CREEP FRACTURE BEHAVIOR OF AUSTENITIC STAINLESS STEELS FROM 550 to 800°C

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ABSTRACT

The constant load creep fracture behavior of class 18/8 austenitic stainless steels has been investigated in two batches of type 316 and one batch of type 304 steel in the interval of temperatures from 550 to 800°C and times to rupture up to 1500 h. It was found, in general, that intergranular fracture occurs below 650°C associated with voids and cracks at the triple points possibly due to grain boundary sliding . This is also the case of type 304 steel at higher temperatures. In type 316 steel transgranular fracture prevails above 700°C in association with an apparent grain boundary strengthening due to carbide precipitation.

Second phase formation, at longer exposure times and higher temperatures, changes to intergranular the fracture mode of one of the 316 steels. Even though an intense surface oxidation was observed at higher temperatures, this does not seems to affect the general fracture behavior.

The mechanisms and effects which are probably responsible for these fracture characteristics are discussed.

KEY WORDS

Creep fracture modes; stainless steel fracture; high temperature fracture; precipitates and voids influence on ductility; creep ductility of stainless steel.

INTRODUCTION

Several research works have been dedicated to limited aspects of the creep fracture behavior of austenitic stainless steels. Garofalo (1968) observed intergranular fracture associated with voids and grain boundary sliding in type 316 steel. This characteristics were also found by White and Le May (1974) which presented evidences for second phases interface with the austenitic matrix as preferred sites for void nucleation. These authors (1974) have also shown that precipitation of carbides would make difficult for grain boundary sliding to occur and therefore to favor transgranular fracture. Surface oxidation of this steel at low stress could aid to the rupture through the nucleation and propagation of cracks.

Bhargava, Moteff and Swindeman (1976) proposed that the presence of carbide would be associated with a loss of ductility. Morris (1978) concluded that over a practical range of applied stresses and grain sizes, the creep rupture of type 316 steel takes place by tearing between grain boundary triple point; cracks formed as a result of grain boundary sliding. Sikka (1978) investigated the influence of aging prior creep deformation, and nitrogen content on the creep ductility of types 304 and 316 steels. Lai and Wickens (1979) suggested that the precipitation process in type 316 steel creep tested between 550 and 675°C will continue throughout the time the steel is in service. Moreover they (1979) found that the strength of the sigma/austenite interface is an important factor in determining the long term creep rupture. In this case failure occurs either by void formation and coalescence or by transgranular cracking.

EXPERIMENTAL PROCEDURES

The steels were received in the form of extruded bars and their chemical analysis as well as grain size is given in Table 1. They were annealed for one hour at 1100°C, resulting in a equiaxed structure with round inclusions showing some tendency to align themselves parallel to the longitudinal axis of the bar. Cylindrical specimens having a gage diameter of 6 mm and a gage length of 40 mm were creep tested from 550 to 800°C at constant loads and times to rupture not exceeding 1500 h.

TABLE 1 Chemical Analysis (Weight Percent) and Average Grain Size (μm) of Steels Used

AISI TYPE	C	Mn	Si	P	S	Cr	Ni	Mo	Cu	Average Grain Size μm
316 (A)	0.07	1.72	0.50	0.026	0.014	18.3	12.5	2.55	0.13	140
316 (B)	0.05	0.73	0.40	0.035	0.018	17.3	12.2	2.07	0.16	80
304	0.025	0.75	0.86	0.04	0.01	19.5	9.35	0.24	0.15	300

RESULTS

At relatively lower creep loads and below 650°C the fracture characteristics of all three types of steel studied presented common aspects:

a) Absense of necking and a relatively low reduction of area which increase with the applied load;
b) Wedge-type voids at triple points as illustrated in Fig. 1;
c) Fracture surface approximately normal to the load axis. The surface showed, predominantely, faced areas typical of intergranular fracture, Fig. 2.

