

Radiation-induced segregation and precipitation in highly neutron-irradiated Be

Michael Klimenkov^{a,*}, Vladimir Chakin^a, Ramil Gaisin^a, Uta Jäntschi^a, Michael Rieth^a, Viacheslav Kuksenko^b

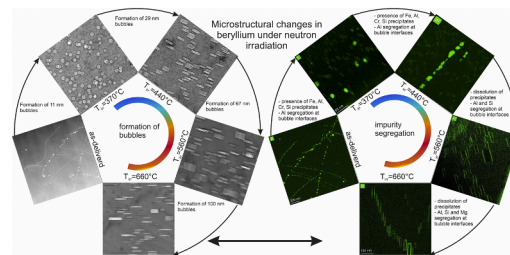
^a Karlsruhe Institute of Technology (KIT), Institute for Applied Materials, 76021 Karlsruhe, Germany

^b United Kingdom Atomic Energy Authority, Culham Science Centre, Abingdon, Oxfordshire OX14 3DB, United Kingdom

HIGHLIGHTS

- Temperature-dependent segregation of Al, Si, and Mg at gas bubbles in neutron-irradiated Be.
- Radiation-induced dissolution of impurity precipitates at $T_{irr} > 500$ °C.
- Quantification of impurity element concentrations in the Be matrix.

GRAPHICAL ABSTRACT



ARTICLE INFO

Keywords:

Radiation-induced segregation
Precipitation
Be
Neutron irradiation

ABSTRACT

This work presents a systematic analytical transmission electron microscopy (TEM) investigation of impurity particles and radiation-induced segregation (RIS) in neutron-irradiated beryllium (Be). Spherical Be pebbles (Ø 1 mm) were irradiated within the HIDOBE-02 experiment in the High Flux Reactor (HFR, Petten) at temperatures of 370, 440, 560, and 650 °C, reaching damage levels of 21, 29, 34, and 37 dpa, respectively. In the unirradiated state, nanoscale precipitates with a complex composition of Fe, Al, Mg, Mn, Si, and Cr are observed, appearing as chain-like or clustered distributions within the Be matrix. Under irradiation, these precipitates progressively dissolve above ~500 °C, significantly altering impurity redistribution and segregation behavior at defect sinks. Scanning TEM with energy-dispersive X-ray spectroscopy (STEM-EDX) reveals a strong temperature dependence of bubble chemistry. At ≤500 °C, bubble surfaces exhibit predominantly Al segregation. At 560 °C, Si enrichment becomes pronounced, while at 650 °C, segregation layers consist of Al, Si, and Mg in approximately equal proportions. Mg and Si decorate both basal and pyramidal facets, whereas Al segregates preferentially to basal planes. Additionally, the concentration of dissolved impurity elements in the Be matrix was quantified. These findings demonstrate that impurity redistribution during neutron irradiation is highly temperature dependent and governed by the interplay of RIS and defect-mediated radiation-enhanced diffusion, ultimately controlling gas bubble chemistry in Be.

* Corresponding author at: Karlsruher Institut für Technologie: Karlsruher Institut für Technologie, Eggenstein-Leopoldshafen, Germany.

E-mail address: michael.klimenkov@kit.edu (M. Klimenkov).

<https://doi.org/10.1016/j.jnucmat.2026.157041>

Received 20 April 2026; Received in revised form 26 July 2026; Accepted 29 August 2026

Available online 30 August 2026

0022-3115/© 2026 The Authors. Published by Elsevier B.V. This is an open access article under the CC BY license (<http://creativecommons.org/licenses/by/4.0/>).

1. Introduction

Beryllium (Be) is considered one of the most efficient neutron multipliers for fusion blanket applications due to its favorable neutronic properties, in particular the low threshold energy of the $^9\text{Be}(n,2n)$ reaction, which enables efficient neutron multiplication even with moderated neutrons [1,2]. In fusion blanket systems, Be can be employed either in the form of pure Be or as Be-containing intermetallic compounds (beryllides). In particular, pure Be in the form of 1 mm pebble beds is foreseen as neutron multiplier in several Test Blanket Module (TBM) concepts developed within the ITER program. These include the European helium-cooled pebble bed (HCPB) TBM, the Chinese helium-cooled ceramic breeder (HCCB) TBM, and the Japanese water-cooled ceramic breeder (WCCB) TBM, where Be pebble beds are used as neutron multiplier material [3–5]. Be pebble multipliers remain under consideration in Chinese fusion reactor concepts such as CFETR, where pebble-bed multiplier configurations similar to those used in ITER TBM designs are being explored [6].

Be pebbles have also long been considered as neutron multiplier material for HCPB breeding blanket concept developed for the European DEMO reactor [7]. During operation of fusion reactors, the Be pebbles are subjected to intense irradiation by 14 MeV neutrons, leading to significant radiation damage. The production of helium (He) and tritium (^3H) in Be has been estimated at 25,700 appm and 640 appm, respectively, corresponding to the expected end-of-life conditions of the HCPB blanket [1]. This blanket design allowed for a maximum operating temperature of approximately 650 °C.

Currently, the HCPB blanket concept has evolved toward the use of more radiation-resistant neutron multipliers, such as titanium beryllide, typically in the form of massive blocks rather than pebble beds [8–11]. Nevertheless, research on radiation damage in Be pebbles remains highly relevant, as these studies provide fundamental insights into radiation-induced processes occurring in Be-based materials under fusion neutron irradiation. Such data are essential for understanding helium generation, swelling, microstructural evolution, and mechanical degradation mechanisms in Be and Be-containing compounds.

Long-term irradiation experiments in high-flux nuclear research reactors provide valuable data on the microstructural evolution of Be pebbles under conditions relevant to fusion reactors, namely irradiation temperature, damage dose, and He/ ^3H generation. However, such experiments do not include the contribution of 14 MeV neutrons characteristic of fusion environments. High-temperature neutron irradiation studies of Be, followed by transmission electron microscopy (TEM) investigations, have been reported previously [12–16]. These studies consistently demonstrate the formation of He- and ^3H -filled bubbles. At the lowest irradiation temperatures, the bubbles exhibit a uniform, nearly spherical morphology. Locally, for example near grain boundaries, bubbles may also adopt a flattened hexagonal shape. With increasing temperature, anisotropy inherent to the hcp crystal lattice of Be promotes the development of flattened, hexagonal bubbles. Electron energy-loss spectroscopy (EELS) has been applied to detect and quantify the presence of He and ^3H within the bubbles [12,15,16].

The performance of Be in neutron irradiation environments is strongly influenced by the presence of impurities, which can be introduced during fabrication, alloying, or by residual contaminants in the raw material. In commercial nuclear Be grades, impurity atoms such as aluminium (Al), iron (Fe) and silicon (Si) are among the most significant metallic contaminants. For instance, specification limits for a “Grade A” Be ingot include Fe up to ~0.110 wt%, Al up to ~0.087 wt%, and Si up to ~0.060 wt% [17]. It is also noted that Be oxide (BeO) and other metallic impurities (Al, Fe, Si) can lead to degradation of the desirable properties of the Be matrix.

Phases containing Fe and Al also play an important role in determining the mechanical behaviour of Be. Jones and Weiner [18], using X ray diffraction on commercially produced Be subjected to various heat treatments, identified a correlation between strain induced precipitation

of the Be₃(FeAl) phase and reductions in both tensile ductility and creep strength. Punni and Cox [19] examined the corrosion performance of several Be grades and found that Fe/Al/Be inclusions act as preferential initiation sites for pitting corrosion. In addition, Morozumi et al. [20] demonstrated that Fe–Be rich precipitates restrict dislocation motion, thereby increasing the hardness of Be containing iron.

The impurity atoms can influence the behavior of Be under irradiation in several ways. The impurities alter the microstructure by segregation on the grain boundaries, precipitation of intermetallic phases (for example, Al-Fe-Be phases). Al is considered as one of the most harmful impurities in Be, since, in the form of free Al or Al-Be eutectic, it may wet grain boundaries and cause serious degradation of ductility at elevated temperatures (400 °C and above) [21]. The possibility of a low melting point Be-Al-Si eutectic in highly alloyed Be has been suggested [22], which might affect the stability of BeO particles [23], or, if present at grain boundaries, influence fracture behaviour [21,22]. Microstructural examination of industrial purity Be irradiated at 50 °C with high energy protons (0.46 dpa) indicates that Al segregated to grain boundaries during irradiation [24].

