Journal of Nuclear Materials 386–388 (2009) 793–796

Contents lists available at ScienceDirect

Journal of Nuclear Materials

journal homepage: www.elsevier.com/locate/jnucmat

Erosion of a TiAl intermetallic alloy under conditions simulating plasma disruptions

B.A. Kalin, V.L. Yakushin *, V.I. Polsky, P.S. Dzhumaev, V.T. Fedotov, O.N. Sevryukov, M.Yu. Golikov, M.V. Zolotarev

Moscow Engineering Physics Institute (State University), 31 Kashirskoe sh., Moscow 115409, Russia

A B S T R A C T

Radiation erosion and thermal stability of TiAl-based intermetallic alloys produced by vacuum-arc melting, compacting of microcrystal powders with binder and impregnantion by melt, and their brazed joints with bronze have been investigated under irradiation by high-temperature pulsed hydrogen plasma flows (the flow energy density $Q = 0.2-0.9$ MJ/m^2, the pulse duration 15 μs, the number of pulses 1–21) which simulate the expected plasma disruptions in a fusion reactor. It has been found that the erosion coefficients and thermal stability of alloys are determined by the way of their fabrication, and compacted intermetallides have a higher thermal stability in comparison with the cast ones. The brazed joints of the intermetallic compound with bronze under irradiation by pulsed hydrogen plasma up to the energy density $Q = 0.75$ MJ/m^2 have a high thermal stability and formation of cracks was not observed.

1. Introduction

One of the main problems to create a fusion reactor is the choice of plasma-facing materials (PFM) for different in-vessel components. The selection of PFM is determined mainly by plasma compatibility and erosion lifetime. At the current stage of developing the ITER, armor materials chosen for its components (Be, carbon fibre composite and W) have a number of known disadvantages [1]. Therefore for the next generation of reactors, for example the DEMO, it is necessary to look for new PFM.

Alloys based on titanium aluminide are being considered [2–5] as an alternative material as they have good temperature strength and oxidation resistance, as well as low neutron induced radioactivity. But unfortunately, they possess poor ductility at low to intermediate temperature and low thermal conductivity. Increasing the characteristics of heat resistance and resistance to creeping and cracking of alloys based on γ-TiAl and γ-TiAl + α_2-Ti_3Al can be attained by development of composite materials [5].

The present work's purpose was to discover the main features of radiation erosion and heat resistance of composite materials based on the intermetallic compound of a Ti–Al system which were obtained by various methods, as well as, their brazed joints with bronze under irradiation by high-temperature pulsed hydrogen plasma simulating the expected plasma disruptions.

2. Experimental procedure

Samples based on the intermetallic compound of a Ti–Al system were made by methods of traditional casting technique, by sintering of rapidly quenched microcrystalline intermetallic powders with a more fusible binder, and by impregnation of sintered intermetallic samples. Powders of intermetallic alloys were obtained by rapid quenching of a melt on a rotating and cooled cupper disk-refrigerator followed by relaxation annealing to decrease their plasticity and reduce to powder the obtained ribbon fragments in a planar ball mill.

The samples of composite materials, based on $Ti_{66}Al_{34}$ and $Ti_{54}Al_{34}V_{12}$ (in wt%) intermetallic compounds, were obtained by sintering in vacuum of titanium aluminide microcrystalline powder with particles less than 20 μm in size using a binding powder made from a $Ti_{68.5}V_{28}Be_{3.5}$ alloy in amounts of 6 wt%. To obtain porosity-free compacted samples based on a $Ti_{66}Al_{34}$ intermetallic alloy, rapidly quenched powders with a particle size less than 45 μm and in the 45–80 μm range were molded in the form of tablets followed by vacuum sintering at 1100 °C for 1 h. The used regimes of pressing and sintering allowed preserving the initial microstructure of separate particles which was formed owing to rapid hardening of a melt. The obtained tablets of the intermetallic matrix and the necessary quantity of a rapidly quenched 'impregnant-alloy' powder ($Ti_{60}V_7Cr_{19}Be_{14}$ and $Ti_{65}Be_{35}$, in at.%) were placed into a vacuum heater in a quartz capsule, heated to 1100 °C and held for 30 min.

