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APPLIED PHYSICAL CHEMISTRY
PROGRESS REPORT

October 1991 - September 1992

by

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ABSTRACT

This document reports on the work done in applied physical chemistry at the Chemical Technology Division (CMT), Argonne National Laboratory (ANL), in the period October 1991 through September 1992. This work includes research into the processes that control the release and transport of fission products under accident-like conditions in a light water reactor, the thermophysical properties of the metal fuel in the Integral Fast Reactor under development at ANL, and the properties of candidate tritium breeding materials in environments simulating those of fusion energy systems.

SUMMARY

We are experimentally determining the thermophysical properties of core-concrete mixtures expected to occur during the molten core-concrete interaction phase of severe accidents in light water reactors (LWRs). During this report period, our experiments were focused on an important property of core-concrete mixtures, i.e., their viscosities at high temperatures (1700-2600°C). Our measurements of this property were significantly higher than calculated values made with a thermal-hydraulic code (CORCON) used to compute the consequences of hypothetical severe accidents at nuclear reactors. This discrepancy was attributed to the viscosity subroutine in CORCON assuming that the core-concrete mixture at temperatures below 2000°C is a Newtonian fluid with no solids, but our experimental observations indicated that this is not the case. The results of this study will provide improved correlations for viscosity versus composition for a broad range of core-concrete mixtures.

Measurements and calculational analyses are being performed to provide needed information on the thermodynamic and transport properties of Integral Fast Reactor fuel. One study is addressing the phase relations with fuel-cladding systems over a wide temperature range. To that end, improvements were made to the existing Fe-Zr phase diagram, and phase diagrams were derived for the Fe-U-Zr system. This work is an extension of our previous assessments of the U-Zr, Fe-U, and Pu-U phase diagrams. We are also exploring the feasibility of examining IFR materials by use of synchrotron radiation sources. The compositional and structural information obtained by such an analysis could help in understanding and

predicting intermetallic compound formation, local melting, and fuel redistribution in IFR fuel.

A critical element in the development of the fusion reactor is the blanket for breeding tritium fuel. Studies are underway with the objective of determining the feasibility of using lithium-containing ceramics as breeder material. Diffusivity and desorption rate constants were determined for Mg-doped and undoped LiAlO_2 samples at 528-785°C. The diffusivity was found to follow a simple Arrhenius-type temperature dependence, with the diffusivity determined for the doped and undoped materials being the same within experimental error. This suggests that impurities which create lithium vacancies have little or no effect on tritium diffusion in lithium aluminate over the temperature range investigated (500-800°C). The doped material appeared to have slightly higher desorption rate constants than the pure material. Taken together, the data indicate that, for sample grain radii less than 100 μm , the tritium transport will be controlled by desorption. In other work, tritium release models were employed in analyzing data from an in-pile tritium release experiment using a lithium oxide breeder blanket. The results from this analysis are being used to refine the models.

I. OVERVIEW

The program in applied physical chemistry involves studies of the thermodynamic, thermophysical, and transport behavior of selected materials in environments simulating those of fission and fusion energy systems. These experimental and calculational studies have the following objectives: (1) determine the thermophysical properties of the core-concrete mixtures expected to occur during the molten core-concrete interaction phase of severe accidents in LWRs, (2) provide needed information on the thermodynamic and transport properties of the clad U-Pu-Zr fuel in the IFR, and (3) evaluate the feasibility of using lithium-containing ceramics as the blanket material for breeding tritium fuel in a fusion reactor.

II. LIQUIDUS-SOLIDUS TEMPERATURES AND VISCOSITIES OF CORE-CONCRETE MIXTURES

(M. F. Roche)

Thermophysical properties of molten mixtures of urania, zirconia, and concrete were measured in this research program. Such mixtures are expected to occur during the molten core-concrete interaction (MCCI) phase of severe accidents in water-cooled nuclear reactors, e.g., core meltdown. The MCCI phase is preceded by a series of other phases: loss of coolant, heatup and degradation of the reactor core caused by heat produced by the radioactive decay of fission products, melting of the core debris (uranium, fission products, Zircaloy cladding, control rods, and structural materials) through the stainless-steel reactor vessel, and deposition of the core debris and molten steel on the concrete basemat beneath the vessel, where the MCCI phase begins. During the MCCI phase, the concrete basemat reacts with the hot (~2700°C) core debris and molten steel to form a multiphase, molten mixture, which is vigorously stirred by the gaseous products (carbon dioxide, carbon monoxide, steam, and

hydrogen) from the decomposition and reduction of the concrete basemat. A knowledge of the thermophysical properties of the solids, liquids, and gases that exist during the MCCI phase is crucial for understanding and modeling the consequences of a severe reactor accident.

The results of our research will be incorporated in thermal-hydraulic codes such as CORCON,¹ which is an integral part of the MELCOR code² employed to compute the consequences of hypothetical severe accidents at nuclear reactors. In addition, our experiments are designed to aid researchers in ANL's Reactor Engineering Division in analyzing results from their large-scale (~500 kg) MCCI experiments, which are being performed under international sponsorship (the Aerosol Containment Experiment Consortium).

In earlier experiments,³ sponsored by the Electric Power Research Institute (EPRI), we conducted source-term measurements of the vaporization of refractory fission products from core-concrete mixtures. The experiments in this report are concerned with two other important properties of the core-concrete mixtures: liquidus-solidus temperatures and viscosities. The liquidus-solidus temperatures were measured by differential thermal analysis (DTA) in a program sponsored by the Nuclear Regulatory Commission. (Liquidus temperatures are the onset of solidification as a material is cooled from the liquid state; solidus temperatures are the onset of melting as a material is heated.) The viscosities and, in some cases, liquidus temperatures were measured by rotational viscometry in a program sponsored by the ACE Consortium.

