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# A high density composite fuel with integrated burnable absorber: $U_3Si_2-UB_2$

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## Abstract

$UB_2-U_3Si_2$  composite fuel pellets have been manufactured and characterised. Both main phases were observed to be stable in contact with one-another up to the maximum processing temperature of 1500°C. The solubility and behaviour of hydrogen in  $UB_2$  was predicted and compared to previous work on the gas' behaviour in  $U_3Si_2$ .  $H_2$  was found to be insoluble in  $UB_2$  to any great extent, a clear distinction to the behaviour with  $U_3Si_2$ .

*Keywords:* Nuclear, Accident Tolerant Fuel, Burnable Absorber

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## 1. Introduction

1           The nuclear accident at Fukushima Daiichi has led to extensive research into so-called Ac-  
2           cident Tolerant Fuels (ATFs). These new combinations of fuel and cladding materials aim to  
3           offer significant advantages over the current  $UO_2$  fuel and zirconium cladding materials used  
4           predominantly in light water reactors. Potential advantages of the new systems include reduced  
5           fuel centre-line temperatures, higher power to melt ratios and improved oxidation performance  
6           under accident conditions with an associated reduction in hydrogen production.  
7

8           A leading candidate ATF system is  $U_3Si_2$  fuel pellets within a silicon carbide or chromium  
9           coated zirconium-alloy cladding. This has been the subject of considerable recent academic and  
10          industrial interest, which has culminated in a series of test irradiations at the Idaho National  
11          Laboratory [1, 2].  $U_3Si_2$  has a higher uranium density than  $UO_2$ , and an improved thermal  
12          conductivity. In concert with an advanced cladding solution such as SiC, it therefore offers both  
13          economic and safety advantages over  $UO_2$ /zirconium. However, recent results suggest that the  
14          performance of  $U_3Si_2$  in high temperature steam may be unacceptable, leading to the energetic  
15          pulverisation of the fuel pellet and rapid oxidation resulting in severe fuel failure [3].

16          This rapid pulverisation is thought to be a result of the unexpectedly high volume change  
17          which occurs when  $U_3Si_2$  reacts with hydrogen to form  $U_3Si_2H_2$ , experimentally observed and  
18          predicted by atomistic modelling to be around 10% [4, 5], resulting in a drastic increase in  
19          reactive surface area and further oxidation and hydrogen production. When considering steam  
20          reactions, hydrogen is formed from an initial oxidation reaction, and so in practice limiting either  
21          the oxidation or hydride-formation reactions could greatly improve steam performance in this  
22          fuel material.

23  $UB_2$  enriched in  $^{11}B$  has been the source of burgeoning interest as a fuel material, as a result  
24 of its physical and neutronic properties. Recent modelling and experiment suggests that  $UB_2$  may  
25 be a significant ATF candidate in its own right [6, 7, 8] with a uranium density of  $11.6 \text{ g/cm}^3$  [9].

26 Furthermore, there remains the need to develop a burnable absorber (BA) concept for  $U_3Si_2$   
27 pellets. Gadolinium compounds have been shown to cause degradation of the  $U_3Si_2$  phase, form-  
28 ing both uranium metal and  $USi$  phases; gadolinium is therefore not a credible candidate for  
29 homogeneous inclusion within  $U_3Si_2$  [10].

30 One potential option is to utilise  $UB_2$  as a BA within  $U_3Si_2$ . It is possible to vary the enrich-  
31 ment of  $^{10}B$  such that the material will act as a BA in some core regions and a simple fuel in  
32 other regions.

33 Atomic scale modelling (Section 2.2) suggests that the addition of  $UB_2$  to  $U_3Si_2$  may both al-  
34 low for reactivity control and may potentially inhibit a mechanism that leads to the adverse steam  
35 reaction observed by Sooby-Wood et al. [3], thus preventing or slowing the rapid pulverisation  
36 of the material.