Under neutron irradiation impurity atoms may themselves be activated to produce secondary radio nuclides (for example Fe impurities leading to long-lived Co-60 upon (n,γ) activation). The impurities may modify defect formation, migration, and clustering under irradiation (for example by acting as sinks or nucleation sites for vacancies, He, ^3H , and transmutation products). It is essential herewith to understand how Si, Al, and Fe impurities in Be influence (i) the initial microstructure and impurity distribution in the as-fabricated material, (ii) their interaction with irradiation-induced defects such as vacancy clusters and He- and hydrogen-filled bubbles, and (iii) the long-term evolution of material properties, including swelling, embrittlement, and thermal conductivity, under neutron irradiation.

This study addresses the role of impurity atoms in neutron-irradiated Be. A combination of microstructural characterization, impurity quantification, and irradiation experiments and/or modelling is employed to assess the impact of impurity atoms on defect evolution and the resulting material behaviour. The results provide input for the selection and optimization of Be materials for application in high-performance nuclear systems.

The paper presents TEM analyses of Be pebbles irradiated in the High Flux Reactor (HFR), Petten, within the HDOBE-02 experiment [25,26], along with corresponding observations of unirradiated reference material [14,27]. Particular emphasis is placed on the composition and spatial distribution of impurities in Be pebbles, providing a basis for understanding their role in the microstructural evolution of neutron-irradiated Be.

2. Experimental

2.1. Materials

The Be pebbles investigated in this study with a diameter of about 1 mm were fabricated by NGK Insulators Ltd., Japan using the rotating electrode method [14,28]. According to the certificate from NGK, these Be pebbles contain impurity elements that composition is listed in Table 1. Be possesses a hexagonal close-packed (hcp) crystalline structure with lattice parameters $a = 0.229$ nm and $c = 0.358$ nm.

2.2. Irradiation parameters

Neutron irradiation was performed in the High Flux Reactor (HFR),

Table 1
Content of the main impurity elements in Be pebbles (appm).

Element	O	Fe	Al	Si	Mg	Cr	Mn
Content	1300	170	160	100	70	30	20

Petten, within the HIDOBE-2 campaign. The Be specimens were placed in 13 holders. These holders, also called drums, were loaded into the stainless steel holder tube surrounded by an inert gas of helium (or helium with 0.1 % H₂). Each drum is designed for a specific temperature and to contain a specific set of specimens including Be pebbles. The irradiation parameters of the pebbles are listed in Table 2.

The operational temperatures within the specimen holder during irradiation were monitored by means of 24 thermocouples. All the thermocouples were used to continuously monitor and record the temperature of the drums with the specimens. The average irradiation temperatures of Be pebbles were approximately 5–30 °C lower than the target temperatures. The deviations from the target temperatures are $\pm$ (1.1, 6.1, 5.5, 4.5) % or $\pm$ (4, 27, 31, 30) °C for the target irradiation temperatures, T_{irr} = 375, 470, 590, 675 °C, respectively [29].

The deviations from the target temperatures are $\pm$ (1.1, 6.1, 5.5, 4.5) % or $\pm$ (4, 27, 31, 30) °C for the target irradiation temperatures, T_{irr} = 375, 470, 590, 675 °C, respectively. The damage doses were calculated for Be according to the NRT (Norgett-Robinson-Torrens) standard, based on the exact position of the Be pebbles along the height of the reactor channel used for irradiation of the HIDOBE-02 capsule. The FISPACT code [18,19] was used to calculate gas concentrations. The total uncertainty for a confidence level of 95 % is 9.8 % for both the amounts of ⁴He and ³H.

2.3. Sample preparation and TEM study

The Be pebbles were first embedded in epoxy resin and then polished to obtain a flat and damage-free surface. Thin lamellae were subsequently prepared using the FEI Helios and Zeiss Auriga (CrossBeam / DualBeam FIB-SEM) focused ion beam (FIB) technique and mounted on molybdenum TEM grids. The electron-transparent region of each lamella typically had a lateral size of 4–8 μ m, allowing precise extraction of regions of interest while minimizing material loss. Final thinning was performed at low ion energies to reduce FIB-induced surface damage. Be exhibits relatively high electron transparency at an accelerating voltage of 200 kV, owing to its large inelastic mean free path of approximately 160 nm. As a result, microstructural features can be imaged in Be foils with thicknesses of up to ~400 nm. The thickness of prepared lamella was in the range from 170 nm to 250 nm.

The prepared lamellae were examined using a Talos 200F high-performance 200 kV TEM. Talos 200F transmission electron microscope (ThermoFisher Scientific Inc.) operated at 200 kV and equipped with a Super-X EDX detector system for advanced chemical analysis. It uses four silicon drift detectors (SDDs) arranged symmetrically around the specimen to maximize the collection solid angle and improve X-ray detection efficiency compared to single-detector EDX system. The analytical investigations were performed in a scanning TEM mode using a beam size of 1.5 nm. Images and two-dimensional elemental distribution maps were acquired using the Velox TEM software.

The color intensity in the elemental maps reflects the net X-ray counts, i.e., the intensity of the corresponding EDX line after background correction. To improve the visibility of the elemental distributions, the brightness and contrast of the maps were adjusted before exporting to

CorelDRAW graphic design software. The maps were then annotated with the required markers, and the final figures were exported as image files. Exporting reduces the image size by a factor of two and smooths the images. This processing does not affect the underlying elemental distribution.

3. Results

3.1. Microstructure of unirradiated Be

The microstructure of the unirradiated Be used in this study has been described previously in Refs. [8,11,12]. The material is coarse-grained, with grain sizes typically exceeding 10 μ m and occasionally reaching several hundred micrometres, i.e. up to approximately half of the pebble diameter (~500 μ m). In some cases, a pebble is composed almost entirely of a single large grain containing only a few smaller grains. A significant population of impurity particles is observed both inside the grains and along grain boundaries. Their size ranges from a few nanometres to approximately 100 nm. Some precipitates appear as isolated particles, while others form chain-like arrangements extending over several micrometres.

Fig. 1 shows a representative HAADF-STEM image of these impurity precipitates together with the corresponding EDX elemental maps. The precipitates contain predominantly Fe, Mn, Mg, Cr, Al, and Si. They are preferentially aligned along linear features in the material, which are presumed to be line dislocations. The homogeneous distribution of the O-EDX signal in Fig. 1a is attributed to the surface oxidation of lamella. Since no local enrichment of oxygen is observed at the precipitates, the oxygen signal is not related to these particles. This observation indicates that the investigated precipitates are most likely intermetallic Be compounds rather than oxide phases. In several smaller precipitates, a Cr- and Mg-rich core is observed, indicating a chemically heterogeneous internal structure (Fig. 1e,f).

The average chemical composition measured for all particles shown in Fig. 1, excluding Be and O, is approximately 34 Fe, 31 Al, 21 Si, 6 Mg, 5 Mn, and 3 Cr (at.%). This composition largely corresponds to the average composition of impurities in the sample, calculated based on the data shown in Table 1. However, analyses performed at different locations within the sample reveal a significant compositional variability among the precipitates. Larger particles, with typical sizes of about 70 nm, contain regions enriched in Fe–Mn–Cr, Mg–Si, Fe–Si, and Fe–Al. In contrast, particles forming chain-like arrangements are predominantly composed of Fe–Al–Si. These chain-like particles frequently contain a small core enriched in Cr, Mn, and Mg, indicating a chemically heterogeneous internal structure and suggesting a multi-stage formation process. As demonstrated in previous studies, metallic impurity particles in Be correspond to beryllide-type intermetallic phases [14]. These intermetallic compounds may occur as individual phases or as composite clusters comprising multiple beryllide phases. Such clusters can form around impurity agglomerates and may reach sizes of several hundred nanometers [12,14].

3.2. Microstructure of be after neutron irradiation at different temperatures

Precipitates in the irradiated Be were observed only after irradiation at relatively low temperatures 370 °C and 440 °C (Figs. 2 and 3). Fig. 2a shows the distribution of Fe–Al-containing precipitates in the vicinity of a grain boundary located in the lower part of the image. The lamella of the 370 °C sample was wedge-shaped, with an average thickness of 170 nm in the investigated area. For the 440 °C sample, the lamella thickness was approximately 200 nm. The corresponding Fe (Fig. 2b) and Al (Fig. 2c) elemental maps were acquired from the area highlighted by the square in Fig. 2a. Iron is detected exclusively within the precipitates, whereas pronounced segregation of Al is observed at irradiation-induced bubbles that are oriented on edge with respect to the electron beam. No

Table 2
Irradiation parameters of Be pebbles.

