

The effect of carbon on high temperature deformation of Fe-40Al- x C (at.%) ternary alloys

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Received 20 September 2007, received in revised form 14 January 2008, accepted 14 January 2008

Abstract

Mechanical properties of Fe-40Al alloys with C contents of 0.07–1.9 at.% were studied by uniaxial compressive creep and constant strain rate tests at temperatures from 600 °C to 750 °C. The dependence of both the creep resistance and the constant strain rate characteristics on the carbon content is not monotonous; the worst mechanical properties are observed in the alloys with 0.75 and 1.7 at.% C. The final heat treatment consisting in annealing at 1100 °C for 8 hours with quenching to oil has only a small influence on the creep rate. The stress and temperature dependence of the creep rate was determined by stepwise loading and evaluated in terms of the stress exponent n and the activation energy Q , respectively. These quantities can be interpreted by means of dislocation motion controlled by climb and by the presence of second phase particles. Deformation in the alloys with 0.75 and 1.7 at.% C occurs by the dislocation motion as in the single-phase FeAl matrix, eventually modified by the presence of dissolved carbon atoms (solid solution hardening). The dislocation motion is obstructed by tiny precipitates, probably of carbide κ , in the alloy with 0.07 at.% C and by particles of Al_4C_3 in the alloy with 1.9 at.%. Precipitates of carbide κ do not improve creep resistance efficiently at moderate carbon additions.

Key words: iron aluminides, creep, high temperature deformation

1. Introduction

The Fe-Al based alloys with 40 to 45 at.% of aluminium were studied extensively in the early fifties of the last century in the former Czechoslovakia [1, 2]. The research resulted in the development of the alloy with commercial name Pyroferal. The alloy was frequently used to replace deficient heat-resistant nickel-chromium steels in many industrial applications (blades in roasting furnaces, cementation pots, annealing boxes, fire-grate bars etc. [3]). Its composition is given in Table 1. Note that the composition reminds that of the recently studied FeAl-based alloys with different additions of carbon [4-6]. The main task of present experiments is to approach step by step the composition of the “old intermetallic”. Using this pro-

Table 1. The composition of Pyroferal

	Al	C	Si	Mn
wt. %	29.0–31.0	1.0–1.2	max. 0.5	max. 0.7
at. %	44.4–46.5	3.4–4.0	max. 0.7	max. 0.5

cedure, the effect of individual additives should be understood. The effect of carbon content in the ternary Fe-40Al- x C is the first we decided to investigate. The results of the phase analyses and of microstructure observations were published in the preceding paper [7]. In the present paper, the results of high temperature mechanical tests (constant strain-rate test and creep test) are summarized.

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Table 2. Composition of the alloys used (at.%/wt.%)

Alloy	Denoted	Al	C	Si
Fe-40Al-0.1C	D	41.05/25.2	0.07/0.02	
Fe-40Al-0.8C	E	40.5/24.9	0.75/0.19	
Fe-40Al-1.7C	F	40.05/24.8	1.7/0.44	
Fe-40Al-1.9C	H	41.3/26.08	1.9/0.52	1.2/0.8

