Interdiffusion Between Zr Diffusion Barrier and U-Mo Alloy

K. Huang, Y. Park, D.D. Keiser Jr., and Y.H. Sohn

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U-Mo alloys are being developed as low-enrichment uranium fuels under the Reduced Enrichment for Research and Test Reactor (RERTR) program. Significant reactions have been observed between U-Mo fuels and Al or Al alloy matrix. Refractory metal Zr has been proposed as barrier material to reduce the interactions. In order to investigate the compatibility and barrier effects between U-Mo alloy and Zr, solid-to-solid U-10wt.%Mo versus Zr diffusion couples were assembled and annealed at 600, 700, 800, 900, and 1000 °C for various times. The microstructures and concentration profiles due to interdiffusion and reactions were examined via scanning electron microscopy and electron probe microanalysis, respectively. Intermetallic phase Mo_2Zr was found at the interface, and its population increased when annealing temperature decreased. Diffusion paths were also plotted on the U-Mo-Zr ternary phase diagrams with good consistency. The growth rate of interdiffusion zone between U-10wt.%Mo and Zr was also calculated under the assumption of parabolic diffusion and was determined to be about 10^3 times lower than the growth rate of diffusional interaction layer found in diffusion couples U-10wt.%Mo versus Al or Al-Si alloy. Other desirable physical properties of Zr as barrier material, such as neutron adsorption rate, melting point, and thermal conductivity, are presented as supplementary information to demonstrate the great potential of Zr as the diffusion barrier for U-Mo fuel systems in RERTR.

Keywords diffusion barrier, diffusion couples, interdiffusion, metallic alloys, multicomponent diffusion, phase equilibria

1. Introduction

Uranium-molybdenum (U-Mo) dispersion and monolithic fuels in aluminum (Al) alloys are being developed as a low-enrichment metallic fuel system under the Reduced Enrichment for Research and Test Reactor (RERTR) program.^[1-3] Metallurgical interactions occur between the U-Mo fuels and Al alloy matrix during processing and irradiation due to interdiffusion.^[4-6] Efforts^[7-15] have been made to document complex interactions and deleterious effects such as thinning of the cladding layer, formation of phases with poor irradiation behavior and relative low melting point, and thermal cracks.

Refractory metal Zr has been proposed as barrier layer between the U-Mo fuel and Al alloy matrix to reduce the interactions, since the diffusion of U in Zr is slow,^[16-18] and the fabrication of Zr barrier is compatible with the current hot-rolling techniques used by Idaho National Laboratory

(INL).^[6] Furthermore, interaction products that form between U-Mo fuel and Zr appear to be stable during irradiation.^[19] In an as-processed U-10wt.%Mo monolithic fuel plate in AA6061 matrix with Zr diffusion barrier,^[6] multiple phases including αU , Mo_2Zr , Zr solid solution, γUZr , and UZr_2 were observed at the interface of U-Mo with Zr.

This study systematically investigated the interdiffusion and reaction between U-Mo alloy with Zr, without the influence of multiple and complex hot-rolling process on the interfacial microstructure and phase constituents. Solid-to-solid isothermal diffusion method was used from 600 to 1000 °C. The microstructure and concentration profiles of interdiffusion zone were examined by scanning electron microscopy (SEM) and electron probe microanalysis (EPMA), respectively. Diffusion paths at each temperature were determined and superimposed on the isothermal U-Mo-Zr ternary phase diagrams. The growth rate of interdiffusion zone was also calculated and compared to those of U-Mo versus Al and U-Mo versus Al-Si diffusion couples reported in the literature.^[9,20] Other desirable physical properties of Zr as barrier material are also presented.

