Intermetallic formation and interdiffusion in diffusion couples made of uranium and single crystal iron

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ABSTRACT

We studied the interfacial phase formation and diffusion kinetics in uranium-iron diffusion couples. A comparison was made between polycrystalline uranium (U) bonded with polycrystalline iron (Fe^{P}) and polycrystalline uranium bonded with single crystalline Fe (Fe^{SC}). After thermal annealing at 575°C, 600°C, 625°C and 650°C, respectively, diffusion and microstructures at the interface were characterized by scanning electron microscopy and transmission electron microscopy. The presence of grain boundaries in iron has a significant influence on interface reactions. In comparison with U- Fe^{P} system, interdiffusion coefficients of the U- Fe^{SC} system are significantly lower and were governed by much higher activation energies. Integrated interdiffusion coefficients and intrinsic diffusion coefficients were obtained. The intrinsic diffusion coefficients show faster diffusion of iron atoms in both U_6Fe and UFe_2 intermetallic phases than uranium.

Keywords: FCCI, intermetallic, grain boundaries, interdiffusion coefficients, Kirkendall interface, intrinsic diffusion coefficients

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1 Introduction

Fuel-cladding interactions occur when the original gap between nuclear fuel and fuel cladding is closed due to radiation induced dimensional changes, especially void swelling in fuel. The physical contact under stress, high temperature, and neutron irradiation accelerate interface compound formation between fuel and clad, which lead to further degradation of mechanical and thermal properties. Solid-to-solid diffusion couples, fabricated by chemical interaction between fuel and cladding materials and thermal annealing for prolonged time period, are ideal to study fuel clad chemical interactions (FCCI) [1-7]. In addition to forming distinctive intermetallic phases to verify phase diagram prediction, annealing experiments under different temperatures enable the calculation of the diffusion kinetics, which are valuable for comparisons with modeling [5, 6]. Because of the complexity in 5-f actinides, the kinetics data is important for those heavy elements having strong Mott effects [8]. To represent nuclear fuel, uranium, uranium dioxide, or uranium alloys are often used [1-6, 8]. To represent the current and future cladding materials, Fe, Ni, Zr, Fe-Cr or Fe-Ni-Cr alloys are often used [1-6]. Diffusion kinetics are often obtained through elemental mapping by using scanning electron micrograph (SEM) and energy-dispersive X-ray spectroscopy (EDS) technologies [6]. By using focused-ion-beam (FIB) technique, it is feasible to sample the region of interest for transmission electron microscopy (TEM) characterization, to reveal microstructural evolution paths towards phase diagram prediction [7].

For U-Fe diffusion systems, only two intermetallic phases are observed below 719 °C, namely UFe_2 and U_6Fe , as shown in the phase diagram in Fig. 1 [9]. With increasing annealing time, interface dynamics change from Fick's law dominated interdiffusion to Gibbs phase law dominated compound formation, which is featured by step height like elemental distributions

[10]. Huang *et al.* carried out a systemic study of intermetallic growth and interdiffusion in uranium and poly-crystal iron system, and obtained integrated interdiffusion coefficient and growth constants of the UFe_2 and U_6Fe intermetallic phases [6]. However, direct experimental comparison with modeling results is still difficult due to the fact the roles of grain boundaries in interface reactions and intermetallic diffusion are unclear. As a matter of fact, grain boundary effect represents a general issue for studies on any diffusion couples made of polycrystalline materials. Although it is predictable grain boundaries will enhance diffusion [11], the level of enhancement is determined by details of grain boundary configurations and densities, which is not generally considered by modeling studies which often focus more on intrinsic properties without boundary complexity. In this work, we studied intermetallic phase kinetics in the U-Fe binary system in the absence of grain boundary effects by using single crystal Fe, in order to meet the needs from model validation and verification.