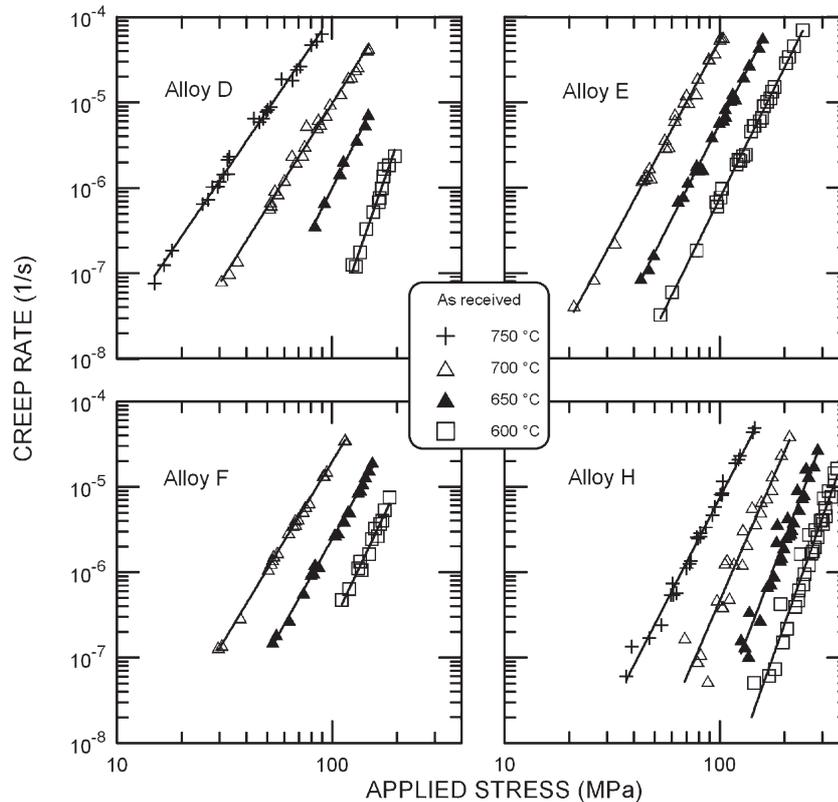


Fig. 1. Dependence of creep rate on applied stress in alloys D to H at different temperatures.

2. Experimental procedure

The alloys D to F were melted and cast as slabs $30 \times 50 \times 100$ mm in a vacuum furnace at the Institute of Modelling and Control of Forming Processes, Technical University, Ostrava, Czech Republic. The slabs were hot rolled at 1200°C in several steps (20 % deformation in each step) to obtain sheets 8 mm thick. Alloy H was cast in the foundry Kdynium, Strakonice, Czech Republic. It is available as casting only (the rolling is impossible) and in an annealed form (see below). The composition of tested materials is in Table 2.

Specimens for mechanical tests were spark machined cylinders (diameter 5 mm for constant strain rate tests, diameter 6 mm for creep tests, height 12 mm). Constant strain rate tests were performed in uniaxial compression at temperature 600°C and (engineering) strain rate $6 \times 10^{-5} \text{ s}^{-1}$. Two types of uniaxial compressive creep tests were performed: (i) at

constant stress on a special cantilever machine with load compensation by Hofman's principle [8] and (ii) at constant load at simple cantilever machine. The former type was performed at 600°C and 120 MPa. The step-wise loading was used in the latter type of tests: in each step, the load was changed to a new value after stationary creep rate had been established. The terminal values of the true stress and the true strain rate were evaluated for the respective step. This type of testing was primarily intended to estimate the stress sensitivity of the creep rate and its temperature dependence. All tests were performed in protective argon atmosphere. The temperature was kept constant within 1 K. The specimens were tested in two different states: (a) as received after casting and rolling (alloys D, E and F) or after casting (alloy H), respectively, and (b) after annealing at 1100°C for 8 hours with quenching to oil.

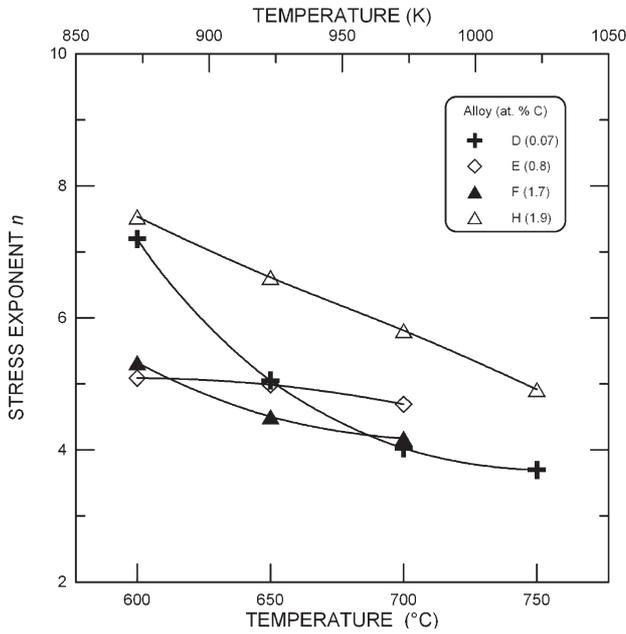


Fig. 2. Temperature dependence of stress exponent n in as-received state.