2. Experimental Procedure

All metallographical preparation and assembly of diffusion couples were carried out under an argon (Ar) atmosphere inside a glove box to minimize oxidation of the U-Mo alloy and Zr. U-Mo rods were cast using high-purity depleted uranium (DU) and Mo via arc melting. They were melted three times to ensure homogeneity and then drop-cast to form

K. Huang, Y. Park, and Y.H. Sohn, Department of Mechanical, Materials and Aerospace Engineering, Advanced Materials Processing and Analysis Center, University of Central Florida, Orlando, FL 32816, USA; and D.D. Keiser Jr., Nuclear Fuels and Materials Division, Idaho National Laboratory, Box 1625 Idaho Falls, ID 83415, USA. Contact e-mail: Yongho.Sohn@ucf.edu.

Table 1 Temperature and time of diffusion anneal for U-10wt.%Mo vs. Zr diffusion couples

Temperature, °C	Time, h
1000	96
900	240
800	480
700	720
600	960

rods with 12.7 mm diameter. Zirconium rods of 99.99% purity (12.7 mm diameter) were acquired from a commercial source (Alpha-Aesar). The homogeneity in composition, phase constituents, and microstructure of the U-Mo and Zr were examined by x-ray diffraction (XRD; Rigaku DMAX-B) and scanning electron microscopy (SEM; Hitachi 3500N) equipped with x-ray energy dispersive spectroscopy (XEDS).

Both the U-Mo alloy and Zr rods were sectioned into 3 mm thick disks, and the surfaces were metallographically polished down to 1 μ m using diamond paste. The prepared surfaces were then placed in contact with each other and held together by two clamping disks with stainless steel rods to form a jig. The jig assembly was wrapped in Ta foil, encapsulated in a quartz capsule, and sealed under an Ar atmosphere after repeated vacuum (10^{-6} Pa) and H₂ purge. The couples were annealed using a Lindberg/Blue three-zone tube furnace at the temperatures listed in Table 1. After annealing, the diffusion couples were quenched by breaking the quartz capsule in cold water. Each diffusion couple was then mounted in epoxy, cross-sectioned, and polished using 1 μ m diamond paste for microstructural examination and compositional analysis.

For each diffusion couple, SEM was performed to examine the interdiffusion microstructure. Electron probe microanalysis (JEOL Superprobe 733) was used to obtain the concentration profiles of U, Mo, and Zr from each couple using a point-to-point scan with a 5-10 μ m step size and an accelerating voltage of 20 keV. The pure metals, DU, Zr, and Mo were used as standards. The intensities of the U-M α , Zr-L α , and Mo-L α x-rays were converted to compositions via ZAF correction.

3. Results and Discussions

3.1 Composition of U-Mo Alloy

The composition of U-Mo alloy was examined by EPMA at 10 random locations. Its average value and standard deviation were determined as reported in Table 2. The averaged composition, 20.91 at.% Mo is very close to the target composition of U-10wt.%Mo fuel, which is 21.61 at.% Mo. The standard deviation of the 10 measured points was 0.15 at.%.

3.2 Microstructure, Concentration Profiles, and Diffusion Paths

Backscattered electron (BSE) micrographs and concentration profiles from the U-Mo versus Zr diffusion couples

Table 2 Composition of the starting U-Mo alloy determined by electron probe microanalysis

Point No.	Composition of Mo, at. %
1	20.82
2	21.22
3	20.89
4	20.71
5	20.74
6	20.85
7	21.01
8	20.85
9	20.96
10	21.03
Average composition	20.91
Standard deviation	0.15