## 37 **2. Methodology**

### 38 *2.1. Fabrication Methodologies*

39 The primary goal of the experimental trials conducted at Manchester was to demonstrate  
40 the feasibility of  $U_3Si_2/UB_2$  composite materials under ‘standard’ sintering treatment. In this  
41 context, ‘standard’ refers to utilising similar powder preparation and sintering parameters to  
42 those employed in existing  $U_3Si_2$  studies [1].

#### 43 *2.1.1. Arc Melting*

44 Early trials of producing  $U_3Si_2$ -boron composites proceeded via the use of arc melting; combin-  
45 ing elemental boron or a combination of elemental boron and silicon with  $U_3Si_2$  and observ-  
46 ing the material phases produced on cooling. Both methodologies produced similar results to  
47 previous work involving the arc melting of gadolinium with uranium silicide ??, with the  $U_3Si_2$   
48 phase breaking down to form silicon rich regions, alongside the formation of a uranium metal  
49 phase. This programme was halted after a small number of trials, and cold pressing and sintering  
50 of composite discs was investigated for further trials.

#### 51 *2.1.2. Powder Production*

52 In order to introduce  $UB_2$  in a lower temperature and more controlled regime,  $UB_2$  and  $U_3Si_2$   
53 powders were combined and used to produce small discs.  $UB_2$  powder was produced by high  
54 energy ball milling an ingot of  $UB_2$  produced directly from arc-melting uranium metal and ele-  
55 mental boron. During this process, crystalline boron was seen to react particularly energetically  
56 to the process, and often shattered preventing its inclusion within the arc melted button. As  
57 such, boron was added until the required mass was reached for stoichiometry, assuming no ura-  
58 nium was lost during the process. Uranium silicide powder was produced following the method  
59 detailed within [10].

60 Tungsten carbide milling vessels and media were required to successfully mill  $UB_2$ ; the use  
61 of hardened steel tooling resulted in the impregnation of  $UB_2$  into both the vessel and milling  
62 media and the loss of a large volume of material. To counteract this, 0.01 g polyethylene-glycol  
63 (PEG) was added to each 50 ml milling vessel for future runs, which proved successful with

64 tungsten carbide mills but made little difference to the performance of hardened steel milling  
65 media.

66 Previous experience at Manchester has shown that the age of  $U_3Si_2$  powder can have a sig-  
67 nificant impact on the sintering performance of green pellets produced from it, assumed to be  
68 through the formation of a thin silicon oxide layer on the powder surface. This has been observed  
69 even in very low  $O_2$  environments ( $<1$  ppm  $O_2$ ). As the powder employed in these preliminary  
70 studies was around two weeks old, it was expected that discs produced from this material would  
71 be of reasonably low density and show large volumes of porosity.

### 72 2.1.3. Cold Press and Sintering

73 Several composite mixtures were produced ranging from pure  $U_3Si_2$  to pure  $UB_2$ . In order  
74 to observe phase instability when either  $U_3Si_2$  or  $UB_2$  was the dominant phase, 10 wt.% and  
75 90 wt.%  $UB_2$  mixtures were studied, as well as a 50 wt.%  $UB_2$  mixture. Powder milling, mixing  
76 and pressing were all performed within an atmosphere controlled argon glovebox, typically less  
77 than 1 ppm  $O_2$ .

78 Powders were mechanically mixed by hand to maintain the stoichiometry of the desired mix-  
79 tures, as it was observed during milling that each compound behaved differently within stainless  
80 steel and tungsten carbide milling vessels.  $U_3Si_2$  behaves well in stainless steel, but a large  
81 fraction becomes unrecoverable within tungsten carbide vessels.  $UB_2$  responds in the opposite  
82 fashion. Hence hand mixing via rolling and shaking within a quartz vessel was used to blend  
83 the powders. This occasionally resulted in variations in local stoichiometry between the different  
84 uranium phases, particularly the 50 wt.%-50 wt.% mixture of  $UB_2$  and  $U_3Si_2$ . It proved rela-  
85 tively easy to identify these regions and remove them from further analysis, as they showed a  
86 significantly altered microstructure.