3. Results

The applied stress dependences of the creep rate in the as-received materials are given in Fig. 1. The dependence can be described at a given temperature by the power function

$$\dot{\epsilon} = A\sigma^n, \quad (1)$$

where $\dot{\epsilon}$ is the creep rate, σ is the applied stress and A is a temperature dependent constant. The power n is decreasing with increasing temperature in all studied alloys. Its values are given in Fig. 2.

Values of logarithm of creep rate found by fitting by means of Eq. (1) were plotted as function of reciprocal temperature. The apparent activation energy of creep

$$Q = -R \left(\frac{\partial \ln \dot{\epsilon}}{\partial \frac{1}{T}} \right)_\sigma \quad (2)$$

was then calculated by linear regression (R is the universal gas constant and T is the absolute temperature). The results are given in Fig. 3. The activation energy is decreasing with the increasing applied stress. This observation is connected with the above-shown decrease of the stress exponent with the increasing temperature. The activation energy extrapolated linearly to zero applied stress can be separated to two different groups: (i) it is equal to 350 kJ mol^{-1} and 383 kJ mol^{-1} in the alloys E and F, respectively, and

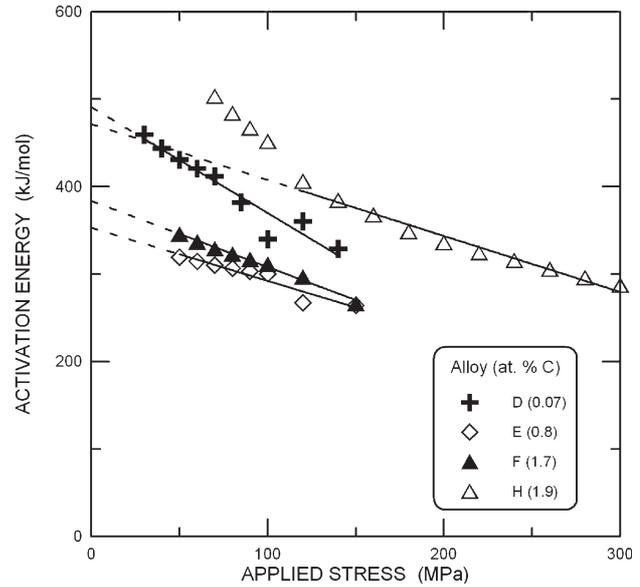


Fig. 3. Applied stress dependence of activation energy of creep rate in as-received state.

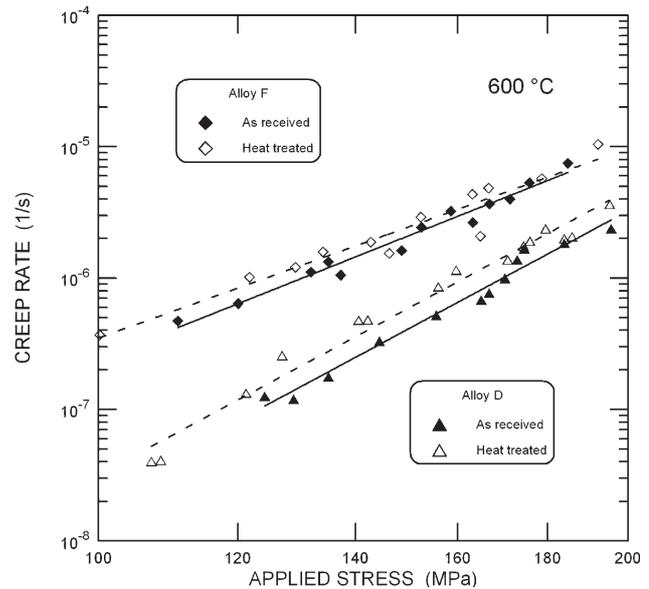


Fig. 4. Applied stress dependence of creep rate in alloys D and F in as-received and heat-treated states.