Fig. 1 Backscattered electron micrograph and concentration profiles from U-Mo vs. Zr diffusion couple annealed at 1000 °C for 96 h

annealed at 1000, 900, 800, and 700 °C are presented in Fig. 1 to 4. Figure 5 shows the interfacial microstructure developed at 600 °C where the couple had a negligible interaction. The large dark precipitates near the interface in Fig. 1 to 4 were identified as Mo₂Zr using semiquantitative composition analysis by XEDS and the Mo-Zr equilibrium binary phase diagram presented in Fig. 6.^[21] When the diffusion anneal temperature decreased from 1000 to 700 °C, the population of the Mo₂Zr precipitates increased (i.e., an increase in volume fraction of Mo₂Zr). The U-Mo versus Zr ternary diffusion couples examined in this study have the extra degree of freedom based on Gibbs phase rule that allows the development of two-phase layer that includes Mo₂Zr phase. Small dark precipitates present throughout the U-Mo alloy are identified as Zr-rich phase, which is composed of > 95 at.% Zr and a few at.% of U and Mo. This tiny Zr-rich precipitate should be the α Zr, formed during quenching. The U-Zr solid solution eutectoidally decomposed into α Zr and Mo₂Zr upon cooling according to the U-Zr phase diagram shown in Fig. 7.^[21] Unfortunately,

Fig. 2 Backscattered electron micrograph and concentration profiles from U-Mo vs. Zr diffusion couple annealed at 900 °C for 240 h

Fig. 3 Backscattered electron micrograph and concentration profiles from U-Mo vs. Zr diffusion couple annealed at 800 °C for 480 h

the α Zr precipitates are too small in size, and the eutectoid structure cannot be observed clearly.

Concentration profiles determined by EPMA show an uphill diffusion of U on the U-Mo side. Scatter in concentration profiles is observed near the interface because of the Mo_2Zr precipitates. When diffusion anneal temperature decreased to 600 °C, U-Mo and Zr diffusion couple exhibited a negligible interaction despite the 960 h of diffusion anneal. A very small intermediate phase with thickness of about 1-2 μm was observed discretely at the interface as seen in Fig. 5. The composition of these intermediate phases has not been determined accurately since the size is too small.

There is an allotropic transformation between α Zr and β Zr at 863 °C according to U-Zr phase diagram shown in Fig. 7.^[21] Therefore, both α Zr and β Zr exist on the Zr side at 800 and 700 °C as shown in Fig. 3 and 4, while only β Zr exists at 1000 and 900 °C as shown in Fig. 1 and 2. A

Fig. 4 Backscattered electron micrograph and concentration profiles from U-Mo vs. Zr diffusion couple annealed at 700 °C for 720 h

Fig. 5 Backscattered electron micrograph from U-Mo vs. Zr diffusion couple annealed at 600 °C for 960 h

marker plane was clearly observed in diffusion couple annealed at 1000 °C as marked in Fig. 1. However, unfortunately, markers could not be identified in other diffusion couples.

U-Mo-Zr ternary phase diagrams have been determined by Ivanov and Bagrov for isothermal temperature of 500, 575, 600, 625, 650, 675, 700, 750, and 1000 °C.^[22,23] The concentration profiles of diffusion couples were plotted as diffusion paths on U-Mo-Zr ternary phase diagrams shown in Fig. 8. Since there is no available ternary phase diagram at 900 and 800 °C, the diffusion paths are presented on the isotherms at 1000 and 750 °C, respectively. Solid circles fitted with solid lines in Fig. 8 present the composition measured by EPMA in single-phase regions. The experimental data within the two-phase region with Mo_2Zr precipitates had scatters in composition. Therefore, diffusion path in two-phase region is estimated based on average composition by considering the composition and volume fraction of U-Zr solid solution and precipitates, which is shown as dot line in Fig. 8. In general, the diffusion paths

Fig. 6 Equilibrium binary phase diagram of Mo-Zr system^[21]

Fig. 7 Equilibrium binary phase diagram of U-Zr system^[21]

constructed based on EMPA measurement were consistent with the ternary phase diagrams determined by Ivanov and Bagrov.^[22,23]

Based on the shape of the diffusion paths and concentration profiles, it is clear that Zr diffuses slowly on the U-Mo alloy side, and Mo is the slow-moving specie on the Zr side. On the U-Mo alloy side, gradients of concentration for Mo and U develop even without a significant diffusional penetration of Zr as shown in Fig. 1 to 4. This is also reflected on the diffusion paths in Fig. 8 wherein the diffusion paths take off toward the pure U corner. While the Mo moves down the concentration gradient, U moves up toward higher U concentration. This result suggests that Zr is an excellent diffusion barrier to retain the high U-density in the U-Mo fuel. With an increase in Zr concentration, diffusion path enters the two-phase (Mo_2Zr and $\gamma\text{-U}$) region.

