Some Oxide Ceramics as Reactor Materials

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Although the conventional advantages of ceramics lie in their favourable high temperature properties, reactor technology has not yet made use of these properties, mainly because of their uncertain behaviour under irradiation. The fabrication and properties of beryllia, uranium dioxide and thoria, and of mixed oxide systems, are discussed.

INTRODUCTION

Ceramics, principally carbides and oxides, are currently being considered for use in nuclear reactors as:

(i) Moderators. Graphite is a ceramic moderator. Beryllia is a better moderator, not sensitive to oxidation at high temperatures, which is an advantage over graphite.

(ii) Fuel-Element Material. The use of ceramic fuels is projected for high temperature systems for which metallic fuels are not available, or for other systems where ceramics are desirable replacements for metals, usually because of chemical inertness and structural stability.

(iii) Future uses may include canning material and control rod material.

Oxides have the advantage over carbides of not being sensitive to oxidation at elevated temperatures; coolant compatibility problems are therefore simplified.

The oxides considered here are principally beryllia, uranium dioxide, and thoria. Some discussion is included on wholly ceramic dispersion type fuel element systems.

BERYLLIA

Introduction

Beryllia is an attractive possibility as a moderator on nuclear grounds, and because of its refractory nature and chemical inertness, has special advantages for high temperature systems. The resistance to attack of dense material by liquid metals at moderate temperatures is also a desirable property for application in liquid metal-cooled or liquid metal-fuelled reactors.

In 1944, a nuclear power reactor containing pebbles of beryllia and uranium dioxide in a pebble-bed cooled by molten bismuth was proposed (Daniels 1957). A modified proposal for a high temperature, helium-cooled, beryllia-moderated power reactor to be built at Oak Ridge was accepted, and a research programme started (Daniels 1957; Howe 1956). The design consisted of beryllia tubes impregnated with uranium dioxide and suspended inside beryllia moderator bricks. The project was later abandoned because of insufficient experience in the fabrication of complicated ceramic bodies, and in the engineering use of ceramics. Present reactor uses are restricted to low temperature applications as reflector (David 1957) in the U.S.A.E.C. bulk shielding facility (swimming pool), the Low Power Water Boiler (LOPO) and the High Power Water Boiler (HYPO) and as moderator (McCreight and Sowman 1956), in the Los Alamos Homogeneous Aqueous Reactor (HAR) and in the Oak Ridge Aircraft Reactor Experiment (ARE) (Bettis et al. 1957). Although in the latter case, the moderator is cooled by liquid sodium, the sodium is enclosed in stainless steel tubing and does not come into contact with the beryllia. These applications are based on the nuclear properties of beryllia; advantage has not yet been made of its favourable high temperature properties.

The available data on the physical, chemical and mechanical properties, and on the fabrication and uses of beryllia, have been reviewed recently by White (1955).

Physico-chemical properties

Beryllia crystallises with a slightly distorted hexagonal wurtzite structure (Jeffrey et al. 1956). No phase changes have been observed up to the melting point of 2,570°C. The theoretical density is 3.025 gm. cm.$^{-3}$. Thermodynamically beryllia is a very stable oxide, with a free energy of formation of —140,000 cal. per mole. at 0°C.

Fabrication

The techniques of crucible production by slip-casting and sintering high-fired (1,700°C) beryllia are fairly well established (Norton 1947; Murray et al. 1954a). The latter workers describe the Harwell procedure in which beryllia containing 0.8 to 2 per cent. impurities was sintered at 1,700-1,800°C after slip-casting, resulting in zero porosity, translucent crucibles. During this study it was observed that the purer material sintered less readily. The practice of hot-pressing beryllia and other refractory oxides as developed at Harwell has also been described (Murray et al. 1954a & b; Murray 1954). Beryllia can be hot-pressed in graphite dies at 1,600-1,800°C and 2,000lb./sq. in. to 98 per cent. theoretical density.

