Fractographs of the specimens subjected to intergranular cracking showed coarse slip steps. It appeared that these steps originated from corrosion of chemically reactive micro-cracks produced by dislcoations with a large Burgers vector near grain boundaries.

A mechanism of intergranular SCC was proposed by considering the results obtained. This supports basically the theory set forth by Robertson and Tetelman.

Electrochemical and XPS Studies on Corrosion Behavior of Amorphous Ni-Cr-P-B alloys

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Boshoku Gijutsu (Corros. Eng.), **26** (1977), 445.

Amorphous Ni-Cr-P-B alloys, which possess a high corrosion resistance and are apt to form the amorphous structure, have been prepared by rapid quenching from melts. The alloys containing chromium of 7 and 9 at. % are not corroded in 10% $FeCl_3 \cdot 6H_2O$ at 30°C. Spontaneous passivation occurs on these alloys in 2N H_2SO_4 and 1N HCl, and a critical potential for pitting is not observed on the anodic polarization curves even in 1N HCl. Simultaneous determination of the thickness and composition of a passive film and the composition of the surface of substrate alloy directly under the passive film has been carried out by X-ray photoelectron spectroscopy. The passive film formed in 2N H_2SO_4 consists mainly of hydrated chromium oxy-hydroxide similar to those formed on corrosion-resistant crystalline alloys containing chromium and the composition of the surface of substrate alloy directly under the film is not different from that of bulk alloy. On the contrary, the composition of passive film formed in 1N HCl can be expressed as $(Ni^{II}_{0.26} Cr^{III}_{0.74})$ $(PO_4)_{0.74}O_x(OH)_{0.52-2x}(1.74+x)H_2O$ and hence phosphate comprises 81% of constitutents of the passive film. In addition, the composition of the surface of substrate alloy directly under the film changes from that of bulk alloy. In conclusion, the amorphous Ni-Cr-P-B alloys with chromium of 7 to 9 at. % have a high corrosion resistance in strongly acidic solutions and the great resistance to pitting corrosion, but may not be always stable for a long time because of the change in alloy composition directly under the film.

Kinetics of Evaporation of Manganese, Copper and Sulfur from Iron Alloys in Vacuum Induction Melting

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The evaporation-rate constant K_i (kg·m^{-2}·s^{-1}) for each of manganese, copper and sulfur in the vacuum induction melting of iron alloys has been determined. The rate constant K_v or K_v^s (kg·m^{-2}·s^{-1}) for the vaporization of these elements at the vacuum-melt interface is calculated. In each of the Fe-C-Mn, Fe-C-Cu and Fe-C-Si-S systems, the value of K_i at a lower stirring rate is about one half of that at a higher stirring rate; hence mass transfer in the melt must be involved in the

rate-determining step for their evaporation at the lower stirring rate. In the Fe-Cu and Fe-Mn systems, the value of K_{Cu} at 1973 K is obviously greater than that at 1873 K, but those of K_{Mn} at both temperatures are close to each other. The rates of evaporation of manganese must be determined predominantly by the rate of its transfer in the melt in view of the values of ρK_m at both temperatures, which were calculated on the basis of the equation $K_i = \rho K_m K_v / (\rho K_m + K_v)$, where ρ (kg·m^{-3}) is the density of the melt and K_m (m·s^{-1}) is the mass-transfer coefficient in the melt. In the Fe-Mn, Fe-Cu and Fe-Sn systems at 1873 K, most of the reported experimental values of K_{Mn}, K_{Cu} and K_{Sn} agree well with their respective values calculated from the equation by taking $\rho K_m = 6.8 \times 10^{-1}$ (kg·m^{-2}·s^{-1}), but the present experimental values of K_{Cu} are smaller than the calculated value. The plots of K_S / f_S vs. a_{Si} in the Fe-Si-S system are linear at two different stirring rates. The rate constants k_1 and k_2 given by this linear relation at a lower stirring rate are smaller than those at a higher stirring rate, respectively, and must involve K_m. In the desulfurization of Fe-S, Fe-C-S, Fe-Si-S and Fe-C-Si-S alloys at temperatures in the range 1688 ~ 1873 K, the relation between K_S and K_v^s is represented by a curve calculated from the equation by taking $K_v = 0.24$ K_v^s and $\rho K_m = 2.5$ regardless of the temperatures and the alloy systems. This value of ρK_m is much greater than the value of ρK_m for Fe-Mn, Fe-Cu and Fe-Sn systems. Oxygen and sulfur hardly affect the rate of evaporation of copper from molten iron at 1873 K.

Microstructures of Deformation and Fracture of Cementite in Pearlitic Carbon Steels Strained at Various Temperatures

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In order to make clear the effect of temperature on deformation and fracture of cementite in steels, observations by transmission electron microscopy were made on cementite in carbon steels strained in tension at various temperatures ranging from -196 to 700°C. Hardly any plastic deformation of cementite was detected at -196 and -78°C. It was confirmed that above room temperature cementite in steel can deform due to dislocation slip and that the deformation becomes easier as the temperature increases. Slip in cementite at room temperature and at 300°C seems to be confined to only (100) or (001). Dislocations observed at room temperature and at 300°C were mostly isolated and straight. Above 400°C (100), (010), (001), and some {110} planes are all operative slip planes. Dislocation loops, dipoles, cusps, and networks were frequently found. These observations indicate that double slipping and the interaction of dislocations can occur and that the deformability of cementite above 400°C is very large. An appreciable degree of dynamic recovery was detected above 500°C. The fracture of cementite at -196 and -78°C occurred in a cleavage manner along some crystallographic planes such as (110), (100) or (210). Above room temperature fracture occurred along an activated slip plane and was preceded by some amount of slip on that plane.