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CORRECTION

To the article "Development of Dense and Hard Materials Based on Oxide – Non-Oxide Compounds with Added Intermetallic Components During Spark Plasma Sintering," by A. V. Hmelov, Vol. 62, No. 5, pp. 570 – 586, January, 2022, DOI 10.1007/s11148-022-00645-5

Replace the text on page 580, paragraph 3, column 2 through the first paragraph on page 584 with

The development of the physico-mechanical properties of the samples, containing Ti_2AlNb, NiTi, NiNbZr, and NiVTa as additives, is consistent with the sintering of these composi-tions in the temperature range from 1200 to 1600°C. Thus, the most active increase in E, K_{1c}, and HV is shown by the NiTi- and NiVTa-containing samples, as opposed to a smooth increase of the values of these properties in case of the Ti_2AlNb- and NiNbZr-containing samples.

A gradual development of the physico-mechanical properties of the Ti_2AlNb-containing sample in the temperature range from 1200 to 1600°C is caused by a non-uniform sintering of a variety of areas with different density and different sizes, with partial filling of the pores (see Fig. 5a, a_1), by a relatively monodispersed granular composition of the crystalline phases (see Fig. 6) and the formation of the narrow boundary layers of c-ZrO_2 and c-BN in combination with a slightly broad intermediate layer of Ti_2AlNb (see Fig. 8a – a_2). This cause incomplete dissipation of stresses at the boundaries of variously-dispersed particles and distribution of plastic properties in the narrow layers of c-ZrO_2 and c-BN, non-uniform increase of rigidity and hardness, reinforcing and strength-hening at the boundary areas of particles of sample. This is related with the displacement of disloca-tions of the low-density near a broader boundary layer of Ti_2AlNb (see Fig. 10a_{1-2}) in combination with a variety of fine, point-type dislocations, layered dislocations of the uniform and dense structu-res near the narrow boundary layers of c-ZrO_2 and c-BN (see Fig. 10a_{1-2}, a_{1-3}). This is explained by a higher mobility and weaker compaction at the dislocation plate joints in the displacement of dislocations (see Fig. 10a_{1-2}), incomplete interaction of the point-type dislocations with the local regions of stresses at the boundary of the point-type and displacement of dislocations (see Fig. 10a_{1-2}), as well as non-uniform reinforcing and strengthening of the globular formations within layered dislocations (see Fig. 10a_{1-3}). In this sample, narrow microcracks propagate in a tortuous path over long distances in greater quantities (see Fig. 10a_1, a_{1-1}), mainly near the displacement of dislo-cations, and partially at the boundary of the point-type and displacement of dislocations with the Ti_2AlNb boundary layer at 1500°C (see Fig. 10a_{1-2}). A less tortuous propagation path of the wide microcracks is observed in the sample at 1300°C (see Fig. 10a, a_{0-1}), where the reinforcing and strengthening of the structure of variously-wide boundary layers of c-ZrO_2, c-BN, Ti_2AlNb is lower.

The most active growth of the properties of the NiTi-containing sample is associated with the uniform and densely sintered microstructures (see Fig. 5b, b_1), polydisperse grain composition of the crystalline phases (see Fig. 6), narrow boundary layers of c-ZrO_2, c-BN and NiTi (see Fig. 8b – b_2). This accelerate dissipation of stresses at the boundaries of variously-dispersed particles and distribution of plastic properties in the narrow layers of c-ZrO_2, c-BN, NiTi, increase the rigidity and hardness, reinforcing and strengthening of the boundary of particles of sample. This is related with the formation of ideal dislocations and combination of dislocations in the toward of the narrow boundary layers of c-ZrO_2, c-BN, NiTi (see Fig. 10b_{1-2}, b_{1-3}), which actively interact with the va-rious-sized particles of mullite, (Ti,Mo)(C,N), c-ZrO_2, c-BN, and NiTi (see Fig. 6) as well as local regions of stresses around these particles, promoting a uniform dissipation of stresses in these dis-locations and boundary layers. This is explained by strong compaction (low mobility) and enhanced elastic properties at the joints of various-sized plates of such dislocations. Due to the most uniform and complete dissipation of stresses in these dislocations, distribution of plastic properties in the boundary layers are absent the microcracks localization sites. Thus, are no microcracks in this sample at 1500°C (see Fig. 10b_1) compared to the tortuous propagation path of a long narrow microcrack at 1300°C (see Fig. 10b, b_{0-1}).