2. Experimental procedure

We used depleted uranium and single crystal iron to fabricate U-Fe^{SC} diffusion couples. A single crystal iron bar of 99.994%+ was obtained from Princeton Scientific Corp, NJ. High-purity depleted uranium were melted three times using arc melting to ensure the homogeneity and drop-cast to form rods. The as-cast uranium is polycrystalline. The single crystal iron bar and uranium bar were cut into pieces of ~3mm thickness, and metallographically prepared with SiC paper down to a grit of 1200 using the isopropanol alcohol as lubricant. The as-polished uranium pieces were further pickled in nitric acid (30% nitric acid and 70% methanol by volume) solution to remove oxide layer before the diffusion couple assembly. Polycrystalline uranium and single crystal iron were stacked in a stainless steel (316) jig, and applied pressure

about 5 MPa so that contact was achieved between them, as shown in Fig. 2. Diffusion couples were then put inside a fused quartz capsule using an oxygen-propane torch. Prior to sealing, the capsules were pumped and purged with 95% Ar and 5% H gas mixture to reach a vacuum better than 10^{-6} torr. The annealing was performed by using a MTI-OTF-1200X furnace with a typical temperature fluctuation less than 1°C, at 575°C for 168 hours, 600°C for 144 hours, 625°C for 120 hours and 650°C for 120 hours for each sample. The heating rate was controlled to be about 5°C/min and the cooling rate is about 2°C/min. Figure 1 shows U-Fe phase diagram [9], with our annealing temperature range marked. The highest annealing temperature of 650°C was selected to be lower than the eutectic point close to 66 uranium at.% to avoid liquid formation. Furthermore, this temperature is also lower than 665°C, the transition temperature from α U to β U. Therefore, no phase transformation is expected during the cooling stage.

Under the hypothesis that atom diffusion on a free surface can be substantially different from that in bulk, all current structural characterization avoids the surface region. The annealed diffusion couples were cut into two pieces with the cutting plane perpendicular to the interface plane and followed by metallographic preparation and polishing on the freshly obtained surface as described in the paragraph above.

The cut and polished surfaces were used for elemental mapping across the interface using SEM with a FEI Quanta 600. For atomic scale characterization, cross-sectional TEM specimens were prepared by using focused-ion-beam (FIB) lift out technique in a FEI Quanta 3D FEG dual beam SEM/FIB microscope. A FEI TECNAI TF30-FEG STwin TEM operated at 300 kV was used for TEM characterization.

3. Results and discussions

3.1 Phase formation and structures

Fig. 3a-d plot the uranium and iron elemental distributions in 575°C, 600°C, 625°C and 650°C annealed samples, respectively, obtained by using line scan EDS. The step height-like elemental distributions confirm the intermetallic phase formations. Figure 3e and 3f show typical back-scattered-electron (BSE) images of U-Fe^{SC} diffusion couples annealed at 625°C and 650°C. The contrast in BSE images reflects the difference in the average atomic number. Two brighter phases on the left side are uranium-rich phases and two darker phases on the right are iron-rich. From the left to the right, the corresponding phases are α U, U₆Fe, UFe₂, and α Fe (see Figs. 3e and 3f). Clearly, the width of the UFe₂ intermetallic is much less than that of the U₆Fe intermetallic.

TEM was also performed to verify phase formations, as well as to investigate the microstructures of each phase. The TEM micrographs of the 625°C the U-Fe^{SC} diffusion couples are shown in Fig. 4a-c. The micrograph of Fig. 4a corresponds to the U/U₆Fe interface, while Fig. 4b corresponds to a region containing the U₆Fe, the UFe₂ and Fe phases. The phases are distinguished by the contrast and further confirmed by the corresponding electron diffraction patterns. Furthermore, the grain size for each phase is observed to vary. α U and the U₆Fe appear to have large grains, comparable to the lamella dimension of 10 μ m. The Fe phase remains single-crystal at 625°C. In a comparison, the UFe₂ intermetallic phase has a smaller grain size on the order of 1 μ m, as shown in Fig. 4c. In Fig. 4b, there is a 1 μ m wide region (labeled as D) between U₆Fe and UFe₂. The corresponding convergent beam electron diffraction (CBED) from this region show a fluorite crystal structure expected from UO₂.

Scanning transmission electron microscopy (STEM) and EDS were used to characterize the compositions of the interfacial region. As shown in Fig. 5a, from left to right, the STEM

micrograph shows alternating bright and dark contrast, corresponding to phase configurations of U_6Fe - UO_2 - UFe_2 -Fe. Point EDS spectrum (as shown in Fig. 5b) obtained from the layer between U_6Fe and UFe_2 shows prominent peaks of O and U, and tiny Fe peaks (<4%). The Cu peak is from the TEM grid. Based on CBED pattern and composition analysis, we conclude that this layer is UO_2 . Formation of this oxide layer is believed to occur during the polishing and bonding process, due to the extreme reactivity of metallic uranium with oxygen (even in an Ar ambient). The layer thickness is approximately 1 μm in all samples, regardless of annealing temperature. The UO_2 layer is very stable. Previous studies showed no chemical interaction at UO_2/Fe interface after 1000-hour annealing at 1000 °C [12]. As to be discussed, the existence of UO_2 will influence interface reaction by introducing an incubation time period for diffusion and this effect must be considered in data interpretation. On the other hand, high stability of UO_2 makes it ideal to represent the original interface between iron and uranium, i.e., the Kirkendall plane in diffusion analysis.

