the dependence of $w(l)$ on the particle radius is more intricate and largely determined by the relation between diffusion and kinetic components of the process.

NOTATION

r, z, θ, spherical coordinates; R_0, radius of a spherical particle; n, distance from the surface of the sphere along the normal; w_r, radial component of the velocity of the liquid; w_θ, tangential component of the velocity of the liquid; w_0, velocity of the particle relative to the liquid; P, pressure; ρ, density of the liquid; ν, kinematic viscosity; D, diffusion coefficient; α, thermal diffusivity; T, temperature; C, concentration; C^*, equilibrium concentration; T_0, initial temperature; C_0, initial concentration; k_0, n_0, kinetic constants in the growth rate; A, B, constants in the temperature dependence of the equilibrium concentration; $w(l)$, linear growth rate; ψ, flow function; ω, vorticity; g, heat of crystallization; L, scale of turbulent fluctuations; ζ, drag coefficient for a particle in the liquid; $k(T)$, m, and n_1, constants in the equation for the growth rate; ρ_s, density of the solid phase; ρ_g, gas density; φ, solid content; ε_k, specific kinetic energy of the three-phase system; $N_{Re} = R_0 W_0/\nu$, Reynolds number; $N_{Sc} = \nu/D$, Schmidt number; and $N_{Pr} = \nu/\alpha$, Prandtl number.

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MECHANISM OF MASS TRANSFER AFFECTING LIQUID METALS

IN SINTERED COMPOSITE MATERIALS

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It is demonstrated that mass transfer of the liquid phase in sintered composite materials can occur as a result of restructurization of the refractory skeleton.

According to established concepts, mass transfer during interaction of sintered composites with liquid metals occurs by the diffusion mechanism and in porous bodies due to action of capillary forces [1, 2].

In this report attention will be drawn to the existence of still another mode of mass transfer in sintered composites, namely mass transfer caused by restructurization of the refractory skeleton.

In the experimental study conditions eliminating the possibility of mass transfer by diffusion and capillary forces were created. Compact spherical specimens, 20 mm in diameter,
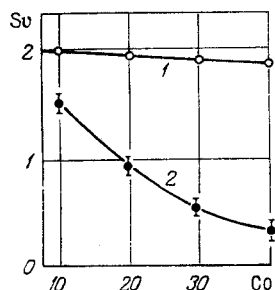
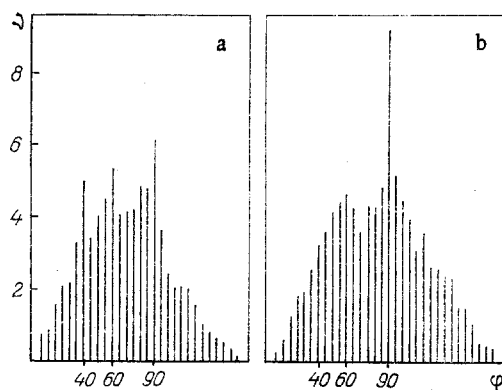Fig. 1   Fig. 2

Fig. 1. Specific surface area (1) and contact area (2) of WC particles ($Sv \cdot 10^{-4}$ cm^2/cm^3) as functions of the cobalt concentration (vol.%) in specimens.

Fig. 2. Frequency distribution curves for random sections of dihedral angles (φ, deg), with the cobalt concentration: a) 10 vol.%; b) 40 vol.%.

of a sintered hard alloy containing 5.9 wt.% Co and 94.1 wt.% WC were immersed at 1370°C in a cobalt melt saturated with tungsten and carbon. In order to ensure the identical chemical composition of the liquid phase in the specimens and in the surrounding cobalt melt, an excess amount of tungsten carbide (38 wt.%) was dispersed in the latter in the form of 1-2 μm large crystals.

Constant temperature, pressure, and composition of the liquid phase within the volume of specimens as well as within the surrounding cobalt melt ensured equal chemical potentials of the components and eliminated diffusional mass transfer, while the absence of capillary pores containing a gaseous medium eliminated the action of capillary forces.

