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Study of the electrical discharge coating degradation

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The peculiarities of formation and destruction of coatings formed by electrical discharge machining (EDM) have been investigated. The kinetic theory of solid-state strength has been used to explain the experimental data. The important role of the thermal fluctuational process in the kinetics of coating formation by EDM has been shown.

1. Introduction

The process of coating metal surfaces by electrical discharge machining (EDM) utilizes the effect of highly-concentrated energetic flows in solids [1], the energy of discharge, $W \sim 0$, $1 \rightarrow 10$ J being available for a time, $t \sim 10^{-5} \rightarrow 10^{-4}$ sec at a frequency of 100 Hz. An area of several mm^2 of the sample can be coated during one impulse. Scanning techniques are used to coat larger surfaces.

It has been shown that the thickness of the coating formed by EDM is limited to some 200 μm, due to cracking [2, 3].

In the attempt to increase practical coating thickness [4], evidence has shown the importance of residual tensile stresses (RTS) in the coating. The purpose of this paper is to study how these stresses and the weight gain varies with machining time and to account for the limited coating thickness observed.

2. Experimental procedure

The investigations were carried out with an energy level, $W = 0.314$ J. A low carbon steel material was used as a substrate and coated with chromium, titanium, iron, nickel and tungsten. These materials possess different physical, mechanical and thermophysical properties. The residual tensile stress σ, and the weight gain Δm have been determined according to standard techniques, the error being not more than 5%.

3. Results and discussion

In Fig. 1 the typical curves illustrating the effect of time on σ and Δm are shown. Both the $\sigma = f(t)$ and the $\Delta m = g(t)$ plots have similar forms but the maxima are displaced along the x-axis.

These curves show that the process of electrical discharge coating can be divided into several time intervals:

(1) Ot_1 – linear increase of stress connected with linear change of weight gain;

(2) t_1t_2 – decrease of growth rate of the residual tensile stress level;

(3) t_2t_4 – drop in σ, decrease of the growth rate of the thickness;

(4) t_4t_5 – intensive cracking in coating, followed by the decrease of weight gain.

In the region Ot_2 the value of σ increases linearly until t_1; then slows down before the maximum at t_2. During these time intervals all the materials

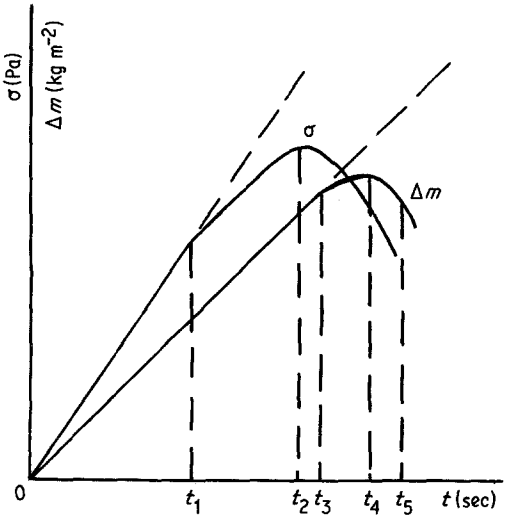

Figure 1 Variation of σ and Δm with machining time.