This is consistent with the observation that Mo_2Zr precipitates in γ -matrix exist at the interface. The U-Mo-Zr isotherms in Fig. 8 clearly show that a decrease in the temperature from 1000 to 700 °C reduces the U-Zr solid-solution phase field and expands the two-phase (Mo_2Zr and γU) field. Correspondingly, the volume fraction of Mo_2Zr increased as shown in Fig. 1 to 4.

Toward the Zr end, diffusion path passes through the U-Zr solid solution and ends at pure βZr at higher temperatures of 1000 and 900 °C, as presented in Fig. 8(a) and (b). At 800 and 700 °C, the diffusion path goes through U-Zr solid solution and jumps into αZr with negligible solubility for either U or Mo as presented in Fig. 8(c) and (d). The diffusion paths within the βZr , for example at 1000 and 900 °C, take off from pure Zr toward the U-corner of the isotherm and suggest that Mo is the slow-moving specie in the βZr solid solution. Presence of Mo plays a significant role on the diffusion path and solubility of Zr in U-Zr solid solution, especially at 700 °C. Based on the U-Zr binary phase diagram shown in Fig. 7, the solubility range of Zr in U-Zr solid at 700 °C should be 50-87 at.%. However, the addition of Mo changes the shape of diffusion paths dramatically and alters the diffusion paths to pass just a small U-Zr solid-solution region as shown in Fig. 8(d). This causes the solubility range of Zr in the U-Zr solid solution to be much smaller, which is 85-90 at.%.

3.3 Growth Rate of Interdiffusion Zone

For Zr to be an effective diffusion barrier between U-Mo fuel and Al matrix, the overall magnitude of diffusional interaction must be small. Figure 5 shows that negligible interdiffusion was observed even after 960 h of isothermal anneal at 600 °C. To compare and contrast the interdiffusion behavior of diffusion couples U-Mo versus Zr to U-Mo versus Al (or Al-Si alloy), parabolic growth rate, K , of interdiffusion zone was calculated at 1000, 900, and 800 °C. Thickness of interdiffusion zone (IZ) is simply defined as the distance between U-10wt.%Mo and pure Zr where concentration gradient becomes negligible (terminal ends where $\partial C/\partial x = 0$). Growth rates of interdiffusion zone were calculated by assuming $K = (\Delta x)^2/2t$, where, Δx is the thickness of IZ, and they are reported in Table 3 for 1000, 900, and 800 °C.

Figure 9 shows that the temperature dependence of the IZ growth rate, K , obeys the Arrhenius relation in the temperature range from 800 to 1000 °C very well. The pre-exponential factor, K_0 , and activation energy, Q_K , of the growth rate were determined and are also reported in Table 3. Growth rates at 700, 600, and 500 °C were then calculated following simple Arrhenius extrapolation as reported in Table 3 and presented in Fig. 9. Table 4 lists the growth rate determined from diffusion couples, U-Mo versus Al and U-Mo versus Al-Si, examined by Perez et al.^[9,20] The growth rate was calculated from the thickness of the complex intermetallic layer. Tables 3 and 4 demonstrate that the growth rates of interdiffusion zone between U-10wt.%Mo and Zr is 2-3 orders of magnitude slower than those between U-10wt.%Mo and Al or Al-Si alloys.