In an earlier report,⁴ DTA measurements of the $\text{UO}_2\text{-CaO}$ phase diagram were described. Solidus and liquidus temperatures were also presented for three types of concrete (limestone, limestone-sand, and siliceous) and for their mixtures at 27.5 wt% calcined concrete with 72.5 wt% $\text{UO}_2\text{-ZrO}_2$ (1.6:1 mole ratio of UO_2 to ZrO_2). In addition, viscosity data for some mixtures of urania and zirconia with these concretes were presented.

During this report period, we continued DTA measurements of liquidus-solidus temperatures and rotational-viscometry measurements of viscosity on mixtures of concretes with $\text{UO}_2\text{-ZrO}_2$. In the DTA experiments, the samples, which weigh up to 20 g, were held in covered molybdenum crucibles (9 cm³) that rest on a sample thermocouple within a high-temperature furnace cavity. A gas of Ar-3% H_2 (at a pressure of about 0.03 MPa) was added to the furnace cavity to minimize vaporization losses to less than 2% of the sample mass. The output from the sample thermocouple was compared with that from a control thermocouple which drove the furnace, via a programmable power supply, between two temperature limits at a prescribed linear rate (e.g., 900 to 2400°C limits at a rate of 30°C/min). The peaks in the resulting DTA curves were related to the liquidus and solidus temperatures.

In the viscosity experiments, the samples, which weighed 150 to 200 g and had a volume of about 30 cm³, were heated to temperatures as high as 2600°C within the hot zone of a molybdenum-tungsten alloy (70Mo-30W) furnace tube. The furnace tube was continuously purged with Ar-3% H_2 gas to ensure retention of the urania in the U^{4+} state. A Brookfield programmable viscometer equipped with a 70Mo-30W alloy spindle that was immersed in the molten sample was employed to measure the sample's viscosity as a function of spindle rotation rate. The viscosities ranged from about 0.05 to 500 Pa·s (50 to 500,000 cP),

depending on the sample composition, rate of spindle rotation, and temperature. Liquidus temperatures were detected by a low sample viscosity (under 0.2 Pa·s or 200 cP), as well as a viscosity that was constant with spindle rpm (Newtonian behavior).

Viscosities were measured for a mixture of siliceous concrete (27.5 wt%), urania (56.6 wt%), and zirconia (15.9 wt%) at intervals of 50 to 70°C over the temperature range 1731-2322°C. The solidus and liquidus temperatures for this particular mixture were determined by DTA to be 1139 and 2276°C, respectively. Therefore, the viscosity measurements extended from above the liquidus well into the liquidus-solidus temperature range. The viscosity measurements plotted in Fig. 1 indicate Newtonian behavior (i.e., viscosity independent of spindle rotation rate) at 2171°C. Our other measurements indicated Newtonian behavior over the temperature range 2094-2322°C. This behavior is expected of a single-phase fluid or a fluid with only a minor volume fraction of solids. However, as the temperature was decreased further into the liquidus-solidus region, the sample exhibited a type of non-Newtonian behavior called "shear thinning," in which the viscosity decreased significantly with increasing spindle rotation rate (see curves in Fig. 1 at 1835 and 2003°C). This shear thinning was due to the presence of a significant fraction of solids at the lower temperatures.

Fig. 1. Viscosity of Mixture Containing Siliceous Concrete (27.5 wt%), Urania (56.6 wt%), and Zirconia (15.9 wt%) at Three Temperatures and Colgate Toothpaste at Ambient Temperature

The large increase in viscosity with decreasing temperature was another indication of increasing solids content (the viscosity increased by a factor of about 1000 with a temperature decrease of about 600°C).

Figure 1 also shows the viscosity of Colgate toothpaste, diluted slightly with water (5 parts by volume of toothpaste to one part by volume of water). Its viscosity at ambient temperature is nearly identical to that of the urania-zirconia-concrete mixture at 1835°C. This comparison is useful because it is easy to visualize the flow of an off-the-shelf product such as Colgate toothpaste. Toothpaste-like flow has not been considered in any thermal-hydraulic code for nuclear reactor accidents. Indeed, the viscosity of a similar mixture of urania, zirconia, and concrete is calculated in CORCON¹ to be about 0.05 Pa·s (50 cP) at 1500°C. This low viscosity is calculated because the viscosity subroutine in CORCON assumes that the mixture is a Newtonian fluid with no undissolved solids. We could not have conducted rotational viscosity measurements on the mixture of urania, zirconia, and siliceous concrete at 1500°C because the sample would have been essentially solid, with a viscosity in excess of 1000 Pa·s (10⁶ cP). We have conducted measurements like those in Fig. 1, with similar results, on a variety of mixtures of urania and zirconia with limestone, limestone-sand, or siliceous concretes. Since the viscosity affects heat transfer, release of fission products, and a variety of other important processes in the MCCI phase (such as melt spreading and cooling), reliable data are clearly needed for incorporation in the codes. The results of this study will provide improved correlations of viscosity versus composition for a broad range of core-concrete mixtures.

Solidus-liquidus temperatures determined by DTA and viscosity measurements for a UO₂-to-ZrO₂ mole ratio of 1.6:1 are plotted in Fig. 2 as a function of concrete weight percent.

Fig. 2. Liquidus-Solidus Curves from DTA and Viscosity Data for Urania-Zirconia (1.6:1 Mole Ratio) Mixtures with Three Concretes. Arrow on top of symbols indicates "greater than."