Irradiation temperature, °C			Neutron fluence, 10 ²⁵ m ⁻²		D, dpa	⁴ He, appm	³ H, appm
Target	Average	Maximum	Thermal	Fast, >1 MeV			
375	370	410	7.89	10.6	21	3632	367
470	440	477	9.87	14.3	29	4751	502
590	560	599	11.3	16.8	34	5524	596
675	650	682	12.0	18.1	37	5925	644

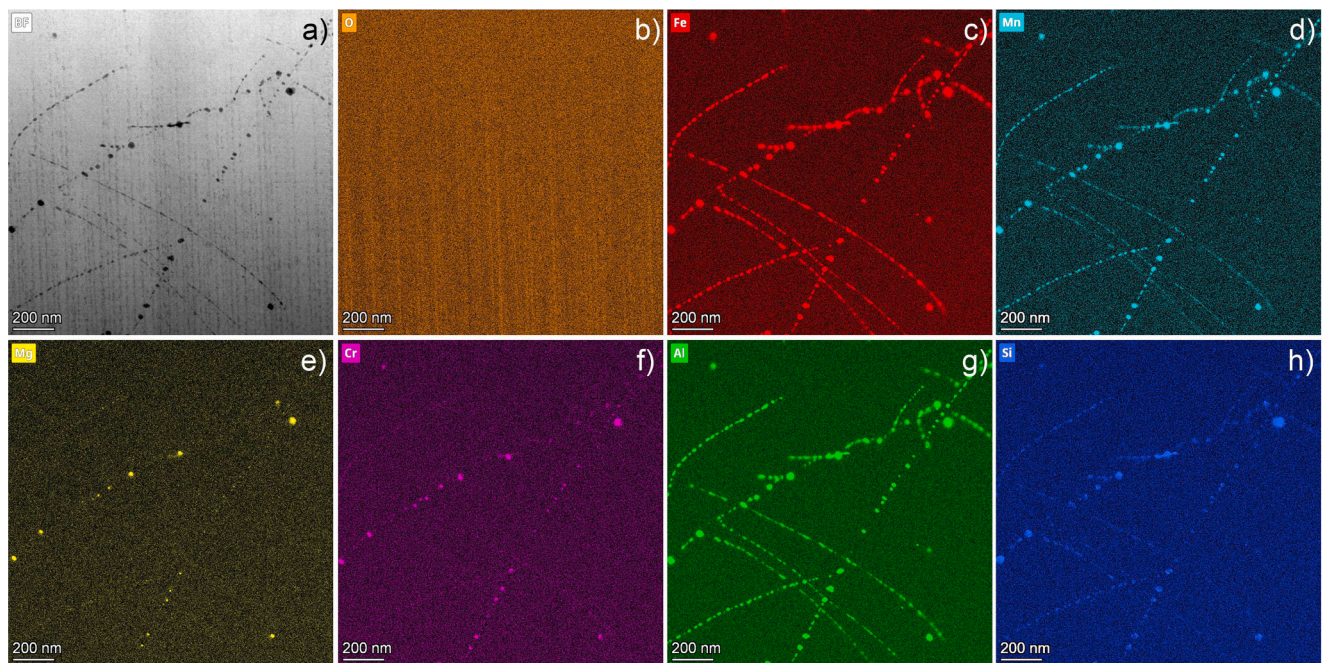


Fig. 1. (a) BF-STEM image of nanoscale impurity precipitates in the unirradiated Be microstructure. (b–h) Corresponding EDX elemental maps showing the distribution of O (b), Fe (c), Mn (d), Mg (e), Cr (f), Al (g) and Si (h), revealing the multicomponent composition of the precipitates.

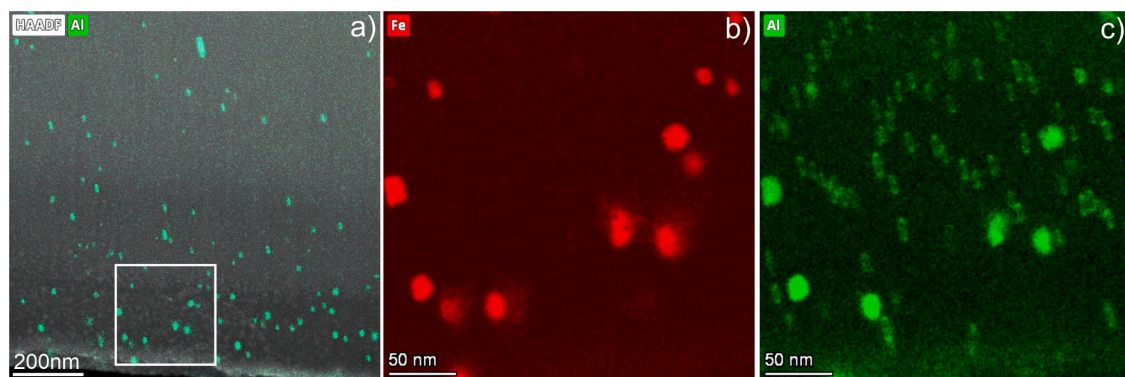


Fig. 2. (a) HAADF-STEM image overlaid with the corresponding Al elemental map of the investigated region in Be irradiated at 370 °C. (b,c) Corresponding EDX elemental maps of Fe (b) and Al (c) acquired from the area marked by the square in (a), showing Fe-containing precipitates and Al segregation at irradiation-induced bubbles.

segregation of other elements was detected within the sensitivity limits of the applied analytical techniques.

The results shown in Fig. 2 represent only one example. Similar to the unirradiated material, the chemical composition of the precipitates exhibits considerable spatial variability across the investigated sample. A comprehensive discussion of this compositional heterogeneity is beyond the scope of the present work. Due to their small size, the bubbles themselves are not clearly resolved in scanning transmission mode; however, their presence is inferred from characteristic contrast variations and associated solute segregation. As demonstrated in earlier studies [12,13,15], irradiation-induced bubbles in Be adopt a flattened, hexagonal, coin-like morphology, which is characteristic of bubble formation in hcp crystal structures. In Fig. 2 and throughout the subsequent analysis, the investigated grains are oriented such that the c-axis of the hcp Be crystal structure lies within the image plane (perpendicular to the electron beam). In this orientation, the coin-shaped bubbles are viewed from the side and have a rectangular projections in the image [15]. The basal planes are aligned parallel to the electron beam, enabling the detection of local EDX signals from segregated elements. The detected

elements that are located on the bubble walls appear as contours in EDX maps. A similar crystallographic orientation was employed in Refs. [12, 15,16] for the detection of hydrogen and helium within the bubbles. At 370 °C, however, bubbles may also exhibit a more uniform morphology with typical sizes of 7–8 nm [15,16]. The preferential segregation of Al to these bubbles suggests a strong interaction between solute atoms and irradiation-induced defects, potentially affecting bubble stability, growth kinetics, and the local chemical environment.

The precipitates in the Be irradiated at 440 °C exhibit a complex Fe–Mn–Al–Si composition, as shown in Fig. 3. The oxygen (O) map highlights bubbles that opened during FIB preparation and exhibit internal oxidation (Fig. 3b). These bubbles were excluded from further analysis because they may also be contaminated by Ga ions and redeposited sputtered species. Closed bubbles, which are unaffected by preparation artifacts, do not appear in the oxygen map in Fig. 3b.

The bubbles formed at the temperature of 440 °C exhibit Al segregation (Fig. 3) similar to that observed after irradiation at 370 °C. Bubbles located in the immediate vicinity of Si-containing precipitates (Fig. 3h) also show very weak Si segregation. Examples of such bubbles