Fig. 8 Diffusion path from U-Mo vs. Zr diffusion couple annealed at (a) 1000 °C for 96 h, (b) 900 °C for 240 h, (c) 800 °C for 480 h, and (d) 700 °C for 720 h. The experimental data measured by EMPA in single phase are presented as solid circles. Solid and dashed lines represent diffusion path determined in single-phase region and estimated in two-phase region, respectively

Table 3 Growth rate of interdiffusion zone determined from the U-10wt.%Mo vs. Zr diffusion couples

Temperature, °C	Thickness of IZ, μm	Annealing time, h	Growth rate (K), m^2/s	$\bar{D}_0$, m^2/s	Q_K , kJ/mol
Measured					
1000	1130	96	1.85×10^{-12}		
900	990	240	5.67×10^{-13}		
800	650	480	1.22×10^{-13}	7.95×10^{-05}	183.83
Calculated					
700	...	...	2.10×10^{-14}		
600	...	...	2.36×10^{-15}		
500	...	...	1.51×10^{-16}

Fig. 9 Arrhenius temperature dependence of growth rate for interdiffusion zone found in the diffusion couples, U-Mo vs. Zr. The measured growth rate at 800, 900, and 1000 °C and calculated values at 500, 600, and 700 °C are represented by solid and empty circles, respectively

Table 4 Growth rate of interdiffusion zone in U-Mo vs. Al and U-Mo vs. Al-Si diffusion couples^[9,20]

Diffusion couple	Annealing temperature, °C	Growth rate (K), m²/s
U-7wt.%Mo vs. Al	600	4.05×10^{-13}
U-10wt.%Mo vs. Al	600	1.69×10^{-12}
U-12wt.%Mo vs. Al	600	7.20×10^{-13}
U-10wt.%Mo vs. Al-2wt.%Si	550	1.45×10^{-14}
U-10wt.%Mo vs. Al-5wt.%Si	550	1.62×10^{-14}

3.4 Other Desirable Barrier Properties of Zr

The ideal barrier materials for metallic fuel system require:

Interdiffusion or reactions between barriers and metallic fuels should be slow.

No or insignificant amount of intermetallic phases should form for better structural integrity. Thermal expansion coefficients and other thermomechanical properties of intermetallic phases may vary significantly. Solid solution is preferred because relevant properties would vary gradually with composition.

Intermetallic phases, if formed due to interdiffusion and reaction, should be stable under irradiation.

Melting point should be high.

Neutron adsorption rate should be low.

Thermal conductivity should be high.

Corrosion resistant should be good.

Results from this study demonstrate that diffusional interaction between U-Mo and Zr is slow compared to U-Mo and Al or Al-Si alloys. For the temperature range investigated in this study, only one intermetallic phase Mo₂Zr exists as precipitates near the interface. However, at the lower

temperature, other intermetallic or solid-solution phases may form according to the U-Mo-Zr ternary phase diagram,^[23] such as, UZr₂, U₄Zr₃Mo, U₆Zr₃Mo, and U₂Mo. In Perez's study,^[6] multiple phases including αU, Mo₂Zr, U-Zr solid solution, and UZr₂ were observed at the interface between U-Mo with Zr in the hot-rolled and annealed U-Mo fuel with Zr as barrier. Therefore, a further investigation of phase equilibrium and diffusional interaction at lower temperature would be required. However, Robinson^[19] has reported that the interaction products containing Zr appear to be stable during irradiation.

As a refractory metal, Zr has a very high melting temperature of 1855 °C, and high thermal conductivity of 22.6 W/m·K.^[21] The neutron absorption rate of Zr is fairly low. The cross section for 2200 m/s neutrons of Zr is 0.185 barn, which is one of the lowest among naturally occurring metals.^[24] Moreover, Zr is highly resistant to corrosion by alkalis, acids, salt water, and other agents.^[25] Diffusional interaction between Zr and Al has been studied by Kidson et al.^[26] and Laik et al.^[27] However, there is discrepancy on the formation and growth rates of intermetallic phases between the two studies. Further investigation into this discrepancy needs to be conducted.