The rigid purity requirements for nuclear reactor uses have directed attention to the fabrication of high purity beryllia, in which near theoretical density is much more difficult to attain. The fabrication by cold-pressing and sintering and by hot-pressing of a high purity beryllia has been studied at Harwell. (Livey et al. to be published). Small additions of calcium oxide and magnesium oxide aid densification, but pure beryllia does not sinter to high density, irrespective of surface activity of the original powder.

The status of beryllia production technology

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in the U.S.A., France and the U.S.S.R. was summarised in papers presented to the 1955 Geneva Conference (Kaufman and Kjellgren; Caillat and Pointud; Meyerson respectively, 1956).

In the U.S.A., a pure grade of Brush beryllia is calcined to 1,150° and hot-pressed in graphite dies at 1,600-2,300° at pressures of 200-500lb./sq. in., producing articles of near theoretical density. Alternately, beryllia is fused, crushed, ground and classified, before cold-pressing, slip-casting or extrusion. Densities achieved on sintering this material range from 2.0 to 2.6 gm. cm.$^{-3}$ (67-87 per cent. theoretical).

In France, a grade of oxide containing less than 0.04 per cent. impurities is used for briquette production by hot-pressing. Beryllium hydroxide is precipitated and calcined under very accurately controlled conditions, then hot-pressed in graphite dies at temperatures up to 1,900°C, and pressures of 2,000-2,500lb./sq. in. Briquettes of rectangular shape 50 by 100 by 100 mm. of mean density exceeding 2.90gm. cm.$^{-3}$ have been produced by this method.

In the U.S.S.R., a pure grade (less than 0.03 per cent. impurities) is fabricated by one of three methods. Tamping of a pre-calcined material (1,200-1,800°C) in graphite dies and sintering at 1,800° produces articles of density 2.1-2.2 gm. cm.$^{-3}$. Hot-pressing in graphite dies at 1,800°C and a pressure of 200-300lb./sq. in. produces bodies of density up to 2.75 gm. cm.$^{-3}$. Hot-pressing under vacuum at 1,850-1,900° and 30-300lb./sq. in. produces bodies of density 2.9 gm. cm.$^{-3}$, which are exceptionally pure and strong, although this is obviously an uneconomic production method.

In general, near theoretical density in highly pure beryllia can be produced readily only by hot-pressing, although for certain applications addition of oxides which assist sintering after cold-forming may be allowable. As hot-pressing is not at present an economical process for large scale production, considerable effort is now being directed to methods of attainment of comparable densities by cold-forming and sintering techniques without resort to additives.

Quirk et al. (1957) have studied the effect of calcination temperature of four grades of beryllia on the density attainable on hydrostatically pressing at 100,000lb./sq. in. and sintering in hydrogen at 1,450°. Calcination at 800° resulted in better than 95 per cent. density for beryllia containing 0.15, 0.25 and 1.0 per cent. impurities, although the purest sample (0.03 per cent. impurities) could be sintered only to 75 per cent. density. Higher calcination temperatures, particularly above 1,200°, reduced attainable densities in each case. The basis of this method of achieving high density after cold-pressing and sintering at such low temperature is in the highly surface-active nature of the low-temperature calcined material. Specific surface areas for 800°C calcined material were 30-80 m^2/g.; for 1,200° calcination, only 2-20 m^2/g. Variations with calcination temperature of the lattice parameter, which passes through a minimum at 1,000°, suggested that strain due to impurities might be important. The reason for the poor sintering behaviour of the purest product is not fully understood. Addition of 1 per cent. magnesia before calcination (to 800°) followed by sintering to 1,450° resulted in better than 90 per cent. theoretical density for all four powders. The high shrinkage on sintering of compacts pressed from very fine surface active powders places a serious size limitation on production of dense bodies by this method.

In the course of a study on the sintering properties of fine magnesia powders, preliminary results on the densification of surface active beryllia powders at temperatures as low as 800°C were obtained (Livey et. al. 1957). Beryllia powder prepared by calcination of beryllium hydroxide at 700° was heated to 800° in a Stellite die under a pressure of 10 tons/sq. in. Although the compact was cracked, the density was 90 per cent. theoretical, and the method is promising.