87 Powders were pressed into 6mm diameter discs using a mechanical press to a pressure of  
88 120 MPa. Die tooling was lubricated with zinc stearate to ease sample removal. Green discs  
89 all appeared to be well formed, although on handling with tweezers some damage occurred,  
90 particularly around the circumference of pellets. This was not unexpected given the use of milling  
91 lubricant, and did not prevent the use of these samples for further testing within this work. Later  
92 pellets were produced with a small mix of binding agent (PolyOx<sup>1</sup>) which greatly improved green  
93 pellet robustness.

94 Composite and  $U_3Si_2$  pellets were sintered within graphite crucibles utilising a graphite-  
95 lined vacuum furnace. Tantalum beads and foil were used to limit contact between fuel materials  
96 and graphite, as well as acting as an oxygen getter during sintering. Vacuum was maintained  
97 at at least  $10^{-3}$  mbar during sintering. Pellets were held at 1500 °C for one hour, with shorter  
98 isothermal steps at 350 °C and 650 °C to remove remaining PEG and binder material. Pure  $UB_2$   
99 pellets were sintered with a maximum temperature of 1800 °C, with a one hour hold time.

100 Sintered pellets were partially friable, as expected due to the age of the silicide powder em-  
101 ployed, and had a relatively low density compared to typical targets for fuel: approximately 70-  
102 80%. This large range in density is due to the non-standard geometry produced during the pro-  
103 cess, and the difficulty in employing pycnometry or immersion density measurements to porous  
104 materials. Density measured by pycnometry was high (above 90%) indicating a large volume of  
105 open porosity, consistent with gas evolution during the sintering stage.  $UO_2$  is typically sintered  
106 with a target density of 95% of theoretical density, which is also a general target for sintered

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<sup>1</sup>Polyethylene Oxide M.W. 100,000

107 U<sub>3</sub>Si<sub>2</sub> in order to accommodate fission induced swelling of the fuel and hence reduce the strain  
 108 in the cladding [11]. The pellets produced here were significantly below this value. This was  
 109 not considered problematic for the current study, as microstructural examination and phase iden-  
 110 tification (Section 3.1) could still be conducted without issue, and both reference silicide and  
 111 composite materials appeared to be broadly similar in sintering behaviour.

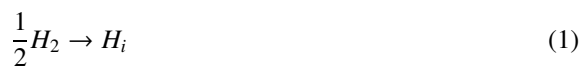
## 112 2.2. Atomic Scale Modelling

113 Density functional theory based calculations were performed at Bangor University using the  
 114 Vienna Ab-initio Software Package (VASP) [12, 13, 14]. Projector augmented-wave potentials  
 115 [15] were used with the GGA-PBE exchange correlation [16] supplied with the software.

116 Simulations of UB<sub>2</sub> were performed using a 4×4×3 super-cell of the *P6/mmm* unit-cell (sim-  
 117 ilar to previous work on the ZrB<sub>2</sub> system [17]). A 4×4×4 gamma centred *k*-point grid was used.  
 118 Simulations were performed under constant pressure allowing both cell volume and shape to  
 119 change. The electronic self consistent field cut-off criterion was set to 10<sup>-4</sup> eV and the geometry  
 120 optimisation cut-off criterion was set to 10<sup>-3</sup> eV. The cut-off energy was set to 400 eV. A first-  
 121 order Methfessel-Paxton [18] method was used for smearing of the bands with a width of 0.1 eV.  
 122 In accordance with the results from Burr *et al.* [7], no Hubbard parameter was applied to the U  
 123 *f*-electrons.

124 In previous work, the solubility of H<sub>2</sub> into U<sub>3</sub>Si<sub>2</sub> was investigated [5], highlighting that the  
 125 reaction was exothermic forming U<sub>3</sub>Si<sub>2</sub>H<sub>2</sub> with an associated swelling of 10%. In this work,  
 126 the solubility of H<sub>2</sub> into UB<sub>2</sub> is considered and compared to the previous results in order to help  
 127 explain the experimentally observed differences in behaviour between UB<sub>2</sub> and U<sub>3</sub>Si<sub>2</sub>.