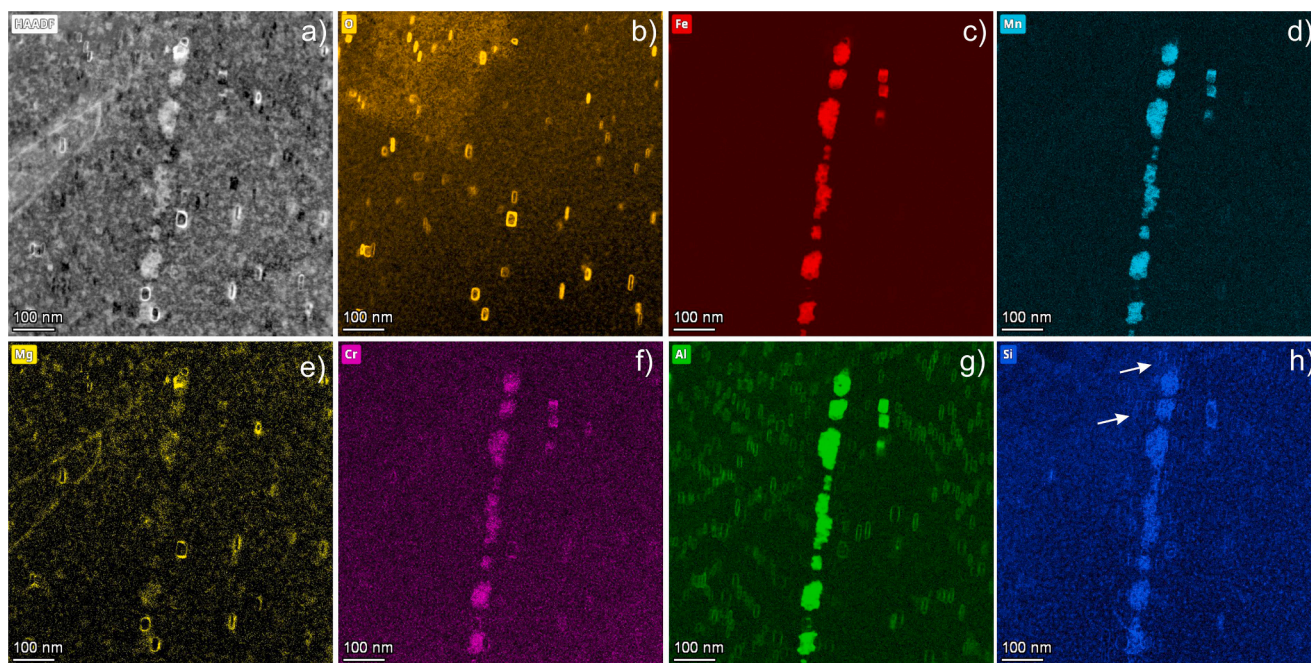


Fig. 3. (a) HAADF-STEM image of the investigated region in a Be irradiated at 440 °C. (b–h) Corresponding EDX elemental maps showing the distribution of O (b), Fe (c), Mn (d), Mg (e), Cr (f), Al (g), and Si (h).

are marked in the Si map by white arrows in the Si map (Fig. 3h). No segregation of Mn (Fig. 3c), Mg (Fig. 3e), or Fe (Fig. 3d) was detected at the bubbles at this irradiation temperature.

When larger areas are considered, the averaged Al and Si concentrations appear relatively homogeneous within the Be matrix; however, slight variations are observed in the vicinity of the precipitates (Fig. 4). As shown in Fig. 4c, the Si signal intensity slightly increases in a region approximately 110–150 nm adjacent to the precipitate row, while the Al signal correspondingly decreases in the same region. This behaviour is consistent with previously reported observations of Si and Al segregation at voids [12]. In the region adjacent to the precipitates, weak Si segregation at the bubbles is observed, whereas bubbles located at larger distances from the precipitates do not show detectable Si segregation. In the same region, Al segregation is also significantly reduced.

No precipitates were detected in the material irradiated at 560 °C (Fig. 5) and 650 °C (Fig. 6). This suggests that radiation-induced dissolution of the initial precipitates occurs at irradiation temperatures above approximately 500 °C. EDX analysis clearly reveals segregation of impurity atoms at the bubbles. With increasing irradiation temperature, the bubble size increases, and the probability that bubbles intersect the foil surface during FIB preparation and subsequently undergo internal oxidation correspondingly increases. This oxidation is clearly visible in

the O maps (Fig. 5b and 6b). The lamella prepared from the 560 °C sample had a thickness of 190 nm, whereas the lamella prepared from the 650 °C sample had a thickness of 230 nm.

In the sample irradiated at 650 °C, the average bubble size (~120 nm) becomes comparable to the foil thickness resulting in a situation where approximately 70 % of the imaged bubbles are open. For this reason, several closed bubbles are marked with arrows in Fig. 5 and 6 to distinguish them from open bubbles. The segregation of impurity elements was analysed only on closed bubbles.

In Be irradiated at 560 °C, segregation of Si and Al with an approximate 1:1 ratio was observed. No segregation of the other major impurity elements (Fe, Mn, and Cr) could be detected within the sensitivity limits of the analysis. Similar behaviour is observed in the material irradiated at 650 °C. However, at 650 °C, segregation of three elements, Al, Si, and Mg, at gas bubbles is detected. While Mg and Si segregate on all bubble surfaces (Fig. 5e,f), Al segregation is observed exclusively on the basal (0001) planes (Fig. 5d).

The evolution of elemental segregation at bubble surfaces as a function of irradiation temperature is summarized in Fig. 7. At low irradiation temperatures of 370 °C and 440 °C, where the precipitates present in the unirradiated material remain stable, the segregated species at the bubbles consist predominantly of Al. At higher irradiation

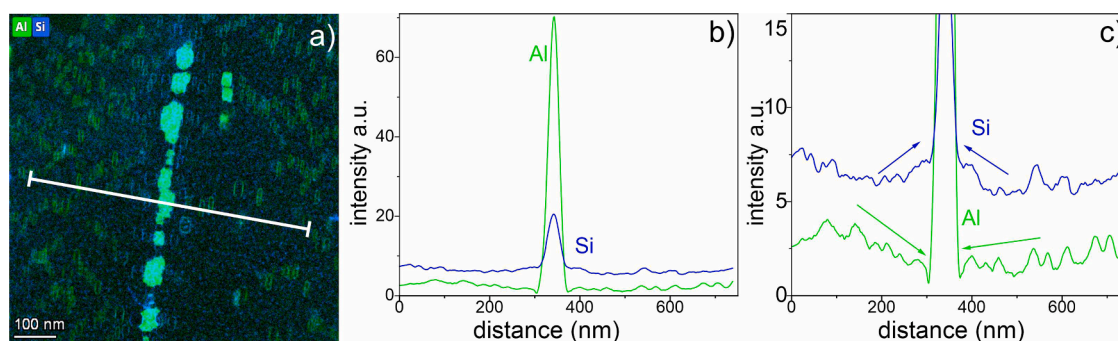


Fig. 4. (a) Overlay of EDX elemental maps of Al and Si obtained from the area with a row of impurity precipitates in a Be. (b,c) Corresponding intensity profiles of Al and Si measured along the line indicated in (a) and integrated over a width of 150 nm.

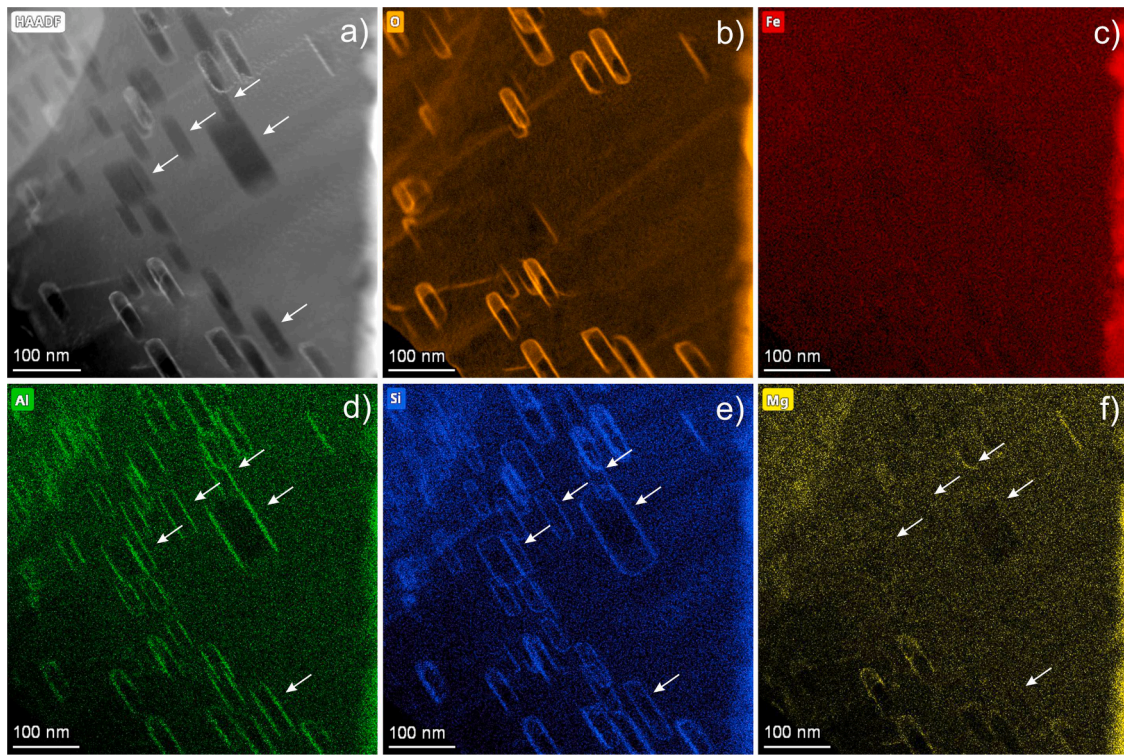


Fig. 5. (a) HAADF-STEM image of the investigated region in a Be irradiated at 560 °C. (b–f) Corresponding EDX elemental maps showing the distribution of O (b), Fe (c), Al (d), Si (e), and Mg (f), illustrating impurity segregation at irradiation-induced bubbles. Closed bubbles selected for the segregation-layer analysis are indicated by arrows.