(ii) it is equal to about 490 kJ mol^{-1} and 470 kJ mol^{-1} in the alloys D and H, respectively. (The data for the applied stress greater than 120 MPa were taken into account for this linear extrapolation in the alloy H. The values of the activation energy about 700 kJ mol^{-1} can be obtained by the extrapolation of low stress data in this alloy.)

The effect of heat treatment on creep properties is illustrated in Fig. 4. The applied heat treatment has only a small influence on the minimum creep rate.

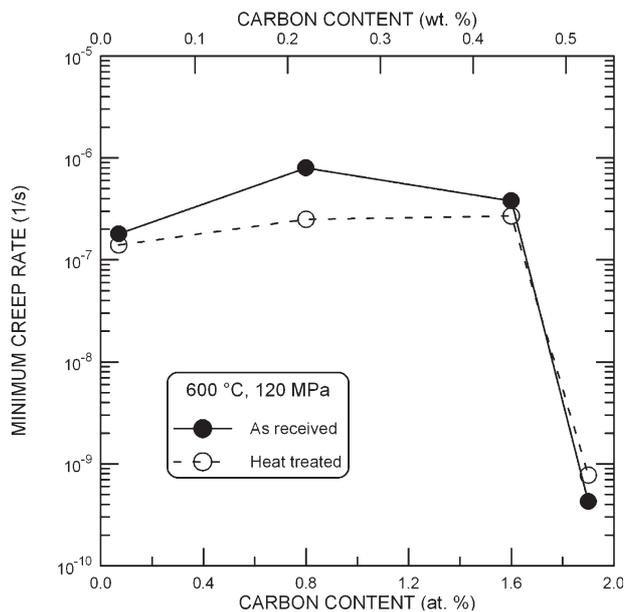


Fig. 5. Dependence of minimum creep rate in constant-stress tests on carbon content.

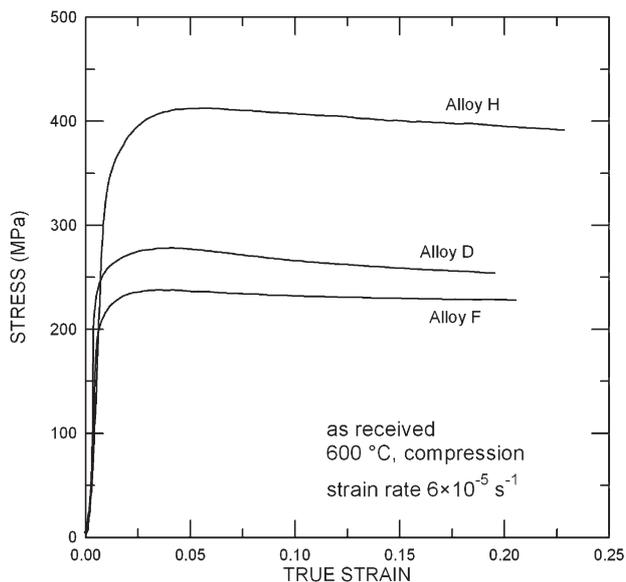


Fig. 6. Examples of constant strain rate tests.

The dependence of the minimum creep rate obtained in constant-stress tests on the carbon contents is given in Fig. 5. The dependence is not monotonous, note a lower creep resistance with the increasing carbon content in the alloys E and F.

The examples of stress vs. strain dependence obtained in constant strain rate test are given in Fig. 6. Two important quantities can be determined from the constant-rate tests: (i) the yield stress and (ii) the maximum value of stress. The yield stress was eval-