128 H solution onto five interstitial sites was considered: the (0, 0,  $\frac{1}{2}$ ), ( $\frac{1}{3}, \frac{2}{3}, 0$ ), ( $\frac{1}{3}, \frac{2}{3}, \frac{1}{4}$ ), ( $\frac{1}{2}, 0, 0$ ) and  
 129 ( $\frac{1}{2}, \frac{1}{2}, \frac{1}{4}$ ) sites. The solution enthalpy was considered through the following simple reaction:



130 The H<sub>2</sub> molecule was modelled as a single dimer in a large cell and the binding energy was found  
 131 to be in good agreement with the experimental observation (-2.27 eV per atom). The temperature  
 132 dependent entropy for the H<sub>2</sub> molecule was taken from the NIST-JANAF tables [19].

## 133 2.3. Atomic scale modelling of hydrogen in UB<sub>2</sub>

134 To understand the potential improvement related to the addition of UB<sub>2</sub> to U<sub>3</sub>Si<sub>2</sub>, atomic  
 135 scale modelling based on density functional theory has been used to assess the propensity for  
 136 hydrogen solution into UB<sub>2</sub> and compare this behaviour with U<sub>3</sub>Si<sub>2</sub>. The experimental results  
 137 suggest that the two main phases do not react with each-other and previous calculations show that  
 138 the potential for defects resulting from stoichiometry deviations are very low in concentration in  
 139 the UB<sub>2</sub> system - and it is therefore only necessary to carry out a simple hydrogen solution  
 140 through interstitial defects computations.

141 Solution of hydrogen into UB<sub>2</sub> was found to proceed endothermically onto all considered  
 142 interstitial sites. The lowest solution enthalpy was computed to be 0.81 eV onto the ( $\frac{1}{3}, \frac{2}{3}, \frac{1}{4}$ ) site.  
 143 Compared to U<sub>3</sub>Si<sub>2</sub>, which reacts exothermically with H<sub>2</sub>, UB<sub>2</sub> is not expected to accommodate  
 144 any significant concentration of hydrogen produced as a result of the oxidation of the fuel or  
 145 the cladding material in the event of cladding failure. In a similar manner to U<sub>3</sub>Si<sub>2</sub>, increases in  
 146 temperature are shown to make the solution of H<sub>2</sub> into UB<sub>2</sub> even less favourable as a result of  
 147 the increasing contribution of the vibrational entropy of the H<sub>2</sub> molecule [5].

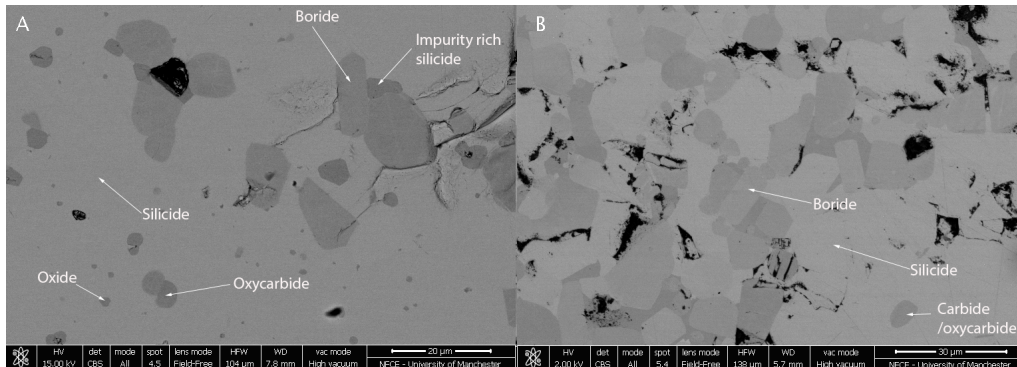


Figure 1: Microstructure of 10wt% (A) and 50wt% (B)  $UB_2$  discs produced using backscatter detector

148 The volume change associated with a single hydrogen atom entering a 144 atom supercell of  
 149  $UB_2$  (0.7 at.%) is 0.2%. Expansion in the  $a/b$ -axis is three times larger than in the  $c$ -axis indicat-  
 150 ing a large anisotropy. Given the concentration of hydrogen in solution within the  $UB_2$  lattice is  
 151 expected to be more than ten orders of magnitude lower than this concentration according to the  
 152 laws of mass action [20], the volume change as a result of hydrogen incorporation is not expected  
 153 to be a factor in the degradation of  $UB_2$  when exposed to aqueous environs.