Besides, the thermal stability of brazed joints of tablets compacted and impregnated by $Ti_{65}Be_{35}$ and CuCrZr–bronze was also investigated. A rapidly quenched amorphous ribbon-type filler

* Corresponding author. Tel.: +7 495 323 9271; fax: +7 495 324 3165.
E-mail address: vlyakushin@mephi.ru (V.L. Yakushin).

0022-3115/$ - see front matter
doi:10.1016/j.jnucmat.2008.12.276

metal STEMET 1202 of the following composition (in wt%) Ti–22Cu–12Zr–12Ni–1V–1.5Be was used as a filler-metal alloy. The intermetallic compound was brazed with bronze at 820 °C for 10 min to exclude the degradation of bronze's mechanical properties.

The samples were irradiated on a Z-pinch type experimental pulsed plasma installation by flows of high-temperature pulsed hydrogen plasma (HTPP) with a maximum ion energy up to 2 keV and an energy density of the incident flow changing in the 0.2–0.9 MJ/m^2 range at a pulse duration of $\tau_p \sim 15$ μs. The number of action pulses N changed from 1 to 21. The irradiation regimes used at the experiments significantly differ from the expected parameters of plasma flows onto the first wall under plasma disruptions in ITER ($Q \sim 2$ MJ/m^2, $\tau_p \sim 0.1$–3 ms). Nevertheless, taking into consideration that the thermal action will be the main factor resulting in erosion of materials under plasma disruptions, to prove the legitimacy of the selection of irradiation regimes to simulate plasma disruptions, it is possible to compare the integral damage coefficient $F = W\tau_p^{1/2}$ ($W = Q/\tau_p$, the flow specific power) for various installations which determines the extent of materials erosion [6]. The coefficient F will change in the (36.5–200) $MW \cdot s^{1/2}/m^2$ range for the expected plasma disruptions onto the first wall in ITER, and in the (58–228.5) $MW \cdot s^{1/2}/m^2$ range for the selected irradiation regimes in the plasma installation. Thus, the selected irradiation regimes by high-temperature pulsed hydrogen plasma make it possible to simulate sufficiently well the action of expected plasma disruptions onto the first wall of the ITER reactor.

The initial and irradiated by HTPP-flows samples were investigated by metallography, scanning electron microscopy, X-ray spectrum microanalysis and qualitative X-ray phase analysis. The erosion coefficient S [$\mu g/(J \cdot cm^2)$] of the irradiated samples was calculated from a mass loss of the target with a weighing accuracy of ±0.01 mg.

3. Experimental results and discussion

Metallographic investigations have shown that the microstructure of the intermetallic alloys of a Ti–Al–V system in the initial cast state is characterized by the presence of columnar dendritic crystals with a distance between the secondary axes of ∼30–40 μm. Particles of titanium aluminide in samples sintered with a binder have a fragmentation shape with sizes of ∼ 10 μm, and the distribution of a binder is sufficiently uniform in the alloy. Qualitative X-ray phase analysis of samples has shown that the main phase of both cast and powder alloy is γ-phase (TiAl). Besides, there are additional lines corresponding to α_2-phase (Ti_3Al) in the spectrum of samples.

Results of metallography and electron microscopy of irradiated samples have shown that the influence of HTPP results in melting and cracking of the surface (Fig. 1). Besides, an outlet of the materials of a binder and an impregnant onto the surface of sintered samples is observed. At that, increasing the HTPP-flow specific power results in formation of a developed relief in the form of drops and waves of the solidified melt. It should be pointed out that under the given irradiation regime there is also cracking of samples of sintered powder tungsten which was investigated as a material for comparison.