One interesting finding from TEM characterization is that both U_6Fe and the UFe_2 show distinguishable grain morphologies: The U_6Fe has grains no smaller than 10 μm while the UFe_2 shows 1 μm grain size. This implies that the grain structures of the two bulk materials, from which the intermetallic phases form, do not have a direct influence on the grain structures of these as-formed intermetallic phases. On the other hand, it can be due to the nature of the intermetallic phases, such as stacking-fault energies and lattice mismatch that result in a residual stress, influencing the intermetallic formation. Furthermore, phase formation conditions, such as the annealing temperatures, bonding pressures and surface textures of the bulk materials, also affect the intermetallic phase formation, at least in terms of grain size [11].

3.2 Integrated interdiffusion coefficients

The BSE micrographs and concentration profiles obtained with EDS, shown in Fig. 3, provide the thickness of intermetallic phases formed along the diffusing direction. We measured and averaged the thickness of each intermetallic phase from BSE micrographs. We also used the TEM micrographs to estimate the thickness of the oxide layer, so that the thickness of the oxide layer was not taken as part of either intermetallic phase. The measured intermetallic phase thickness can be found in Table 1. With these measurements, we calculated kinetic parameters that describe the phase formation and growth process.

We first carried out calculations for integrated interdiffusion coefficients $\tilde{D}^{Int}$ with the Boltzmann-Matano approach [6, 13-15]. The results are listed in Table 1. The approach includes the following steps: Firstly, the position of the so-called Matano plane, x_M , was determined by the concentration profile by:

$$\int_{N_i^-}^{N_i^+} (x_M - x) dN_i = 0 \quad (1)$$

where N_i^+ and N_i^- refer to the mole fraction of type i atom at the ends of the as-bonded diffusion couple[15]. Secondly, the interdiffusion flux through each intermetallic phase was calculated, taking the Matano plane position x_M and the mole fraction of type i atom at the position where the interdiffusion flux is calculated:

$$\tilde{J}(N_i^*) = \frac{1}{2t} \int_{N_i^\pm}^{N_i^*} (x - x_M) dN_i \quad (2)$$

where t is the intermetallic diffusing time.

The integrated interdiffusion diffusion coefficient of phase ν could then be calculated by integrating the flux over intermetallic phase of ν [16]. The intermetallic phases in the U-Fe systems, shown in Fig. 2b-d, have narrow compositional range. As a result, a simplified equation provided by Huang *et al.* was used [6]:

$$\tilde{D}^{Int,(v)} = \Delta x^{(v)} \times \tilde{J}^{(v)} = \frac{\Delta x^{(v)}}{2t} \int_{N_i^+}^{N_i^{(v)}} (x^{(v)} - x_M) dN_i \quad (3)$$

where $\Delta x^{(v)}$ is the dimension of phase v along the direction of diffusion.

The presence of UO_2 layer influences the interfacial diffusion and this effect must be carefully considered in data analysis. The formation of UO_2 is unlikely to be avoided. Previous studies have noticed the existence of such incubation periods at low temperatures due to existence of diffusion barrier [16]. At a time shorter than this incubation period, diffusion is very limited and interfacial compound formation is not initialized. Therefore the diffusing time t in Eq. (2) should be replaced by effective time t_{eff} , the annealing time subtracted by the incubation time period. Experimental determination of the incubation periods at all temperatures, however, is challenging and unrealistic considering a large sample matrix required. To alleviate the issue, we proposed the following way to “correct” the data to exclude/minimize UO_2 effects, based on self-consistency and Arrhenius relationship expected for $\tilde{D}^{Int}$. Based on the observation (to be discussed in Fig. 6) that only the lowest temperature point (575°C) deviates significantly from Arrhenius relationship, we first assume that two highest temperature points (650°C, 625°C) have neglected incubation period. The Arrhenius fitting of $\tilde{D}^{Int}$ at highest temperatures is then extended to two lowest temperature points (600°C, 575°C) and the deviations between the line and the experimental $\tilde{D}^{Int}$ are used to obtain incubation time periods. Since the diffusion length L required to penetrate through a stable UO_2 is the same at all temperatures, the product of incubation time period t_{incu} and corresponding diffusivity D should be fixed at all temperatures. Thus we use $t_{incu} = L^2/2D$ to estimate incubation time periods for two highest temperature points and to recalculate $\tilde{D}^{Int}$ at these two points. This process is then repeated. After a few iteration steps, results are quickly converged.