### 154 3. Experimental Results

#### 155 3.1. Microstructure and Phase Analysis

156 The microstructure of composite pellets and the phases present were examined via electron  
 157 microscopy and X-ray diffraction (XRD) of powdered pellet sections.

158 SEM micrographs of a 90 wt.%-10 wt.% and 50 wt.%-50 wt.%  $U_3Si_2$ - $UB_2$  discs are shown  
 159 in Figure 1 A and B, respectively. Both pellets show  $UB_2$  regions well sintered within a  $U_3Si_2$   
 160 matrix. There are also small inclusions of uranium oxide and uranium oxy-carbide within the  
 161 silicide. From XRD these have been identified as  $UO_2$  and  $UC_{0.75}O_{0.25}$ . The UCO phase is visible  
 162 in  $UB_2$  containing samples following milling, while a small mass of U-metal was observed in  
 163 un-milled  $UB_2$ , suggesting the  $UB_2$  was slightly uranium rich following arc melting, and reacted  
 164 with the carbide milling vessel during powder processing. This phase was identified prior mixing  
 165 with  $U_3Si_2$ , and it is therefore not indicative of a potentially problematic reaction between the  
 166  $UB_2$  and  $U_3Si_2$  phases.

167 Due to both the porosity present within the samples, and the presence of the additional phases,  
 168 the experimental conclusions drawn from these samples are largely limited to the investigation  
 169 of interfacial boundary layers between the  $UB_2$  and  $U_3Si_2$  phases. Previous work on burnable  
 170 poisons in  $U_3Si_2$  has identified that the  $U_3Si_2$  phase can be almost entirely disrupted from small  
 171 quantities of gadolinium [10]. Analysis of the  $U_3Si_2$  phase following introduction of  $UB_2$  is  
 172 therefore important to ensure that it remains stable following pressing and sintering.

173 Energy Dispersive X-ray Spectroscopy (EDX) was employed to determine the microstruc-  
 174 tural segregation of the two phases (Figure 2). Boride and silicide phases appear distinct accord-  
 175 ing to SEM-EDX analysis.

176 Although some regions of Ni and Fe impurities exist, and silicon and boron-poor areas are  
 177 visible, these are not generally co-located at the boundaries between silicide and boride phases.

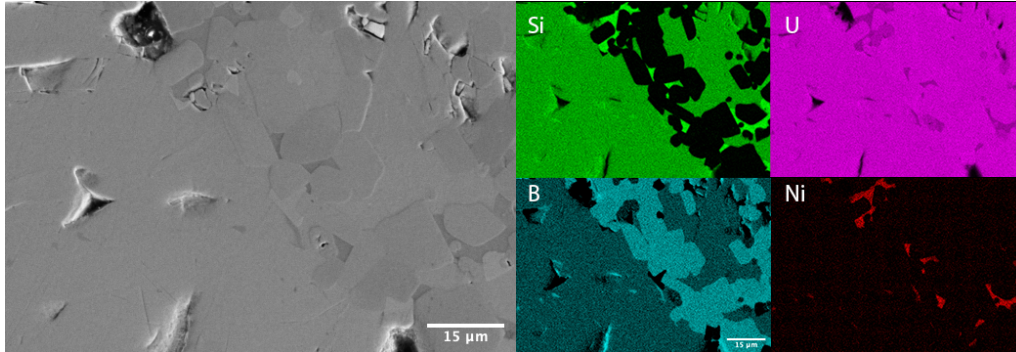


Figure 2: Elemental mapping of 50wt%  $UB_2$  composite pellet, produced from EDX

178 Ni impurities result from the original source uranium, which was nickel plated and mechanically  
 179 ground prior to use. Fe contamination likely resulted from pre-crushing of the arc-melted  $UB_2$   
 180 material, which was achieved via impact with a steel weight. Both are present in Figure 2 in  
 181 concentrations less than 0.5 at%. These impurities appear to cause the formation of silicon-rich  
 182 silicide regions, although these do not appear to be  $USi$  and were not present on XRD analysis.