The values for $\text{UO}_2\text{-ZrO}_2$ (at 0 wt% concrete) were taken from the published phase diagram.⁵ Also shown in Fig. 2 are proposed solidus and liquidus curves. Thermodynamic calculations presented to date^{6,7} are in reasonable agreement with our solidus curves for all three concretes and with our liquidus curves for the siliceous-concrete mixtures, but are 300 to 500°C lower than our liquidus curves for mixtures containing limestone-sand or limestone concretes. The agreement is expected to improve when our revised $\text{UO}_2\text{-CaO}$ phase diagram⁴ is incorporated in the thermodynamic calculations.

III. THERMOPHYSICAL PROPERTY STUDIES

(L. Leibowitz and R. A. Blomquist)

Measurements and analyses are being performed to provide needed information on the thermodynamic and transport properties of Integral Fast Reactor (IFR) fuel clad U-Pu-Zr alloy. In 1992, this effort was focused on determining phase relations in the Fe-Zr and Fe-U-Zr systems and the feasibility of examining IFR materials using synchrotron radiation sources.

A. Phase Relations

The IFR currently being developed at ANL employs a U-Pu-Zr alloy fuel clad in stainless steel. We are, therefore, interested in phase diagrams of fuel and stainless steel components, especially at high temperatures. In previous work, we presented our assessments of the U-Zr,⁸ Fe-U,⁹ and Pu-U¹⁰ phase diagrams. This effort is continuing with our assessment of the Fe-Zr and Fe-U-Zr systems.

The general methods used in our analysis have been described elsewhere.¹¹⁻¹³ The calculations are performed with programs of the F*A*C*T (Facility for the Analysis of Chemical Thermodynamics) computer system and involve determining equations for the Gibbs energies of all phases existing in an alloy system as functions of temperature and composition. Phase diagrams are derived by calculating the lowest common tangents to the Gibbs energy-composition curves. This procedure ensures thermodynamic consistency of the resulting phase diagram and provides a powerful means for assessing conflicting data.

1. Iron-Zirconium System

Two somewhat different Fe-Zr phase diagrams are available in the literature. Massalski¹⁴ included the Fe-Zr diagram derived by Arias and Abriata,¹⁵ shown in Fig. 3, whereas Kubaschewski¹⁶ gave a somewhat different diagram, shown in Fig. 4. Our calculated diagram, shown in Fig. 5, was designed to resolve the conflicts between the two published diagrams. Some of the areas of disagreement are discussed below.

As reflected in Figs. 3 and 4, several intermetallic compounds (e.g., Fe_2Zr , Fe_3Zr , FeZr_4) have been reported for the Fe-Zr system, with some disagreement over what compounds exist.^{15,16} The enthalpy of formation of Fe_2Zr from the solid elements has been measured calorimetrically,¹⁵ and within the error limits of the measurements, it is independent

Fig. 3. Iron-Zirconium Phase Diagram of Arias and Abriata.¹⁵
Reprinted from Ref. 14.

Fig. 4. Iron-Zirconium Phase Diagram of Kubaschewski.
Reprinted from Ref. 16.

Fig. 5. Calculated Iron-Zirconium Phase Diagram

of temperature. We calculated the Gibbs energy of formation (ΔG° , in units of joules per mole of atoms) for the reaction $2/3 \text{ Fe}(\ell) + 1/3 \text{ Zr}(\ell) = 1/3 \text{ Fe}_2\text{Zr}(\text{s})$ to be:

$$\Delta G^\circ = -46442 + 9.8073T \quad (1)$$

where T is the temperature in degrees kelvin. The Fe_2Zr phase was then modeled with a general defect model recently developed for non-stoichiometric phases. This model was developed by Li and Pelton¹⁷ and is similar to the Wagner-Shottky model.¹⁸

The calculated liquidus curve (Fig. 5) is in reasonable agreement with the published diagrams (Figs. 3 and 4); however, as zirconium content increases after the Fe_2Zr phase, the calculated liquidus curve descends more steeply than the published ones. The liquidus of Fe_2Zr has not been firmly established. Because the rate of decrease of the liquidus temperature with composition on either side of the stoichiometric Fe_2Zr composition is determined by the same enthalpy of formation, the liquidus cannot descend much more steeply on one side than the other. We, therefore, believe our calculated liquidus to be reliable.

Both FeZr_4 and FeZr_3 have been proposed to be stable at high temperature.^{15,16} Our assessment of the evidence led us, in the present analysis, to consider only FeZr_3 , which was assumed to be stoichiometric. The following Gibbs energies of formation of FeZr_2 and

FeZr_3 from the liquid elements were derived: for the reaction $1/3 \text{ Fe}(\ell) + 2/3 \text{ Zr}(\ell) = 1/3 \text{ FeZr}_2(\text{s})$,

$$\Delta G^\circ = -38158 + 9.623T \quad (2)$$

and for the reaction $1/4\text{Fe}(\ell) + 3/4\text{Zr}(\ell) = 1/4\text{FeZr}_3(\text{s})$,

$$\Delta G^\circ = -42007 + 16.736T \quad (3)$$

As shown in Fig. 5, the eutectic reported at 928°C ,¹⁵ the eutectoid reported at 795°C ,¹⁶ and the peritectoid reported at 885°C ¹⁵ are well reproduced. In contrast to the results for the published diagrams, the calculated diagram shows FeZr_2 melting congruently at 998°C , with a eutectic at 995°C and 64.4 at.% Zr. That is, the melting is just at the limit between being congruent and incongruent. We find that ΔG° for the decomposition of FeZr_2 to Fe_2Zr and FeZr_3 is only about 800 J/g-atom at 700°C . Small changes in the Gibbs energies of the compounds cause large changes in the calculated eutectoid temperature. A set of Gibbs energies which yields a eutectoid at 775°C (see Fig. 3) can easily be obtained. However, this can only be incorporated into our phase diagram at the expense of changing another calculated invariant temperature (e.g., the 928°C eutectic, which is the melting point of FeZr_2 , or the 796°C eutectoid for FeZr_3) by about 10°C . Accurately measuring a eutectoid decomposition temperature is very difficult because of the very slow kinetics and the small driving force. The measurement technique (anneal and observe), involving relatively short annealing times, tends to give errors suggesting too high a eutectoid temperature.