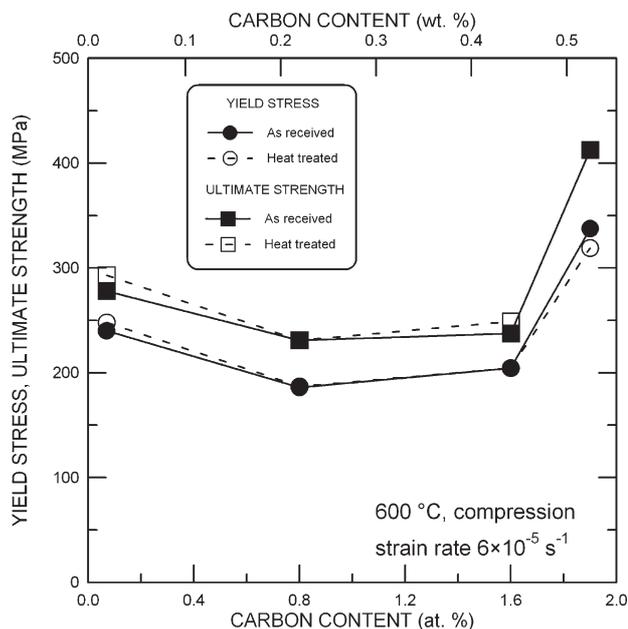


Fig. 7. Dependence of yield stress and ultimate strength on carbon content.

uated as stress causing a permanent strain of 0.002. The maximum value of the stress is equivalent to the ultimate tensile strength (UTS) measured usually in the tensile variant of the test. It will be designated as ultimate strength in what follows. The dependences of both the yield stress and the ultimate strength on the carbon content are given in Fig. 7. They follow the same pattern as the creep resistance depicted in Fig. 5.

4. Discussion

The stress exponents in the alloys E and F (and in the alloy D at higher temperatures) varied between 4 and 5. This is usually taken as an indication that the recovery processes during creep are controlled by dislocation climb in single-phase matrix. On the other hand, the greater values of the stress exponent in the alloy H and in the alloy D at 600°C suggest that the dislocation motion is impeded by dispersed second-phase particles. The observed great values of the activation energy can be explained by the same argumentation: The volume fraction of secondary particles is expected to be temperature dependent and the measured activation energy is thus greater than the value corresponding to the volume diffusion of both Fe and Al in FeAl matrix. This interpretation can be supported by following microstructural observations.

Tiny precipitates are formed in the alloy D as suggested by an appearance of dislocation network (cf. Figs. 8a,b). It is evident that the dislocations are

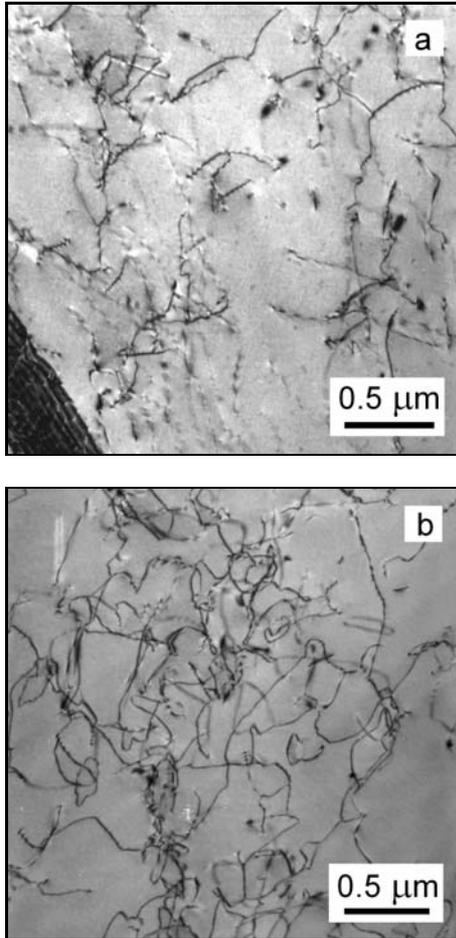


Fig. 8. Dislocation structure after creep at 600°C in alloy D in (a) as received state and (b) heat-treated state.

pinned down by very small particles that obstruct their motion. The type of particles and/or their composition was not possible to determine. It seems probable that they can be assigned as iron carbides.

In alloys E and F, particles of carbide κ (Fig. 9) are formed and the matrix is depleted of dissolved carbon. Though the dislocations react with particles of diameter up to 500 nm, the density of such particles is too low to prevent dislocation motion. The fraction of small particles that react with dislocations is unsatisfactory.

The alloy H with great proportion of carbide Al_4C_3 is characterized by an outstanding creep resistance and in this point it differs from the other investigated alloys. Dendritic structure of this alloy that is formed by needles of Al_4C_3 carbide is dissolved only partially at 600°C (cf. Figs. 10a,b) and it constitutes obstacle to dislocation motion in creep deformation.

