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Knolls
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Irradiation Behavior of Dispersion Fuels

D. W. White

A. P. Beard

A. H. Willis

October 1, 1957

Operated for the
United States Atomic
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General Electric Company
KNOLLS ATOMIC POWER LABORATORY
Schenectady, New York

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D. W. White
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ABSTRACT

The dispersion fuel concept offers a method for minimizing irradiation-induced property changes in nuclear fuel systems by localizing fission damage into regulated domains within the system. The system is thereby provided with a structural framework, or matrix, whose properties are relatively unaffected by the dynamic and impurity effects of fission fragments. Dispersion-type fuel systems whose irradiation behavior has been, or is being, investigated in the United States are listed.

The extent of property retention in a dispersion fuel alloy is dependent on the materials selected for the fuel-bearing and matrix phases and on their microstructural arrangement in the alloy. The consequences of varying the materials and fuel-phase particle size and volume fraction are discussed. From experimental observations it is shown that, by proper control of these parameters within the bounds of existing technology, the dispersion concept does provide improvement in property retention over alternate alloy design concepts.

However, gaining property retention by the use of the dispersion principle requires the acceptance of additional sources of stress and greater irradiation damage in localized regions. A superior fuel system is obtained only when these effects do not limit its performance. However, the flexibility in design, fabrication, and selection of materials offered by the dispersion concept serves to combat these potential limitations.

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IRRADIATION BEHAVIOR OF DISPERSION FUELS*

D. W. White, A. P. Beard, A. H. Willis

I. DEVELOPMENT OF THE DISPERSION FUEL CONCEPT

A continuing objective in the development of fuel materials for use in heterogeneous reactors is the improvement of their physical and chemical properties and the stability of these properties under irradiation. Desirable materials have sufficient strength and ductility to resist failure under the external and severe internal thermal stresses; possess good thermal conductivity to transfer heat to the coolant and to keep thermal stresses to a minimum; and will not corrode if exposed to the coolant. Many fuel materials have adequate initial properties, but these properties change severely under irradiation if the fission products become distributed homogeneously. Uranium metal, which becomes completely embrittled with only small amounts of fission products, ^{1,2} provides an extreme example of such property instability.

Another objective in the development of fuel materials is to minimize dimensional and volume instability resulting from irradiation. Uranium dioxide is an example of a material with good stability in this respect, but it lacks the desirable properties and characteristics mentioned in the previous paragraph.

The dispersion fuel concept, which has been discussed by Howe and Weber,³ Weber and Hirsch,⁴ Howe,⁵ and Billington,⁶ offers a potential technique for achieving both objectives by localizing the fission damage into regulated domains within the fuel system. A dispersion type of fuel alloy is a two-phase alloy with particles of the fuel-bearing phase dispersed in a well-behaved metal matrix. In accordance with the dispersion fuel concept, the alloy ideally should be so designed and fabricated that a continuous network of fission-product-free metal is maintained throughout its service in a reactor. In practice, however, there are fuel alloys with a two-phase, dispersion type of structure in which fission products are found throughout the matrix as a result either of their escape from the fuel-bearing phase during irradiation or of the presence of a small uranium content in the matrix phase. To assess the merits of the dispersion concept fully, we shall consider the behavior under irradiation of both types of alloys - those with heterogeneous and those with homogeneous distribution of fission products.

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*This paper was presented at Paris Conference on Fuel Element Technology, November 18-22, 1957.

A major feature of the dispersion fuel concept is the flexibility that it offers in the selection of the constituent materials and in the microstructural arrangement of these constituents to optimize the properties and behavior under irradiation. The relation of microstructure to property stability and mechanical response of dispersion fuel systems in irradiation will be treated in detail.

The flexibility in the selection of materials for the dispersed, fuel-bearing phase is illustrated by the number of uranium compounds of potential interests which are listed in Table 1. Some of the factors to be considered in choosing a particular uranium compound for a dispersion fuel system will be discussed throughout this report. Foremost among these factors, of course, is the irradiation stability of the compound itself at high temperature.

Materials for the matrix phase are selected on the basis of their physical and chemical properties, since these properties will, to a large measure, govern the properties of the bulk alloy. Such properties may include ductility, high-temperature strength, thermal expansivity, thermal conductivity, and corrosion resistance. One should also consider allotropic transformations within the temperature range of operation and whether the presence of such a transformation will have objectionable effects. Rate of reaction or interdiffusion with the dispersed phase at fabricating and operating temperatures should not be overlooked, as this factor can affect properties even in the absence of irradiation. Of course fabricability, cost, nuclear cross section, and other factors must be considered as with any reactor core material.

Table 2 lists some dispersion-type fuel systems whose irradiation behavior has been, or is being, investigated in the United States. Included in this table are two combinations representing what are commonly referred to as aluminum-uranium alloys and zirconium-uranium alloys. The former, which are used in the fuel elements for the Materials Testing Reactor (MTR), consist microstructurally of a dispersion of the compound UAl_4 in an aluminum matrix.* The latter is normally comprised of UZr_2 particles dispersed in zirconium;* this is one of several fuel systems employed in fuel elements for naval reactors. Table 2 also lists graphite and some ceramic materials as matrix constituents; the dispersion fuel concept can be applied to an all-ceramic system as well as to an alloy. Many of the data and observations on the systems shown in the table are either preliminary or too fragmentary to be discussed; data will be presented where they appear to be significant

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*In these cases, the matrix actually contains a small fraction of a per cent of uranium in solid solution.

to the evaluation of the dispersion concept. In addition, critical use will be made of significant information on the irradiation behavior of dispersion alloys of boron and its compounds. The n, α reaction of B^{10} is analogous to the fission of uranium in that it creates two fragments, He^4 and Li^7 , (with an energy release of about 2.8 Mev) which serve as counterparts of fission fragments.

TABLE 1. Fissile Materials^{4,21}

<u>Material</u>	<u>Density, gm/cc</u>	<u>V_r^*</u>	<u>Melting Point, °C</u>
U	18.9	1.0	1120
UAl ₂	8.1	0.35	~1590
UAl ₃	6.7	0.26	1320
UAl ₄	6.0	0.22	730
UBe ₁₃	4.37	0.15	2000
UC	13.6	0.69	2350-2400
UC ₂	11.7	0.56	2450-2500
UO ₂	10.96	0.53	2750**
U ₆ Fe	17.7	0.91	815(a)
UFe ₂	13.2	0.48	1235
UN	14.3	0.71	2650 $\pm$ 100
UPb	14.5	0.41	1280
U ₃ Si	15.6	0.77	930(b)
U ₃ Si ₂	12.2	0.59	1650
UZr ₂	10.3	0.31	600(c)

*Uranium content per unit volume relative to density of uranium.

**Determined by L. G. Wisnyi and S. W. Pijanowski.³³

(a)Peritectic temperature.

(b)Peritectoid temperature.

(c)Disordering.

TABLE 2.. Dispersion Fuel Systems

<u>Dispersed Phase</u>	<u>Matrix Phase</u>
UO ₂	Stainless steel (austenitic and ferritic), Fe, Nichrome, SiC-Si, MoSi ₂ , Nb, Al, Zr, Zircaloy, graphite, ThO ₂ , BeO, Al ₂ O ₃ , SiO ₂
U	Mg, Th, Zr, Zircaloy
UAl ₄	Al
UZr ₂	Zr, Zircaloy
UC	Stainless steel, Zr, Zircaloy
UN	Stainless steel, Zr
U ₃ Si	Zr
U ₆ Ni	Zr
U ₂ Ti	Zr

The foregoing remarks on the potential merits of the dispersion fuel concept should be tempered by pointing out that there are also potential limitations. The very fact that the fission events and fission products are localized will give rise to local internal stresses. The effect of these stresses on the mechanical response of a dispersion fuel alloy in service will be appraised (pp. 34-38). Volume stability may also be affected, and this possibility will be discussed later.

II. MICROSTRUCTURAL GEOMETRY AND MECHANICAL RESPONSE UNDER IRRADIATION

A basic premise of the dispersion fuel concept is that the retention of the preirradiation properties after irradiation is a desirable characteristic in a fuel element and that this objective can be accomplished by controlling the microstructure of the fuel system to maintain a continuous matrix with no fission fragment damage. The following will be discussed: (a) the influence of the various microstructural parameters on fission product distribution in a dispersion-type system and, in turn, on properties of the system; (b) the properties of some irradiated systems; (c) sources of stress in dispersion fuel systems; and (d) the mechanical response resulting from the interplay between stresses and properties under conditions of fuel element operation.