The obtained $\tilde{D}^{Int}$ values are listed in Table 1 and plotted in Fig. 6. For comparison, the integrated intermetallic diffusion coefficients from previous studies on U-Fe^P system were also plotted [6]. We obtained activation energy of 309 KJ/mole for U₆Fe formation and 414 KJ/mole for UFe₂ formation. These values are significantly higher than previous results from U-Fe (polycrystalline), which are 148 KJ/mole for the U₆Fe and 243 KJ/mol for the UFe₂. Obviously, the interdiffusion coefficients are very sensitive to whether Fe is single crystal or not. Fig. 6 also compares the “raw” $\tilde{D}^{Int}$ data by using total annealing time as t in Eq. (3) and “corrected” $\tilde{D}^{Int}$ by using effective annealing time (considering the incubation time periods obtained from iteration process). The corrected $\tilde{D}^{Int}$ is approximately twice of the raw $\tilde{D}^{Int}$ for 575 °C, The difference between raw and corrected $\tilde{D}^{Int}$ becomes less for higher temperature, i.e. less than 6% for the highest temperature 650 °C.

The as-calculated integrated intermetallic diffusion coefficients are the integration of interdiffusion coefficient, $\tilde{D}$, over an intermetallic phase, v . In a binary system, in which the only kinetic process is diffusing, $\tilde{D}$ is a characteristic value of a component with a certain concentration of the two elements. As a result, the value of integration of $\tilde{D}$ over any given concentration range should only be determined by the composition and physical property, such as microstructures and defect distribution of this range. $\tilde{D}^{Int,(v)}$, then, was believed to be a characteristic value determined solely by the conditions of the intermetallic phase v , as pointed out by Huang [6]. However, it is well known that both diffusion and reaction at the interface govern the interface formation and growth [11, 17]. The concentration profiles in Fig. 3 are attributed to both of the two mechanisms and the as-calculated $\tilde{D}^{Int}$ with Boltzmann-Matano approach is a combination of both diffusion and interface reaction. For a given intermetallic phase v , $\tilde{D}^{Int,(v)}$ is no longer only a characteristic value of the phase, but also determined by

interface conditions, interdiffusion with neighboring phases and growth of neighboring intermetallic phases [17].

With this knowledge, we believe the difference of $\tilde{D}^{Int}$ between the U-Fe^{SC} system and the U-Fe^P system is attributed to both diffusion and interface reaction. For the formation and growth of the UFe₂ phase in U-Fe^P system, the presence of grain boundaries changes the interface condition for UFe₂ to form and grow, i.e., it makes direct contribution to the interface reaction-controlled part of $\tilde{D}^{Int}$ [11]. In polycrystalline iron, grain boundaries are quick diffusion paths for uranium to take when diffusing into iron. The migration energy of uranium atoms at Fe grain boundaries are expected to be less than that in iron bulk, due to reduced stress energy associated with outward relaxation of lattice surrounding uranium atoms [18]. As a result, uranium atoms can diffuse faster and deeper by using grain boundaries to increase interaction with iron domains, which is equivalent to increasing interface/nucleation area. Because of a faster diffusion and enhanced nucleation by grain boundaries in iron, a faster growth of the UFe₂ intermetallic phase zone is expected in polycrystalline iron. The microstructure of the UFe₂ intermetallic phase, such as defects and grain boundaries, has great influences on diffusion-controlled part of $\tilde{D}^{Int}$. Thus, the grain boundaries in iron affect both interface reaction and diffusion components of $\tilde{D}^{Int}$.

One very interesting observation is that the presence of grain boundaries in iron side affects the diffusion of iron into uranium side to form the U₆Fe phase, as shown by the U₆Fe width difference between the U-Fe^{SC} and the U-Fe^P samples. Our hypothesis is that once uranium diffusion is facilitated through boundary-assisted diffusion, uranium vacancy concentration on uranium side expects to be higher in the U-Fe^P sample. Consequently, iron diffusion into uranium, which is mediated through vacancy diffusion mechanism, is enhanced with a wider the

U₆Fe phase. In other words, the U₆Fe and UFe₂ intermetallic phases are closely coupled with each other and both intermetallic phases are influenced by the grain boundaries in iron side.