4. Summary

Diffusional interaction between U-10wt.%Mo alloy and Zr has been systematically studied in the temperature range of 600 to 1000 °C. Mo₂Zr has been found at the interface, and its population increased with a decrease in the temperature of diffusion anneal. The diffusion paths were plotted on U-Mo-Zr ternary isotherms and examined with respect to ternary diffusional interactions. The growth rates of interdiffusion zone between U-10wt.%Mo and Zr were found to be about 10³ times slower than those between U-10wt.%Mo and Al or Al-Si alloys. Other desirable factors of Zr working as barrier material were also presented.

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References

1. D. Keiser, S. Hayes, M. Meyer, and C. Clark, High-Density, Low-Enriched Uranium Fuel for Nuclear Research Reactors, *JOM*, 2003, **55**, p 55-58
2. J.L. Snelgrove, G.L. Hofman, M.K. Meyer, C.L. Trybus, and T.C. Wienczek, Development of Very-High-Density Low-Enriched-Uranium Fuels, *Nucl. Eng. Des.*, 1997, **178**, p 119-126

3. D. Wachs, D. Keiser, M. Meyer, D. Burkes, C. Clark, G. Moore, J.-F. Jue, T. Totev, G. Hofman, T. Wiencek, Y.S. Kim, and J. Snelgrove, High Density Fuel Development for Research Reactors, *Glob. 2007 Adv. Nucl. Fuel Cycles Syst.*, 2007
4. M.K. Meyer, G.L. Hofman, S.L. Hayes, C.R. Clark, T.C. Wiencek, J.L. Snelgrove, R.V. Strain, and K.H. Kim, Low-Temperature Irradiation Behavior of Uranium-Molybdenum Alloy Dispersion Fuel, *J. Nucl. Mater.*, 2002, **304**, p 221-236
5. A. Leenaers, S. Van den Berghe, E. Koonen, C. Jarousse, F. Huet, M. Trotabas, M. Boyard, S. Guillot, L. Sannen, and M. Verwerft, Post-Irradiation Examination of Uranium-7 wt% Molybdenum Atomized Dispersion Fuel, *J. Nucl. Mater.*, 2004, **335**, p 39-47
6. E. Perez, B. Yao, D.D. Keiser, and Y.H. Sohn, Microstructural Analysis of As-Processed U-10 wt.%Mo Monolithic Fuel Plate in AA6061 Matrix with Zr Diffusion Barrier, *J. Nucl. Mater.*, 2010, **402**, p 8-14
7. D.D. Keiser, J.F. Jue, B. Yao, E. Perez, Y. Sohn, and C.R. Clark, Microstructural characterization of U-7Mo/Al-Si alloy matrix dispersion fuel plates fabricated at 500 C, *J. Nucl. Mater.*, 2011, **412**, p 90-99
8. B. Yao, E. Perez, D.D. Keiser, Jr., J.-F. Jue, C.R. Clark, N. Woolstenhulme, and Y. Sohn, Microstructure Characterization of As-Fabricated and 475 C Annealed U-7 wt.% Mo Dispersion Fuel in Al-Si Alloy Matrix, *J. Alloys Compd.*, 2011, **509**, p 9487-9496
9. E. Perez, D. Keiser, and Y. Sohn, Phase Constituents and Microstructure of Interaction Layer Formed in U-Mo Alloys vs Al Diffusion Couples Annealed at 873 K (600 C), *Metall. Mater. Trans. A*, 2011, **42**, p 3071-3083
10. A. Leenaers, S. Van den Berghe, W. Van Renterghem, F. Charollais, P. Lemoine, C. Jarousse, A. Rohrmoser, and W. Petry, Irradiation Behavior of Ground U(Mo) Fuel with and Without Si Added to the Matrix, *J. Nucl. Mater.*, 2011, **412**, p 41-52