The Gibbs energy of formation of Fe_3Zr was derived to give a peritectic temperature of 1482°C and a eutectic at 1335°C and 8.8 at.% Zr. For the reaction $3/4 \text{ Fe}(\ell) + 1/4 \text{ Zr}(\ell) = 1/4 \text{ Fe}_3\text{Zr}$,

$$\Delta G^\circ = -34636 + 6.422T \quad (4)$$

As shown in Fig. 5, our calculations predict that Fe_3Zr undergoes eutectoid decomposition below 1175°C . However, this calculation is very sensitive to small differences among the Gibbs energies of the solid phases. Within the error limits of the optimization, a eutectoid temperature cannot be calculated with any precision.

In sum, all the available data on the Fe-Zr system are generally well correlated by our optimized phase diagram, but some modifications are required to ensure thermodynamic consistency. Experiments are planned to test the reliability of our calculated liquidus.

2. Iron-Uranium-Zirconium System

To derive the Fe-U-Zr phase diagram, we calculated the excess Gibbs energies in the ternary liquid and solid phases from the values for the three binary systems. The method used was either the Kohler equation

$$G^E = (1 - X_{Fe})^2 G_{U/Zr}^E + (1 - X_U)^2 G_{Fe/Zr}^E + (1 - X_{Zr})^2 G_{Fe/U}^E \quad (5)$$

or the Toop equation

$$G^E = \left[\frac{X_{Zr}}{(X_{Fe} + X_{Zr})} \right] G_{U/Zr}^E + \left[\frac{X_{Fe}}{(X_{Fe} + X_{Zr})} \right] G_{Fe/U}^E + (1 - X_U)^2 G_{Fe/Zr}^E \quad (6)$$

These equations have been used extensively for ternary alloy systems.¹¹⁻¹³ The phases considered in the calculation of the ternary diagram are summarized in Table 1. We also assumed that no ternary compounds occur in the system, despite the preliminary report of the existence of such a compound.^{19,20} We are studying the possibility of such a compound.

Table 1. Phases Considered in Calculation of Fe-U-Zr Diagram

Phase	Comment/Assumptions
Liquid	Kohler interpolation was used.
Fe(γ)	Assumed stoichiometric.
Fe(δ)	Dissolves Zr in Henrian solution; no solubility of U.
Zr(β)	Dissolves U in Henrian solution; no solubility of Fe.
U(γ)/Zr(β)	Solid solution of U and Zr; dissolves small amount of Fe in Henrian solution; Toop model was used.
Fe ₂ U	Assumed stoichiometric.
FeU ₆	Assumed stoichiometric.
Fe ₃ Zr	Assumed stoichiometric.
Fe ₂ Zr	Non-stoichiometric in the Fe-Zr binary; assume no solubility of U.
FeZr ₂	Assumed stoichiometric.
FeZr ₃	Assumed stoichiometric.

In developing the Fe-U-Zr phase diagram, we were not certain what assumptions should be made about the mutual solubilities of Fe₂U and Fe₂Zr. Two cases were examined: the phases were assumed to be miscible or to have no mutual solubility. Both of these compounds have the same structure (similar to the Laves phase Cu₂Mg), and the atomic radii of Zr and U are similar. Hence, some mutual solubility is expected.

The liquidus surface of the calculated Fe-U-Zr phase diagram is shown in Fig. 6 for both assumptions. The liquidus surface for both assumptions is dominated by the Fe_2Zr and the $\text{U}(\gamma)/\text{Zr}(\beta)$ phase fields. Small phase fields of Fe_3Zr , FeZr_2 , $\text{Fe}(\gamma)$, and $\text{Fe}(\delta)$ are also seen. In comparing the phase diagram where Fe_2U and Fe_2Zr form an ideal solid solution with that where they are immiscible, we find that the calculated liquidus surface in the former case is up to 80°C higher in the vicinity of the line joining Fe_2U and Fe_2Zr . However, as the distance from this line increases, the changes rapidly become much smaller. In the vicinity of the saddle point at 996°C , for example, the liquidus temperature increases by only 2°C . In uranium-rich solutions (above 70% U), the liquidus increases by less than 5°C . In reality, Fe_2U and Fe_2Zr probably exhibit limited mutual solubility, and the true phase diagram lies between the two extremes. Work is planned to examine experimentally this region of the Fe_2U - Fe_2Zr connection.

B. Synchrotron Studies

A joint working group with representatives from CMT, the ANL Intense Pulsed Neutron Source, and the ANL Fuels and Engineering Division has been studying the feasibility of examining IFR materials by use of synchrotron radiation sources. Synchrotron radiation offers advantages over other existing analytical tools such as electron microscopy and neutron diffractometry. In particular, the powerful Advanced Photon Source (APS), under construction at ANL, has the potential to reveal compositional and structural information on a microscopic scale. This information is important to the understanding and prediction of intermetallic compound formation, local melting, and fuel redistribution in the IFR fuel system.