With increasing applied load and decreasing the time to rupture, the surface becomes more and more ductile in appearance as shown in Fig. 3.

Above 650°C the type 304 steel presented essentially the same features observed at lower temperatures. In this steel at even higher temperatures, such as 750 and 800°C there occurs formation of second phases along the grain boundary and voids between the grains as shown in Fig. 4.

A surface oxidation due to the air exposition during the creep test occurs even at 550°C but becomes progressively intense above 600°C with intergranular penetration in the material, as presented in Fig. 5.

Both types 316-A and 316-B steels above 650°C displayed a marked change in the fracture aspects as compared to the lower temperature behavior.

a) Fracture is preceded by necking;
b) An intense precipitation of carbides and second phases occurs at the grain boundaries as depicted in Fig. 6;

600°C 200 MN/m² 19h   600°C 350 MN/m² 7 h   600°C 280 MN/m² 85 h

(a) (b) (c)

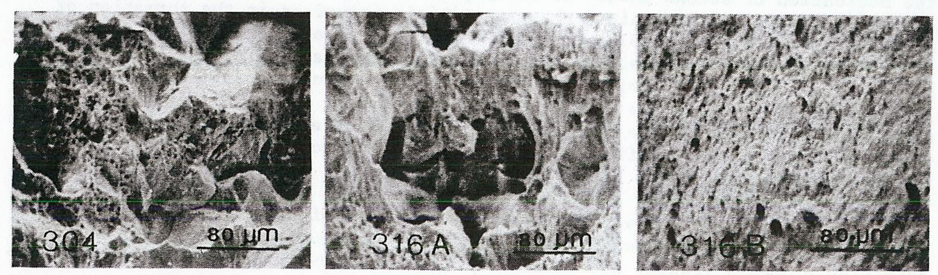
Fig. 1. Wedge-type voids at triple points

600°C 250 MN/m² 16 h   600°C 320 MN/m² 30 h   600°C 280 MN/m 85 h

(a) (b) (c)

Fig. 2. Electron fractography evidences of intergranular rupture

550°C 340 MN/m² 6 h   650°C 240 MN/m² 17 h   600°C 350 MN/m² 1,5h



(a) (b) (c)

Fig. 3. Ductile aspects with decreasing time to rupture

c) The absence of facets, the dimple covered surface (Fig. 7) and the elongated deformed grains (Fig. 6) are evidence of typical transgranular fracture.

An interesting exception occurred in type 316-B steel tested at 800°C for relatively low creep loads. Under this conditions the material showed intergranular fracture associated with intense surface oxidation (Fig.8 (a)) as well as other macro features of intergranular rupture observed below 650°C, such as absence of necking and fracture surface normal to the load axis. Moreover it was observed formation and coalescence of second phases, possibly sigma phase (Lai and Wickens, 1979), with extensive development of voids and microcracks at their interfaces in the grain boundaries as illustrated in Fig.8 (b).

DISCUSSION

The present results have displayed the change in creep fracture behavior which occurred in different types of class 18/8 austenitic stainless steels. Below 650°C all steels presented the same general intergranular features as shown in Fig. 1 and 2. This probably occurs when grain boundary sliding becomes a predominant mechanism (Morris, 1978) which is favored by a strengthening process inside the grain due to the tangled dislocation structure related to dynamic strain aging. The longer the time to rupture the more effective the grain boundary sliding and the dynamic strain aging, which depend, both, on diffusion processes. This explain the higher ductility and increasing amount of surface dimple areas for shorter creep rupture times. The grain boundary sliding promotes voids and cracks formation at the triple points which could then be connected by simple propagation or, at higher creep loads, by ductile coalescente of dimples. The relatively lower ductility of the 304 steel is possibly a consequence of its lower Mo content which makes it less resistant to thermal effects.