Thermal properties

Beryllia is outstanding amongst ceramic oxides in its excellent thermal shock resistance; the main contributing factor to this behaviour is the exceptionally high thermal conductivity. Recent determinations of thermal conductivity on slightly porous samples indicate a strong dependence on the method of fabrication, probably due to differences in pore size and orientation. This effect has been pointed out by Murray (1954), while comparing the conductivities of hot-pressed and slip-cast-sintered beryllia of the same porosities. The hot-pressed samples had a higher conductivity in each case. The thermal shock index $\left(\frac{k\sigma}{E\alpha}\right)$ was determined as a function of temperature for hot-pressed beryllia of 2 per cent. porosity and for slip-cast-sintered beryllia of 13 per cent. porosity. The index fell off sharply with temperature up to 1,000°, but was always approximately twice as high for the hot-pressed as for the sintered specimens.

Two recent determinations of thermal conductivity are summarised in Table 1. Those obtained at MIT (Kingery et al. 1954) were on slip-cast and sintered beryllia of density 2.7-2.86g. cm.$^{-3}$, and those at NBS (Ditmars and Ginnings, 1957) on similarly prepared material of density of 2.6 g. cm.$^{-3}$. Both sets are uncorrected for porosity.

TABLE 1.—THERMAL CONDUCTIVITY OF BERYLLIA

T (°C)	k(cal.sec.$^{-1}$ °C^{-1} cm^{-2} cm.)	
	MIT	NBS
100	.50	.38
200	.40	.27
300	.30	.20
400	.21	.16
500	.15	.13
600	.11	.11
700	.08	.097
750	.07	.092

The lack of agreement between these results may be due to differences in density, in starting material and method of fabrication, and in view of the marked difference in the rates of change of conductivity with temperature, probably in part due to inherent inaccuracies of one or both methods.

Mechanical and high temperature properties

The bend strength of hot-pressed beryllia at room temperature is usually in the range 20,000 to 30,000lb./sq. in. (White 1955), provided that excessive grain growth has not occurred. Loss of strength due to large grain growth in beryllia and other ceramics is a well-known phenomenon. Bend strengths of up to 40,000lb./sq. in. were noted for the dense beryllia prepared from active powder, for average crystal size in the sintered material of less than 20 microns (Quirk et al. 1957). The strength was observed to fall off sharply with increase in crystal size, being only 500lb./sq. in. at 200 microns. The importance of avoiding excessive grain growth, which occurs during high temperature sintering, but is not essential to it, is thus established. Sintering temperatures should be kept as low as possible.

The Russian work (Meyerson 1956) refers to exceptionally strong beryllia produced by vacuum hot-pressing, although no figures are quoted. There is little doubt, however, that there is much scope for improvement in the strength of beryllia by controlling the fabrication process.

It is generally agreed that the mechanical properties of beryllia assume lower values with rise in temperature. Figures summarised by White (1955) indicate an appreciable fall in bend strength, compressive strength and elastic modulus from room temperature to 1,000°C, although unpublished Harwell results (Scott) on hot-pressed material indicate no loss in strength at 1,000°C. Baroody et al. (1951) report that plastic flow occurs in beryllia at 1,000°C. Table 2 shows reported creep rates for beryllia of 30-35 per cent. porosity under a stress of 95lb./sq. in. (Sinott 1949).

TABLE 2.—CREEP OF BERYLLIA

Temperature °C	Creep Rate Per cent. hr. x 10^4
1095	3.1
1150	7.0
1205	14.6
1260	100

Brittan and Sibbitt (1947) give some results on the high temperature behaviour of hot-pressed beryllia. At Battelle, a bend strength of 15,000lb./sq.in. was measured at 1,370°C, and no deformation was observed at 1,630° for several hours under a load of 170lb./sq. in. "Norton" hot-pressed beryllia showed only a slight tensile strength decrease up to 1,460°. Slight expansions on reheating were noted. The high temperature behaviour of beryllia is undoubtedly sensitive to method of fabrication and extent and distribution of porosity; dense hot-pressed material probably creeps negligibly below 1,200°C, although more detailed work is needed.