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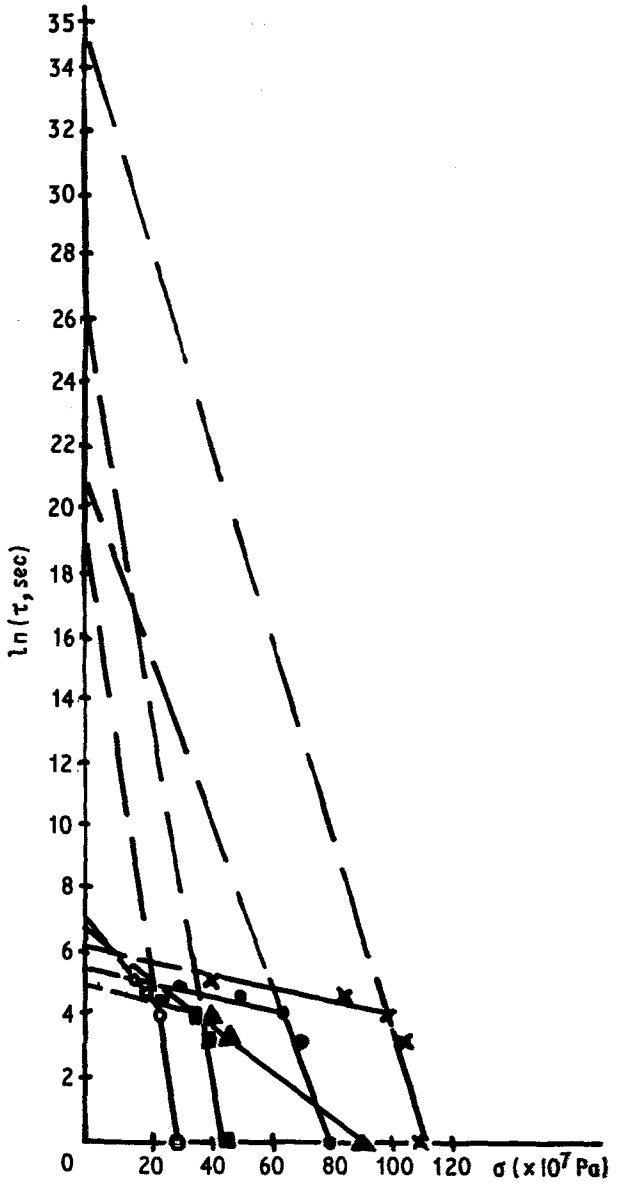

Figure 2 The dependence of durability τ on residual tensile stress value: × – Cr, ● – W, ▲ – Ti, ■ – Ni, ○ – Fe.

(with the exception of chromium) showed linear time dependence of the weight gain.

We may now analyse the process of coating fracture using the kinetic theory of solid-state strength [5]. The time interval $\tau = t_2 - t$ is taken as the durability of the samples (in this expression t is the time corresponding to the given level σ of the residual tensile stress and t_2 is the time corresponding to the maximum of value of the stresses). Obviously, the durability, τ, of the specimen becomes equal to zero as σ tends to σ_{max}.

Transforming the relationship $\sigma = f(t)$ for chromium, tungsten, iron, nickel and titanium into the traditional coordinates of the kinetic theory, one obtains a set of lines, consisting of two straight segments, for nickel, chromium and tungsten as illustrated in Fig. 2. One can use Equation 1 quoted by Regel *et al.* [5]

$$\tau = \tau_0 \exp \frac{U_0 - \gamma\sigma}{kT} \qquad (1)$$

for the analysis of the kinetic fracture layers affected by residual tensile stresses as the relationship $\ln \tau = f(\sigma)$ is linear. In Equation 1 τ is layer durability, $\tau_0 = 10^{-13}$ sec, U_0 is the initial activation energy for fracture, equal to the heat of sublimation, γ is a structure-sensitive coefficient, σ is

TABLE I

Coating material	T_{beg} (K)	T_{fin} (K)	γ_{beg} (10^{-3} J mol^{-1} Pa^{-1})	γ_{fin} (10^{-3} J mol^{-1} Pa^{-1})	γ_{beg} (10^{-22} cm^3)	γ_{fin} (10^{-22} cm^3)
Fe	1340	1030	0.145	0.595	2.4	9.8
Cr	1300	730	0.024	0.195	0.4	3.2
Ni	1460	970	0.054	0.5	0.88	8.2
W	2850	2000	0.051	0.425	0.84	7.0
Ti	1550	1550	0.102	0.102	1.68	1.68

the residual tensile stress in the layer, k is Boltzmann's constant, and T is the test temperature.

Using the experimental results we can determine the intercepts $\ln A$, of the extrapolated straight lines on the y-axis and the slope α of the lines. Knowing $\ln A$ and α one can estimate the values of T and γ in Equation 1 from Equations 2 and 3

$$T = U_0/k \times \ln(A\tau_0) \quad (2)$$

$$\gamma = \alpha kT, \quad (3)$$

where T is the effective temperature of the metal surface at which layer destruction would occur under static conditions. The results of these analyses are given in Table I.