Microstructural Parameters and Fission Product Distribution

Although the fission fragments are born in the uranium-bearing, particulate phase of a dispersion fuel system, some of them will recoil into the matrix phase. (The average recoil range of fission fragments in various materials is $\sim 10\mu$; for the $B^{10}(n,\alpha)Li$ reaction,

the alpha particles have about one-third of this range, while the lithium fragments have a still somewhat shorter range; see Table 3) Many of these impurity atoms are large and incompatible with most of the suitable matrix materials. Consequently the properties of the contaminated region around particles are changed, usually in an undesirable way. To maintain a continuous matrix free from impurities, it is necessary to space the fuel-bearing particles far enough apart that the damaged matrix zones do not touch or overlap.

In Figure 1 are pictured the damaged zones in an irradiated fuel system of stainless steel - UO_2 dispersion; the two UO_2 particles shown are sufficiently spaced to provide a web of undamaged stainless steel between them. This photograph also illustrates the potential improvement offered by the dispersion concept with regard to retention of chemical properties and corrosion resistance. The fission-product-free matrix region was far less sensitive to the metallographic etchant than was the contaminated zone.

TABLE 3. Fragment Recoil Ranges in Various Materials

Material	Average Range					
	For U^{235} Fission Fragments*			For Alpha Particles from B^{10} "Fission"		
	mg/cm ²	μ	in. x 10 ⁴	mg/cm ²	μ	in. x 10 ⁴
U	12.6	6.8	2.7	—	—	—
UO_2	10.0	9.4	3.7	—	—	—
Zr	5.8	9.1	3.6	—	—	—
Fe	5.2	6.7	2.6	—	—	—
Al	3.7	13.7	5.4	1.22	4.8	1.9**
Cu	—	—	—	2.0	2.0	0.9**
Ag	—	—	—	1.8	1.5	0.6**
Pb	—	—	—	2.54	2.0	0.9**
B^{10}	—	—	—	—	5.1	2.0‡
Stainless Steel	—	—	—	—	2.5	1.0‡

*Reference 4.

**Reference 22.

‡Reference 16.

← 50 μ →

Irradiated UO₂ - Stainless Steel Dispersion,
Showing Zones of Fission Fragment Damage
in Steel Matrix Around UO₂ Particles.

(Barney and Wemple, KAPL)

Figure 1

KS - 17732

The microstructure of an irradiated dispersion fuel alloy can be uniquely described by specifying the volume fraction and particle size of the fuel-bearing phase and the distances which the fission fragments will penetrate either phase, as will be shown below. These microstructural parameters can be related to thickness of undamaged matrix between particles. Setting this thickness at a value of zero specifies the limiting conditions for maintaining a continuous undamaged matrix.

While the web thickness of undamaged matrix may be an important variable particularly if this thickness is small, it is not sufficient to describe how the alloy will respond to imposed stresses; additional measures of the extent of property change are needed. One such measure is the volume proportion of matrix remaining undamaged; this factor will be described in terms of the microstructural variables.

If it is reasoned that fractures will develop locally in the most highly damaged regions of the matrix, then the average fission product content of these damaged regions may be a significant quantity. An average concentration per unit volume of total matrix may be equally as suitable a measure of property change if local damage is not limiting. The relation of these two factors to microstructural geometry will be presented, after we have first shown the particle size dependence on fraction of fission fragments escaping into the matrix.

Next, the influence of particle size, particle volume fraction, and fission fragment recoil distance on distribution of fission products in the matrix will be shown qualitatively. The significance of variations in impurity damage to the selection of matrix materials will be discussed. Finally, it will be pointed out how fission product concentration is affected by the choice of material for the fuel-bearing phase and by fuel depletion or burnup.

This analysis, the details of which are presented below, cannot be used as a precise formula for the design of dispersion fuel alloys because it is based on several simplifying assumptions. However, it may serve as a qualitative guide for optimizing property stability through the control of microstructure. Experimental observations to test the analysis will be presented later.

Conditions for Undamaged Matrix between Particles. Equations have been derived (Harrison⁷ and Kernohan⁸) to describe cubical arrays of spherical particles of constant diameter. Such equations are adequate for dilute fuel concentrations, but they can be readily revised

for close-packing arrays to provide relationships suitable for high as well as low fuel concentrations. With a close-packing array of spherical particles of uniform diameter, the distance d between particles is

$$d = \left[\left(\frac{\pi}{3\sqrt{2}V_f} \right)^{1/3} - 1 \right] D$$

where D is particle diameter and V_f is volume fraction of the fuel-bearing phase. Depending upon the size and quantity of the fuel particles and also the distance λ_m that the fission fragments penetrate the surrounding matrix (which depends, of course, upon the particular matrix material selected), the intervening matrix may or may not be continuously damaged. This is indicated schematically in Figure 2A for two different particle diameters, D_1 and D_2 , and graphically in Figure 2B. The distance between adjacent particle surfaces has been subdivided into two parts: $2\lambda_m$, the thickness of the damaged matrix zones around both particles, and d' , the thickness of the intervening region free of fission products. For fixed values of fuel volume fraction and fission fragment recoil range, inter-particle distance is seen to vary linearly with particle diameter. Hence, specifying that there be intervening undamaged matrix ($d' > 0$) places a lower limit on permissible particle size.

These relationships can be expressed in terms of the ratios $D/2\lambda_m$ and $d'/2\lambda_m$, as follows:

$$D/2\lambda_m = \frac{1 + d'/2\lambda_m}{\left[\left(\frac{\pi}{3\sqrt{2}V_f} \right)^{1/3} - 1 \right]}$$

This is exhibited graphically in Figure 3. Both Figures 2B and Figure 3 indicate how, for a fixed value of λ_m , particle size must be increased when volume fraction of the fuel constituent is increased, if a given thickness of undamaged matrix is to be maintained. (Note that volume fraction occupied by spheres of equal size when they are closely packed is 0.74.)

Proportion of Matrix That Is Undamaged. The extent of property change of the dispersion alloy will be influenced by the amount of matrix becoming contaminated with fission products. The proportion of damaged matrix to total matrix is given by V_{dm}/V_m , where V_{dm} and V_m are the respective volume fractions of damaged matrix and total matrix in the alloy; the expression is related to V_f , D , and λ_m as follows:

$$V_{dm}/V_m = \frac{V_f}{1 - V_f} \left[\left(1 + \frac{1}{D/2\lambda_m} \right)^3 - 1 \right]$$

(A) FOR CONSTANT VOLUME FRACTION OF PARTICLE PHASE

(B) FOR VARIOUS VOLUME FRACTIONS (V_f) OF PARTICLE PHASE

EFFECT OF PARTICLE SIZE AND SPACING ON UNDAMAGED
DISTANCE BETWEEN PARTICLES

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Figure 2

The remaining portion of the matrix, the undamaged portion, will be $1 - V_{dm}/V_m$, and its relation to particle size and volume fraction of fuel is shown graphically in Figure 4. Again, the advantage of large particle size is observed, but with respect to undamaged matrix this advantage diminishes with increasing particle size. The points on the dotted line in this graph represent the condition for $d' = 0$. Therefore, the region of interest for dispersion fuel design is to the right of this line.

Figure 3

Fission Product Content in Matrix. Not only will the alloy properties be affected by the proportion of matrix volume becoming damaged but also by the degree of damage within the affected region. The fraction, E_{fp} , of fission fragments escaping from each fuel-bearing particle increases as particle size is reduced and as fission fragment recoil range in the fuel phase, λ_f , is increased. The fraction changes in accordance with the following relationship, reported by Weber and Hirsch:⁴

Figure 4

$$E_{fp} = 3/4 \left(\frac{2\lambda_f}{D} \right) - 1/16 \left(\frac{2\lambda_f}{D} \right)^3,$$

which is given graphically in Figure 5. All of the fission fragments escape from the particle when the particle diameter is equal to or smaller than the recoil range ($D/2\lambda_f \leq 0.5$). Because of differences in recoil range λ_f , there will be complete escape of fission fragments from UO_2 particles up to 9.4μ diameter and from uranium metal particles $<6.8\mu$ as was shown in Table 3. The fuel particles neither gain nor lose atoms when $E_{fp} = 0.5$, a value which corresponds to a $D/2\lambda_f$ ratio of about 1.4, as indicated by the dotted line in Figure 5. The significance of this factor on irradiation stability will be discussed later.

Figure 5

The fission fragments escaping from the particle come entirely from within a thickness, λ_f , of the particle surface. The depletion from this zone, E_{fp} , decreases with increasing particle size in accordance with the relationship

$$E'_{fp} = \frac{E_{fp}}{1 - \left(1 - \frac{2\lambda_f}{D}\right)^3}$$

which is also plotted in Figure 5. The value for E'_{fp} levels off at 0.25 for large values of $D/2\lambda_f$; it is synonymous with the value for E_{fp} when particle radius is equal to or less than the recoil range ($D/2\lambda_f \leq 1$).