The vapour pressure of beryllia is given as 3.6 x 10^{-4}mm. at 2,000°, and 2.4 x 10^{-3}mm. at 2,140° (Brittan and Sibbitt 1947). A more important limitation on the use of beryllia at elevated temperatures is in the reaction with water vapour, which becomes appreciable above 1,300°. (Livey and Murray 1956). Weight losses of 0.5-0.7 per cent. after 2½ hours at 1,400°C in an atmosphere containing water vapour at 90 mm. mercury pressure were quoted by Brittan and Sibbitt (1947).

Permeability to gases is an important property for possible reactor application. Low density products have considerable open porosity, and hence, are quite permeable. Open porosity has been observed to persist to quite low total porosities (Murray et al. 1954). However, it is reported that 96 per cent. density beryllia appears to be "impermeable" in ½in. thicknesses at room temperature, and that Degussa slip-cast beryllia is "gas-tight" at 1,000°C (Brittan and Sibbitt 1947). No quantitative data has been published.

Resistance to liquid metals

Beryllia is reported to be attacked by liquid sodium-potassium at 500°-600°C unless the specimen is very dense and highly "vitrified." (Kelman et al. 1950). Although no reliable figures are available, pure hot-pressed beryllia is probably unattacked by oxygen-free sodium or sodium-potassium at these temperatures. Oxygen in the liquid metal may be a major cause of beryllia attack.

Irradiation behaviour

Present results on the behaviour of BeO under irradiation are incomplete. Irradiation to 10^{19} n.v.t. has been observed to result in serious decreases in compressive strength, a 40 per cent. loss in thermal conductivity, and up to 1 per cent. expansion (Billington 1956). Preliminary British results (Bacon and Wilson 1955) showed that irradiation to 7 x 10^{20} n.v.t. caused lattice expansions of 0.088 per cent. in the c-axis and 0.033 per cent. in the a-axis. This change could be annealed out to the extent of 20 per cent. after eight hours at 500°, and 95 per cent. at 1,100°C. Very much more work is needed, particularly at neutron energies and irradiation temperatures where helium production due to (n, α) and (n, 2n) reactions may be of importance.

Summary

The present situation on the possibilities of the use of beryllia as moderator in high temperature reactor systems can be summarised as follows:—

(i) Beryllia is an efficient moderator from the reactor physics point of view.
(ii) It is compatible with coolant gases to high temperatures, provided water vapour is excluded.
(iii) There are no phase changes up to 2,570°C. There is probably little loss of strength up to 1,000°C or higher.

(iv) In common with most other ceramics, beryllia is brittle, which is an undesirable engineering property.

(v) The thermal shock resistance, although high compared with most other ceramics, may still not be adequate for parts subject to a high heat flux, or to high rates of heating and cooling.

(vi) Reactor design with points (iv) and (v) in mind can minimise these unfavourable properties.

(vii) Fabrication of dense material is at present carried out by hot-pressing, an expensive method for large-scale production. It may be possible in the future to use cold-forming and sintering methods.

(viii) Fabrication of complicated shapes in large sizes is difficult.

(ix) The behaviour under irradiation is not fully established. In particular, the possible serious decreases in thermal conductivity would lower thermal shock resistance.

URANIUM DIOXIDE AND THORIA

Introduction

The three oxides of most interest as fuel or fertile material are uranium dioxide, plutonium dioxide, and thoria. Little unclassified information is available on plutonium dioxide (see however Drummond and Welch 1957).

The uses of uranium dioxide as fuel element material may be described as follows:—

(i) Use in high pressure water systems, based on the excellent corrosion resistance of uranium dioxide.