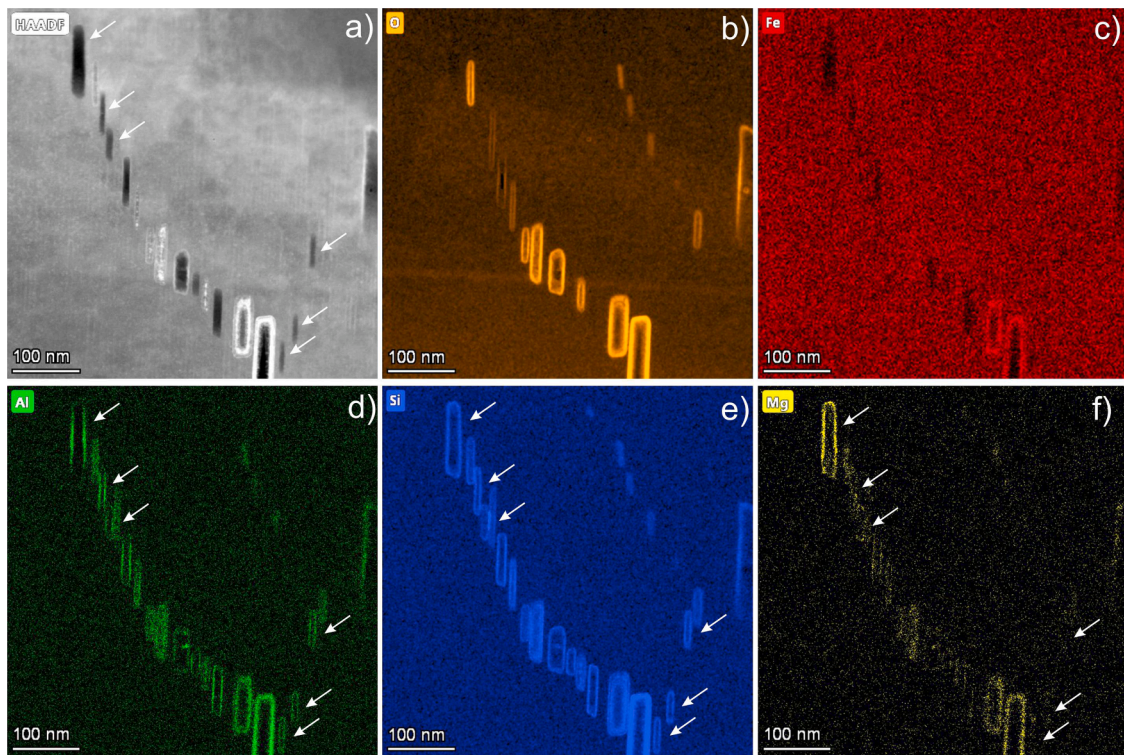


Fig. 6. (a) HAADF-STEM image of the investigated region in a Be irradiated at 650 °C. (b–f) Corresponding EDX elemental maps showing the distribution of O (b), Fe (c), Al (d), Si (e), and Mg (f). Closed bubbles selected for the segregation-layer analysis are indicated by arrows.

temperatures of 560 °C and 650 °C, where the precipitates dissolve, segregation of Si and Mg at the bubbles is also observed. At 650 °C, the

concentrations of Al, Si, and Mg in the segregation layers are approximately the same, with a ratio close to 1:1:1. However, this ratio may

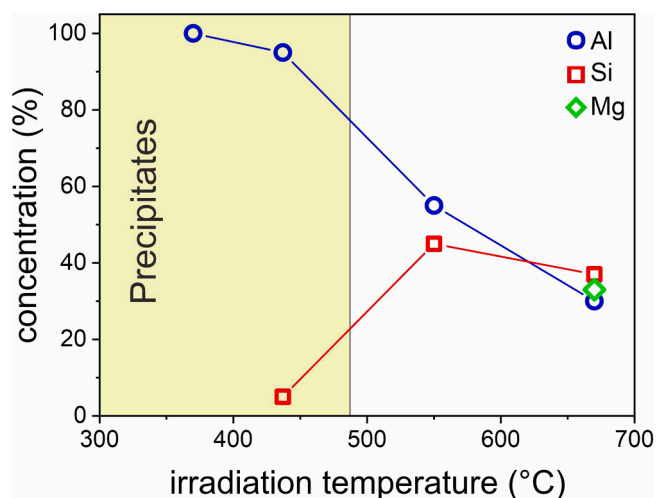


Fig. 7. Temperature dependence of the elemental composition of segregation layers formed at gas bubbles in neutron-irradiated Be.

vary for bubbles of different sizes.

4. Discussion

The presence of multicomponent impurity precipitates in the initial microstructure is expected to influence defect evolution under irradiation. Such particles may act as sinks for point defects or serve as nucleation sites for gas bubbles. Their frequent occurrence along grain boundaries and their chain-like arrangement within grains suggest that they may also affect local transport processes during irradiation. Therefore, the evolution and eventual dissolution of these precipitates under neutron irradiation are likely to play an important role in the redistribution of impurity elements observed in the present study.

Neutron irradiation of Be leads to the formation of bubbles filled with He and $^3\text{H}_2$, which represent the dominant defect structures in irradiated material. Their microstructural characteristics have been investigated in several previous studies [12,15,16]. In Ref. [15], a detailed analysis of bubble properties, including their shape, spatial distribution, and size dependence on irradiation temperature and damage dose, was reported. The bubble size increases markedly with irradiation temperature. Irradiation at 370 °C results in bubbles with an average diameter of ~ 7 nm, whereas irradiation at 650 °C produces bubbles with an average size of ~ 160 nm. Using EELS spectroscopy, it was also demonstrated that these bubbles contain significant amounts of He and ^3H , both produced in Be through transmutation reactions. These gases therefore play a key role in bubble nucleation and growth [8].

Technologically pure Be used in fusion applications contains a non-negligible concentration of impurity elements that form intermetallic precipitates in the as-received material (Table 1) [12,14]. It has been reported that these precipitates dissolve at elevated irradiation temperatures [8]. In Ref. [12], Al, Si, and Mg were observed to segregate at bubble surfaces following irradiation at 670 °C. However, the dissolution behaviour of these precipitates and the associated segregation processes at other irradiation temperatures were not investigated. In the present work, we show that Al, Si, and Mg, which are present in Be as impurities, segregate at bubble surfaces, and that this process strongly depends on irradiation temperature. The apparent presence of Mg signal within the open bubbles or the surrounding, as shown in Fig. 3e, 5f, and 6f, is an artifact introduced during FIB preparation. The observed two-dimensional Mg distribution closely corresponds to the Ga distribution, which is localized in the open bubbles. This artifact arises from the spectral overlap between the Mg $K\alpha$ peak (1.25 keV) and the Ga $L\beta_2$ peak (1.125 keV). Owing to the relatively high Ga signal, the measured intensity of the Mg $K\alpha$ line is artificially increased, leading to a false Mg

signal. Consequently, the analysis of Mg segregation was performed exclusively on closed bubbles, which are not exposed by FIB preparation and therefore are not affected by Ga contamination.

At irradiation temperatures below 500 °C, only Al segregation was detected at the bubble surfaces. Nearly spherical bubbles with sizes of 7–10 nm, formed at 370 °C, contain only trace amounts of Al on their walls, making its detection by TEM difficult. Segregation of Al could be identified only when the bubbles exhibited a flattened hexagonal, coin-like morphology (Fig. 2). Ref. [12] demonstrated that bubbles located near grain boundaries often adopt such a flat, coin-like shape. In this geometry, an impurity layer on the basal planes, possibly only a monolayer thick, is oriented parallel to the electron beam, producing sufficient signal intensity for local EDX analysis. A similar geometric effect applies to the detection of He and $^3\text{H}_2$ within the bubbles.

At an irradiation temperature of 440 °C, Al remains the primary element detected at the bubble surfaces, whereas Si segregation is observed only in the vicinity of precipitates (Fig. 3f). The Si fraction increases to approximately 50 % at 560 °C, whereas at 650 °C the segregation layer consists of Al, Si, and Mg in roughly equal proportions ($\approx 1:1:1$). It was further found that Si and Mg segregate to both basal and pyramidal bubble facets, whereas Al segregates exclusively to basal planes (Fig. 5d).

EDX analysis in TEM was also employed to investigate the distribution of impurity elements in the matrix outside the precipitates. Spectra were acquired from precipitate-free regions, typically covering areas of 1–3 μm^2 , including numerous bubbles in the irradiated material, as well as from smaller areas between bubbles. Fe, Al, and Si, the principal impurity elements listed in Table 1, were reliably detected. The quantification results are summarized in Table 3.