FISSION PRODUCT CONCENTRATION IN MATRIX: RELATION TO SIZE AND VOLUME OF FUEL PARTICLES AND TO DISTANCE BETWEEN DAMAGED MATRIX ZONES

Figure 6

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The fission product content per unit volume of damaged matrix, M_f , is affected by two competing factors. While the number of fission fragments escaping into the matrix (for a given burnup of all atoms in the dispersion alloy)* increases as the particle size is reduced, the volume of the matrix region in which they become distributed also increases. The net effect is given by the relationship

$$M_f = \frac{E_{fp}}{\left(1 + \frac{2\lambda_m}{D}\right)^3 - 1}$$

which is represented by the solid line in Figure 6 for the simplified case of $\lambda_f = \lambda_m$.

*While E_{fp} has been defined as the ratio of the number of fission fragments escaping to the number produced per particle, it can be readily shown to be equivalent to the ratio of the total number of fission fragments released to the matrix to the total number produced throughout the alloy.

It can be concluded from this graph that the fission product concentration in the damaged portion of the matrix is greater in a coarse particle alloy than in a fine particle alloy. However, this is offset by the greater proportion of undamaged matrix in the coarse particle material, as previously shown in Figure 4. A simplified measure of the net effect on property alternation of the over-all matrix phase may be the average fission product content in the total matrix, M_f^i , which is given by

$$M_f^i = E_{fp} \left(\frac{V_f}{1 - V_f} \right).$$

This is plotted (as the dash-dot lines) against particle size in Figure 6 for various volume fractions of fuel phase;* here again λ_f and λ_m are taken to be equal for simplicity. In contrast to the damaged matrix region, the matrix as a whole should, under the assumptions made,** suffer less property change when the fuel-bearing particles are large than when they are small. For fixed particle size and fixed recoil range, fission product content in the matrix increases as the volume fraction of fuel is raised.

An interesting observation is made in Figure 6 by plotting the points corresponding to fixed thicknesses of undamaged matrix between fuel particles. For very small values of $d^i/2\lambda$, the fission product concentration reaches a maximum at $V_f = 0.3$ to 0.4 , which implies that the minimum property change ought to be encountered when the volume fraction of fuel is either above or below this range. However, this variation diminishes rapidly as $d^i/2\lambda$ is increased, and its significance to dispersion fuel design is uncertain.

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*These relationships are not valid when the interparticle spacing is so small that the range of fission fragments recoiling from a particle extends into adjacent particles. The concentration of fission products found in either phase will then be proportional to the volume fraction and the stopping power of that phase.

**Such a criterion as average fission product content in the whole matrix assumes that the properties of the matrix would not be radically different if the impurity atoms were distributed homogeneously rather than concentrated in portions of the matrix. The validity of this assumption will probably vary with the kind of property being measured and with the degree of inhomogeneity in fission product distribution. The problem is quite analogous to measuring the properties of any alloy at the ingot stage, where there may be micro-segregation of alloying elements, rather than after homogenization at a later state.

Distribution of Fission Products in Matrix. The foregoing discussion referred to average fission product concentration in the damaged matrix region or in the matrix as a whole. Actually the concentration will vary within a recoil range of the particle surface in both fuel and matrix phases, somewhat as represented in the upper horizontal row of diagrams in Figure 7. The variation in concentration from the center of the fuel-bearing particle to the extremity of the damaged matrix zone is shown schematically for three different particle sizes, when the fission fragment range is (a) less than the particle radius, (b) greater than the radius but less than the diameter, and (c) greater than the diameter. In the latter case it will be noted that all of the fission fragments escape from the particle and reach a maximum concentration in the matrix at a finite distance from the particle.

The fission product distribution in two adjacent particles of equal diameter in the intervening matrix phase is presented schematically in the lower two horizontal rows of diagrams in Figure 7. The middle row represents one volume fraction of fuel particles while the bottom row represents a higher volume fraction; interparticle spacing is reduced when particle size is decreased or when volume fraction is increased (as was previously shown in Figure 2). The significant observation to be made here is that a homogeneous distribution of fission products is not necessarily reached until the interparticle spacing becomes so very small that the range of the fission fragments escaping from the particle extends into adjacent particles as well as throughout the intervening matrix. At somewhat larger spacings the fission product distribution varies, reaching minimum concentrations at the points midway between adjacent particles for the particle size range down to $D = 2\lambda_f$ or slightly less. For particle diameters $< 0.5\lambda_f$, the peak fission product concentration may be in the matrix rather than in the particles.

While it has been assumed necessary to maintain a matrix zone free from fission products between adjacent particles to fulfill the requirements of the dispersion fuel concept, it is possible that small concentrations of impurity atoms are tolerable, depending on the particular material chosen for the matrix phase. Undoubtedly, there is a difference from one metal to another in the fission product content that can be tolerated without extensive change in properties. For example, uranium metal has been reported to become almost completely embrittled at burnups

Figure 7

as low as 0.001%,¹ while a zirconium - 7% uranium alloy has been observed by R. L. Mehan^{3,4} to retain some measure of ductility at 1.5% burnup of all atoms. Therefore, the criterion for dispersion fuel design should perhaps be modified to specify a continuous matrix phase not in terms of absence of fission product contamination but rather in terms of minimum change of pertinent physical and mechanical properties. It should then be possible to select component material which will permit closer inter-particle spacing and higher volume fractions of fuel-bearing phase than would be allowed under the present design objectives. To this end, it would be very desirable to obtain experimental data on the variation in properties with fission product content for various matrix-phase materials.

Other Factors. The volume fraction of fuel-bearing phase in a dispersion fuel alloy depends of course on the uranium content specified for the fuel element and also on the particular uranium-bearing material selected for the dispersed phase. The more uranium atoms per unit volume of the dispersed material, the smaller will be the volume fraction required in the alloy, a desirable feature from the standpoint of minimizing matrix damage as previously discussed. In this respect, U^{235} metal is the optimum material as the dispersed phase, because it obviously offers the highest possible concentration of fuel. The concentration, V_r , of uranium atoms per unit volume of a compound U_aX_b relative to that for pure uranium increases with density of the compound and with the ratio of a to b, and decreases with atomic weight of X. Weber and Hirsch⁴ have listed values of V_r for a number of uranium compounds, and some of these are presented in Table 1. The most attractive ones from the standpoint of volume fraction required in a dispersion fuel are those with the higher V_r values. On the other hand, selection must also be based on factors such as irradiation stability, compatibility with the matrix material, melting point, and neutron economy.

In all of the relationships between microstructural parameters and fission product distribution presented thus far, fuel burnup has been held to be constant. Since some of the fission fragments recoil from the fuel-bearing particles into the matrix phase, it is to be expected that as more fuel is burned up and more fission products are produced, the properties of the system would change as a result of further contamination. However, according to the dispersion fuel concept, the rate of property change with burnup will be less than in a system having homogeneous fission product damage.

Property Changes Caused by Irradiation

The foregoing analysis of the relationship between microstructural variables and fission product distribution in a dispersion fuel system and the implications of this relationship with respect to property alteration have been based on close-packing arrays of spherical particles of uniform diameter. Such an ideal is of course not realized in actual practice and the departure from ideality will vary with the care and ease of fabrication. The particles are not of uniform size but at best fall within a discrete range of particle sizes and differ by 15 to 20%. Rarely are the particles spherical in shape and they often may have sharp edges or corners. It is difficult to achieve complete uniformity in particle distribution; clustering or alignment (stringing) is often a problem.

Nevertheless, the principles derived for the ideal case should apply qualitatively to the usual cases, and it is relevant to test these principles by actual measurement of properties of irradiated dispersion fuel systems and comparison of the results with micro-structural parameters and fission product distribution. One of the earliest experimental observations of a particle size effect was made by L. P. Hunter,^{6,9,10} who measured the in-pile thermal conductivity of compacts of graphite - U_3O_8 in the temperature range of 600 to 800°C. The samples had been fabricated in two different ways, one of which resulted in a larger UO_2 particle size. After a relatively short exposure in the Oak Ridge graphite reactor, the samples with the fine particles were found to have greatly reduced values (an order of magnitude) of thermal conductivity, while the coarse-particle samples lost only one-third of their original conductivity. Subsequent extension of this work by Berggren and Primak¹⁰ to include lower irradiation temperatures disclosed larger changes in the thermal conductivity as a function of irradiation time, with the same large differences between types of samples still apparent. These studies gave qualitative evidence that particle size has an important influence on radiation damage in a dispersion-type fuel.