We noted that at higher temperatures (>~950°C), the integrated interdiffusion coefficients of the U-Fe^{SC} and the U-Fe^P seem to converge. This is due to the fact that at higher temperatures, diffusion through vacancies within domains becomes more significant. Thus the grain boundary effect is less important at elevated temperatures.

3.3 Intrinsic integrated diffusion coefficients

As shown previously, the interdiffusion coefficient $\tilde{D}$ is a function of the relative diffusion of the two species in a binary intermetallic phase. In order to understand diffusion of a particular species, the intrinsic diffusion coefficient for each phase must be calculated. The relationship between interdiffusion coefficient and intrinsic diffusion coefficients has been reported as [19]:

$$\tilde{D} = D_1 N_2 + D_2 N_1 \quad (7)$$

where D_1 and D_2 are the intrinsic diffusion coefficients of the two species.

The intrinsic diffusion coefficients of both species can be calculated from the concentration profiles if the original interface can be marked [19-21]. With respect to the original interface, commonly referred to as the Kirkendall plane, the intrinsic fluxes, which reflect the motilities of species, can be defined as:

$$J_i = -D_i \frac{\partial C_i}{\partial x} \quad (8)$$

where C_i is the concentration of species i [mol/m³].

This equation of Fick's law is similar to the equation of intermetallic flux and intermetallic diffusion coefficient. The only difference is that the former one is calculated with respect to the Kirkendall plane while the latter one with respect to the Matano plane.

The approach used by Huang, *et al.* [6, 13-16] can be used to calculate integrated intrinsic diffusion coefficients in a binary system like U-Fe without significant changes. The intrinsic fluxes can be obtained by using:

$$J(N_i^*) = \frac{1}{2t} \int_{N_i^*}^{N_i^*} (x - x_K) dN_i \quad (9)$$

a similar equation to Eq. (2). The only difference is the reference point is changed from the Matano plane x_M to the Kirkendall plane x_K . The integrated intrinsic diffusion coefficients, with respect to the intrinsic fluxes, can be calculated with:

$$D_i^{Int,(v)} = J_i^{(v)} \bullet \Delta x^{(v)} \quad (10)$$

In this study, we used the residual UO_2 layer as the marker of the Kirkendall plane. The intrinsic fluxes and integrated intrinsic diffusion coefficients of uranium in the UFe_2 phase and iron in the U_6Fe phase were calculated from the concentration profiles using Eq. (9) and (10). The integrated intrinsic diffusion coefficients of iron in the UFe_2 and uranium in the U_6Fe were calculated with Eq. (7) using results obtained in Section 3.2. The calculated coefficients are listed in Table 2. Equation (7) is a simplified form of the Darken equation. Because of this, the values of D_{Fe}^{Int,UFe_2} and D_U^{Int,U_6Fe} would be overestimated as suggested by Mehrer [22].

Figure 7 shows plots of the integrated intrinsic diffusion coefficients $D_i^{Int,(v)}$ of uranium and iron in both the UFe_2 intermetallic phase and the U_6Fe intermetallic phase in the U-Fe^{SC} system overlapped on their respective integrated interdiffusion coefficients $\tilde{D}^{Int,(v)}$. **Data analysis has considered the time correction due to UO_2 presence, as discussed in Section 3.2. Extracted activation energies were listed in Table 2.** Comparison between the two intermetallic phases

shows systematically lower diffusion coefficients in the UFe_2 phase than in the U_6Fe phase. Comparison between the two species, uranium and iron, shows that, in both intermetallic phases, iron diffuses faster than uranium does and iron has lower activation energies.

In the UFe_2 , the C15 (MgCu_2) type of laves phase was observed to be in agreement with previous studies in NbCo_2 , which has the same crystal structure [23]. In both cases, the faster diffusing atom (Fe/Co) has a smaller atomic size and a higher atomic fraction. In our case, iron has a Goldschmidt radius of 1.28 Å while uranium has a Goldschmidt radius of 1.38 Å. As a result, Fe should have a lower migration energy and higher mobility than uranium does. Such size dependence in diffusion has been reported in other systems as well [24]. On the other hand, compared to uranium, the higher atomic fraction of iron is believed to provide higher thermal vacancy concentration, which enhances vacancy diffusion of iron. The major diffusion mechanism in the NbCo_2 system is vacancy diffusion [23], which we believe is also the major diffusion mechanism in the UFe_2 intermetallic phase.