To test the feasibility of studying radioactive materials using synchrotron radiation, a fuel-cladding diffusion couple was examined at the National Synchrotron Light Source (NSLS) of Brookhaven National Laboratory. The diffusion couple, which was prepared at Purdue University, consists of Fe-Cr and Ni-Cr alloys mated with an unirradiated U-10 wt% Zr fuel sample. This couple was annealed at 700°C for four days. It was subsequently mounted and polished longitudinally and was thoroughly characterized using scanning electron microscopy (SEM) and energy-dispersive X-ray analysis. These examinations revealed complex diffusion zones between the alloys that are only fractions of a millimeter in width.

To prevent the spread of contamination from the diffusion couple, we designed and fabricated a brass holder that will allow X-rays to pass through a beryllium window. The holder allows for precisely orienting the diffusion interfaces and was sealed with silicone rubber. When assembled, the brass holder was placed in a custom-made fixture that was mounted to an NSLS specimen chamber. This fixture was designed to permit the X-ray beam to move incrementally across the diffusion zone, so that as different phase fields are traversed, new diffraction peaks are recorded. Previously, secondary phases were successfully identified in bulk IFR ternary fuel samples using the ANL Intense Pulsed Neutron Source. The same phase-identification technique was used to analyze the X-ray diffraction data from NSLS.

Fig. 6. Liquidus Surface of Calculated Fe-U-Zr Phase Diagram Based on the Assumption of (a) No Solid Solution between Fe_2U and Fe_2Zr and (b) Ideal Solution between Fe_2U and Fe_2Zr . All temperatures in $^{\circ}\text{C}$. E_n = ternary eutectic, e = binary eutectic, P_n = ternary peritectic, and p = binary peritectic.

On examination of the diffusion sample at NSLS, we found that even though the intermetallic phase regions in the diffusion zone were each only a few microns thick, the phases were distinguishable from the X-ray diffraction patterns obtained. In fact, a detailed analysis of the diffraction patterns has verified most of the phases predicted based on binary phase diagrams and SEM studies. Moreover, X-ray fluorescence results using a hard X-ray microscope at the NSLS were found to be qualitatively consistent with SEM results. The working group is planning additional studies at the NSLS using more powerful X-ray beam lines. At that time, an improved sample holder, which could potentially be used with plutonium-bearing samples, will be tested.

IV. FUSION-RELATED RESEARCH

A critical element in development of a fusion reactor is the blanket for breeding tritium fuel. We are conducting experimental and calculational studies with the objective of determining the feasibility of using lithium-containing ceramics (e.g., Li_2O , LiAlO_2 , Li_4SiO_4 , Li_2ZrO_3) as breeder material.

A. Tritium Transport in Single-Crystal LiAlO_2 (J. P. Kopasz and C. A. Seils)

1. Introduction

Lithium ceramics have received considerable attention because of their potential use as tritium breeding materials for fusion reactors. In particular, tritium transport in lithium aluminate has been investigated by several authors.²¹⁻²⁷ However, the mechanism of tritium transport and release from LiAlO_2 and the rate constants for diffusion in, and desorption from, LiAlO_2 are still uncertain. Tritium diffusivity values ranging over six orders of magnitude have been reported. In addition, the question as to whether diffusion or desorption is the rate-limiting mechanism is still being debated.

One reason for the wide range in reported values for tritium diffusion is that measurements were performed under conditions where desorption was actually the rate-controlling mechanism. The method for determining the diffusion and desorption rate constants has generally been to observe tritium release from a lithium ceramic and analyze the time dependence of the release process. The data are usually analyzed based on a diffusion, desorption, or a mixed diffusion-desorption model. However, complementary measurements that would indicate whether release is in the diffusion- or desorption-controlled regime are usually not made. Recently, Quanci²⁸ and Verrall²⁹ have employed mathematical tests based on the initial slopes of plots of the natural log of the fraction released versus time to determine whether release falls in the diffusion or desorption regime. However, the initial release behavior often does not follow the prescribed mathematics due to time delays in the response of the experimental apparatus.

It may be possible to overcome these difficulties simply by examining the tritium distribution within the solid. In the bulk, contributions from desorption are expected to

be small and should be limited to the near-surface region. Therefore, the concentration gradient in the interior should be less sensitive to the desorption rate constant, making it possible to obtain good values for the diffusion coefficients, even in the mixed diffusion-desorption regime.

Another factor contributing to the spread observed in the reported diffusivities is the presence of impurities in the ceramic materials. Impurities can alter the diffusivity by creating vacancies, interstitials, or other defects. In lithium aluminate, it has been postulated that tritium diffusion occurs via a lithium-vacancy tritium complex.³⁰ If this is the case, then impurities that affect the number of lithium vacancies should also affect the tritium diffusivity. The sensitivity of the tritium diffusivity to these impurities depends on the number of lithium vacancies caused by the impurity relative to those defects present in the pure material.

During this report period, we conducted isothermal anneal tests on single-crystal samples (approximately 1.5-mm dia) to determine the diffusivity of tritium in lithium aluminate. Large single crystals were used to minimize the contributions from desorption. Diffusion kinetics becomes more dominant as the crystal radius increases due to the decreasing surface area to volume ratio. However, the release may still be in the mixed diffusion-desorption regime. If so, obtaining the rate constants from the time dependence of the tritium release would be difficult. Therefore, the experimental approach was to section the sample after the anneal and determine the diffusivity from the tritium concentration profile within the sample. Further, to understand the effects of impurities on the tritium transport, the tritium profiles in pure and Mg-doped (0.3%) LiAlO_2 have been determined.

2. Experimental

Single crystals of pure and Mg-doped LiAlO_2 were supplied by the Commissariat à l'Énergie Atomique (CEA). The crystals were grown using the Czochralski technique, and the magnesium dopant level was determined to be 0.3% by spark source spectrometry.³¹ The samples were dried under vacuum at 800°C for 4 h and then packaged in aluminum capsules, which were cold-welded shut in an inert helium atmosphere glovebox. The capsules were irradiated in a cadmium-lined neutron activation tube at the University of Illinois, Urbana-Champaign, for 3 hours at a power level of 1500 kW. The cadmium-lined facility was employed to harden the neutron spectrum and ensure that the tritium production within the single crystal was spatially uniform. The calculated amount of tritium produced is approximately 1 μCi per single crystal.