Above 650°C transgranular fracture prevails in the 316 steels, as presented in Fig. 6 and 7, when deformation occurs mainly inside the grains. In this case the carbides precipitation at the grain boundaries acts as a barrier for sliding and the formation of voids. The dislocation substructure inside the grain is characterized by cells and becomes progressivelly softer with temperature as it changes to subgrains (Kestenbach, Silveira and Monteiro, 1976). All these processes contribute to the highly ductile, dimple covered transgranular rupture. The lower carbon content, Table 1, of type 304 steel would make carbide precipitation at the grain boundary innefective in preventing intergranular fracture at high temperature, Fig. 3 and 4.

The nucleation of second phases at the expense of carbides (White and Le May, 1970) and the role of the sigma phase/austenitic interface in the formation of cracks (Lai and Wickens, 1979) are also important factors in the creep fracture of austenitic stainless steels. This appears to be the case of type 316-B steel at higher temperatures. In this steel, and surprisingly not in type 316-A, the second phase nucleation at the grain boundaries promotes, at 800°C (Fig. 8), a marked intergranular rupture. The reason for this differential behavior is not clear but might be explained by differences in microalloying composition or specifically in the Mn content.

In spite of the heavy oxidation observed (Fig. 5) at high temperature, this does not seem to affect the creep fracture behavior. The surface damage due to the oxidation probably results from local rupture of the oxide passive layer as a consequence of the plastic deformation.

It can be concluded from the above results that it is not possible to predict the creep fracture behavior of class 18/8 austenitic stainless steel using general approach such as fracture mechanism maps (Ashby, Gandhi and Taplin, 1979). The in

750°C 60 MN/m² 288 h   750°C 60 MN/m² 288 h

(a) (b)

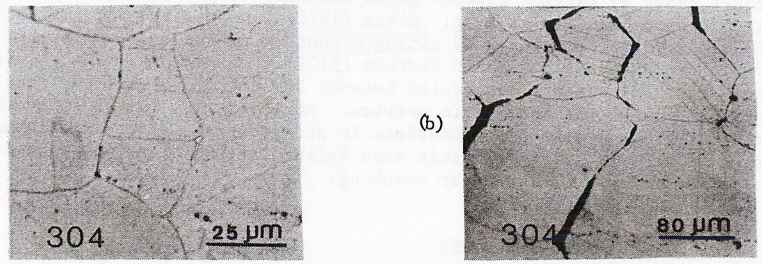


Fig. 4. Second phase (a) and voids (b) at the grain boundaries

750°C 60 MN/m² 288 h   750°C 60 MN/m² 288 h

(a) (b)

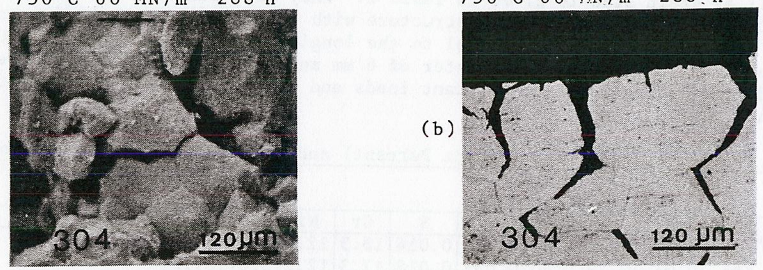


Fig. 5. Oxidation with intergranular penetration

700°C 170 MN/m² 30 h   800°C 120 MN/m² 1 h

(a) (b)

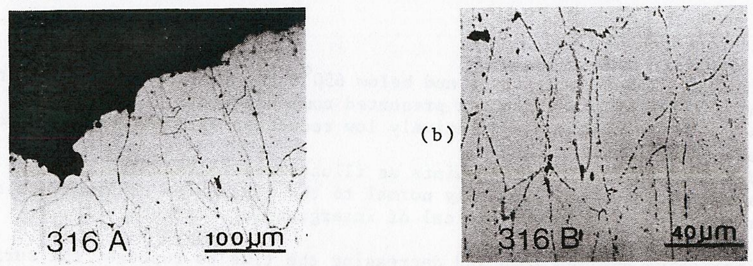


Fig. 6. Precipitation of carbides and second phases at g.b.