183 EDX line-profile analysis of silicide-boride grain boundaries shows a sharp transition be-  
 184 tween the two phases, up to the resolution of the instrument, suggesting that the phases do not  
 185 interact at the pellet sintering temperature employed in this study. Some small quantities of ura-  
 186 nium metal/uranium carbide/oxy-carbide<sup>2</sup> are visible in Figure 2, but these appear to be located  
 187 very close to larger volumes of impurities and oxide grains. Their formation is not consistently  
 188 on boride/silicide boundaries. In addition, silicon rich regions visible in Figure 2 are close to the  
 189 boundary with the  $UO/UC$  phases already identified: most likely formed during the preferential  
 190 oxidation of U over Si. Note that the boron signal visible on Figure 2 within silicon rich regions  
 191 is a result of the overlapping boron, carbon and uranium peaks present at these X-ray energies,  
 192 and below a reasonable limit of detection for boron using this technique

193 Phase analysis was performed via XRD on initial materials, powder mixes and sintered pellets  
 194 which were crushed prior to analysis. XRD patterns of 50-50 composite material are presented  
 195 in Figure 3 as representative. The raw  $UB_2$  material produced shows a small quantity of uranium  
 196 metal present within the material, suggesting that the produced bulk is slightly uranium rich  
 197 ( $UB_2$  is a line compound [7]), bulk  $U_3Si_2$  appears phase pure. Powder blended material show  
 198 small quantities (<5%) of uranium oxy-carbide ( $UC_{0.75}O_{0.25}$ ) and carbide ( $UC_2$  phases), likely  
 199 resulting from the interaction of uranium metal within the initial boride material reacting with  
 200 the tungsten carbide milling vessel during the high energy milling process.

#### 201 4. Discussion

202 The manufacture of test specimens in the  $U_3Si_2-UB_2$  system suggest that these composite  
 203 materials are stable during standard cold press and sintering preparations. It is particularly en-  
 204 couraging to note that the  $UB_2$  rich samples produced sintered to similar densities and physical

<sup>2</sup>EDX is unable to distinguish between uranium and carbon due to overlapping low energy peaks, and surface oxygen is overestimated using the technique