Dependence between the erosion coefficients and the number of irradiation pulses is shown in Fig. 2. The figure shows that the maximum erosion of samples takes place at the starting irradiation stage. At that, a monotonous decrease of erosion is observed for cast samples of a $Ti_{54}Al_{34}V_{12}$ alloy, S-coefficients practically become constant within the error of their determination at $N > 15$, and the erosion coefficient averaged through 21 irradiation pulses

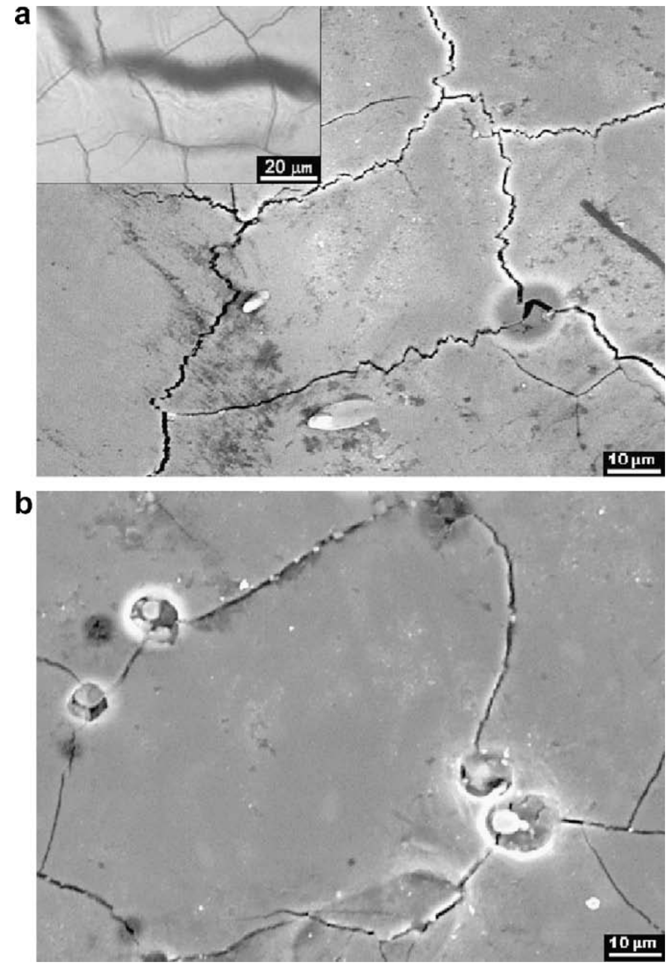

Fig. 1. Surface topography of $Ti_{43}Al_{48}V_9$ intermetallic samples after irradiation by hydrogen plasma flows ($Q = 0.7$ MJ/m^2, $N = 21$): (a) cast alloy, (b) compacted alloy.

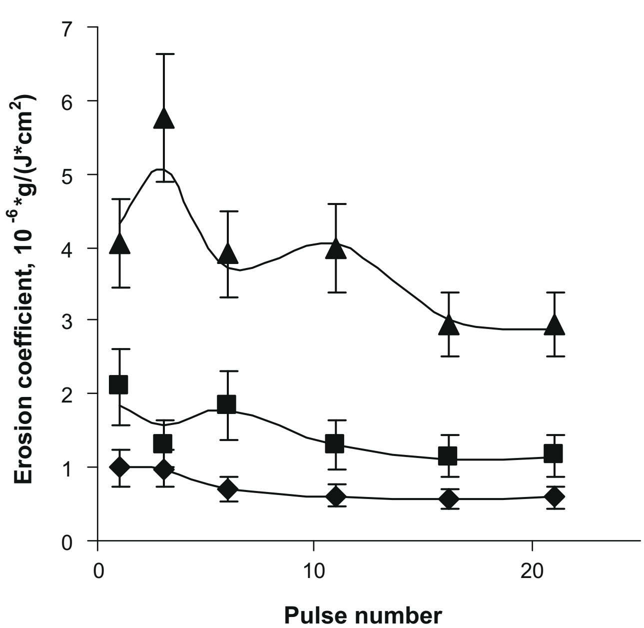

Fig. 2. Dependence between the erosion coefficient of intermetallic alloys and the number of irradiation pulses ($Q = 0.7$ MJ/m^2): ■ – $Ti_{43}Al_{48}V_9$ (cast); ▲ – $Ti_{66}Al_{34}$ + 6 wt% of binder; ♦ – $Ti_{43}Al_{48}V_9$ + 6 wt% of binder.