(ii) Use as replacement for uranium metal fuel in gas-cooled reactors, allowing the temperature to be raised above that of the α-β phase change (660°C). Inertness towards possible metallic canning materials compared with uranium is also an advantage here.

(iii) Use in cermet dispersion-type fuel elements (Howe 1956; Weber and Hirsch 1956); the uranium dioxide may be simply a convenient form in which to disperse the uranium (e.g., Al-UO_2, Geneva reactor) or temperature conditions may be such that only the oxide is compatible with suitable matrix metals.

(iv) Possible use in ceramic dispersion type fuel elements for high temperature reactors.

Uses of thoria may be described as follows:—

(i) Use of thoria in suspension (aqueous, liquid metal), as powder, or in massive form as a convenient and inert form of thorium for breeding by the Th-U233 cycle.

(ii) Use as "inbuilt" breeder in UO_2-ThO_2 fuel elements.

(iii) Possible use in a ceramic dispersion type fuel element, with fissile and fertile material both dispersed in the matrix.

Fabrication

The technology of uranium dioxide and thoria has been reviewed recently (Johnson and Curtis 1956; Murray and Livey 1956). Both are typical ceramics, and can be fabricated by the usual ceramic techniques.

URANIUM DIOXIDE

This oxidises readily on heating in air to give successively non-stoichiometric oxides $UO_2 + x$ and U_3O_8 (with destructive increase in volume); U_3O_8 decomposes above 1,200°C to UO_2, containing a slight excess of oxygen. U_3O_8 formation must be avoided, and consequently UO_2 must be protected from oxidation during the sintering process and in use, except at temperatures above 1,200-1,300°C.

Stoichiometric UO_2 is difficult to sinter to high density in argon, but non-stoichiometric oxide sinters much more readily. The favourable sintering properties of British non-stoichiometric oxide ($UO_{2\cdot13}$) are thought to be due partly to the fine particle size and high specific surface of the material, and partly to the oxygen excess (Williams 1957). Conditions of preparation of UO_2 powder to achieve optimum sintering properties are very critical. Both stoichiometric and non-stoichiometric material can be hot-pressed to high density, but the latter is not suitable for the hot-pressing of large sizes, because reduction to UO_2 occurs in the outer layers adjacent to the graphite die, differential sintering rates are set up, and cracking occurs (Murray and Livey 1956).

Non-stoichiometric oxide is known to be appreciably more volatile at elevated temperatures. Results discussed recently by Williams (1957) indicate that stoichiometric oxide, although more difficult to fabricate, may be preferable for reactor use because volatility and structural changes at high temperatures are minimised.

THORIA

Thoria shows no non-stoichiometry, and is generally more difficult to sinter to high density than non-stoichiometric uranium dioxide. However, starting from very finely ground, pure thoria (99.9 per cent.), zero-porosity translucent crucibles have been prepared at Harwell by slip-casting and sintering at 1,825°C (Murray et al. 1953).

Striking improvements in sintering behaviour have been obtained by addition of ½ w/o calcium oxide before sintering at 1,700°C. The density attainable was improved from 86 per cent. to 96 per cent. of the theoretical by this means (Johnson and Curtis 1954). Effort has recently been directed to obtaining this effect by the use of other additives which on nuclear grounds may be less undesirable. Calcium fluoride, strontium oxide, and vanadium pentoxide have similar, but smaller effects, than calcium oxide. Addition of beryllia is only very slightly beneficial; uranium dioxide retards sintering (Curtis and Johnson 1957; Arenberg et al. 1957).

Some properties of uranium dioxide and thoria are presented in Table 3, with corresponding properties of beryllia included for comparison. Figures quoted on fabricated material are representative ones for material of at least 90 per cent. theoretical density.

TABLE 3.—PROPERTIES OF BeO, UO_2 and ThO_2 (a, b, c).