Later, a more detailed study^{6,8,10} of the particle size effect was made by R. H. Kernohan of Oak Ridge National Laboratory (ORNL) in cooperation with L. D. Loch, R. J. Harrison, and G. B. Treff of Battelle Memorial Institute (BMI). These investigators used samples of graphite- UO_2 , some of which were made with enriched oxide while others contained normal UO_2 . The UO_2 particle size was quite uniform within each sample but ranged among the samples from ~600 to <40 μ . The samples, which contained 5 wt % uranium, were irradiated to 0.1% burnup of the uranium at a temperature not exceeding 90°C. Comparison of postirradiation properties with those in the unirradiated condition revealed that the 2×10^{18} nvt exposure to neutrons had produced appreciable increases in elastic modulus and electrical resistivity and decreases in thermal conductivity in all samples. However, in the samples having enriched UO_2 with particle sizes <44 μ the changes were two to three times greater than those exhibited by other samples with larger particle size and with normal UO_2 ; this would indicate that fission fragment escape from the UO_2 had played an important role in damaging the graphite matrix. While the other samples did not show significant differences in property changes among themselves as irradiated, subsequent annealing produced less recovery in the samples with smaller particle size. Thus the permanent type of damage caused by fission product contamination was demonstrated and again the effect of particle size on fission product content in the matrix was exemplified.

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Because of their application in fuel elements for such reactors as the MTR, the LITR (at Oak Ridge), and various swimming pool reactors, the aluminum-uranium alloys containing up to 30% uranium have received considerable attention by people engaged in the field of radiation damage. Since these alloys consist primarily of a dispersion of UAl_4 particles in an aluminum matrix, it is appropriate to cite one or two pertinent observations from the data reported by Billington.^{6,10} After a burnup of 0.2% of all atoms, the electrical resistivity had increased by about 25% for a 5.7% uranium-aluminum alloy and by about 50% for a 15% uranium-aluminum alloy. Similarly, the increase in yield and ultimate tensile strengths and decrease in uniform elongation after a given nvt exposure were greater in the 15% alloy than in the 5.7% alloy. These observations are consistent with the prediction that an increase in volume fraction of the dispersed, uranium-bearing phase (UAl_4 in this case) promotes greater fission product contamination of the matrix phase.

While the experimental observations cited thus far appear to verify the previously discussed concepts on the relation between microstructural parameters, fission product distribution, and property retention, they (as well as numerous other observations not mentioned) were made on samples having relatively low burnup. It is of interest to examine what happens to the properties of dispersion fuel systems at higher burnups. Most of the pertinent observations in this connection are to be found in work on fuel elements of stainless steel - UO_2 and on dispersion systems containing boron or a boron compound as the dispersed phase. While the latter systems represent nuclear absorbers rather than nuclear fuels, they serve as examples of the dispersion concept.

Stainless Steel - UO_2 . In work at ORNL by M. J. Feldman and co-workers,^{10,11,12} dispersions of up to 30 wt % UO_2 in austenitic stainless steel (Type 304 or Type 347) matrices were irradiated to high burnup, as high as 40% of the uranium in some cases. The UO_2 particle size ranged from <3 to 105μ . Definite particle size effects on mechanical properties of the irradiated samples were noted. For example, for samples having equivalent fuel loading and irradiation history, those with the 3μ particle size showed the highest increase in hardness. The elements with small UO_2 particle size also were more brittle on post-irradiation bend testing than those with large particle size. These effects are shown graphically in Figure 8 (reported by Feldman¹²); hardness measurements are plotted which were made before irradiation, after irradiation to 30% burnup of the uranium, and after postirradiation annealing for stainless steel - 25% UO_2 dispersion fuel element samples having three different particle size ranges. Comparing the results before and after

ANNEALING OF UO_2 -STAINLESS STEEL PLATE - TYPE
FUEL ELEMENTS (FELDMAN,¹²)

Figure 8

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irradiation it is noted that there is a very definite correlation between hardness change and particle size; the largest hardness increase corresponds to the smallest particle size. Some of the hardness increase is caused by neutron bombardment alone, as indicated by the change shown for the stainless steel cladding. The differences between this change and the total changes represent the respective amounts of fission damage in the three samples.

The results shown in Figure 8 for the postirradiation study are very interesting in that hardnesses actually increase upon annealing at 400°C and then decrease progressively with higher annealing temperature. This behavior is suggestive of an aging phenomenon, followed by over-aging at high temperature, wherein the impurity fission product atoms may be playing the role of the alloying elements of conventional age-hardening alloys.

The ORNL workers also observed an increase in the incidence of cracking and failure during irradiation and during postirradiation bend testing with decrease in particle size and with increase in burnup. This again demonstrated the embrittling effect of fission product contamination of the stainless steel matrix. A further interesting observation was that some specimens having electrolytic iron as the matrix phase showed greater tendency for cracking upon irradiation and postirradiation bending than

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did the stainless steel matrix samples with equivalent UO_2 particle size, fuel loading, and irradiation history. This may possibly be another example of the differences between materials in sensitivity to the effects of fission product contamination.

From work performed at this Laboratory on stainless steel - UO_2 , there are also illustrations of the effect of UO_2 particle size on post-irradiation properties. Figure 9 shows bend test curves obtained by W. D. Valovage and R. A. Siergiej¹³ for stainless steel - 25% UO_2 samples before and after irradiation to uranium burnups of 28 and 40% in two cases and 10 to 20% in most other cases. The bend tests were performed both at room temperature and at two different elevated temperatures, 1200 and 1400°F. The ductility differences between the large particle (105 to 210 μ) and smaller particle (44 μ) samples after irradiation are readily apparent. Even after 40% burnup, the large particle samples exhibited greater ductility than smaller particle size samples after 28% burnup. These differences persist at the elevated temperatures, although the actual ductility decreases with temperature both before and after irradiation, and thus reflects an inherent characteristic of some of the stainless steels. Table 4, which lists tensile test data on stainless steel - 25% UO_2 (105 to 210 μ particle size) at three different temperatures after various burnups, also illustrates the reduction in ductility at higher temperatures. This observation points up a factor to be considered in the selection of a matrix material for a dispersion fuel alloy - namely, the need to take into account the properties over the entire temperature range of fuel element operation.

Boron Dispersion Systems. Perhaps the most striking comparison between a dispersion-type alloy and a homogeneously damaged material with respect to property retention is found in the work of W. K. Anderson, D. N. Dunning, W. A. Neisz, and G. F. McKittrick¹⁴ on zirconium-boron and titanium-boron alloys. Bend test curves are presented in Figure 10 and 11, respectively, for samples of these two systems irradiated to 1 to 2% burnup of all atoms. The microstructures of the melted alloys were such that the distribution of fission products should be homogeneous, while the dispersion alloys, which were synthesized from powders, had boron particles of 105 to 420 μ in size, sufficiently large to cause relatively little matrix damage. None of the samples failed on pre-irradiation bend testing. After irradiation, however, the melted alloy samples failed at low deflections both at room temperature and at 600°F, while the dispersion alloys endured considerable deflection at room temperature before failure and exhibited no failure at 600°F. Although the irradiation conditions labeled on these curves show somewhat higher burnups for the melted alloys than for the dispersion alloys, other data indicate that the embrittlement of the melted alloys occurs early in their irradiation.

BEND TESTS OF 25 wt% UO_2 STAINLESS STEEL DISPERSION FUEL ELEMENTS AT R.T., 1200°F AND 1400°F (VALOVAGE & SIERGIEJ,¹³)

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Figure 9

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TABLE 4. Tensile Tests of 25 Wt % UO₂* - Stainless Steel Dispersion Alloys¹³

<u>Burnup of U, %</u>	<u>Temperature, °F</u>	<u>Ultimate Strength, psi</u>	<u>Yield Strength,** psi</u>	<u>Total*** Elongation, %</u>
0	RT	46,000	23,000	16
3.2	RT	48,000	44,700	8
15.3	RT	46,700	42,700	4
21.3	RT	49,500	46,500	2
0	1000	25,050	13,800	9
6.7	1000	32,300	28,200	4
10.5	1000	33,200	31,300	3
19.4	1000	33,700	31,200	2
0	1200	20,070	12,400	11
6.6	1200	21,800	19,200	3
12.0	1200	20,900	19,100	3
18.5	1200	18,500	18,500	2
0	1400	13,800	10,400	8
6.6	1400	15,000	15,000	2
12.0	1400	9,900	8,700	6

*105 to 210μ UO₂ particle size.