The next series of tests was performed with cylindrical hard-alloy specimens, 20 mm high and 35 mm in diameter, at the base of which a truncated spherical cavity 4 cm^3 in volume had been cut out. In a vacuum furnace at 1370°C into such a cavity were successively poured 3.5 cm^3 doses of cobalt melt. Each dose, after being held for 3-5 min, became completely absorbed by the specimens. A specimen with a volume of 15.63 cm^3 thus absorbed altogether 14 cm^3 of liquid cobalt. Following the absorption of each cobalt dose, the volume of the specimens was found to have increased correspondingly. These tests demonstrate that not selective mass transfer of the components (Co, W, C) from the melt to the specimens but rather absorption of the entire solution containing 64 wt.% Co + 33.8 wt.% W + 2.2 wt.% C has taken place, with the volume of the specimens becoming 1.89 times larger. The volume of the spherical specimens has increased correspondingly.

The change in the specific surface area of WC particles and of the WC—WC contact area were determined in an examination of the specimens by methods of stereometric metallography [3], also the dihedral angles between carbide particles were measured [4], while the concentration of binding metal in the volume of specimens was determined by x-ray spectral analysis with a Mikroskan-5 instrument.

These studies revealed that liquid cobalt melt had intensively impregnated spherical as well as cylindrical specimens, with the concentration of binding metal in these specimens becoming 4 times higher (increasing from 10 to 40 vol.%), with the specific surface area of carbide particles remaining almost the same, and with the contact area decreasing as the cobalt concentration in the specimens increased (Fig. 1). An analysis of the frequency distribution curves characterizing random sections of dihedral angles in specimens (Fig. 2) revealed that particles of tungsten carbide in the specimens in the initial state had formed dihedral angles of 40, 60, and 90° in approximately equal numbers. As a result of impregnation with liquid cobalt melt, the fraction of 40 and 60° dihedral angles decreased appreciably while the fraction of 90° dihedral angles increased appreciably.

In tests involving interaction of spherical specimens with cobalt melt the concentration of binding metal also became 4 times higher. The equality of cobalt, tungsten, and carbon

concentrations in the liquid phase of specimens and in the melt, respectively, excluded the possibility of diffusional mass transfer of these components, while the absence of a gaseous phase in the pores of specimens and at their surface eliminated the action of capillary forces. Thus, both these mechanisms of mass transfer could not noticeably contribute to the change of cobalt concentration in the specimens.

The 1.89 times increase of the volume of specimens, the decrease of contact area between refractory particles, and the attrition of 40 and 60° dihedral angles indicate a restructurization of the carbide skeleton in these specimens and a formation of a new stereometric configuration in which 90° dihedral angles are preferred. This skeleton restructurization is accompanied by a decrease in the free energy of the system [5]. Crystals of tungsten carbide have the shape of elongated prisms [6]. Calculations indicate that the new orientation of such crystals with preferred formation of 90° dihedral angles is possible only during enlargement of the volume of specimens, which has also been confirmed in these tests. The presence of an extra volume of cobalt melt provided this possibility and, meanwhile, the restructurization of the carbide skeleton facilitated an influx of liquid melt into the volume of specimens, new doses of liquid cobalt favorably contributing to the subsequent reorientation of particles.

The aggregate of experimental data demonstrate that mass transfer of liquid cobalt melt into sintered specimens occurs as a result of processes associated with restructurization of the refractory skeleton, i.e., with recrystallization of tungsten carbide particles. Studies have revealed that the recrystallization mechanism of mass transfer can be the predominant one in several sintered composites such as TiC–Ni, WC–Ni, WC–TiC–Co, and others.

Mass transfer of cobalt melt into our specimens occurred as a result of flow of the liquid phase through channels formed by refractory particles. The velocity of the cobalt melt was 10^{-5} m/sec, the size of the channels was 10^{-6} m, the density of the melt was 10^{4} kg/m^3, and its dynamic viscosity was 10^{-3} kg/m·sec. These parameters yield a Reynolds number $N_{Re} = 10^{-4}$. Such a flow, therefore, is laminar and subject to Darcy's law [7].

