



Fig. 4. Kinetics of static oxidation at 1000 °C for a Ni–19.4%Cr foam in air and for bulk Ni–20%Cr in pure oxygen [25], with weight gain normalized by (a) sample weight and (b) sample surface area.

much higher surface area, the Ni–19.4%Cr foam exhibits much faster oxidation kinetics than Ni–20%Cr solid when normalized by weight (Fig. 4(a)). However, no difference is observed, within experimental error, in the oxidation kinetics normalized by surface areas, as shown in Fig. 4(b), indicating that the oxidation mechanism for the foam in air is the same as that for the bulk alloy in pure oxygen.

The high-temperature oxidative resistance of metals depends on the formation of a surface layer (scale) that acts as a barrier to oxygen transport. Shown in Fig. 5 is a non-porous, protective scale of oxide layer, which formed on the surface of a Ni–19.4%Cr foam exposed in air for 55 h. The possible oxide species that can develop on Ni–Cr alloys include NiO, NiCr₂O₄, and Cr₂O₃, the prevalence of each varying with oxidation temperature and treatment [24]. However, as temperature increases above ~800 °C, the formation of chromium oxide (Cr₂O₃) becomes dominant [22,26]. The morphology and dimension of the scale formed on the surface of a Ni–19.4%Cr foam (5–10 μm) at 1000 °C for 55 h is similar to that reported for a Ni–20%Cr bulk alloy (~8 μm)

Table 1
Microhardness for struts of heat-treated foams

	Microhardness (HV)
Pure Ni	74.4 ± 3*
Ni–12.3Cr	116 ± 9
Ni–19.4Cr	120 ± 13
Ni–27.2Cr	166 ± 14

* As-received foams.

at 900 °C for 66 h [22]. It can thus be safely assumed that the dense oxide layer in Fig. 5 is predominantly Cr₂O₃. Its thickness is about 10% that of the average wall thickness.

3.3. Mechanical properties at ambient temperature

The results of the micro-hardness tests conducted at room temperature are summarized in Table 1. All indentations were made at least 20 μm away from the nearest cell strut edge. As expected from the solid-solution strengthening effect of chromium in nickel, the strut hardness increases with increasing chromium content.

Compressive stress–strain curves at ambient temperatures for Ni foams with 0–27.2%Cr are shown in Fig. 6(a), where the foam stress is the applied load divided by the specimen cross-sectional area. The pure Ni foam was annealed at 1000 °C for ~30 min before testing to produce a microstructure comparable to that of the alloyed Ni–Cr foams. All foams display the behavior typical of a ductile metallic foam [27]: linear elasticity at low stresses, followed by a long plastic collapse region where the stress rises slowly, and finally a densification regime where the cell walls start to contact and the stress rises steeply. It is apparent from Fig. 6(a) that increasing the chromium content of the foams leads to an increase in the plastic collapse stress, or foam yield-stress, corresponding to the end of the elastic region and start of the plastic collapse region, and determined graphically from the experimental curves as shown in Fig. 6(a).

3.4. Creep properties

The Ni–21%Cr and Ni–29.6%Cr foams displayed a primary creep phase with decreasing strain rate, followed by a secondary creep regime with a minimum strain rate constant over an extended period of time. Fig. 7(a) shows the minimum strain rate $\dot{\epsilon}$ plotted against the applied stress σ for these foams tested at 825 °C. The creep data follow a power-law:

$$\dot{\epsilon} = K\sigma^n \exp\left(\frac{-Q}{RT}\right) \quad (1)$$

where K is the Dorn constant, n the stress exponent, Q the creep activation energy, R the gas constant, and T the temperature. The best-fit stress exponents are 3.7 for the Ni–21.0%Cr foam and 3.3 for the Ni–29.6%Cr foams, which are lower than the value reported for Ni–20%Cr bulk alloy ($n = 4.6$) in the temperature range of 680–1160 °C [28]. The narrow range of stresses (0.1–0.2 MPa) measured is due to the following

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