The single crystals were transferred from the aluminum capsules to a sample holder in an inert atmosphere glovebox. The sample holder was designed with two compression fittings, which enabled isolation of the sample from the atmosphere while it was being transferred from the glovebox to the annealing apparatus. The annealing apparatus was preheated to the desired temperature (500-800°C), and a helium purge flow established. The sample was then introduced to the furnace and annealed for a specified time. The released tritium was reduced in a zinc bed, detected with a proportional counter, oxidized over a CuO bed, and then trapped in ethylene glycol. After the anneal, the sample was quenched by sliding it

out of the furnace into a cool zone in the apparatus, causing a drop in sample temperature of 200°C in about 3-5 min. Cooling continued until the sample reached room temperature.

The tritium profile was measured by stepwise dissolution of the sample by the following procedure. After cooling, the sample diameter was measured three times with a spring-loaded dial micrometer. The sample was then placed in a vial containing 1 mL of an HCl/HF acid solution (6 N HCl, 0.5 N HF), which was placed in a thermostatted bath at 50°C. After approximately 30 min in the bath, the vial was removed and 2 mL of water added. A 1 mL aliquot of this solution was removed and placed in a scintillation vial along with 18 mL of scintillation cocktail. The remaining solution was pipetted out of the vial and discarded. The vial and crystal were washed four times with 1 mL of water. The crystal was then removed from the vial and blotted dry, and its diameter was measured three times with a dial micrometer. The crystal was then placed in a vial with 1 mL of acid solution, and the above steps repeated until the crystal size was too small to handle. At that point the crystal was left in the acid solution overnight, then 2 mL of water was added to the acid solution, and a 1 mL aliquot taken for scintillation counting. From these measurements, tritium profiles within the crystals were obtained.

3. Calculations

Tritium release from single-crystal lithium aluminate can be modeled by assuming diffusion in the bulk with desorption occurring at the surface. The differential equations governing the tritium transport are

$$\frac{\partial C}{\partial t} = D \left(\frac{\partial^2 C}{\partial r^2} + \frac{2}{r} \frac{\partial C}{\partial r} \right) \quad (7)$$

for $0 \leq r < a$, and

$$\frac{\partial C}{\partial r} + \frac{K}{D} C = 0 \quad (8)$$

for $r = a$. In these equations, C = concentration at radial position r , K = desorption rate constant, D = diffusivity, a = crystal radius, and t = time. The solution to these equations for the case of a constant initial concentration C_i is given by Carslaw and Jaeger.³²

$$C = \frac{2hC_i}{r} \sum e^{(-D\alpha_n^2 t)} \frac{a^2 \alpha_n^2 + (ah - 1)^2}{\alpha_n^2 [a^2 \alpha_n^2 + ah(ah - 1)]} \sin(a\alpha_n) \sin(r\alpha_n) \quad (9)$$

where $h = K/D$ and α_n = roots of $a\alpha_n \cot(a\alpha_n) = 1 - ah$.

Although one can determine the diffusivity and desorption rate constant given the tritium concentration at the surface, the tritium concentration at the center, and the initial tritium concentration, a better method was to fit the total observed tritium profile to Eq. 9. This approach minimizes errors due to measuring the very small crystals left at the end of the dissolution. Diffusivities were obtained from the initial concentration (determined from the sum of the tritium collected in the ethylene glycol traps during the anneal and the tritium remaining in the sample determined by sectioning) and a minimization routine that fit the concentration data to Eq. 9.

4. Results and Discussion

Representative plots showing the observed tritium profile and the optimized fit of Eq. 9 to the data are shown in Figs. 7 and 8 for pure samples and in Figs. 9 and 10 for Mg-doped samples. The diffusivities and desorption rate constants obtained from these fits are plotted versus the inverse of the annealing temperature in Fig. 11.

Fig. 8.

Observed and Calculated Tritium Profile for Pure LiAlO_2 Annealed at 606°C for 64500 s ($D = 6.39 \times 10^{-9} \text{ cm}^2/\text{s}$, $K = 1.37 \times 10^{-7} \text{ cm/s}$)

Fig. 7.

Observed and Calculated Tritium Profile for Pure LiAlO_2 Annealed at 783°C for 4500 s ($D = 3.24 \times 10^{-8} \text{ cm}^2/\text{s}$, $K = 9.0 \times 10^{-6} \text{ cm/s}$)

Fig. 9.

Observed and Calculated Tritium Profile for Mg-Doped LiAlO_2 Annealed at 750°C for 18000 s ($D = 3.9 \times 10^{-8} \text{ cm}^2/\text{s}$, $K = 9.0 \times 10^{-6} \text{ cm/s}$)

Fig. 10.

Observed and Calculated Tritium Profile for Mg-Doped LiAlO_2 Annealed at 572°C for 61200 s ($D = 1.2 \times 10^{-9} \text{ cm}^2/\text{s}$, $K = 1.2 \times 10^{-6} \text{ cm/s}$)

Fig. 11.