Recrystallization of refractory particles generates an internal pressure within the volume of specimens, called the migration pressure [8]. On the basis of all these data, a differential equation has been constructed which describes mass transfer of the liquid phase due to recrystallization of the refractory skeleton [9]. This equation of migration can be used for describing mass transfer in the case of equal chemical potentials of the components within the volume of a sintered product piece and within the surrounding melt, respectively.

When the chemical potentials of the components are not the same within the shrinking melt and within the volume of the product piece, as during interaction of liquid copper with specimens of WC–Co alloy, for instance, then mass transfer of the components and of the liquid phase on the whole occurs by the diffusion mechanism and the recrystallization mechanism simultaneously. The mass transfer under these conditions is described by a set of differential equations describing migration and diffusion.

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TURBULENT BOUNDARY LAYER OF POLYMER SOLUTIONS WITH

A FLOW-RETARDING PRESSURE GRADIENT

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A description of the mean velocity distribution in the near-wall turbulent boundary layer of polymer solutions with a flow-retarding pressure gradient is given by using dimensional analysis and similarity.

Certain Peculiarities of Turbulent Boundary Layer Similarity

As is known [1] mean velocity distribution and drag law for turbulent flows is described with satisfactory accuracy by using dimensional analysis and similarity without detailing the turbulent model. The main reason for such simplicity is associated with the separation of the shear turbulent flow in a domain with substantially different scales, which are autonomous to a considerable degree and allows local description.

There are two characteristic length scales in the boundary layer on a smooth plate around which a viscous fluid flows with the constant velocity U: the boundary layer thickness δ and the "viscous sublayer thickness" scale $\delta_\nu \equiv \nu/u_*$ between which there exists the inequality $\delta \gg \delta_\nu$ for a high degree of development of the turbulence. Using the lowest approximation for juncture of the asymptotic expansions in the small parameter δ_ν/δ, a logarithmic formula for the mean velocity is derived successfully [1]. The derivation relies on the assumption of locality. It is considered that for a given x (x is the coordinate along the surface in the flow direction) the mean characteristics depend on $\delta(x)$ and $u_*(x)$ and are independent of their derivatives.

In the case of a boundary layer with a significant pressure gradient $\alpha \equiv dp/dx$, a third length scale $\delta_\alpha \equiv u_*^2/\alpha$ becomes important (we neglect compressibility of the fluid and consider the density equal to one). It characterizes the length within which the friction stress on the streamlined surface $u_*^2(x)$ varies substantially. Then the assumption of locality also implies slowness of the change in the velocity U(x). Taking into account the steady flow equation $\alpha(x) = -U(x)dU(x)/dx$, the constraint on the derivative dU/dx can be given the form $\alpha\delta \ll U^2$.

Another important hypothesis is the assumption of autonomy of certain subdomains of the boundary layer. It results in the possibility of an independent analysis of the mean characteristics in these subdomains.

The autonomy condition actually reduces to similarity in the Ryenolds number $U\delta/\nu \gg 1$ for the outer flow zone; viscosity plays no part here. The assumption of autonomy in the inner subdomain (close to the surface being streamlined) denotes independence of the distribution from the external parameters U, δ, α. This can be expected when the inner and outer scales are not commensurate, at least the length $(\delta_\nu \ll \delta)$, the time $(\nu/u_*^2 \ll \delta/U)$, and the smallness of the change in the stress change u_*^2 in the "viscous" length $(\alpha\nu/u_* \ll u_*^2)$. This latter assumption can be given the form of the condition $\delta_\nu \ll \delta_\alpha$.

For a sufficiently strong pressure gradient, its influence will penetrate deeply, and $\delta_\alpha \ll \delta$, as will also be taken into account later. If the time scales associated with the effect of viscosity, the pressure gradient, and the external flow $\nu/u_*^2 \ll u_*/\alpha \ll \delta/U$ are also incommensurate, then the existence of an autonomous intermediate pressure subdomain can also be expected. Together with those considered earlier, these conditions result in the requirement $u_*U \ll \alpha\delta \ll U^2$, which imposes strong constraints on the possibility of an autono-

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