**0.2% offset.

***~2-in. gage length.

BEND TEST RESULTS ON 0.35 WT % B¹⁰ ZIRCALOY 2 ALLOYS

Figure 10

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Asanovich et al.¹⁵ in a study on irradiation effects in dispersion alloys of various boron-containing materials in titanium and Zircaloy-2 matrices observed a curious effect in Zircaloy-2 - 8.4% normal B₄C samples. After irradiation to 20 to 31% of the B¹⁰ present, the ductility in bending was 25% higher than before irradiation. No explanation for this behavior was apparent.

While the studies on zirconium-base and titanium-base boron dispersions demonstrated the effect of a dispersion-type structure in minimizing property alteration by irradiation, such was not the case with stainless steel - boron dispersions according to Eichenberg.¹⁶ Samples prepared with boron particle sizes and interparticle spacings that were presumably sufficient to preserve a continuous undamaged stainless steel matrix were found to have become almost completely embrittled by irradiation, even at exposures as low as 0.8×10^{20} nvt, as evidenced by hardness, tensile, and impact results before and after irradiation. The cause for this behavior could not be explained.

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BEND TEST RESULTS ON TITANIUM B¹⁰ ALLOYS

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Figure 11

Other Observations. Evidence for the effect of large volume fraction of fuel-bearing phase in a dispersion structure is found in studies by Freshley and Last¹⁷ of a magnesium-uranium fuel element. The samples contained 50 vol % uranium metal, in the form of large particles, surrounded by a matrix of magnesium or magnesium-silicon alloy. Obviously, with such a large fraction of uranium phase the interparticle spacings were extremely small. This condition was reflected in the observation that there was almost complete loss of ductility as a consequence of irradiation at 0.1% burnup of the total uranium atoms. However, the elements had merit even at 2% burnup from other standpoints (p. 39).

The foregoing examples of the effects of microstructure of dispersion fuel systems on property alteration by irradiation are not meant to be exhaustive. There are other data that could be cited, but those presented serve as highlights to exemplify one of the major features of the dispersion fuel concept.

Sources of Stress

The mechanical behavior of any system depends upon the nature of the stresses to which the system is exposed and the response of the material to them. The principal sources of stress in a fuel element are the gross differential thermal stresses which arise from the dissipation of heat from the composite element to the coolant. These stresses depend upon the power per unit length of fuel, the shape of the fuel element, and the mechanical and thermal properties of the fuel. In a dispersion-type element there are additional sources of stress which must be tolerated. These are:

1. Differential thermal stresses of a local nature arising from the inhomogeneous distribution of heat sources within the system. For a fixed power per unit length of a fuel element, this stress increases with increasing particle size.
2. Differential volume changes caused by mismatch in thermal expansion coefficients or thermal conductivities between the fuel and matrix phases, or which result from phase transformations.
3. Differential volume changes caused by the localized accommodation of fission products or the agglomeration of the rare gas atoms to form a gas phase.

The response to the imposed stress will depend upon the mechanical properties of the component phases, the interactions between these phases, and the changes in these properties induced by irradiation. Figure 12 provides an example of interparticle cracking which results from local stresses in a stainless steel - UO_2 dispersion fuel. In this particular sample the cracks were not observed in the irradiated condition, but rather were formed in subsequent annealing. In the annealing treatment, interaction between fission gas atoms probably produced additional stresses around the UO_2 particles.

Dispersion fuel alloys inherently have higher temperatures in the fuel constituent than do homogeneous alloys, where heat generation is uniform. The magnitude of this temperature will depend upon the thermal conductivity of the fuel phase and upon the thermal resistance at the fuel-matrix interface, which is influenced by the differential expansion between the two phases. Increased fuel temperatures are undesirable if the increased mobility of the atoms within this phase causes agglomeration of the fission gases and the development of gas pressures. However, the flexibility of the dispersion concept offers the possibility of selecting alternate constituents which may reduce this problem. Current indications are that the rate of gas agglomeration is lower in the materials with a higher melting point.

← 250 μ →

Cracks between UO_2 Particles in 25 wt. % UO_2 -
Stainless Steel, Irradiated to 8% Burnup of The
Uranium and Annealed 500 Hours at 1200 - 1650°F.

(Barney and Wemple, KAPL)

Figure 12

Stresses caused by volume increases in the dispersed phase can be relieved by introducing porosity in and around the fuel-bearing phase. Such porosity would be advantageous provided that it does not result in higher fuel temperatures and gas atom interactions. (The advantageous use of porosity will be discussed later in connection with the volume changes in the stainless steel - UO_2 fuel element.) It is doubtful, however, that sufficient porosity could be introduced into a dispersion element to accommodate very large amounts of released gases and still derive the benefits in mechanical behavior of the dispersion alloy.

Particle temperatures can also be decreased by selecting a matrix material with a high thermal conductivity. For example, samples of magnesium - 25 vol % uranium dispersion alloy contained in stainless steel jackets were irradiated by Hirsch and Ropitsky³⁵ to 2.5% total atom burnup (10% of the uranium atoms); no dimensional instability was observed. The surface temperature of the samples was 300°C , and the surface heat flux, 250 watts/cm². Another example is the utilization of a ferritic rather than an austenitic matrix in stainless steel - UO_2 dispersion elements to decrease the difference in thermal expansion coefficients of UO_2 and matrix, and thereby decrease the fuel temperature.

Stress concentrations may result from irregularity in particle shape and from alignment of particles in stringers. The directional cracking of a bend test specimen at a point removed from the load is illustrated in Figure 13, where the failure lies in a plane parallel to the neutral axis in the plane of stringering. J. G. Morgan and P. E. Reagan of ORNL observed severe cracking and cladding separation in nickel - 10.3% BN dispersion alloys with a badly stringered structure at about 38% burnup of the B^{10} . On the other hand, at the same burnup of the B^{10} , only microcracking was observed in an iron - 7.6% CaB_6 alloy that had a more random dispersions of particles. The boron content per unit volume in these alloys was similar.

An effect of localized stress may also be indicated by Asanovich et al.¹⁵ who reported microcracking in dispersions of enriched boron and boron compounds in titanium and Zircaloy-2 matrices, but none in natural boron dispersions with equivalent total atom per cent depletions. The local stresses causing the cracking may have stemmed from the higher burnup of the enriched particles.

Mechanical Response

The mechanical behavior of a system will depend upon the magnitude and the state of stress, the mechanical properties of the components, and the effects of environment upon these variables. Normally, through the judicious selection of the cause of failure, reasonable estimates of the

30 wt. % UO_2 - Stainless Steel, Having
Fine ($< 3\mu$), Stringered UO_2 Particles.
Irradiated to 9% Burnup of Uranium
Atoms. Cracks Induced by Bending.
(ORNL.)

Figure 13

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life of a system can be made. However, if large changes in properties occur, such estimates are difficult. The introduction of irradiation as an environmental condition and the changes in properties caused by the fission products create new situations about which little is known. Predictions about the mechanical response of such systems are, therefore, uncertain.

Along these lines, one should observe the conflicting reports of the embrittlement^{1,2} of alloy fuel materials as indicated by postirradiation testing, and concurrently, the reports of large volume increases in these same materials. Roberts and Cottrell¹⁸ nicely resolved this anomaly in uranium metal by demonstrating experimentally that this material, which is brittle in postirradiation tests, does creep at low temperatures and at low stress levels during irradiation. Konobeevsky et al¹⁹ also mention induced plastic behavior in alpha uranium upon the instant of irradiation. In interpreting the volume changes in zirconium-uranium alloys, Willis²⁰ has suggested that a similar plasticity is induced even in fuel materials that lack the gross crystallographic anisotropy of alpha uranium. Such a phenomenon should enhance ductility during irradiation. The absence of fission in the matrix phase of a dispersion fuel alloy might then be disadvantageous.

If strain is the significant parameter, the criterion for satisfactory performance of a material is that the maximum strain to be endured should be less than the strain to failure, or limiting strain. Inherent in the dispersion fuel concept is the assumption that this strain ratio is minimized by maintaining, as nearly as is possible, the properties of the unirradiated matrix material. While such a conclusion is consistent with the decrease in ductility observed in postirradiation tests of homogeneously damaged alloys, it may be contradictory to the above-mentioned processes for the enhancement of ductility. However, the versatility provided by introducing porosity in dispersion fuel alloys and thereby reducing the maximum strain should also be noted as being the favor of the dispersion principle.

III. DIMENSIONAL STABILITY AND FISSION GAS EFFECTS

While a dispersion-type design may improve the response of a fuel system to its environment, it also increases the fission product damage in localized regions of the system. Therefore, an improved material results only when these latter effects do not become limiting. Although the source of instability of selected components can be established only by direct testing, some potentially major problems which may arise should be discussed.