Conclusion

In this work, interdiffusion and interface compound formations in the U-Fe^{SC} system in the temperature range between 575°C and 650°C were studied and compared with published data obtained from the U-Fe^{P} system. The difference in integrated interdiffusion coefficients of the U_6Fe and the UFe_2 intermetallics show that the grain boundaries play an important role in influencing interface reactions and intermetallic phase formations. The intrinsic integrated diffusion coefficients of the U-Fe^{SC} system were also calculated by using the residual UO_2 layer as a marker of the Kirkendall interface. In both intermetallic phases, the U_6Fe and the UFe_2 , iron

diffuses faster in the temperature range of this study, while uranium has higher activation energies of intrinsic integrated diffusion coefficients.

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Captions:

Figure 1. Phase diagram of Fe-U binary system. The black dashed lines on the phase diagram indicate diffusion couple annealing temperatures ranging between 575°C and 650°C.

Figure 2. A schematic drawing of the diffusion couple assembly.

Figure 3. (a-d) Concentration profiles obtained with SEM/EDS for diffusion couples annealed at different temperatures and (e, f) BSE image of 625°C and 650°C diffusion couples.

Figure 4. TEM micrographs of 625°C diffusion couple: (a-b) interface regions of U-U₆Fe and U₆Fe-UFe₂-Fe, with diffraction patterns obtained and indexed shown by (A-F); and (c) the grain structure of the UFe₂ intermetallic phase.

Figure 5. (a) STEM micrograph of U₆Fe-UO₂-UFe₂-Fe interface and (b) EDS spectrum of the UO₂ layer.

Figure 6. Integrated interdiffusion coefficients for intermetallic phases in U-Fe^{SC} with or without considering incubation time periods from UO₂. As a comparison, data obtained from various studies on U-Fe^P are plotted (Ref. [6]).

Figure 7. Integrated intrinsic diffusion coefficients of uranium (triangles) and Fe (circles) in U₆Fe (empty) and UFe₂ (solid) of U-Fe^{SC}.

Figure 1

Figure 2.

Figure 3.

Figure 4

Figure 5.

Figure 6.

Figure 7.

Table 1. Intermetallic phases, phase thickness, integrated interdiffusion coefficients $\tilde{D}^{Int,(v)}$, and corresponding activation energies Q in U-Fe^{SC} system.

Intermetallic phase	Annealing temperature (°C)	Thickness measurement (μm)	Raw $\tilde{D}^{Int,(v)}$ ($\times 10^{-16}$ at. frac. m ² /s)	Corrected $\tilde{D}^{Int,(v)}$ ($\times 10^{-16}$ at. frac. m ² /s)	Q (KJ/mol)
U ₆ Fe	575	4.75	0.0236	0.0749 ± 0.0208	309
	600	12.98	0.207	0.285 ± 0.055	
	625	27.57	1.123	1.29 ± 0.15	
	650	44.12	2.93	3.11 ± 0.17	
UFe ₂	575	0.31	0.00101	0.00227 ± 0.00029	414
	600	1.30	0.0114	0.0129 ± 0.0011	
	625	3.02	0.0693	0.0718 ± 0.0022	
	650	7.11	0.303	0.306 ± 0.003	

Table 2. Integrated intrinsic diffusion coefficients $D_i^{Int,(v)}$ in U-Fe^{SC} system and the extracted activation energies Q.

Intermetallic phase	Species	$D_i^{Int,(v)} (\times 10^{-16} \text{ at. frac. m}^2/\text{s})$				Q (KJ/mol)
		575 °C	600 °C	625 °C	650 °C	
U ₆ Fe	U	0.0162 ±	0.0747 ±	0.370 ±	1.28 ± 0.07	375
		0.0045	0.0147	0.041		
	Fe	0.0847 ±	0.320 ±	1.45 ±	3.42 ± 0.18	305
		0.0235	0.061	0.16		
UFe ₂	U	0.000942 ±	0.00616 ±	0.0364 ±	0.197 ± 0.002	465
		0.000120	0.00051	0.0011		
	Fe	0.00492 ±	0.0264 ±	0.143 ±	0.524 ± 0.005	384
		0.00063	0.0022	0.004		