11. X. Liu, T.C. Lu, Z.H. Xing, and D.Z. Qian, Effects of Different Irradiation Conditions on Swelling Performance of U(10)Mo-Al Dispersion Fuel, *Rare Metal Mater. Eng.*, 2011, **40**, p 1125
12. F. Mazaudier, C. Proye, and F. Hodaj, Further Insight into Mechanisms of Solid-State Interactions in UMo/Al System, *J. Nucl. Mater.*, 2008, **377**, p 476-485
13. C. Komar Varela, M. Mirandou, S. Aricó, S. Balar, and L. Gribaudo, Interdiffusion Between U(Mo, Pt) or U(Mo, Zr) and Al or Al A356 Alloy, *J. Nucl. Mater.*, 2009, **395**, p 162-168
14. Y. Kim, G. Hofman, H. Ryu, and S. Hayes, Irradiation-Enhanced Interdiffusion in the Diffusion Zone of U-Mo Dispersion Fuel in Al, *J. Phase Equilib. Diffus.*, 2006, **27**, p 614-621
15. H. Ryu, J. Park, C. Kim, Y. Kim, and G. Hofman, Diffusion Reaction Behaviors of U-Mo/Al Dispersion Fuel, *J. Phase Equilib. Diffus.*, 2006, **27**, p 651-658
16. L. Pavlinov, A. Nakonechnikov, and V. Bykov, Diffusion of Uranium in Molybdenum, Niobium, Zirconium, and Titanium, *Sov. At. Energy*, 1965, **19**, p 1495-1497
17. G.B. Fedorov, E.A. Smirnov, F.I. Zhomov, V.N. Gusev, and S.A. Paraev, Diffusion of Uranium in Refractory Body-Center-Cubic Metals, *At. Energy*, 1971, **31**, p 1280-1282
18. T. Ogata, M. Akabori, A. Itoh, and T. Ogawa, Interdiffusion in Uranium-Zirconium Solid Solutions, *J. Nucl. Mater.*, 1996, **232**, p 125-130
19. A.B. Robinson, D.M. Wachs, D.E. Burkes, and D.D. Keiser, US RERTR Fuel Development Post Irradiation Examination Results, *Proceedings of the 30th International Meeting on Reduced Enrichment for Research and Test Reactors (RERTR)*, Washington, DC, 2008
20. E. Perez, D.D. Keiser, Jr., B. Yao, and Y.H. Sohn, Interdiffusion in Diffusion Couples: U-Mo vs Al and Al-Si, *RERTR 2009 31st International Meeting on Reduced Enrichment for Research and Test Reactors (RERTR)*, Beijing, China, 2009
21. E.A. Brandes and G.B. Brook, *Smithells Metals Reference Book*, 1992
22. O.S. Ivanov and G.N. Bagrov, Isothermal Cross Sections of the Triple System Uranium-Molybdenum-Zirconium at 1000 °C-625 °C, *Struct. Alloys Certain Syst. Cont. Uranium Thorium*, 1963, p 131-153
23. O.S. Ivanov and G.N. Bagrov, Isothermal Cross Sections at 600 °C, 575 °C, and 500 °C, Polythermal Sections, and the Phase Diagram of the Triple System Uranium-Molybdenum-Zirconium, *Struct. Alloys Certain Syst. Cont. Uranium Thorium*, 1963, p 154-176
24. V.F. Sears, Neutron Scattering Lengths and Cross Sections, *Neutron News*, 1992, **3**, p 26-37
25. J.R. Davis, *Properties and Selection: Nonferrous Alloys and Special-Purpose Materials, Vol 2, ASM Handbook*, 1992
26. G.V. Kidson and G.D. Miller, A Study of the Interdiffusion of Aluminum and Zirconium, *J. Nucl. Mater.*, 1964, **12**, p 61-69
27. A. Laik, K. Bhanumurthy, and G.B. Kale, Intermetallics in the Zr-Al Diffusion Zone, *Intermetallics*, 2004, **12**, p 69-74