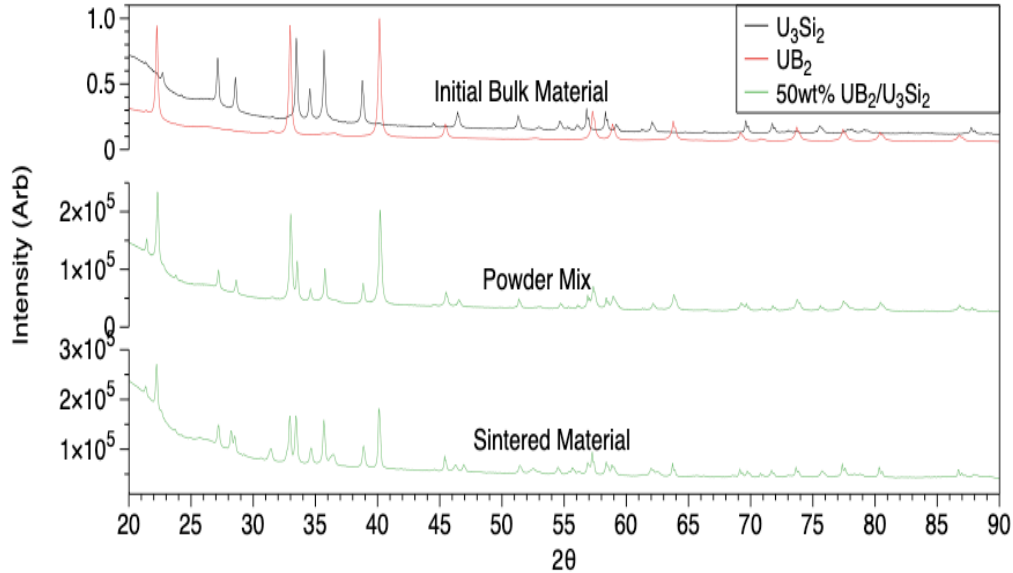


Figure 3: XRD Patterns from initial bulk materials (top), powder blends (middle) and sintered material (bottom). Uranium metal peaks are denoted with an asterisk

205 robustness as the U<sub>3</sub>Si<sub>2</sub> material utilised in the study. A key challenge with employing UB<sub>2</sub> as a  
 206 monolithic fuel form is expected to be that of sintering; UB<sub>2</sub> has a melting point of over 2400 °C,  
 207 and similar diboride compounds such as ZrB<sub>2</sub> typically do not fully densify without the use of  
 208 sintering aids or pressurised sintering techniques [22]. The composites produced here appear to  
 209 have successfully incorporated UB<sub>2</sub> grains at lower temperatures than are necessary for sintering  
 210 of pure UB<sub>2</sub>. Although several minor phases and inclusions were noted within the material mi-  
 211 crostructure, the source of these appears to be understood and future studies would likely reduce  
 212 or remove these via an alternative uranium source and coated steel components.

213 Modelling of hydrogen solubility within UB<sub>2</sub> demonstrates that the reaction is endothermic  
 214 and has a predicted volume change of less than 10% of that predicted for U<sub>3</sub>Si<sub>2</sub>. The inclusion  
 215 of a UB<sub>2</sub> phase may therefore offer some improvements to the steam performance of U<sub>3</sub>Si<sub>2</sub> fuel,  
 216 without compromising on the improved thermal performance or uranium density which are key  
 217 drivers of U<sub>3</sub>Si<sub>2</sub> development. This will rely on the interaction of the phases within a com-  
 218 plex steam environment and detailed experimental data is required to properly assess potential  
 219 improvements.

220 The formation of composite fuel pellets is encouraging beyond any potential water perfor-  
 221 mance improvements; UB<sub>2</sub> may be a promising fuel material in its own right, but has historically  
 222 been considered commercially unattractive due to the need to procure enriched boron for fabri-  
 223 cation and as such has been the subject of limited study. Limited modern atomistic modelling  
 224 studies [6] and historic experimental investigations [23] demonstrate attractive thermal properties  
 225 for example, and it has a higher uranium density than U<sub>3</sub>Si<sub>2</sub> and UO<sub>2</sub>. The use of smaller weight  
 226 percentages of UB<sub>2</sub> within a U<sub>3</sub>Si<sub>2</sub> matrix goes some way towards addressing this concern, while  
 227 arriving at a fuel form which potentially retains the benefit of both materials, and solves the issue



228 of BA addition to  $U_3Si_2$ .

## 229 5. Conclusions

230 Several composite fuel discs were produced, containing a range of  $U_3Si_2$  and  $UB_2$  phases.  
231 Despite the incorporation of some impurities and minor phases, the two materials appear to re-  
232 tain phase stability during sintering, and composite material sintering performance was similar  
233 to that of reference  $U_3Si_2$  discs produced for this study. The successful incorporation of a boron  
234 compound into a  $U_3Si_2$  pellet provides one solution to the need to add burnable absorber ma-  
235 terial to fuel pellets. Hydrogen solubility in  $UB_2$  is predicted to be low indicating a potential  
236 improvement of the material's behaviour when exposed to steam compared to monolithic  $U_3Si_2$ .