Temperature Dependence of Rate Constants for Tritium Transport in Pure and Mg-doped LiAlO_2

The diffusivity was observed to follow a simple Arrhenius-type temperature dependence, with the diffusivity determined for the doped and undoped materials being the same within experimental error. This suggests that impurities which create lithium vacancies have little or no effect on tritium diffusion in lithium aluminate over the temperature range investigated. The calculated diffusivity using data from the pure and doped material is given by $\ln D = -2.6 \pm 0.7 - (1.50 \pm 0.07) \times 10^4/T$, yielding an activation energy of 125 ± 5 kJ/mol. The diffusivity determined in these samples is higher than most of the reported values in the literature. However, it is comparable to the values given by Botter²³ and Bruning et al.,²¹ which are several orders of magnitude above the other literature values. The activation energy is substantially larger than the 81 kJ/mol reported by Bruning.²¹

The desorption rate constants determined by the sectioning method exhibit some scatter. The scatter may be due to either experimental difficulties in determining the concentration at the surface or more complex surface desorption processes with different mechanisms contributing at different temperatures. The latter explanation would be in agreement with previous temperature-programmed desorption (TPD) experiments, which indicate several desorption mechanisms²⁷; however, more data are needed to clarify this issue. Magnesium doping appears to have some effect on the desorption rate constant. The doped material appears to exhibit slightly larger desorption rate constants than the pure material. For the doped material, the desorption activation energy calculated from our data is 93 kJ/mol, while for the pure material the activation energy is 111 kJ/mol. These values are in good agreement with activation energies reported from our earlier TPD work.²⁷

Using the values determined for D and K from these large single crystals, and the rule that desorption will be rate controlling when the ratio aK/D is less than or equal to one, we determined that for grain radii less than 100 μm , the tritium transport will be in the desorption-controlled regime (see Fig. 12). Thus, for in-pile tests where $a < 100$ μm , these results suggest tritium release should be desorption controlled. Diffusivities determined under desorption-controlled release conditions will be in error.

Fig. 12.

Ratio of Desorption to Diffusion Rates for Pure LiAlO_2 ($aK/D < 1$, desorption controlled; $aK/D > 10$, diffusion controlled)

B. Data Analysis of BEATRIX-II Dynamic Tritium Release Experiments
(J. P. Kopasz, H. M. Attaya,* and M. C. Billone†)

One of the key issues in designing a tritium breeding blanket for a fusion device is the tritium inventory. Dynamic tritium release experiments and tritium modeling efforts are being used to address this issue for ceramic tritium breeding materials. One such experiment is BEATRIX II, an International Energy Agency sponsored collaboration (Pacific Northwest Laboratory, U.S.; Japan Atomic Energy Research Institute; and Atomic Energy of Canada, Ltd.). It is intended to simulate tritium release from lithium oxide in a tritium breeding blanket during neutron irradiation. The tritium release is monitored in real time during changes in temperature or purge gas chemistry. From the changes in the monitored tritium release, we calculated changes in the tritium inventory. These results can then be compared with predictions from our tritium release models.

A set of mathematical models incorporating tritium diffusivity, desorption, adsorption, and solubility has been incorporated into the TIARA³³ and DISPL2³⁴ computer codes. These codes were developed at ANL and are used to calculate tritium inventory and release under steady state and transient conditions, respectively. The revised models were used to calculate the tritium inventory changes and tritium release curves under the conditions of the BEATRIX II experiment. The predicted inventory changes and those determined for the tritium release from the BEATRIX II experiment are given in Table 2. The model predictions are within about a factor of two of the observed inventory changes for temperature change tests when the experimental error due to uncertainties in the baseline is taken into account. This is a reasonable agreement, especially considering the low tritium inventories present in the samples (<1 ppm or 116 mCi). However, for the gas change tests, the model predictions differ from the observed inventory changes by several orders of magnitude. Part of this error is believed to be due to changes in tritium adsorption in the experimental system when the purge gas is changed.

A more stringent test on the tritium release models is to match the time dependence of the tritium release curves in the BEATRIX II experiment. The predicted tritium release curve and the observed curve were compared for a temperature decrease and a temperature increase of 91°C. The model predicts a large decrease in release rate for the temperature decrease (81%), followed by an increase to steady state. It also predicts a very large increase in the tritium release rate for the temperature increase (460%), followed by a decline to steady state. The data show a much smaller decrease (8%) and increase (14%) with the changes in temperature. Also, these data exhibit double peaks: a sharp peak believed to be associated with HT and a broad peak associated with HTO. The double peaks are believed to be due to system effects and not related to the kinetics of the release from lithium oxide. These system effects make it unlikely that the transient release curves will be reproduced by a model that does not include the mechanisms of HT/HTO adsorption and desorption that occur on the tubing and other components of the experimental apparatus.

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Table 2. Experimental and Calculated Tritium Inventory Change in BEATRIX-II Lithium Oxide Sample for Temperature and Hydrogen Pressure Changes in the Purge Gas

Temp., °C		H ₂ Press., Pa		Changes in Tritium Inventory, mCi	
Initial	Final	Initial	Final	Model	Exp.
598	562	150	150	+6.7	+28±54
562	513	150	150	+27.4	+26±27
513	559	150	150	-26.5	-24±19
559	596	150	150	-7.6	-1±36
596	513	150	150	+34.1	+42±10
513	596	150	150	-34.1	-55±22
599	508	150	150	+40.2	+75±5
508	598	150	150	-39.9	-85±15
594	511	150	150	+36.3	+63±13
511	594	150	150	-36.3	-88±13
602	510	150	150	+38.4	+80±10
510	602	150	150	-38.4	-100±10
596	596	150	0	+9.5	+15000
596	596	0	15	-4.0	-3000

End-of-life inventories are being determined for BEATRIX II samples. Inventories will be determined at several radial positions corresponding to several different temperatures. Our future work in this area will be directed toward predicting the end-of-life inventory, including a radial profile, for these samples and improving the model for gas change tests.