Generally, two types of dimensional instability have been observed in irradiated fuel materials:

1. Anisotropic changes which cause deformation but which result in little or no volume changes, and
2. Volume increases associated with the accumulation and interaction of the fission product atoms.

Anisotropic Dimensional Stability

The deformation of polycrystalline alpha uranium on thermal cycling and the irradiation growth of textured uranium and of graphite provide well-known examples of dimensional instability. Here the crystallographic anisotropy is the source of instability.

Such anisotropy in the fuel-bearing phase should not be limiting in dispersion alloys, even if uranium metal is the fuel material. The fuel particles should be randomly oriented throughout the element in a gross sense and the matrix should constrain the growth of individual particles. The contribution of such anisotropic deformation of the dispersed phase to the matrix stresses should be small.

Excellent dimensional stability is reported by Freshley and Last¹⁷ for dispersions of 50 vol % uranium in magnesium to 2% burnup of the uranium; their finding appears to substantiate the above conclusions.

Volume Stability

If a given crystalline structure is maintained, then volume increases should be anticipated in irradiated systems containing fissile atoms unless:

1. Free storage space exists in the material with which to accommodate the increased volume (e.g., as in B_4C , as indicated by Tucker and Senio,²⁵)
2. Chemical effects reduce the average size of the impurity atoms formed (e.g., as in UO_2 , as suggested by Howe and Weber,³) or
3. There is accommodating microscopic porosity within the element so that the volume increase of the components need not produce net external dimensional changes.

The typical rates of volume increase per per cent burnup of the total atoms are: 3% for zirconium-uranium alloys containing up to 40% uranium,²⁶ <1% for UO_2 ,²⁷ and 2 to 4% for boron alloyed with titanium or zirconium.¹⁴

In addition, abnormally large volume increases (swelling) should be anticipated where the inert gas atoms interact to form gas pockets. These rare gases constitute about one-eighth of the fission yield and amount to the formation of 25 to 30 cc (NTP) of gas formed per gram of uranium fissioned. If the size of the gas pockets becomes sufficiently large that surface tension forces do not play an important role in minimizing the resultant gas volume, swelling results. The rate and magnitude of the swelling will then depend upon the creep strength of the material surrounding the gas pockets. Rates of volume increase as high as 300% per per cent burnup have been obtained under such conditions.

Sources of Instability in Dispersion-Type Fuels. In a dispersion fuel alloy where the fissionable material is confined to the dispersed phase, the fission density and the resultant impurity atom concentration are increased locally. Both effects may alter the volume stability of the system.

From the standpoint of total volume increase of the system, the accommodation of the fission products should be no more serious a problem in dispersion systems than in homogeneous alloys. (However, the localized nature of these volume increases represents an additional source of differential stress, as previously discussed.)

Difficulties may, however, be encountered in dispersion systems because burnup in the fuel-bearing phase is higher than in a homogeneous system. The burnup which can be tolerated in the fissile phase should be questioned because both the high fission product concentration and the degradation of the parent phase may lead to rare gas atom agglomeration. Fission products are incompatible with most materials, and the effects of adding increasing amounts of these impurity atoms on the stability of materials is yet unknown. Accelerated swelling must be considered a potential problem unless it can be demonstrated that the average size of the gas pockets does not increase with increasing burnup. Should the size of these gas-filled pockets increase to the point where surface tension forces do not limit the contained gas volume, swelling must be expected.

Interactions between fuel and matrix are also important in dispersion alloys. Gross dissimilarities in properties between the dispersed and matrix phases may result in higher particle temperatures and fission gas agglomeration. Chemical reactions between the fuel and matrix phases may occur in irradiated systems even where no reaction occurs in the unirradiated system. An example of such a reaction was found by Barney and Wemple²⁷ in some irradiated stainless steel - UO_2 samples, in which the stainless steel matrix contained silicon and

manganese. The UO_2 particles in these samples were surrounded by a layer of foreign substance after irradiation; this phenomenon was not observed in samples with no silicon or manganese in the stainless steel matrix. The investigators interpreted the phenomenon to have been caused by a reaction between the oxygen released by the fissioned UO_2 and these impurities. Of course, the operating temperature must be sufficiently high for such diffusion processes to occur.

Other sources of irradiation instability which are unique to dispersion alloys may also be encountered. A dispersion alloy containing dispersed particles that have a diameter approximately equal to the fission fragment recoil range is unique in that the total number of atoms in the fuel-bearing phase need not increase. In fact, there may be a net flux of atoms out of the fissionable phase when the ratio $D/2\lambda_f$ is < 1.4 , as was shown in Figure 5. Such depletion in itself may lead to instability. The stability of such a fissionable phase cannot be demonstrated by irradiation in the bulk form under such circumstances. The technology of such small fuel particles may be decidedly different from that of bulk materials. With a depletion of atoms in the particles, the mobility of the contained gas atoms may be increased as a result of the increased concentration of vacancies, thereby causing swelling. On the other hand, the fission-spike-induced homogenization process postulated by Bleiberg, Jones, and Lustman²⁸ should reduce or eliminate differences in the concentration of vacancies within the recoil range of the fission events. Such a competitive phenomenon would render stable sufficiently small particles (those with about the dimensions of a fission spike). Therefore, there may exist a range of particle sizes which leads to instability, above and below which the system is stable.

Anomalous swelling observed in this Laboratory in some irradiated zirconium-uranium alloys may provide a valid indication of such a particle size effect. Four compositions were investigated: 7, 12, 22, and 40% uranium. In this range of composition, the zirconium-uranium system is composed of two phases, UZr_2 (approximate composition of the epsilon phase) and alpha zirconium (containing about 0.3 at.% uranium in solid solution). An increase in uranium content results in an increase in the volume fraction of the uranium-rich phase, as shown in Table 5.

TABLE 5. Volume Increases in Irradiated Zirconium-Uranium Alloys

<u>U, wt %</u>	<u>Burnup, At. %</u>	<u>U Burnup, %</u>	<u>% Volume Change Per At. % Burnup*</u>	<u>Calculated Vol % U-Rich Phase (UZr₂)</u>
7	0.94-2.0	34-71	3-4	9
22	0.5 -1.2	5-10	~3	35
40	1.5 -4.5	22	2-3	73
12	2.5	49	7.8	17.4
22	3.2	33	5.3	35

*Potential error $\pm 0.5\%$.

With the exception of 12 and 22% compositions irradiated to 3.2 and 2.5 total at. % burnup, zirconium-uranium alloys below 1100°F have exhibited only the predictable rate of volume increase of 3 to 4% for each per cent of the total atoms fissioned.²⁶ This observation is based on many experiments, which include 7 and 40% alloys irradiated to burnups as high as 2 and 4.5%, respectively.

The UZr₂ particle size (which is less than the range of fission fragments in the 7, 12, and 22% alloys) as well as volume fraction increases with greater uranium content in the alloys. Therefore, the apparently poorer irradiation stability of the alloys of intermediate composition can be interpreted in terms of the particle size effect mentioned above. That is, the 12 and 22% alloys fall in a range of particle size and interparticle spacing which apparently leads to instability, while above or below this range the system is more stable. If the instability were solely caused by the degradation of the uranium-rich phase, then the 7% alloys, with the highest proportion of uranium atoms fissioned (50 to 70%), should have been the least stable. If the number of fission product atoms per unit volume limited stability, the 40% alloy, which was irradiated to 4.5% burnup of all atoms, should have been unstable.

Fission Gas Retention in Dispersion-Type Fuels. If the fission gases collect in pockets within the fuel-bearing phase, one is concerned with the amount of gas, the size of the pockets, and the associated swelling of the fuel element. While the escape of these gases from the fuel phase relieves this problem, it certainly aggravates the net swelling problem unless the gases are concurrently released from the fuel element. However, interconnected porosity caused either by fabrication or interparticle fracturing during irradiation is undesirable since it increases the potential consequences of a cladding defect. Of course, the consequences of such a defect will depend upon whether the fission products freely diffuse to the coolant or whether the coolant enters the element and leaches fission products or increases the severity of the defect.

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Data on the release of fission products from fuel elements are sparse and uncertain and only recently have been subjected to any rigid examination. The extent of the fission product release will depend upon particle size and the ability of both matrix and fuel-bearing phase to retain fission product atoms.