The two stage nature of the curves $\ln \tau = f(\sigma)$ (Fig. 2) indicates that a change of material composition and structure is occurring during the tests [5]. It should be noted that at equal regimes of coating formation the effective temperature T determined from the linear sections of the broken-line $\ln \tau = f(\sigma)$ (Fig. 2) depend on the type of coating. The temperature increases together with the increase of sublimation specific heat. One can see that in the given process, heat removal is carried out at the expense of evaporation but not of heat conductivity. This fact can, obviously, be explained by the extremely high density of electrical discharge energy transferring to the surface [6] (the same occurs in the laser effect on solid). Apparently, the high density of crystal lattice defects appearing during spark discharge prevents the effective heat removal from the metal surface and promotes its accumulation within the surface coating layers. As the possibility of evaporation increases it will cause a decrease in temperature at long processing times (see Table I). (A considerable part of the material is evaporated and this causes a decrease in the weight gain.)

Variation of the structure-sensitive coefficient γ for tungsten, chromium and nickel is also connected with the variation of phase composition. The increase of γ ($\gamma_{fin} > \gamma_{beg}$, see Table I)

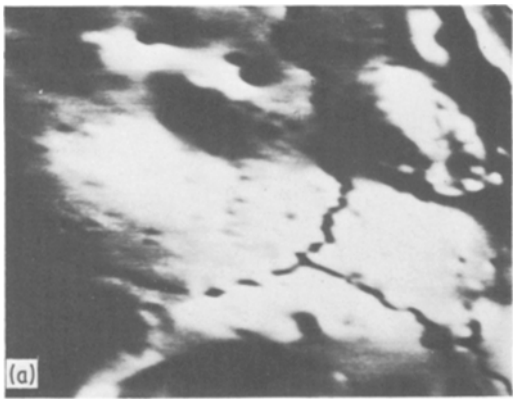

Figure 3 The micrographs showing the metal coating on low-carbon steel substrate: (a) Fe, (b) Cr, (c) W (× 2300).

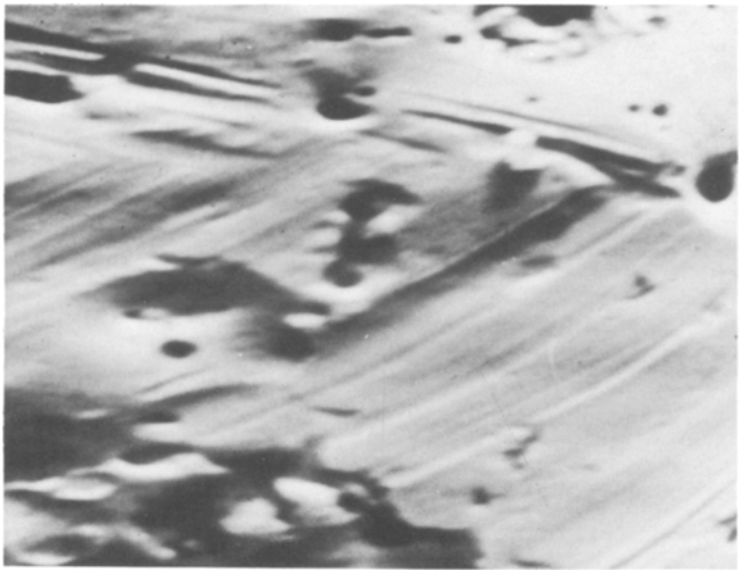

Figure 4 The micrograph showing Ni-coating on steel (X 3200).

indicates increasing heterogeneity and imperfection of coating during the growth process. Scanning electron microscopy of the coatings revealed different features of stress relaxation and fracture process depending on the material and time of coating. For chromium and tungsten coatings cracking appeared mainly at grain boundaries, the crack density increasing with machining time. For iron and nickel coatings, evaporation tracks are seen to form grooves on the block boundaries, the cracks in these materials are corrugated (Figs. 3a, b and c), and both coatings exhibit slip bands (Fig. 4). The presence of these slip bands indicates that at long machining time, plastic deformation prevents the development of cracks. The maximum level of RTS in iron and nickel coatings is two times lower than in other materials which appears to confirm the effect of plastic deformation. Probably, as evaporation and plastic deformation are less effective mechanisms of stress relaxation than simple cracking it accounts for the abrupt drop in the residual stresses value in chromium and tungsten coatings and the smooth drop in iron and nickel.

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