CORRECTION

To the article “Stimulation of Spark-Plasma Sintering of Mixtures of Oxide–Non-oxide Components by Adding a Solid Solution of TaB₂–NbC and through a Nickel Melt in Mixtures of Metal Powders,” by A. V. Hmelov, Vol. 62, No. 1, pp. 74 – 88, May, 2021

Please replace the following text:

Literature review — pp. 74 – 75

The main method for the production of **hard** and impact-resistant ceramics and cermets is related to the spark-plasma **solid-state** sintering of ceramic and metalloceramic compositions with additives of metal powders or metal powder mixtures [1 – 5]. The nonuniform evolution of this sintering affects the phase composition and microstructure of boundary layers of oxide-**non-oxide** and oxide-**non-oxide**-metal crystalline phases and, as a result, causes variations in the elasticity modulus, fracture toughness, and hardness of these materials [2 – 5]. The degree of crystallinity, brittleness or density of crystalline phases, homogeneity of the microstructure, **and** width and ratio of brittle and elastic properties of boundary layers are responsible for the relationship between the phase composition, microstructure, and properties of the boundary layers of crystalline phases [2 – 5]. Less dense crystalline phases, in particular *h*-BN, β-WC, β-WB, and β-Zr, Ta form nonuniform, broad, and brittle intermediate layers, while amorphous carbon is extensively incorporated and accumulated between grains. As a result, the plastic properties of these boundary layers are degraded; regions of dislocations and microcrack boundary layers are generated, and the boundary structure of the material grains becomes brittle [2, 3, 5]. Dense crystalline phases, e.g., *c*-BN, β-MoC, β-MoB, and β-Ta, Zr, form uniform, narrow, and strong boundary layers that promote uniform and complete distribution of the plastic properties and propagation of microcracks **by** a tortuous **path with** slow velocity or hinder the advance of microcracks in boundary layers; as a result, the boundary structure of the material grains is **reinforced** and strengthens [2 – 5]. These properties of the boundary layers are important because they determine how extensively and evenly ceramic and cermet materials absorb, accumulate, and dissipate impact energy and the ratio of absorption and dissipation/accumulation of different amounts of impact energy [2 – 5]. The method used for reinforcing and strengthening the boundary structure of grains is by adding *c*-ZrO₂ or solid solution TiC–ZrC to spark-plasma sintered ceramic and cermet compositions [6 – 8]. Such additives have low efficiency because they have dense structures and cause nonuniform and incomplete solid-state diffusion at the boundaries of the **sintered** grains [6 – 8]. Also, the **reinforcing** and strengthening of the grain boundary structure with added solid solution TiC–ZrC depend considerably on

Figure 7 on page 80 is incorrect. The correct figure is

Fig. 7. Elemental composition and peak intensities of crystalline phases in samples sintered at 1500 °C.

the structure and percent content of an actual component in this solid solution [7, 8]. A method for **reinforcing** and strengthening the grain boundary structure by producing solid solution $\text{ZrB}_2\text{--TaSi}_2$ during spark-plasma sintering of the ceramic composition $\text{ZrB}_2\text{--SiC--TaSi}_2$ is known [9]. The sintering evolves and the solid solution forms simultaneously, which facilitates more extensive and complete incorporation of the solid-solution crystalline phase into the grain boundary structure to form uniform and narrow boundary layers between **non-oxide** crystalline phases and solid solutions in addition to **reinforcing** intermediate layers between solid-solution crystalline phases [9]. This mechanism of **reinforcing** and strengthening of the grain boundary structure is realized through solid-state sintering and depends on the structure and percent content of an actual component in the solid solution [9]. Another method for **reinforcing** and strengthening the grain boundary structure relates to spark-plasma sintering of cermet compositions using a nickel melt with added molybdenum powder [10–12]. As a result, diffusion through the formed low-melting eutectic is accelerated and evolves evenly to produce solid solutions of cermet and metallic phases of various compositions and stoichiometries [10–12]. The percent ratio of Ni and Mo in the sintered cermet compositions is important and determines the type, mechanism, rate of sintering, and the extent of side processes, in particular recrystallization and coalescence of particles of solid solutions of cermet and metallic phases [10–12]. Depending on the sintering specifics, embrittlement (crumbling) **with forming** microcracks occurs, or the material grain boundary structure is **reinforced** and strengthens [10–12].

Experimental part — p. 78

Figure 6 shows the phase composition of samples spark-plasma sintered in the range 1200–1600°C. Samples with Ni, Ta and Ni, Zr mixtures were characterized by extensive mullitization and **higher** crystallization of (Nb,Ta)C, B

than the less developed crystalline (Ta,Nb)B,C in the range 1400–1600°C. Mullite appeared because of active dissolution of Al_2O_3 and SiO_2 in the nickel melt with the formation and structuring of stoichiometric mullite via this melt. Higher crystallization of (Nb,Ta)C, B is explained by the **more intensive** dissolution of the less dense (Nb,Ta)C, B in the nickel melt in the range 1400–1600°C. Extensive crystalline Ni(Nb,Ta) is observed in the samples, the development of which is similar to **the intensity** of crystalline (Nb,Ta)C, B. Crystalline Ni(Nb,Ta) **differs** from the composition of (Nb,Ta)C, B in the range 1400–1600°C (Fig. 6a–d). In particular, **Ni(Nb,Ta)** contains only metallic components.

Experimental part — pp. 84, 86

The sample with the $c\text{-ZrO}_2$, Zr mixture has lower crack resistance. Extensive propagation of microcracks **by** a long linear **path** and formation of **small** damages and microcracks (Fig. 13d₁), related to uneven and incomplete interaction of highly disperse NiZr and NiZr_2 particles with local stressed regions along the (Nb,Ta)C, B boundary layer and its embrittlement, were observed in it. The combination of dislocations, in contrast to **layered dislocations**, was associated with **higher reinforcing** and strengthening of the structure, i.e., more developed plastic properties, than **in** layered dislocations. As a result, the stresses are more evenly dissipated in combined dislocations than in layered dislocations where this process **is** nonuniform and limited. Thus, the plastic properties and hardness evolved **mostly** at the boundaries of NiZr, NiZr_2 particles than at the boundaries of NiZr, NiZr_2 , (Nb, Ta)C, B particles. **Narrower** and short microcracks advanced tortuously through combined dislocations. Broader microcracks propagated linearly and rapidly through layered dislocations (Fig. 13d₁₋₁). The crack resistance of the sample with the $c\text{-ZrO}_2$, Zr mixture at 1500°C was slighter greater than that of the sample at 1300°C and had **more** damages and many microcracks (Fig. 13d, d₁).