Neisz and McKittrick²⁹ report loss of corrosion resistance and spalling of the surface of zirconium-boron alloys and surface blistering of Zircaloy-2 clad dispersion alloys after irradiation. Such blistering was also observed in unclad titanium-boron dispersions and is shown in Figure 14. These pictures were taken from the work of Anderson and Dunning,¹⁴ who noted that such blisters could be correlated with sub-surface boron particles. It is evident that the thin web of titanium matrix at the surface did not provide sufficient strength to restrain local swelling of those particles.

The importance of creep strength in resisting swelling may be inferred from the results of W. V. Johnston,³⁰ who studied swelling and fission gas release during postirradiation annealing of zirconium - 8% uranium alloy samples. As shown in Figure 15, Johnston found that the rate of volume increase is dependent upon fuel burnup as well as on temperature. The irradiated samples were annealed for 500 hr at the three temperatures noted; <1% of the fission gases was released from the samples at the temperatures studied. The activation energy for the swelling rate indicated that the deformation was controlled by the creep strength of the material.

The effect of interconnected porosity in allowing fission gas release and relief of the swelling problem was observed by BMI and this Laboratory in a cooperative program for conducting postirradiation annealing tests on stainless steel - 25% UO_2 samples having about 8% porosity. These samples, which had been irradiated to 9 to 15% burnup of the uranium, were observed to release up to 15% of the fission gases at annealing temperatures as low as 900°C with virtually no volume change. These results, however, probably do not reflect fission gas diffusion within the UO_2 particles. As indicated in Figure 5, approximately this amount of gas would have recoiled from the particles (which were 50 to 100 μ in size) during irradiation. Furthermore, L. G. Wisnyi³⁶ has found by postirradiation heating that fission gases are not released from bulk UO_2 until temperatures much higher than 900°C are reached. Because of the large degree of porosity in the stainless steel - UO_2 samples studied cooperatively with BMI and this Laboratory, it was not possible to deduce whether the fission gases that had recoiled from the UO_2 particles escaped from the fuel elements by permeation through interconnected voids and cracks or by diffusion through the stainless steel.

5 x

(a)

18 x

(b)

100 x

(c)

Surface Blistering Due to Subsurface Boron Particles in Unclad 1 wt. % B¹⁰ - Titanium
Dispersion Alloy Plates.
(Anderson, Dunning and McKittrick, KAPL)

Figure 14

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DENSITY CHANGES IN ZIRCALOY-2 CLAD, 8 wt% URANIUM-ZIRCONIUM ALLOYS ON POSTIRRADIATION ANNEALING - (JOHNSTON³⁰)

Figure 15

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UNCLASSIFIED

The extent of fission gas release from stainless steel - UO_2 is greatly reduced when interconnected porosity is absent or when the UO_2 particles are sufficiently large that a significant fraction of fission fragments do not recoil from them. Reporting on fission gas release from stainless steel - 25% UO_2 fuel elements during irradiation to burnups of 20 to 40% of the uranium atoms, Barney and Seymour²⁴ observed that only small amounts of gas (<10%) were released at irradiation temperatures of 400 to 650°C (matrix temperature). Even in elements that had purposely defective cladding and interconnected porosity the gas release was small, indicating the retentivity of UO_2 for fission gases. However, elements irradiated at a matrix temperature of about 900°C ruptured and released large amounts of gas (45 to 60%). Of course, the temperature of the UO_2 particles was considerably higher than that of the matrix.

There are other examples of the ability of uranium compounds in dispersion fuel systems to retain fission gases up to relatively high temperatures. Personnel at ORNL found that 99.8% of the generated gases in thorium-uranium dispersions containing 2.5% uranium were retained at temperatures of 1840°F.³¹ M. B. Reynolds³⁷ also observed no gas release from UAl_4 particles dispersed in aluminum to about the decomposition point of UAl_4 .

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Additional Observations. The stainless steel - UO_2 dispersion system is being investigated for several applications because of its good irradiation stability at relatively high temperatures. As a result, much of the data substantiating the dispersion concept have been obtained from investigations on this system.

Barney and Seymour,²⁴ who investigated this system with UO_2 particle sizes ranging from 44μ to $105-210\mu$, report excellent stability to burnups of as high as 40% of the uranium and at temperatures of 400 to 650°C . The volume expansion rate of the UO_2 was deduced to be $<1\%$ for each per cent burnup of the uranium atoms;²⁷ this would indicate that the stresses which result from the accommodation of fission product atoms should be small. In addition, swelling caused by gas agglomeration was not observed, although the UO_2 particle temperature was considerably higher than the 400 to 650°C cited for the matrix.

In view of the problems that have been encountered in metallic fuel materials, UO_2 might be termed surprisingly stable. Commenting on the irradiation stability of UO_2 , Weber³² notes that the fission product atoms should become oxidized by the released oxygen and suggests that the resultant mixed oxides may be analogous to naturally occurring mineral oxides. As such, they may be inherently more compatible and stable than mixtures of foreign atoms in a metal lattice.

Barney and Wemple²⁷ have observed the following irradiation effects in UO_2 particles in stainless steel - UO_2 dispersion fuels:

1. The tendency for cracks initially present in the particles to heal upon irradiation. This effect can be seen in Figure 16 by comparing the fragmented, unirradiated particles (Figure 16a) with the irradiated particles (Figure 16b, c, and d). Similar observations have been made by A. E. Richt of ORNL.
2. The sintering and rounding of UO_2 particles during irradiation, as demonstrated in Figure 17. The material shown here initially possessed a considerable amount of porosity (5 to 10%), which was located principally in and around the UO_2 particles. With such a low starting density, the particles contracted away from the stainless steel matrix on sintering. Such behavior would be expected to raise the particle temperature and accelerate further sintering. The material shown in Figure 16 had a higher initial density than that in Figure 17 and, therefore, exhibited less sintering.

← 100 μ →

(a) Unirradiated.

Unetched.

← 100 μ →

(b) Irradiated to 12% Burnup of Uranium Atoms
at 450 - 600°C. Lightly Etched.

25 wt. % UO_2 - STAINLESS STEEL

(Barney and Wemple, KAPL)

Figure 16

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3. The formation of pores in the UO_2 , as shown in Figure 16b, c, and d, and Figure 17. Observe the size of the pores in Figure 16c, which represents an irradiation to 30 to 40% burnup at 450 to 600°C, and compare these with the larger pores in Figure 16d, where less than half this burnup but a considerably higher temperature, 1100 to 1700°C, is represented.

Thus, while good irradiation stability of some stainless steel - UO_2 dispersion fuels has been demonstrated, physical changes in the UO_2 particles have also been observed. The healing of cracks initially present in the UO_2 , sintering of UO_2 particles, and the formation of pores in the UO_2 may influence this stability. The significance of these changes on the irradiation stability of the stainless steel - UO_2 dispersion system should be investigated further to establish more fully the limitations of the system.

Merits of Dispersion-Type Fuels. The foregoing discussion merely indicates precautions that must be considered in the design of a dispersion fuel alloy. Where irradiation instability limits the design, there is no advantage (and there may be disadvantages) in the dispersion concept. If these factors are not limiting, the dispersion concept has significant merits, particularly in the flexibility of design, since it offers the following advantages over the customary approach to fuel design:

1. A wider selection of fuel materials, many of which may be less sensitive to impurity atom damage. Uranium dioxide is a good example of such a superior fuel-bearing material. Even though its uranium content per unit volume is less than that of uranium metal, it does offer a higher fuel operating temperature, a lower expansion rate for the accommodation of fission product atoms, and higher fuel burnups (40 to 50% uranium burnup and perhaps more) without gross instability.
2. A wider selection of matrix materials, and therefore, optimization of the gross properties of the fuel system. In this respect, note that the fuel composition, particle size, and matrix composition are limited only by the conditions required for fabrication and the existence of suitable materials. Within these bounds and the limitations imposed by irradiation, any combination of fuel and matrix material is practical.

100 μ

50 μ

(c) Irradiated to 30 - 40% Burnup of Uranium Atoms at 450° - 600°C. Unetched.

(d) Irradiated to 14% Burnup of Uranium Atoms at 1100 - 1700°C. Unetched.

25 wt. % UO_2 - STAINLESS STEEL

(Barney and Wemple, KAPL)

Figure 16 (continued)

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3. A convenient method for introducing porosity into the fuel material to alleviate the strains produced by fission product accommodation and potentially to allow higher fuel burnups. The effect of porosity in minimizing the volume increase of the composite element appears to have been adequately demonstrated in stainless steel - UO_2 dispersions. Here, 98% dense alloys exhibited some volume changes (~ 1 to 2%), while those having 90 to 95% composite densities had none at even higher burnups.