The Fe signal, detected at all irradiation conditions and in the as-delivered material, was used as an internal reference and normalized to unity. It should be noted that the absolute Fe concentration in solid solution may vary with irradiation temperature. Therefore, the reported contents of the other impurity elements (X) represent relative concentrations with respect to Fe, expressed as the C_X/C_{Fe} ratio. A fully quantitative compositional analysis including Be was not performed for several reasons: (i) the Be signal cannot be reliably detected and therefore cannot be used for normalization; (ii) uncertainties exist in the local foil thickness; and (iii) the EDX signals of impurity elements are very weak, in some cases approaching the detection limit. A dash (“–”) in Table 3 indicates that the EDX signal of the corresponding element was below the detection limit. This does not necessarily imply the absence of the element in the matrix but rather reflects the sensitivity limits of the applied analytical technique.

After irradiation at 370 °C and 440 °C, Al and Si are clearly detected in the Be matrix at relative concentrations approximately 5–7 times higher than that of Fe (Table 3). At 440 °C, Si is mainly distributed in the matrix and shows similar concentrations in large analysed areas and in

Table 3

Relative concentrations of impurity elements dissolved in the Be matrix. The data are given as C_X/C_{Fe} ratios (in at.%), where X = Al, Si, or Mg. The Fe signal was used as an internal reference and normalized to unity. Columns marked l.a. (large area $\sim 2\mu\text{m}^2$) and b.b. (between bubbles) correspond to spectra acquired from larger matrix regions containing many bubbles and from smaller regions between bubbles, respectively. A dash (–) indicates that the corresponding element was not detected or that the signal was below the detection limit.

T_{irr} (°C)	as-delivered	370 °C		440 °C		560 °C		650 °C	
element (X)									
Fe	1	1		1		1		1	
		l.	b.	l.	b.	l.a.	b.	l.a.	b.
		a.	b.	a.	b.		b.		b.
Al	–	3	–	2	–	1.2	0.2	0.5	–
Si	–	7	–	11	9	1.7	1.5	1.5	0.3
Mg	–	–	–	–	–	–	–	0.2	–

regions between bubbles. In contrast, a significant fraction of Al is segregated at the bubbles. Owing to the high bubble number density at 370 °C, reliable measurements in the inter-bubble regions were not feasible.

For samples irradiated at 560 °C and 650 °C, the relative concentrations of Al and Si in the matrix decrease significantly. This trend is consistent with the dissolution of Fe-containing precipitates and the increasing segregation of impurity elements at bubble surfaces. At 560 °C, most of the Si remains in the Be matrix, particularly in regions between bubbles, whereas at 650 °C Si begins to segregate more strongly to the bubble surfaces. In contrast, Al is predominantly located at the bubbles at all irradiation temperatures. Mg was not detected in the matrix at the highest irradiation temperature and is therefore likely segregated at the bubbles. Overall, these observations indicate a progressive redistribution of impurity elements with increasing irradiation temperature.

These observations are consistent with irradiation-enhanced diffusion and radiation-induced segregation processes, which are known to modify phase stability and microchemical distributions in irradiated alloys. The progressive enrichment and redistribution of Al and Si in the matrix with increasing temperature indicate that thermally assisted, radiation-driven transport plays a dominant role in the observed compositional evolution.

Following the emergence in the mid-1970s of experimental evidence showing segregation of impurity atoms around voids in irradiated metals and alloys, the theory of radiation damage underwent significant improvement. This development explains the mechanisms of radiation-induced segregation (RIS) and radiation-induced precipitation (RIP) by assuming that the dissolved element (SA) can form complexes with either an interstitial atom (I) or a vacancy (V) of the matrix, i.e., “SA-I” and “SA-V” complexes, respectively [30–38]. The binding energies of these complexes allow them to exist within the temperature range of 0.2–0.6 T_m , which coincides with the swelling regime. The maximum binding energy occurs approximately in the middle of this interval, at about 0.4 T_m . These complexes are mobile and can migrate through the matrix either toward or away from structural sinks. Their stability and mobility depend on several factors, the most important being the binding energies and the irradiation conditions (temperature, neutron flux, and fluence). It should also be noted that interstitial configurations in Be have high migration energies and are unstable; therefore, only vacancy-type defects (SA-V) will be considered here (Table 4).

The atomic radius of impurity and matrix atoms, along with their binding energy, are crucial for determining whether RIS leads to enrichment or depletion of impurity atoms at structural defects [30–32]. A solute with an atomic radius smaller than the matrix should increasingly segregate on sinks such as surfaces, grain boundaries, dislocations, dislocation loops, voids, bubbles, while a solute with a larger atomic radius, conversely, should be uniformly distributed in the matrix.

Table 4

Conditions for radiation-induced segregation (RIS) in irradiated Be [15,39]: “segr.” denotes segregation at bubble surfaces, whereas “neutr.” indicates no significant influence on RIS. The relationships summarized in the table are assumed to be valid for all irradiation temperatures considered in this study. The binding energies were taken from ref [15].

Element	Content, appm	T_m , °C	0.2–0.6 T_m , °C	Atomic radius, nm	Binding energy, eV	RIS in Be
Be	rest.	1287	39–663	0.112	-	-
Al	376	660	86–287	0.143	0.6–0.7	segr.
Si	305	1412	64–738	0.111	0.8–1.0	segr.
Mn	104	1244	30–637	0.127		weak segr.
Mg	116	649	–89–280	0.160	0–0.3	segr.
Fe	1050	1536	89–812	0.126	~0	neutr.
Cr	102	1860	154–1007	0.128	0–0.3	neutr.

However, experimental results often show the opposite behavior, indicating that the “SA-V” binding energy is more influential in the RIS process.

During neutron irradiation of Be, two concurrent processes govern defect and impurity evolution: displacement damage caused by elastic neutron-nucleus interactions and the generation of He and ^3H by transmutation interactions. Displacement cascades produce vacancy supersaturation regions, where excess vacancies interact with impurity atoms to form mobile “SA-V” complexes. Both free vacancies and vacancy-impurity complexes subsequently migrate toward defect sinks. Neutron irradiation also leads to the formation of He- and ^3H -filled bubbles, which represent the dominant sink population in irradiated Be. As a result, high-binding-energy “SA-V” complexes are expected to migrate preferentially along bubble surfaces, enhancing impurity segregation at these sinks. At the maximum irradiation temperature of 650 °C ($\approx 0.6 T_m$), thermal segregation becomes significant, and the mobility of Mg increases substantially, promoting its accumulation at sinks. In contrast, Fe and Cr are expected to play a largely neutral role in RIS in Be. Their relatively low mobility limits their participation in vacancy-mediated transport, and they can additionally act as nucleation sites for defect sinks, which may partially suppress the segregation of other impurities during irradiation.

In the initial unirradiated state, the impurity elements considered in this study are present in complex precipitates, with exception of Fe that is homogeneously distributed within the matrix (Section 3.1). Under neutron irradiation, these precipitates undergo radiation-induced dissolution (RID). Several mechanisms contribute to this process, including ballistic disordering of chemical phases within displacement cascades, radiation-enhanced diffusion (RED), preferential solute transport of Al, Si, and Mg toward bubbles via radiation-induced segregation (RIS), and a reduction in the thermodynamic stability of the precipitate phases. The sequence of dissolution of individual elements reflects their mobility, their binding energy with vacancies, and the thermodynamic stability of the intermetallic phases in which they are incorporated.

At irradiation temperatures of 370 °C and 440 °C, displacement cascades partially disrupt the precipitates, with Al and Si preferentially released into the matrix. At higher irradiation temperatures (560 °C and 650 °C), the combined effects of cascades and RED lead to a complete loss of ordering in the precipitates. Under these conditions, Al, Si, and Mg segregate at the He- and ^3H -filled bubbles, whereas Fe, Cr, and Mn remain dispersed in the matrix. The absence of Fe, Cr, and Mn segregation at bubble surfaces can be attributed to three factors. First, these elements exhibit weak binding to vacancies in Be and are therefore not efficiently transported by vacancy fluxes to defect sinks, remaining near their original precipitate sites. Second, Fe–Be, Cr–Be, and Mn–Be bonds are stronger than Al–Be, Si–Be, and Mg–Be bonds, resulting in greater thermodynamic stability of their intermetallic compounds. Third, Fe, Cr, and Mn appear to interact weakly with bubble surfaces and are thus prone to re-enter the matrix rather than being retained at bubble interfaces.