It could be concluded, that the deformation in the alloys E and F occurs by the dislocation motion as in the single-phase FeAl matrix, eventually modified by the presence of dissolved carbon atoms (solid solution

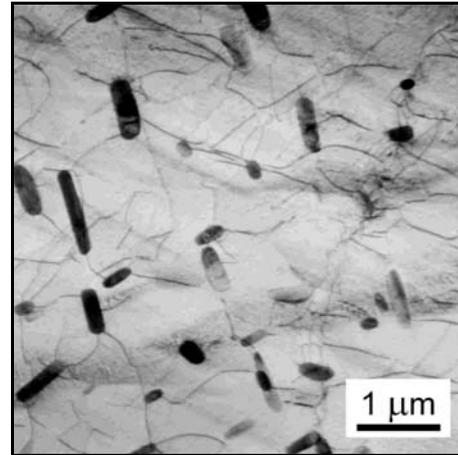


Fig. 9. Dislocation structure after creep at 600°C in alloy E in as received state.

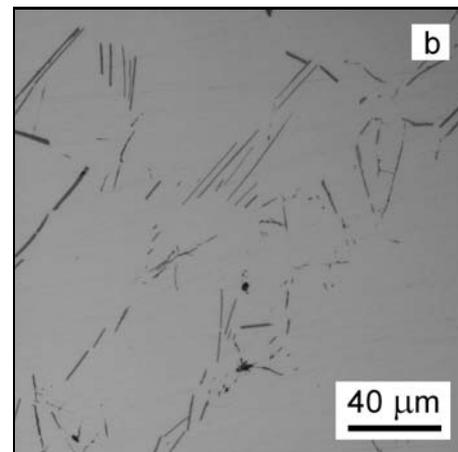
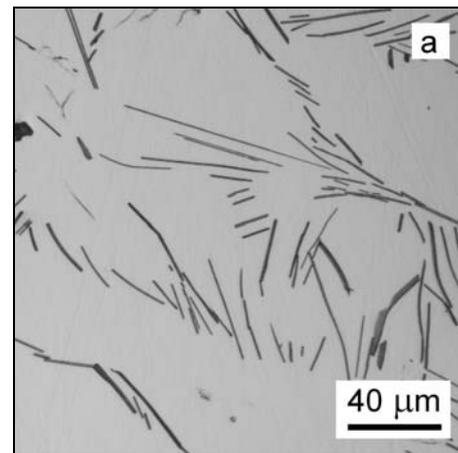


Fig. 10. Dendritic structure in alloy H in (a) as cast state and (b) after creep at 600°C.

hardening). The dislocation motion is obstructed by tiny precipitates, probably of carbide κ , in the alloy D and by particles of Al_4C_3 in the alloy H.

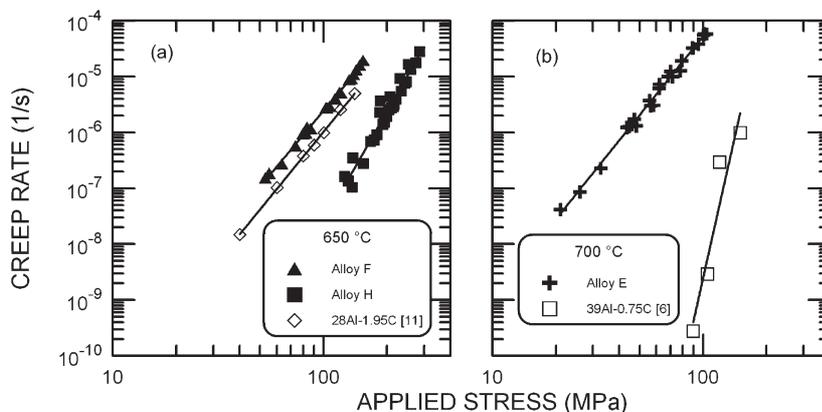


Fig. 11. Comparison of stress dependence of creep rate in present alloys and in Fe-Al alloys with carbon additions published in Refs. [11] (a) and [6] (b).