IV. CONCLUSIONS

The improvement of the postirradiation properties of a fuel system by the control of microstructural factors in accordance with the dispersion concept has been borne out experimentally. With such improvement in properties, contingent upon the stability of the components, the dispersion concept provides a technique for the design of potentially superior fuel elements. On the other hand, the dispersion fuel concept may not represent a unique solution to the complete loss of ductility characteristic of irradiated uranium. Measurable post-irradiation ductility has been observed in some homogeneously damaged alloys. Nevertheless, maximum retention of ductility and other properties will probably be found only in alloys designed along the lines of the dispersion fuel concept.

One should observe that the dispersion concept suggests that where strains must be tolerated the ratio of the existing strain to limiting strain is minimized by maintaining the properties of the unirradiated materials. While this conclusion is consistent with the comparative decreases in ductility observed in postirradiation tests, it should be noted that processes for the enhancement of ductility of fission-product-damaged materials during irradiation have also been proposed. Moreover, the localized increase in fission density and fission product concentration that are characteristic of dispersion alloys are also potentially disadvantageous in that they may cause both local stresses and irradiation instability. The dispersion concept is practical only where these effects do not limit the stability of the alloy. Nevertheless, the flexibility offered by the dispersion fuel principle both in selecting the constituent materials and in introducing porosity may combat these limitations.

The successful performance of

1. The aluminum - 15% uranium alloy (aluminum- UAl_4 dispersion) fuel element in the MTR at burnups of up to 50% of the uranium in endurance tests,²³

$\leftarrow 50\mu \rightarrow$

Irradiated to 21% Burnup of Uranium Atoms at
650 - 830°C.

25 wt. % UO_2 - STAINLESS STEEL

(Barney and Wemple, KAPL)

Figure 17

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2. The stainless steel - 25% UO_2 dispersion fuels at high temperatures (650°C) for burnups of 20 to 50% of the uranium,^{11,24}
3. The aluminum - 54% UO_2 fuel element in the Geneva reactor,¹¹ and
4. Other systems, including boron dispersions,

demonstrate the applicability of dispersion fuel systems.

Obviously a considerable amount of additional information is necessary to utilize fully the merits of dispersion elements. In an engineering sense this does not prevent one from designing an improved fuel element on the basis of the dispersion concept but it does restrict one in establishing the optimum parameters. Further work is needed to develop fully the potentialities of the dispersion concept. To date, a majority of the investigations have been cursory, with emphasis on the development of a satisfactory composite fuel element. Many aspects have not been adequately investigated, such as the allowable range of particle sizes, the criteria for proper selection of fuel and matrix materials and the effect of fission product impurity atoms on their properties, and the importance of porosity in and around the fuel-bearing phase or in the matrix. There is need for a true comparison of homogeneous alloys and dispersion alloys having the same components irradiated near their limit of applicability; a dispersion of U^{235} in a matrix of U^{238} , compared with the normal homogeneous solution of the two, would be interesting to study in this connection. The influence of particle size on matrix strength and the mode of failure is yet to be completely evaluated.

Dispersion alloys as a class of dense solid fuel materials must be compared to the alternate type, homogeneous alloys. The validity of the comparison will depend upon the knowledge of the sources of instability in both types of systems. From this viewpoint a better understanding of the effect of fission product concentration on materials and the mode of interactions between gas atoms is needed.

REFERENCES

1. Reynolds, M. B. The Diffusion of Fission Gases in Metallic Uranium, Nuclear Engineering and Science Congress (1955).
2. Hueschen, R. E., Kemper, R. S., and Kelley, W. S., The Effects of Irradiation on the Tensile Properties of Uranium, 2nd Nuclear Engineering and Science Conference, Paper No. 57-NESC-13 (1957).
3. Secret Report.
4. Weber, C. E., and Hirsch, H. H., Dispersion-Type Fuel Elements, International Conference on the Peaceful Uses of Atomic Energy, Paper No. P/561 (1955).
5. Howe, J. P., The Metallurgy of Reactor Fuels, International Conference on the Peaceful Uses of Atomic Energy, Paper No. P/825 (1955).
6. Billington, D. S., Radiation Damage in Reactor Materials, International Conference on the Peaceful Uses of Atomic Energy, Paper No. P/744 (1955).
7. Harrison, R. J., Effect of Particle Size on Fission Fragment Damage for Particles of Fissionable Material Dispersed in a Matrix, U. S. Atomic Energy Commission, BMI-846 (1953).
8. Confidential Report.
9. Secret Report.
10. Billington, D. S., Irradiation Effects in Reactor Materials, Nuclear Metallurgy, Vol. III, 31-74, AIME-IMD Special Report Series No. 3 (1956).
11. U.S. Atomic Energy Commission, TID-7515 (Pt. 2), 8-36 (1956).
12. Feldman, M. J., Thesis, University of Tennessee Metallurgy Department (1956).
13. Valovage, W. D., and Siergiej, R. A., Mechanical Properties of Stainless Steel - Uranium Dioxide Dispersion Fuels, U.S. Atomic Energy Commission, KAPL-1590, to be issued.
14. Anderson, W. K., and Dunning, D. N., Absorber Materials for Reactor Control, Paper No. 57-NESC-66, 2nd Nuclear Engineering and Science Conference (1957).
15. Confidential Report.
16. Eichenberg, J. D., The Effect of Irradiation on the Properties of Boron-Stainless Steel Powder Dispersion Samples, U.S. Atomic Energy Commission, WAPD-160 (1957).

17. Confidential Report.
18. Roberts, A. C., and Cottrell, A. H., Creep of Alpha Uranium During Irradiation with Neutrons, Philosophical Magazine 1, Series 8, 711-718 (1956).
19. Konobeevsky, S. T., Pravdyuk, N. F., and Kutaitsev, V. I., International Conference on the Peaceful Uses of Atomic Energy, Vol. 7, 433 (1955).
20. Willis, A. H., Dimensional Changes in Some Irradiated Fuel-Bearing Alloys, to be published.
21. Loch, L. D., Engle, G. B., Snyder, M. J., Duckworth, W. H., Survey of Refractory Uranium Compounds, U. S. Atomic Energy Commission, BMI-1124, (1956).
22. Aron, W. A., Hoffman, B. G., and Williams, F. C., Range Energy Curves, U. S. Atomic Energy Commission, AECU-633 (1950).
23. Bartz, M. A., Radiation Damage Observations at the MTR, U. S. Atomic Energy Commission, TID-7515 (Pt. 1), 19-48 (1956).
24. Barney, W. K., and Seymour, W. E., Irradiation of Stainless Steel - Uranium Dioxide Dispersion Fuel Element, KAPL-1688, to be issued.
25. Tucker, C. W., and Senio, P., The Stability of Boron Carbide after 15 Atomic Per Cent Burnup, U. S. Atomic Energy Commission, KAPL-1091, (1954).
26. Willis, A. H., The Irradiation Stability of Low Weight Per Cent Uranium - Zirconium Alloys, presented at AIME Meeting, November, 1957.
27. Barney, W. K., and Wemple, B. D., Metallography of Irradiated UO₂-Containing Fuel Elements, KAPL-1836, February 1958.
28. Bleiberg, M. L., Jones, L. J., and Lustman, B., Phase Changes In Pile - Irradiated Uranium-Base Alloys, J. Appl. Phys. 27, 1270 (1956).
29. McKittrick, G. F., Neisz, W. A., Development of Zircaloy-Clad, Discrete Burnable Poison Elements for S3G/S4G, U. S. Atomic Energy Commission, KAPL-1726 (1957).
30. Johnston, W. V., The Effects of Post-Irradiation Annealing of Uranium - Zirconium Alloys, U. S. Atomic Energy Commission, KAPL-1562 (1956).

32. Weber, C. E., Effects of Radiation on Nuclear Core Components, Colloquium on Effects of Irradiation on Materials, Sponsored by Office for Naval Research and Glen L. Martin Co. (1957).
33. KAPL-1702, Thermal Stability of UO₂, L. G. Wisnyi and S. W. Pijanowski, to be issued.
34. R. L. Mehan, unpublished data (KAPL).
35. H. H. Hirsh and P. Ropitsky, KAPL, unpublished data.
36. Wisnyi, L. G., unpublished data.

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Throughout this report the authors refer to the studies and observations of many different investigators. Much of this work has not yet appeared in the open literature; it has either been reported only in AEC literature or not reported at all. It is requested that any future reference to such work be accredited to the particular investigator associated with it and not to the authors of this report.

This report is by no means an exhaustive study of observations on irradiation effects in dispersion fuels. To have been so would have made it even more voluminous than it now is. The authors have selected the material from the viewpoint of indicating the merits and potential limitations of the dispersion fuel concept. As a result, some observations and data have been omitted because they did not lend themselves to this appraisal and critique.