700°C 220 MN/m² 4 h   800°C 120 MN/m² 4 h

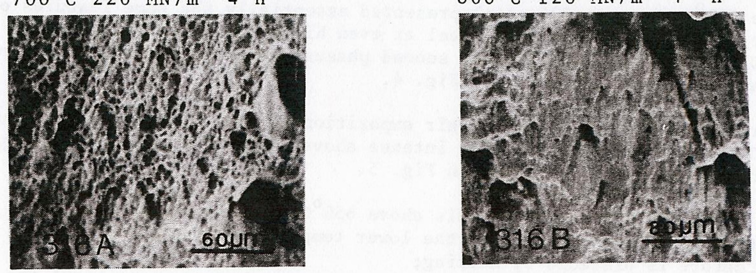
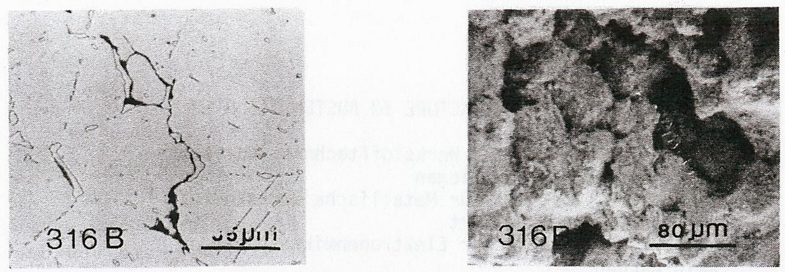


Fig. 7. Electron fractographies of ductile conditions

800°C 40 MN/m^2 1320 h   800°C 40 MN/m^2 1320 h



(a)   (b)

Fig. 8. Intergranular fracture at high temperature

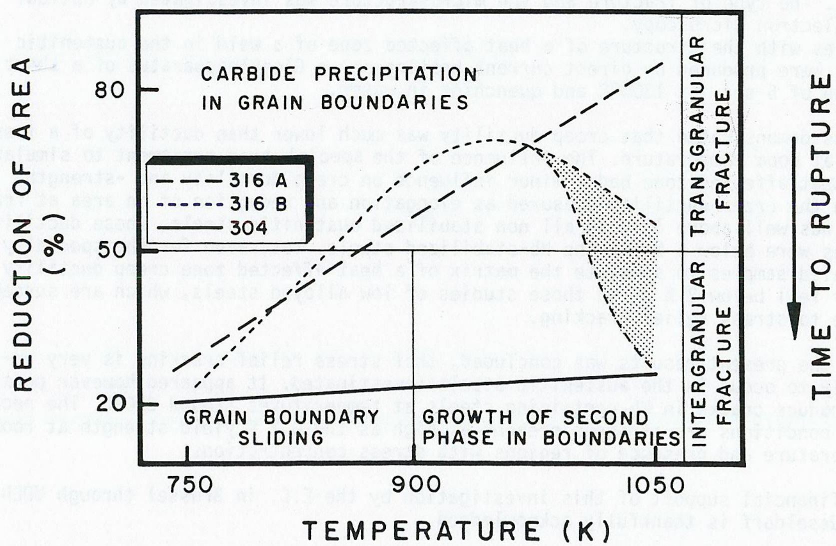


Fig. 9. Schematic integrated diagram of creep fracture behavior of austenitic stainless steels. The scales correspond to approximate values

fluence of voids from grain boundary sliding, solute atom effects, cracks at second phase austenitic interfaces, and dislocation substructures are important factors to condition the high temperature rupture behavior of this class of steels. The diagram of Fig. 9, for the steels studied, is an example of an integrated picture to provide more general comprehension of the creep fracture behavior of austenitic stainless steels.

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