is equal to ~0.6 μg/(J · cm^2). The found dependence is due to a change in the structural-phase state and the element composition of the surface layers under the action by HTPP-flows [2]. In case of irradiation of compacted alloys (especially those of the system TiAl + 6% of binder), a periodic change of erosion coefficients is observed at the starting irradiation stages. It is conditioned by outlet of more fusible (T_m~1150 °C) material of a $Ti_{68.5}V_{28}Be_{3.5}$ binder along the grain boundaries and forming microcracks onto the target surface and its evaporation under subsequent irradiation. Fig. 2 shows that cast Ti–Al–V intermetallic alloys possess a higher resistance to mass loss under irradiation by HTPP-flows in comparison with that of samples compacted with a binder. It is necessary to mention that the erosion coefficient of sintered intermetallic alloys of the Ti–Al system is significantly higher than that of alloys additionally alloyed by vanadium. Higher erosion coefficients of the Ti–Al alloys in comparison with those of the Ti–Al–V system were also found for cast samples. This is conditioned by the influence of vanadium on the structural-phase state of alloys and apparently on increasing their thermal conductivity and is in a good agreement with the results obtained earlier [2].

The thinning depth of investigated samples under irradiation was calculated from results of the target mass loss (Fig. 3). The figure shows that the value of thinning monotonically increases with the number of pulses. From extrapolation estimations for 500 expected plasma disruptions in ITER, this value will not exceed 100 μm. However, it is necessary to note that the expected duration of plasma disruptions in ITER will be an order of magnitude bigger than in our experiments. That is why the calculated value of thinning will be of the order of 1 mm, which is significantly less than the thickness of armor materials.

The front structure of irradiated samples was investigated to discover the spreading character and penetration depths of cracks after irradiation by HTPP-flows (Fig. 4). It has been found that the forming cracks predominantly develop at right angle to the irradiated surface and, as a rule, their spreading finishes at the boundaries of grains. At that, the penetration depth of cracks for cast samples can attain ~70 μm and in some cases formation of tangential cracks (Fig. 4(a)) is observed. For samples compacted by sintering, the maximum depth of spreading the cracks does not exceed 15 μm and their spreading is blocked by the surface layer modified by the action of plasma flows. The structural-phase state and chemical composition of the surface layer are changed as a result of modification. Formation of similar cracks will not significantly influence, with a high probability, the operational properties of armor material.

