium phase diagrams. The Zr-Cu and Ca-Al systems are biased toward the high-melting-point phase side, Pd-Si and Zr-Ni systems toward the low-melting-point phase side, while the Hf-Be system is intermediate. In the Zr-Cu system, experiments by Wang et al. [31] confirmed that within the Cu51Zr14–Cu10Zr7 composition range, compositions with higher amorphous forming ability are also biased toward the high-melting-point phase (Cu51Zr14) side. Based on Cahn' s thermodynamic metastable equilibrium theory [32], Dubey and Ramachandrarao [33] studied numerous eutectic phase diagrams and first theoretically predicted the asymmetry of amorphous forming ability around eutectic points. Later, Highmore and Greer [34] proposed that near eutectic compositions in metastable phase diagrams correspond to large amorphous forming ability. Based on the suggestion of Highmore and Greer [34], and noting the single-peak melting characteristic of alloys near the U66.7Co33.3 composition, it is speculated that this composition falls within the metastable eutectic region bounded by dashed lines in [Figure 6: see original paper]. Obviously, the liquidus line near the metastable eutectic point extends to lower temperatures, which can result in higher Trx values, consistent with the Trx data trends in .

Furthermore, considering the crystal structures of U6Co and UCo phases, the former has complex atomic occupancy with significant chemical and topological short-range order (atomic clusters), while the latter has significant chemical disorder in specific atomic occupancy, and its bcc crystal structure is relatively simple with lower packing density [22]. Alloy thermodynamics studies [35,36] show that in a given system, the formation enthalpy of intermetallic compounds typically displays negative extreme values. Therefore, in amorphous alloy research, investigators [37~40] believe that stable atomic clusters exist in melts near intermetallic compound compositions. The presence of these atomic clusters will affect the viscous flow of undercooled liquids. From a purely kinetic viewpoint, amorphous solids are treated as configurationally frozen liquids, where Tg corresponds to a shear viscosity of 10^{13} Pa · s. Then, for alloys on the U-rich side of the eutectic point U62.5Co37.5 in the U-Co equilibrium phase diagram, atomic clusters with local structures similar to the U6Co phase may exist in their undercooled liquids. These clusters will increase the resistance to viscous flow in undercooled liquids compared to completely disordered undercooled liquids.

Consequently, only by increasing temperature under higher thermal activation conditions can the viscosity of undercooled liquids be reduced, meaning that Tg for U-Co alloys near the U-rich side of U62.5Co37.5 composition will gradually increase. Meanwhile, the presence of these short-range orders will also lower the liquidus temperature to some extent [37~39]. In summary, the undercooled liquids of U-rich alloys near the right side of U62.5Co37.5 are relatively favorable for amorphous formation from both metastable thermodynamic and kinetic perspectives.

Rapid quenching experiments on U-Co alloy melts show that: at high U contents (x$ $83.3), supersaturated α-U phases are the competing phases for amorphous formation; at low U contents (x=50–55), UCo phases are the competing phases, and the phase compositions at these compositions are basically single-phase. Since this is near congruent crystallization, the compositional conditions for nucleation of such crystalline phases are very easily satisfied and difficult to control in rapid quenching kinetics. In the U content range of 62.5%–78%, the competing phases for amorphous formation are UCo and U6Co phases. Since the compositions of these two phases deviate significantly from the alloy nominal composition, their nucleation and growth require long-range diffusion of constituent atoms. Therefore, the crystallization kinetics of alloys in this composition range are relatively difficult, thus favoring amorphous formation. Comparing the two competing phases UCo and U6Co, although the latter has more favorable compositional conditions for nucleation, its crystal structure is extremely complex, and atomic rearrangement kinetics during nucleation and growth are relatively slow. On the other hand, since the alloy composition is near the U6Co side, dense atomic clusters related to this phase easily form in the alloy, expelling more excess volume, which may in turn favor nucleation of the UCo phase with simpler structure and lower packing density. Qualitative analysis indicates that the UCo phase nucleates relatively easily and competes with amorphous formation, which is consistent with the phase identification results in this experiment.

Furthermore, since the U raw material contains large amounts of U-C compound impurities, the effect of heterogeneous nucleation on amorphous formation must be considered. Due to the presence of such impurities, the UCo phase competing with amorphous phases nucleates extremely easily. Therefore, suppressing its crystal growth becomes key to improving the amorphous forming ability of U-Co alloys. To this end, it is necessary to maximize the cooling rate of U-Co alloy melts. In view of this, rapid quenching experiments were conducted on x=66.7 U-Co alloys using Cu roller linear velocities higher than 80 m/s. However, the precipitation of UCo phase still could not be suppressed experimentally, indicating that the elemental diffusion kinetics required for crystal growth in U-Co alloys are extremely rapid. Based on research results on elemental diffusion in amorphous alloys [41], it is speculated that this is likely related to the decoupled rapid diffusion behavior of small Co atoms in undercooled melts.

To improve amorphous forming ability, multicomponent alloying was performed on the U-Co system. By adding appropriate elements to destabilize the UCo

phase and increase atomic diffusion difficulty, single-phase amorphous alloys were successfully prepared using low-purity U raw materials. Related research results will be reported in subsequent work.

Conclusions

(1) In the U-Co system, the phase composition of slowly cooled UxCo100-x (x=50~87.5) alloy ingots is basically consistent with the equilibrium phase diagram.

(2) During rapid solidification, U-Co alloys obtained obvious compositional and structural metastability. UCo phases formed in the x=50-55 range; amorphous phases formed in the x=58.5-78 range, with the x=62.5 alloy showing poor amorphous forming ability.

(3) U-rich alloys at the eutectic point U62.5Co37.5 have greater amorphous forming ability, with optimal amorphous ability appearing at the U66.7Co33.3 composition.

(4) The corrosion potential of U-Co amorphous alloys is about -50 mV, their corrosion resistance is significantly superior to depleted uranium, and a positive correlation exists between alloy corrosion resistance and amorphous forming ability.

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Note: Figure translations are in progress. See original paper for figures.