The NiNbZr-containing sample shows a less active increase of the physico-mechanical properties in the temperature range from 1200 to 1600°C. This is explained by a non-uniform and weakly sintered microstructure (see Fig. 5b, b_1), less polydisperse grain composition of the crystalline phases (see Fig. 6), more wide boundary layers of β-Si_3N_4, Ni(Nb,Zr), Ni(Zr,Nb) and $Ni_{45}Nb_{35}Zr_{20}$ (see Fig. 8c_1 – c_3). In a result, non-uniformly dissipate the stresses at the boundaries of variously-dispersed particles and distribute the plastic properties in the narrow boundary layers with the formation of various areas of brittleness, in particular smaller in the narrow boundary layer of the Ni(Nb,Zr) and larger in the wide boundary layers of the β-Si_3N_4, Ni(Zr,Nb), $Ni_{45}Nb_{35}Zr_{20}$ of the sample (see Fig. 8c_2, c_3). This causes lower rigidity and hardness, reduced reinforcing and strengthening at the boundary particles areas. This is caused by layered dislocations with different uniformity and density near the various-width boundary layers of β-Si_3N_4, Ni(Nb, Zr), Ni(Zr, Nb), and $Ni_{45}Nb_{35}Zr_{20}$ (see Fig. 10$c_{1\text{-}2} – c_{1\text{-}4}$). As a result, form a highly non-uniform stress areas in these dislocations and boundary layers, which are caused by different mobility and compac-tion at the joints of the plates and globular formations of such dislocations. This is more pronounced in the layered dislocations with a non-uniform, low-density structure with unevenly compacted glo-bular formations near a wide, non-uniform boundary layer of $Ni_{45}Nb_{35}Zr_{20}$ (see Fig. 10$b_{1\text{-}4}$), less actively in the layered dislocations with a partially dense structure with slightly compacted wide plates near the uniform, equally wide boundary layers of β-Si_3N_4 and Ni(Nb,Zr) between them a joint (see Fig. 10$b_{1\text{-}2}$) during a gradual deformation shift from the β-Si_3N_4 layer to the Ni(Nb,Zr) layer due to a high density of the layered dislocations and greater hardness of the Ni(Nb,Zr) layer and minimal in the layered dislocations of the uniform and dense structure with strongly compacted globular formations near the non-uniform, various-width boundary layers of β-Si_3N_4 and Ni(Zr,Nb) (see Fig. 10$c_{1\text{-}3}$). The arrangement of the more dense and hard Ni(Nb,Zr) structure between the less dense and brittle structures of Ni(Zr,Nb) and $Ni_{45}Nb_{35}Zr_{20}$ of different densities (see Fig. 8c_3) doesn't reinforce or strengthen the Ni(Zr,Nb) and $Ni_{45}Nb_{35}Zr_{20}$ structures. This is caused by a higher mobility at the globular formation joints in the layered dislocations near the $Ni_{45}Nb_{35}Zr_{20}$ boundary layer (see Fig. 10$c_{1\text{-}4}$), partially at the wide plate joints of the layered dislocations near the Ni(Nb,Zr) boundary layer (see Fig. 10$c_{1\text{-}2}$), and by incomplete, weak reinforcing and strengthening of the Ni(Zr,Nb) and Ni(Nb,Zr), Ni(Zr,Nb) and $Ni_{45}Nb_{35}Zr_{20}$ structures (see Fig. 8c_3). In such sample, wide microcrack propagates tortuously over long distances (see Fig. 10c_1, $c_{1\text{-}1}$) near the layered dislocations of non-uniform, low-density structures with the $Ni_{45}Nb_{35}Zr_{20}$ boundary layer (see Fig. 10$c_{1\text{-}4}$), partially near the layered dislocations with boundary layers of the β-Si_3N_4 and Ni(Nb,Zr) boundary layers at 1500°C (see Fig. 10$c_{1\text{-}2}$) with higher tortuousty of microcracks around fragile indentation press with a variety of fine chips (see Fig. 10c) due to reinforcing, strengthhening the boundaries of β-Si_3N_4 and *c*-BN, *c*-BN, and Ni(Zr,Nb)/Ni(Nb,Zr) at 1500°C compared to the tortuous-linear propagation path of the longer microcrack with the existence of different density and particle sizes of the *c*-BN, Ni(Nb,Zr), Ni(Zr,Nb) and $Ni_{45}Nb_{35}Zr_{20}$ (see Fig. 10$c_{0\text{-}1}$) with relatively rectilinear microcracks path around indentation press (see Fig. 10*c*) due to minimal reinforcing and strengthening the boundaries of these particles at 1300°C.