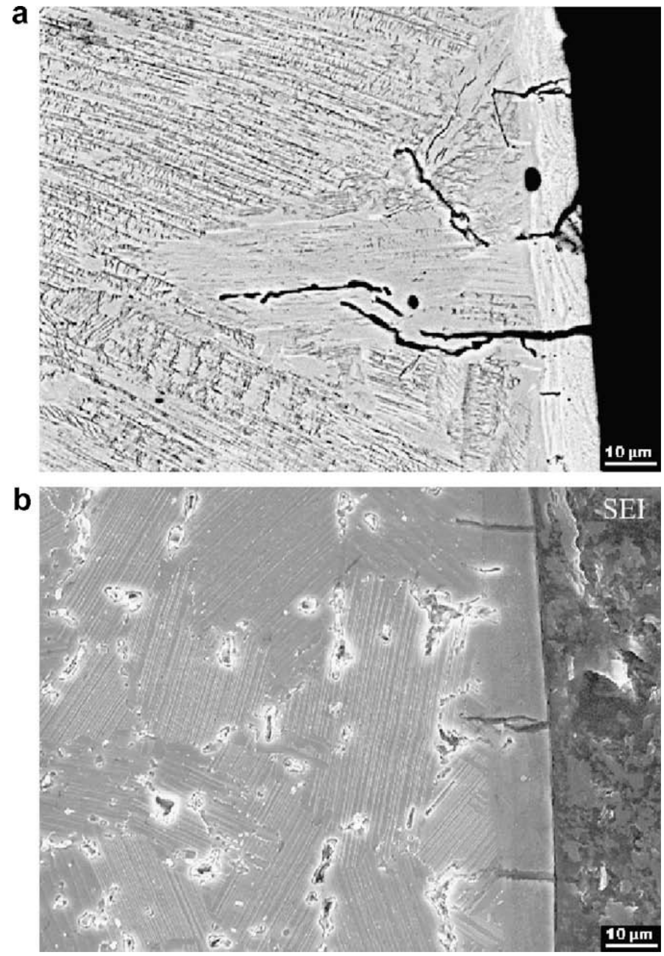

Fig. 4. Microstructure of the front sections of $Ti_{43}Al_{48}V_9$ intermetallic alloys after irradiation by HTPP-flows (Q = 0.7 MJ/m^2, N = 21): (a) cast alloy; (b) compacted alloy + 6 wt% of binder.

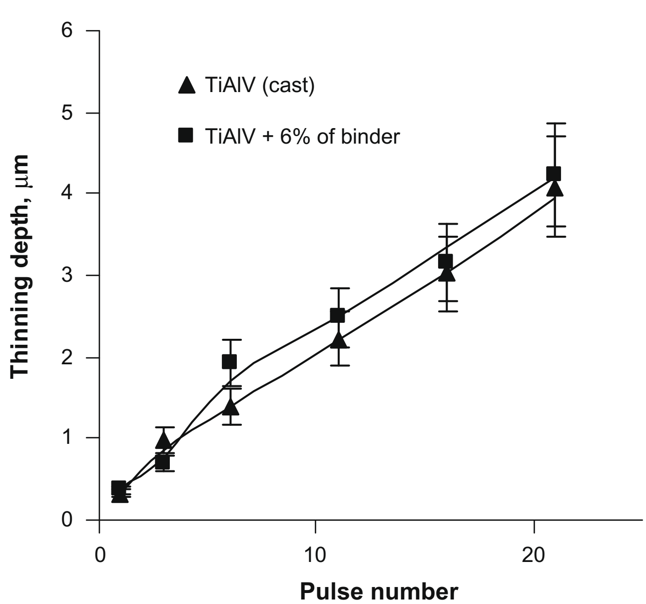

Fig. 3. Dependence of the thinning extent of TiAlV intermetallic compound irradiated by pulsed hydrogen plasma flows (Q = 0.7 MJ/m^2) from the pulse number.

Fig. 5 shows dependence of the erosion coefficient of the brazed sample's front surface (impregnated intermetallic compound $Ti_{66}Al_{34}$–bronze) from the plasma flow specific power. It is seen from the figure that the maximum erosion is observed at the starting irradiation stage. This is due to evaporation of a more fusible material of the impregnant-alloy ($Ti_{60}V_7Cr_{19}Be_{14}$ or $Ti_{65}Be_{35}$) from the surface layers that goes out onto the surface through thin initial micropores and technological microcracks. Further irradiation significantly decreases, as the pores heal, the outlet of a binder material, which stipulates decreasing the erosion coefficient. At that, separate erosion fluctuations are observed at the expense of microcracks' formation in a $Ti_{66}Al_{34}$ tablet and owing to outlet of the impregnant-alloy onto a surface through these microcracks and its evaporation. It is necessary to note that the area of a brazed joint for all the irradiation regimes up to the energy density of a plasma flow Q = 0.75 MJ/m^2 has a high thermal stability and formation of cracks was not observed. This proves a principle possibility to use brazed joints, obtained with the use of rapidly-quenched amorphous ribbon filler metals of the type STEMET 1202, for working under hard conditions in the units of fusion installations.