These observations suggest that vacancy defects may promote the RIS of Al, Si, and Mg in Be, analogous to defect-assisted solute transport observed in other alloy systems, such as Cr and Fe in ferritic steels [26–28], Cr and Nb in vanadium alloys [29–30], Fe in chromium alloys [31], Si in Ni alloys [32], and Re in Mo alloys [23,34]. However, a dedicated theoretical investigation is required to quantitatively assess this mechanism. Finally, it is plausible that Al atoms preferentially interact with basal planes of bubble surfaces, leading to their segregation at these interfaces, whereas interaction with pyramidal planes may be energetically unfavourable and effectively repulsive. Such anisotropic segregation behaviour could arise from differences in surface energy, atomic packing, or local strain associated with crystallographic orientation of the bubble interfaces.

5. Conclusions

This study provides detailed nanoscale characterization of impurity particles and irradiation-induced defects in Be irradiated as part of the HIDOBE-02 programme at 370, 440, 560, and 650 °C to damage doses of 21, 29, 34, 37 dpa, respectively.

In the as-delivered state, Be contains a heterogeneous distribution of Fe-, Al-, Mn-, Cr-, and Si-containing intermetallic precipitates. Neutron irradiation at temperatures of 370 °C and 440 °C leads to the dissolution of Al, and to a lesser extent Si, into the matrix. In contrast, irradiation at higher temperatures of 560 °C and 650 °C results in the complete dissolution of these precipitates and the subsequent redistribution of all impurity elements within the Be matrix.

A clear dependence of local composition near the bubbles on irradiation temperature was established. At low temperatures (370–440 °C), only Al segregates at bubble surfaces, with Si enrichment detectable only near residual Si-bearing precipitates. At intermediate temperatures (560 °C), Si enrichment becomes substantial, and at the highest irradiation temperatures (650 °C), a mixed segregation layer of Al, Si, and Mg forms, with Mg and Si decorating all bubble facets and Al confined to basal planes. Fe and Cr, despite being common impurities, do not exhibit measurable bubble segregation at any irradiation condition studied.

A consistent model of radiation-induced segregation of impurities in neutron-irradiated Be is proposed, describing the obtained results from the point of view of the known RIS theory. The distinct facet-dependent segregation behavior suggests differences in defect binding and mobility, likely mediated by irradiation-induced vacancies. The observed temperature-dependent impurity redistribution provides new insight into the mechanisms governing bubble evolution and impurity transport in neutron-irradiated Be. These findings contribute to improved understanding of Be behavior under high-dose neutron irradiation and support the optimization of Be-based materials for advanced fusion applications.

CRedit authorship contribution statement

Michael Klimenkov: Writing – original draft, Visualization, Validation, Investigation. **Vladimir Chakin:** Writing – review & editing, Investigation, Formal analysis. **Ramil Gaisin:** Validation, Formal analysis. **Uta Jäntschi:** Investigation. **Michael Rieth:** Validation, Supervision, Resources. **Viacheslav Kuksenko:** Writing – review & editing, Validation, Methodology.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Acknowledgements

This work has been carried out within the framework of the EUROfusion Consortium, funded by the European Union via the Euratom Research and Training Programme (Grant Agreement No 101052200 — EUROfusion). Views and opinions expressed are however those of the author(s) only and do not necessarily reflect those of the European Union or the European Commission. Neither the European Union nor the European Commission can be held responsible for them.

Contribution of Viacheslav Kuksenko has been part-funded by the UK EPSRC Energy Programme [grant number EP/W006839/1].

Data availability

Data will be made available on request.

References

- [1] F.A. Hernández, P. Pereslavytsev, G. Zhou, Q. Kang, S. D'Amico, H. Neuberger, L. V. Boccaccini, B. Kiss, G. Nádasi, L. Maqueda, I. Cristescu, I. Moscato, I. Ricapito, F. Cismonti, Consolidated design of the HCPB Breeding Blanket for the pre-conceptual design phase of the EU DEMO and harmonization with the ITER HCPB TBM program, *Fusion Eng. Des.* 157 (2020) 111614, <https://doi.org/10.1016/j.fusengdes.2020.111614>.
- [2] R. Gaisin, R. Rolli, M. Podoinikov, V. Kuksenko, K. Seemann, T. Bergfeldt, V. Chakin, P. Vladimirov, Microstructure and high-temperature mechanical performance of industrial-scale titanium and chromium beryllides for fusion applications, *J. Alloys. Compd.* 1044 (2025) 184503, <https://doi.org/10.1016/j.jallcom.2025.184503>.
- [3] G. Federici, L. Boccaccini, F. Cismonti, M. Gasparotto, Y. Poitevin, I. Ricapito, An overview of the EU breeding blanket design strategy as an integral part of the DEMO design effort, *Fusion Eng. Des.* 141 (2019) 30–42, <https://doi.org/10.1016/j.fusengdes.2019.01.141>.
- [4] G. Federici, C. Bachmann, L. Barucca, W. Biel, L. Boccaccini, R. Brown, C. Bustreo, S. Ciattaglia, F. Cismonti, M. Coleman, V. Corato, C. Day, E. Diegele, U. Fischer, T. Franke, C. Gliss, A. Ibarra, R. Kembleton, A. Loving, F. Maviglia, B. Meszaros, G. Pintsuk, N. Taylor, M.Q. Tran, C. Vorpahl, R. Wenninger, J.H. You, DEMO design activity in Europe: progress and updates, *Fusion Eng. Des.* 136 (2018) 729–741, <https://doi.org/10.1016/j.fusengdes.2018.04.001>.
- [5] L.M. Giancarli, M. Abdou, D.J. Campbell, V.A. Chuyanov, M.Y. Ahn, M. Enoda, C. Pan, Y. Poitevin, E.R. Kumar, I. Ricapito, Y. Strebkov, S. Suzuki, P.C. Wong, M. Zmitko, Overview of the ITER TBM Program, *Fusion Eng. Des.* 87 (2012) 395–402, <https://doi.org/10.1016/j.fusengdes.2011.11.005>.
- [6] H. Cheng, B. Gong, B. Zhou, J. Yan, L. Wang, L. Zhang, Y. Feng, X. Wang, Experimental investigation on pressure drops of purge gas helium in packed pebble beds for, *Nucl. Fusion Blanket Energ.* 17 (2024) 1309, <https://doi.org/10.3390/en17061309>.
- [7] F. Cismonti, L.V. Boccaccini, G. Aiello, J. Aubert, C. Bachmann, T. Barrett, L. Barucca, E. Bubelis, S. Ciattaglia, A. Del Nevo, E. Diegele, M. Gasparotto, G. Di Gironimo, P.A. Di Maio, F. Hernandez, G. Federici, I. Fernández-Berckeruelo, T. Franke, A. Froio, C. Gliss, J. Keep, A. Loving, E. Martelli, F. Maviglia, I. Moscato, R. Mozzillo, Y. Poitevin, D. Rapisarda, L. Savoldi, A. Tarallo, M. Utili, L. Vala, G. Veres, R. Zanino, Progress in EU breeding blanket design and integration, *Fusion Eng. Des.* 136 (2018) 782–792, <https://doi.org/10.1016/j.fusengdes.2018.04.009>.
- [8] P.V. Vladimirov, V.P. Chakin, M. Dürrschnabel, R. Gaisin, A. Goraieb, F.A. H. Kolmalez, M. Klimenkov, M. Rieth, R. Rolli, N. Zimmer, S. Udatsev, M. Kolmakov, A. Vechkutov, E. Frants, Development and characterization of advanced neutron multiplier materials, *J. Nucl. Mater.* 543 (2021) 152593, <https://doi.org/10.1016/j.jnucmat.2020.152593>.
- [9] R. Gaisin, Y. Frants, M. Kolmakov, B. Zorin, M. Kylyshkanov, M. Podoinikov, S. Udatsev, A. Vechkutov, V. Chakin, P. Vladimirov, Beryllium intermetallics: industrial experience on development and manufacture, *Nucl. Mater. Energy* 35 (2023) 101444, <https://doi.org/10.1016/j.nme.2023.101444>.
- [10] V. Chakin, R. Rolli, H.-C. Schneider, R. Gaisin, P. Vladimirov, M. Klimenkov, M. Dürrschnabel, N. Zimmer, M. Rieth, B. Gorr, F.A. Hernández, D. Radloff, A. Fedorov, M. Zmitko, M. Nakamichi, S. Udatsev, High-dose neutron irradiation of beryllium and titanium beryllide: summary and outlook, *Nucl. Mater. Energ.* 42 (2025) 101910, <https://doi.org/10.1016/j.nme.2025.101910>.
- [11] L.V. Boccaccini, F. Arbeiter, P. Arena, J. Aubert, L. Bühler, I. Cristescu, A. Del Nevo, M. Eboli, L. Forest, C. Harrington, F. Hernandez, R. Knitter, H. Neuberger, D. Rapisarda, P. Sardain, G.A. Spagnuolo, M. Utili, L. Vala, A. Venturini, P. Vladimirov, G. Zhou, Status of maturation of critical technologies and systems design: breeding blanket, *Fusion Eng. Des.* 179 (2022) 113116, <https://doi.org/10.1016/j.fusengdes.2022.113116>.
- [12] N. Zimmer, P. Vladimirov, M. Klimenkov, V. Kuksenko, Investigation of a high-dose irradiated beryllium microstructure, *J. Nucl. Mater.* 540 (2020) 152374, <https://doi.org/10.1016/j.jnucmat.2020.152374>.
- [13] M. Klimenkov, V. Chakin, A. Moeslang, R. Rolli, TEM study of beryllium pebbles after neutron irradiation up to 3000appm helium production, *J. Nucl. Mater.* 443 (2013) 409–416, <https://doi.org/10.1016/j.jnucmat.2013.07.050>.
- [14] M. Klimenkov, V. Chakin, A. Moeslang, R. Rolli, TEM study of impurity segregations in beryllium pebbles, *J. Nucl. Mater.* 455 (2014) 660–664, <https://doi.org/10.1016/j.jnucmat.2014.08.030>.
- [15] M. Klimenkov, P. Vladimirov, U. Jäntschi, V. Kuksenko, R. Rolli, A. Möslang, N. Zimmer, New insights into microstructure of irradiated beryllium based on experiments and computer simulations, *Sci. Rep.* 10 (2020) 8042, <https://doi.org/10.1038/s41598-020-64654-5>.
- [16] M. Klimenkov, P. Vladimirov, J. Hoffmann, N. Zimmer, A. Möslang, V. Kuksenko, First simultaneous detection of helium and tritium inside bubbles in beryllium, *Micron* 127 (2019) 102754, <https://doi.org/10.1016/j.micron.2019.102754>.
- [17] M. Uchida, M. Uda, T. Iwadachi, M. Nakamichi, H. Kawamura, Elemental development of beryllide electrode for pebble production by rotating electrode method, *J. Nucl. Mater.* 329–333 (2004) 1342–1346, <https://doi.org/10.1016/j.jnucmat.2004.04.195>.
- [18] A.W. Jones, R.T. Weiner, The effect of heat-treatment on the mechanical properties of some commercial beryllium, *J. Less Common Met.* 6 (1964) 266–282, [https://doi.org/10.1016/0022-5088\(64\)90123-7](https://doi.org/10.1016/0022-5088(64)90123-7).
- [19] J.S. Punni, M.J. Cox, The effect of impurity inclusions on the pitting corrosion behaviour of beryllium, *Corros. Sci.* 52 (2010) 2535–2546, <https://doi.org/10.1016/j.corsci.2010.03.024>.
- [20] S. Morozumi, N. Tsuno, S. Koda, Precipitation in beryllium-iron alloys, *Trans. JIM* 10 (1969) 64–69, <https://doi.org/10.2320/MATERTRANS1960.10.64>.