Greater ingrowth of the physico-mechanical properties of the NiVTa-containing samples is associated with relatively uniform and densely sintered microstructures (see Fig. 5*d*, d_1), more polydisperse grain composition of the crystalline phases (see Fig. 6), and more narrow boundary layers of the β-Si_3N_4, Ni(Ta,V) and Ni(V,Ta) (see Fig. 8$d – d_3$). As a result, more uniformly dissi-pate the stresses at the boundaries of variously-dispersed particles and distribute the plastic pro-perties in the boundary layers. This increase the rigidity and hardness, reinforcing and strengthening at the boundaries of the various-sized particles of the sample. The reason is mainly the combination and partially layering the dislocations of high, slightly low uniformity and density with about the equal-width plates near the uniform, narrow boundary layers of β-Si_3N_4, Ni(Ta,V), and Ni(V,Ta) (see Fig. 10$d_{1\text{-}2}$, $d_{1\text{-}3}$). This considerably reduces the initiation of non-uniform areas of stresses in such dislocations and boundary layers due lower mobility and greater compaction at the plate joints of these dislocations. These processes are lower in the layered dislocations with a non-uniform, low-density structure with weakly compacted wide plates near the nonuniform, wide boundary layer of $Ni_{17}V_{61}Ta_{22}$ (see Fig. 10$d_{1\text{-}4}$). At the same time, a double joint appears between the boundary layers of β-Si_3N_4 and Ni(Ta,V) (see Fig. 10$d_{1\text{-}2}$) during an active deformation shift from the β-Si_3N_4 layer to the Ni(Ta,V) layer due to the low density of the layered dislocations and low hard-ness of Ni(Ta,V). As a result, this sample is characterized by different microcrack types and propa-gation paths, in particular branching and tortuousty with retarding of microcracks propagation at 1500°C (see Fig. 10d_1, $d_{1\text{-}1}$, $d_{1\text{-}1\text{-}1}$). A similar tortuous path of microcrack propagation with small quantity of microcracks around indentation press at 1500°C (see Fig. 10d_1) as at 1300°C (see Fig. 10*d*) with small microcrack length is visible. In the middle of the microcrack is located the bridge of Ni(V,Ta), which partially inhibits the microcrack propagation at 1300°C (see Fig. 10$d_{0\text{-}1}$). Thus, branching of a wide microcrack with a tortuous propagation path (see Fig. 10$d_{1\text{-}1}$) is developed near the layered dislocations with a non-uniform, low-density structure of the $Ni_{17}V_{61}Ta_{22}$ boundary layer and the layered structure of Ni(Ta,V) (see Fig. 10$d_{1\text{-}4}$). This is caused by a different density, activity of cracking of $Ni_{17}V_{61}Ta_{22}$ and Ni(Ta,V) structures, wherein primary — $Ni_{17}V_{61}Ta_{22}$, secondary — Ni(Ta,V) with the absent of the mechanism of reinforcing and strengthening of such structures via microcrack branching [5]. This is cause the embrittlement and destruction of boun-dary areas of these structures and corresponding boundary layers. Incorporation of the dense Ni(V,Ta) structure into the brittle $Ni_{17}V_{61}Ta_{22}$ structure (see Fig. 8d_3) doesn't reinforce and strengthen the $Ni_{17}V_{61}Ta_{22}$ structure. This is related with a higher mobility at the wide plate joints in the layered dislocations near the $Ni_{17}V_{61}Ta_{22}$ boundary layer (see Fig. 10$d_{1\text{-}4}$), an incomplete compaction and strengthening of the Ni(V,Ta) and $Ni_{17}V_{61}Ta_{22}$, Ni(Ta,V) and Ni(V,Ta) structures (see Fig. 8d_3). The inhibition of the narrow microcrack propagation along a tortuous path with a complete stop near the Ni(V,Ta) particles (see Fig. 10$d_{1\text{-}1\text{-}1}$) is actively near the combination of dislocations of a uniform and dense structure with completely densified small plates and boundary layers of β-Si_3N_4 and Ni(V,Ta) (see Fig. 10$d_{1\text{-}3}$). This related with an intensive interaction of the propagating micro-crack, combination of dislocations with various-sized β-Si_3N_4 and Ni(V,Ta) particles (see Fig. 6), the active dissipation of stresses in front of the microcrack with stress reduction at the microcrack — Ni(V,Ta) particles interface due to higher reinforcing and strengthening of small plate joints (see Fig. 10$d_{1\text{-}2}$), various-sized β-Si_3N_4 and Ni(V,Ta) particles (see Fig. 6) and enhanced plastic proper-ties at the small plate joints in the combination of dislocations (see Fig. 10$d_{1\text{-}3}$), at the boundaries of various-sized β-Si_3N_4 and Ni(V,Ta) particles (see Fig. 6). As a result, the combination of disloca-tions (see Fig. 10$d_{1\text{-}3}$) and various-sized particles of β-Si_3N_4 and Ni(V,Ta) (see Fig. 6) mostly rein-force and strengthen the structure of the boundary layers of β-Si_3N_4 and Ni(V,Ta), as opposed to layered dislocations with the boundary layers of β-Si_3N_4 and Ni(Ta,V) (see Fig. 10$d_{1\text{-}2}$). The above-described processes along with results of physico-mechanical properties of the samples have a different effect on the linear correlation of E and K_{1c} in the temperature range from 1200 to 1600°C (Fig. 11).