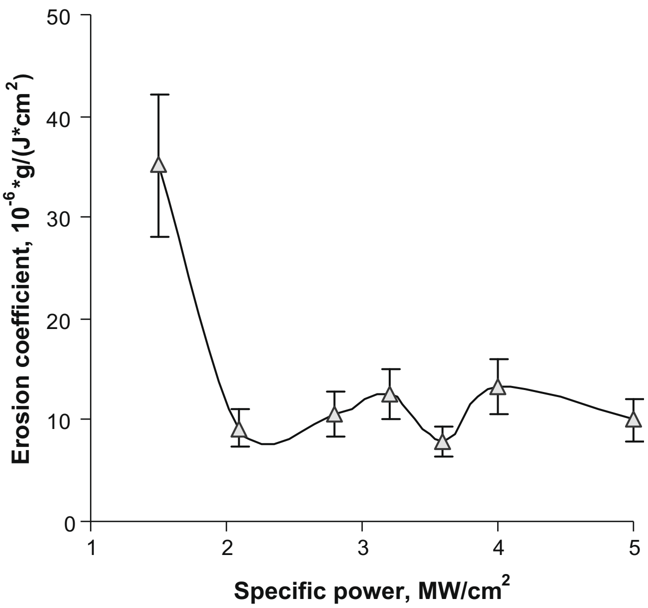

Fig. 5. Influence of the power density of a plasma flow on the erosion coefficient of the front surface of brazed 'intermetallic compound–bronze' joints.

4. Conclusion

The experimental investigations and analysis of the results obtained have shown that the exposure of intermetallic alloys of the Ti–Al and Ti–Al–V systems to the action of pulsed hydrogen plasma flows with an energy density of flow changing within 0.2–0.9 MJ/m^2 and at a pulse duration of 15 μs leads to a surface melting and formation of microcracks. At that, an outlet of the material of a binder or an impregnant onto the target surface, its evaporation and formation of a developed relief under solidification are observed for samples compacted by sintering and impregnated by a melt.

It has been found that the erosion coefficients of intermetallic alloys are determined by their initial composition and fabrication technology. It has been revealed that the maximum erosion of samples takes place at the starting irradiation stage and the erosion coefficients for the cast alloys are less than those of the compacted alloys with the use of fusible binder.

It has been discovered that the penetration depth of cracks is determined by the fabrication technology of an intermetallic compound. At that, this value for cast Ti–Al–V samples reaches ∼70 μm and in some cases formation of tangential cracks is observed, while for sintered samples the maximum penetration depth does not exceed 15 μm. The latter proves a significant increase of the thermal stability of compacted intermetallic compounds and a possibility to preserve their operational properties as the first wall armor material.

The erosion of samples of double-layer brazed joints preferably takes places at the expense of evaporation of the impregnant material from a compacted intermetallic coating $Ti_{66}Al_{34}$ owing to its starting technological porosity in a volume. In this case, the brazed joint zone between the intermetallic screen and the substrate from bronze under irradiation by hydrogen HTPP-flows up to the flow energy density Q = 0.75 MJ/m^2 has a high thermal stability and formation of cracks was not observed.

References

[1] G. Kalinin, V. Barabash, A. Cardella, et al., J. Nucl. Mater. 283–287 (2000) 10.
[2] V.L. Yakushin, B.A. Kalin, A.V. Shul'ga, et al., J. Nucl. Mater. 258–263 (1998) 1127.
[3] O. Okado, K. Nakata, K. Fukai et al., Effects of Mn and Si additions on microstructural development in TiAl intermetallic compounds irradiated with He-ions, Ibid. 258–263 (1998) 1750.
[4] T. Kondo, Y. Watanabe, Y.S. Yi, A. Hishinuma, An evaluation of potential material-coolant compatibility for applications in advanced fusion reactor, Ibid. 258–263 (1998) 2083.
[5] B.A. Kalin, V.L. Yakushin, et al., Voprosy atomnoy nauki i tekhniki 3–4 (2002) 39. in Russian.
[6] M. Fujitsuka, H. Shinno, T. Tanabe, H. Shiraishi, J. Nucl. Mater. 179–181 (part A) (1991) 189.