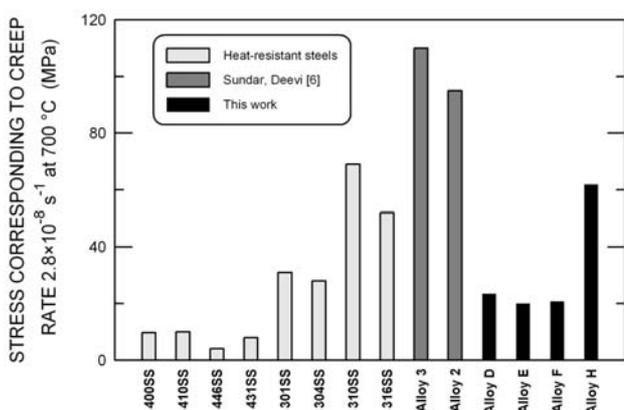


Fig. 12. Comparison of the creep resistance of present alloys with that of several grades of commercial steels and of Fe-Al-C alloys by Sundar and Deevi [6].

A decrease of elevated temperature mechanical properties with increasing carbon content was reported by Baligidad et al. [9, 10] also in Fe₃Al-based alloys. The reduction in creep resistance with an increase in carbon addition from 0.27 to 1.3 wt.% was attributed to the precipitation of carbon as soft graphite [9]. The amount of carbon in the alloys studied in the present contribution is too low to accept this explanation and in fact the graphite phase is not observed here. On the other hand, the decrease of the ultimate tensile strength and of the yield strength at 600 °C with the increase of carbon addition from 0.27 to 0.50 wt.% [10] may be related to the same reasons that worsen mechanical properties of the alloys E and F with respect to the alloy D.

A quantitative comparison of the present results with the data reported by the Baligidad's group can be done for the results of compressive creep tests at 650 °C given in Ref. [11] (cf. Fig. 11a). Creep rate of the reported alloy with 0.5 wt.% of C (about 1.95 at.%) is comparable with the creep rate in the present al-

loy F with 1.7 at.% of C and is faster than the creep rate in the present alloy H with 1.9 at.% of C but the difference in aluminium content should be also noted.

Comparison with the literature data of carbon-alloyed iron aluminides of the same class of aluminium content can be done for the data given by Sundar and Deevi [6] for the alloy with 39.33 at.% of Al and 0.75 at.% of C at 700 °C, Fig. 11b. Creep resistance of the alloy of Sundar and Deevi is substantially superior to that observed in the comparable alloy E of the present paper. This is certainly due to the additions of carbides-forming elements as Mo, Zr and Ti in the alloy of Sundar and Deevi. The excellent creep resistance of this alloy – comparable with or better than that of various superalloys (cf. Fig. 12b in [6]) – confirms the effectiveness of alloying of FeAl-based alloys with carbon. The Sundar's and Deevi's comparison with several grades of commercial heat-resistant steels at 700 °C (Fig. 12a in [6]) can also be extended to the present results. Though the stress that results in a creep rate of $2.8 \times 10^{-8} \text{ s}^{-1}$ in the alloy H of the present study (60 MPa) is lower than that in Sundar's and Deevi's FeAl-based alloys, it is surpassed by the steel 310 SS only of the selected eight heat-resistant steels (Fig. 12).

5. Conclusions

- The values of the stress exponent and of the activation energy of creep can be explained by dislocation motion controlled by climb and by the presence of second-phase particles.

- Deformation in the alloys E and F occurs by the dislocation motion as in the single-phase FeAl matrix, eventually modified by the presence of dissolved carbon atoms (solid solution hardening).

- The dislocation motion is obstructed by tiny precipitates, probably of carbide κ , in the alloy D and by particles of Al₄C₃ in the alloy H.

- Precipitates of carbide κ do not improve creep resistance efficiently at moderate carbon additions.
- Creep resistance of the present alloy H is better than that of several grades of austenitic, ferritic and martensitic stainless steels and it is surpassed by the steel 310 SS only.

Acknowledgements

The paper is based on work supported by the Grant Agency of the Czech Republic within the project 106/05/0409.

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