## 237 6. Data Availability

238 The raw/processed data required to reproduce these findings cannot be shared at this time as  
239 the data also forms part of an ongoing study

## 240 References

- 241 [1] J.M. Harp, P.A. Lessing, B.H. Park, J. Maupin, DOE-NE Report, INL/CON-13-28403 (2013).  
242 [2] F. Cappia, J.M. Harp, J. Nucl. Mater. 518 (2019) 62-79.  
243 [3] E. Sooby-Wood, J.T. White, C.J. Grote, A.T. Nelson, J. Nucl. Mater. 501 (2018) 404-412.  
244 [4] S. Mäsková, K. Miliyanchuk, L. Havela, J. Nucl. Mater., 487 (2017) 418-423.  
245 [5] S.C. Middleburgh, A. Claisse, D.A. Andersson, R.W. Grimes, P. Olsson, S. Maskova, J. Nucl. Mater. 501 (2018)  
246 234-237.  
247 [6] E. Jossou, L. Malakkal, B. Szpunar, D. Oladimeji, J.A. Szpunar, J. Nucl. Mater. 490 (2017) 41-48.  
248 [7] P.A. Burr, E. Kardoulaki, R. Holmes, S.C. Middleburgh, J. Nucl. Mater. 513 (2019) 45-55.  
249 [8] E. Jossou, D. Oladimeji, L. Malakkal, S. Middleburgh, B. Szpunar, J. Szpunar, J. Nucl. Mater. 494 (2017) 147-156.  
250 [9] S.J.L. Snelgrove, G.L. Hofman, C.L. Trybus, T.C. Wiencek, OSTI Report, DE97001401; NTIS; US Govt. Printing  
251 Office Dep (1996).  
252 [10] J. Turner, J. Buckley, G. Phillips, T.J. Abram, J. Nucl. Mater., 509 (2018) 204-211.  
253 [11] S.C. Middleburgh, R.W. Grimes, K.H. Desai, P.R. Blair, L. Hallstadius, K. Backman, P. Van Uffelen, J. Nucl.  
254 Mater., 427 (2012) 359-363.  
255 [12] G. Kresse and J. Hafner. Ab initio molecular dynamics for liquid metals. Phys. Rev. B, 47:558, 1993.  
256 [13] G. Kresse and J. Hafner. Ab initio molecular-dynamics simulation of the liquid-metal-amorphous-semiconductor  
257 transition in germanium. Phys. Rev. B, 49:14251, 1994.  
258 [14] G. Kresse and J. Furthmüller. Efficiency of ab-initio total energy calculations for metals and semiconductors using a  
259 plane-wave basis set. Comput. Mat. Sci., 6:15, 1996.  
260 [15] P. E. Blochl. Projector augmented-wave method. Phys. Rev. B, 50:17953, 1994.  
261 [16] J. P. Perdew, K. Burke, and M. Ernzerhof. Generalized gradient approximation made simple. Phys. Rev. Lett.,  
262 77:3865, 1996.  
263 [17] S.C. Middleburgh, D.C. Parfitt, P.R. Blair, R.W. Grimes, J. Am. Ceram. Soc. (2011) 94 2225-2229.  
264 [18] M. Methfessel, A. Paxton, Phys. Rev. B 40 (1989) 3616-3621.  
265 [19] M.W. Chase Jr., C.A. Davies, J.R. Downey Jr., D.J. Frurip, R.A. McDonald, A.N. Syverud, NIST JANAF Thermo-  
266 chemical Tables (1985) 20899,  
267 [20] A. Choneos, R.W. Grimes, Appl. Phys. Lett., 91 (2007) 192106.  
268 [21] J.M Harp, P.A. Lessing, R. E. Hoggan, J. Nucl. Mater., 466 (2015) 728-738  
269 [22] Chamberlain, A. L., Fahrenholtz, W. G., Hilmas, G. E., J. Am. Ceram. Soc., 89 (2006) 450-456  
270 [23] Flotow, H. E., Osborne, D. W., OHare, P. A. G., Settle, J. L., Mrazek, F. C., Hubbard, W. N., The Journal of  
271 Chemical Physics, 51(2) (1969), 583592.