	BeO	UO_2	ThO_2
Crystal structure	Wurtzite (hexagonal)	Fluorite (cubic)	Fluorite (cubic)
X-ray density ($g.cm^{-3}$)	3.025	10.95	10.15
Melting point (°C)	2570	2800	3200
Colour	White	Brown-red	White
Thermal conductivity ($cal.sec^{-1}$ $cm^{-2}cm$ $°C^{-1}$			
20°C	0.5-0.6	0.02	0.02
1000°C	0.05	0.008	0.007
Modulus of elasticity (lb./sq. in.) 20°C	50×10^6	25×10^6	21×10^6
Coefficient of thermal expansion ($°C^{-1}$) 0-1000°C	9.3×10^{-6}	10×10^{-6}	9×10^{-6}
Bend strength 20°C (lb./sq. in.)	20,000	16,000	12,000

(a) White, 1955; (b) Johnson, et al, 1957; (c) Livey, 1957.

Discussion

One of the most serious disadvantages of uranium dioxide as a fissile material, which is necessarily subject to high heat fluxes, is its low thermal conductivity and hence sensitivity to thermally induced stresses. The advantages of dispersion in a high conductivity material (metallic or beryllia), particularly for highly rated fuel elements, are thus apparent. However, massive uranium dioxide as fuel is being intensely investigated in Canada, U.S.A. and the U.K., and it seems likely that the low thermal conductivity will not prove as big a disadvantage as might be supposed.

Thoria, as fertile material, may not be subject to such severe thermal stresses as fissile uranium dioxide but the low thermal conductivity may be a disadvantage here also. Chemical processing of sintered thoria is difficult, and it may be advantageous from this point of view to use cold-consolidated powder, which could be processed more readily. However, heat production during breeding, combined with low thermal conductivity, may result in high centre temperatures and sintering. The advantages of avoiding sintering in thoria for future chemical processing are shown in the time required for nitric/hydrofluoric acid dissolution of thoria calcined to various temperatures (Curtis and Johnson, 1957), given in Table 4.

TABLE 4.—DISSOLUTION OF THORIA.

Calcination Temperature (°C)	Dissolution Time (hours)
700	1½
1,200	2½
1,630	5½

CERAMIC DISPERSIONS

Introduction

If reactor requirements are such that a ceramic oxide fuel is desirable, there are two major alternatives:—

(i) Use of UO_2 fuel, natural or slightly enriched, with its attendant disadvantage of low thermal conductivity and probable cracking during use or

(ii) use of a dispersion type fuel, in which enriched UO_2 is dispersed in a ceramic matrix.

The matrix should have a higher thermal conductivity than uranium dioxide, should be capable of fabrication to low porosity, should be reasonably stable under neutron irradiation, and should not react with uranium dioxide under fabrication or operating conditions.

Some possible systems will now be considered.

Uranium dioxide-thoria

The fabrication of prototype Borax IV fuel elements from this material has been described previously. (Handwerk and Noland, 1957). This is not a true dispersion type system, because uranium dioxide and thoria form a complete series of solid solutions (Lambertson, et. al. 1953). The equilibrium state of uranium dioxide-thoria bodies is thus a uniform solid solution, which forms readily during fabrication. There is no improvement in thermal conductivity over uranium dioxide; moreover, these mixtures do not sinter as readily as either thoria or non-stoichiometric uranium dioxide.

However, the system has some advantages, viz.:—

(i) UO_2-ThO_2 can be fabricated by the air-sintering of U_3O_8-ThO_2 mixtures; air-sintering is not possible with UO_2.

(ii) The ThO_2 is present as an "in-built" breeding material.

(iii) UO_2-ThO_2 bodies are stable in air; excess oxygen can be accommodated interstitially in the solid solution fluorite lattice. For less than 50 w/o UO_2 in the solid solution, no additional phases have been observed on oxidation (Anderson, et al. 1954).