- [21] D. Webster, Beryllium Science and Technology: Volume 1, Springer-Verlag New York Inc.
- [22] M. Weisz, J. Mallen, J.L. Baron, Possibilit  d'apparition d'une phase liquide a partir de 430 C dans le beryllium commercial, *J. Nucl. Mater.* 10 (1963) 56–59, [https://doi.org/10.1016/0022-3115\(63\)90116-8](https://doi.org/10.1016/0022-3115(63)90116-8).
- [23] D. Webster, The effect of low melting point phases on the elevated temperature microstructural stability of hot pressed beryllium, *Met. Trans A* 6 (1975) 803–808, <https://doi.org/10.1007/BF02672303>.
- [24] V. Kuksenko, K. Ammigan, B. Hartsell, C. Densham, P. Hurh, S. Roberts, Irradiation effects in beryllium exposed to high energy protons of the NuMI neutrino source, *J. Nucl. Mater.* 490 (2017) 260–271, <https://doi.org/10.1016/j.jnucmat.2017.04.011>.
- [25] S. van Til, J. Hegeman, H.L. Cobussen, M.P. Stijkel, Evolution of beryllium pebbles (HIDOB ) in long term, high flux irradiation in the high flux reactor, *Fusion Eng. Des.* 86 (2011) 2258–2261, <https://doi.org/10.1016/j.fusengdes.2011.01.079>.
- [26] V. Chakin, A. Moeslang, P. Kurinskiy, R. Rolli, H.-C. Schneider, E. Alves, L.C. Alves, Tritium permeation, retention and release properties of beryllium pebbles, *Fusion Eng. Des.* 86 (2011) 2338–2342, <https://doi.org/10.1016/j.fusengdes.2011.01.091>.
- [27] V. Chakin, R. Rolli, A. Moeslang, M. Klimenkov, M. Kolb, P. Vladimirov, P. Kurinskiy, H.-C. Schneider, S. van Til, A.J. Magielsen, M. Zmitko, Tritium release and retention properties of highly neutron-irradiated beryllium pebbles from HIDOB -01 experiment, *J. Nucl. Mater.* 442 (2013) S483–S489, <https://doi.org/10.1016/j.jnucmat.2013.03.032>.
- [28] E. Ishitsuka, H. Kawamura, N. Sakamoto, & K. Nishida, Process for preparing metallic beryllium pebbles. US Patent 5958,105. Process for preparing metallic beryllium pebbles. US Patent 5958,105.
- [29] Kamer S., High DOse irradiation of Beryllium (HIDOB ) (351-01, 351-02), Design and Safety Report NRG Report, May 20th, 2005.
- [30] R.A. Johnson, N.Q. Lam, Solute segregation under irradiation, *J. Nucl. Mater.* 69–70 (1978) 424–433, [https://doi.org/10.1016/0022-3115\(78\)90258-1](https://doi.org/10.1016/0022-3115(78)90258-1).
- [31] H. Wiedersich, P.R. Okamoto, N.Q. Lam, A theory of radiation-induced segregation in concentrated alloys, *J. Nucl. Mater.* 83 (1979) 98–108, [https://doi.org/10.1016/0022-3115\(79\)90596-8](https://doi.org/10.1016/0022-3115(79)90596-8).
- [32] P.R. Okamoto, N.Q. Lam, H. Wiedersich, R.A. Johnson, Effects of defect trapping and solute segregation on defect recombination rates and void swelling in irradiated alloys, *J. Nucl. Mater.* 69–70 (1978) 821–824, [https://doi.org/10.1016/0022-3115\(78\)90352-5](https://doi.org/10.1016/0022-3115(78)90352-5).
- [33] N.Q. Lam, P.R. Okamoto, H. Wiedersich, Effects of solute segregation and precipitation on void swelling in irradiated alloys, *J. Nucl. Mater.* 74 (1978) 101–113, [https://doi.org/10.1016/0022-3115\(78\)90538-X](https://doi.org/10.1016/0022-3115(78)90538-X).
- [34] W.G. Wolfer, L.K. Mansur, The capture efficiency of coated voids, *J. Nucl. Mater.* 91 (1980) 265–276, [https://doi.org/10.1016/0022-3115\(80\)90226-3](https://doi.org/10.1016/0022-3115(80)90226-3).
- [35] E.A. Little, Void-swelling in irons and ferritic steels, *J. Nucl. Mater.* 87 (1979) 11–24, [https://doi.org/10.1016/0022-3115\(79\)90122-3](https://doi.org/10.1016/0022-3115(79)90122-3).
- [36] E.A. Little, D.A. Stow, Void-swelling in irons and ferritic steels, *J. Nucl. Mater.* 87 (1979) 25–39, [https://doi.org/10.1016/0022-3115\(79\)90123-5](https://doi.org/10.1016/0022-3115(79)90123-5).
- [37] V. Chakin, V. Kazakov, Y. Goncharenko, Z. Ostrovsky, Formation of the σ -phase in Cr-Fe alloys under irradiation, *J. Nucl. Mater.* 233–237 (1996) 573–576, [https://doi.org/10.1016/S0022-3115\(96\)00029-3](https://doi.org/10.1016/S0022-3115(96)00029-3).
- [38] F. Morito, V.P. Chakin, M.I. Danylenko, A.V. Krajnikov, Radiation-induced strengthening in EB welds of Mo-Re alloys during high temperature neutron irradiation, *J. Nucl. Mater.* 417 (2011) 976–979, <https://doi.org/10.1016/j.jnucmat.2010.12.183>.
- [39] E. Clementi, D.L. Raimondi, W.P. Reinhardt, Atomic screening constants from SCF functions. II. Atoms with 37 to 86 electrons, *J. Chem. Phys.* 47 (1967) 1300–1307, <https://doi.org/10.1063/1.1712084>.