REFERENCES

1. R. K. Cole, Jr., D. P. Kelly, and M. A. Ellis, *CORCON-Mod2: A Computer Program for Analysis of Molten Core-Concrete Interactions*, U.S. Nuclear Regulatory Commission Report NUREG/CR-3920 (1984).
2. R. M. Summers, R. K. Cole, Jr., E. A. Boucheron, M. K. Carmel, S. E. Dingman, and J. E. Kelly, *MELCOR 1.8.0: A Computer Code for Nuclear Reactor Severe Accident Source Term and Risk Assessment Analyses*, U.S. Nuclear Regulatory Commission Report NUREG/CR-5531 (1991).
3. M. F. Roche, L. Leibowitz, J. L. Settle, C. E. Johnson, R. C. Vogel, and R. L. Ritzman, *Nucl. Technol.* **96**, 96 (1991).
4. J. E. Battles et al., *Nuclear Technology Programs Semiannual Progress Report, April-September 1991*, Argonne National Laboratory Report ANL-93/21, p. 10 (1993).
5. E. M. Levin and H. F. McMurdie, *Phase Diagrams for Ceramists, 1975 Supplement*, American Ceramic Society, Columbus, OH (1975).
6. P. Y. Chevalier, *J. Nucl. Mater.* **186**, 212 (1992).
7. R. G. J. Ball and M. A. Mignanelli, "The Calculation of Phase Equilibria of Oxide Core-Concrete Systems," presented at the OECD CSNI Specialist Meeting on Core Debris-Concrete Interactions, Karlsruhe, Germany, April 1-3, 1992.
8. L. Leibowitz, R. A. Blomquist, and A. D. Pelton, *J. Nucl. Mater.* **167**, 76 (1989).
9. L. Leibowitz and R. A. Blomquist, *J. Nucl. Mater.* **184**, 47 (1991).
10. L. Leibowitz, R. A. Blomquist, and A. D. Pelton, *J. Nucl. Mater.* **184**, 59 (1991).
11. C. W. Bale, A. D. Pelton, and W. T. Thompson, *F*A*C*T Guide to Operations*, McGill University/Ecole Polytechnique, Montreal (1985).
12. A. D. Pelton, in *Physical Metallurgy*, 3rd ed., Eds., R. W. Cahn and P. Haasen, North-Holland Publishing, New York, p. 328 (1983).
13. A. D. Pelton, in *Materials Science and Technology*, Vol. 5, Eds., R. W. Cahn, P. Haasen, and E. J. Kramer, VCH Weinheim, Germany, p. 1 (1991).
14. T. B. Massalski, Ed., *Binary Alloy Phase Diagrams*, 2nd ed., American Society for Metals, Metals Park, OH (1990).
15. D. Arias and J. P. Abriata, *Bull. Alloy Phase Diagrams* **9**, 597 (1988).

16. O. Kubaschewski, *Iron-Binary Phase Diagrams*, Springer-Verlag, New York, p. 175 (1982).
17. L. Li and A. D. Pelton, Ecole Polytechnique, Montreal, private communication (1992).
18. H. Schmalzried, *Solid State Reactions*, Academic Press, New York, pp. 35-51 (1974).
19. L. R. Kelman, in *Plutonium 1965*, Eds., A. E. Kay and M. B. Waldron, Chapman and Hall, London, p. 525 (1967).
20. C. M. Walter, L. R. Kelman, and S. T. Zegler, *Annual Progress Report for 1965, Metallurgy Division*, Argonne National Laboratory Report ANL-7155 (1966).
21. D. Bruning, D. Guggi, and H. R. Ihle, *Fusion Technol.* **1**, 543 (1982).
22. R. Clemmer et al., *The TRIO Experiment*, Argonne National Laboratory Report ANL-84-55 (1984).
23. F. Botter, in *Proceedings of the 14th SOFT, September 1986, Avignon, France*, Pergamon Press, New York, p. 1023 (1986).
24. K. Okuno and H. Kudo, *J. Nucl. Mater.* **138**, 357 (1986).
25. T. Kuraswa, H. Watanabe, G. W. Hollenberg, Y. Ishii, A. Nishimura, H. Yoshida, Y. Naruse, M. Aizawa, H. Ohno, and S. Konishi, *J. Nucl. Mater.* **141-143**, 265 (1986).
26. S. Tanaka, A. Kawamoto, M. Yamawaki, T. Terai, Y. Takahashi, H. Kawamura, and M. Saito, *Fusion Eng. Des.* **8**, 155 (1989).
27. J. P. Kopasz, C. A. Seils, and C. E. Johnson, *J. Nucl. Mater.* **191-194**, 231 (1992).
28. J. F. Quanci, Princeton University Dissertation, Department of Chemical Engineering (1989).
29. R. Verrall, *Proceedings of the International Workshop on Ceramic Breeder Blanket Interactions*, November 22-23, 1991, Clearwater, FL, p. 91 (1991).
30. J. P. Kopasz, S.-W. Tam, and C. E. Johnson, *J. Nucl. Mater.* **179-181**, 816 (1991).
31. N. Roux, Centre d'Etudes de Saclay, Commissariat à l'Energie Atomique, France, personal communication (1990).
32. H. S. Carslaw and J. C. Jaeger, *Conduction of Heat in Solids*, Clarendon Press, Oxford, p. 238 (1959).

33. M. C. Billone, H. Attaya, and J. P. Kopasz, *Modeling of Tritium Behavior in Li₂O*, Argonne National Laboratory Report ANL/FPP/TM-260 (1992).
34. G. K. Leaf and M. Minkoff, *DISPLI: A Software Package for One and Two Spatially Dimensioned Kinetics-Diffusion Problems*, Argonne National Laboratory Report ANL-84-56 (1984).

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