Beryllia-thoria

This system has the advantage of a high conductivity matrix material; no solid solutions or compounds are known, and thus a dispersion of uranium dioxide in a beryllia matrix should be stable if liquid phase formation is avoided. A eutectic in this system has been observed at 2,150° (Lang, et al. 1956). A liquid phase will begin to form at this temperature, setting the upper limit for use of beryllia-uranium dioxide bodies at somewhat less than 2,150°.

Extensive work was carried out on beryllia-uranium dioxide dispersions at Battelle, Oak Ridge and Argonne as part of the Daniels Pile Project (Howe, 1956). Brittan and Sibbitt (1947), describe the fabrication of 2 w/o and 10 w/o dispersions of uranium dioxide in beryllia. These were made by hot-pressing a mixture of fine beryllia and U_3O_8 powders at 1,800°C, and 95 per cent. theoretical density could be achieved this way. The U_3O_8 breaks down to uranium dioxide which is present in the compact as a fine micron-size dispersion. The thermal conductivity, melting-point and bend-strength were reported to be the same as for hot-pressed beryllia.

These bodies were unstable in air, and crumbled to powder at 700° due to U_3O_8 formation. Above 1,000°C volatilisation of oxidised specimens occurred. Impregnation of porous beryllia with UO_3 and subsequent decomposition is also described.

Preliminary results on the behaviour of beryllia-uranium dioxide bodies under irradiation have been reported (Billington, 1956). After one in 2 x 10^5 uranium atoms had fissioned the thermal conductivity of 2 w/o and 10 w/o uranium dioxide in beryllia had dropped by a factor of six without signs of saturation. The linear dimensions increased by one per cent. and the compressive strength and elastic modulus decreased by 30 per cent. However, severe damage from fission product recoil atoms to the beryllia matrix would be expected in these fine (less than 2 microns) dispersions of UO_2.

Recoil atom escape and also the volume per cent. damage to the matrix in such systems generally can be decreased by increasing the particle size of the dispersion to 50-100 microns. (Weber and Hirsch, 1956). Methods of producing uranium dioxide grains of such sizes have been summarised by Johnson, et alia (1957).

Beryllia-uranium dioxide

The additional advantages of this dispersion are:

(i) thoria is present as an "inbuilt breeder" and

(ii) bodies made from this material are stable to oxidation.

This system has been studied at Harwell (Reeve and Williams, 1958). Because a eutectic has been observed in the beryllia-thoria system at approximately 2,200°C and almost the same beryllia composition as for the beryllia-uranium dioxide eutectic, viz., 60-70 mol.% BeO (Dukelski, 1906), and because uranium dioxide and thoria form a complete series of solid solutions, it was expected that the ternary system would behave as a pseudobinary system. This has been confirmed. The maximum allowable temperature for use of beryllia-uranium dioxide-thoria would be somewhat less than 2,150°, because it has been observed that at 2,150° a liquid phase is beginning to penetrate the beryllia matrix. This is obviously undesirable in a dispersion-type fuel element. Fabrication of such dispersions in beryllia to near theoretical density is easier to achieve than for pure beryllia.

Other oxides as matrix

Other possible matrix oxides are magnesia, alumina, silica, and zirconia. Thermal conductivities are inferior to beryllia; possible advantages are superior resistance of the matrix to neutron irradiation, and superior retention of fission product gases. At present not enough data is available on which to base a choice. Some data on fission product retention in magnesia-uranium dioxide and alumina-silica-uranium dioxide bodies have been discussed by Williams (1957). Dispersions of 100 micron uranium dioxide in magnesia of 75 per cent. theoretical density released only 20 per cent. of the xenon produced at 800°C. However, as magnesia and uranium dioxide have some mutual solid solubility, this is not an ideal dispersion system since solution may occur slowly during high temperature use. Xenon release from a 100 micron uranium oxide dispersion in an impervious alumina body containing some silica bond was negligible at 800°C. However, uranium dioxide was found to be present in solution in the glassy bond, which is not desirable in a dispersion type fuel.

Silica and zirconia both suffer from phase changes, and there is little to recommend them at present, although fused silica is reported to be stable under irradiation (Simon, 1957).

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