A R C H I V E S O F M E T A L L U R G Y A N D M A T E R I A L S
Volume 52 2007 Issue 2

R. ŁYSZKOWSKI*, J. BYSTRZYCKI*

INFLUENCE OF TEMPERATURE AND STRAIN RATE ON THE MICROSTRUCTURE AND FLOW STRESS OF IRON ALUMINIDES

WPŁYW TEMPERATURY I SZYBKOŚCI ODKSZTAŁCENIA NA MIKROSTRUKTURĘ I NAPRĘŻENIE PŁYNIĘCIA STOPÓW Fe-Al

Iron aluminide alloys are promising advanced structural materials for high temperature applications. They offer a combination of such properties as excellent corrosion and wear resistance, lower density and cost advantage over the Fe-Cr or Fe-Cr-Ni stainless steels. The subject matter of this work focuses on studying characteristics of hot deformation of iron aluminides by using a compression test in the temperature range 600-1100°C and the true strain rate range 0.001-100 s^{-1}. The flow stress has been found to be strongly dependent on the temperature as well as on the strain rate. At strain rates higher than about 10 s^{-1} or temperatures lower than 800°C, the materials exhibited flow softening type of stress-strain curves, while at lower strain rates and temperatures higher than 800°C, the flow curves were of steady-state type. At lower strain rates (<1 s^{-1}) and higher temperatures (>800°C) the material undergoes dynamic recrystallization to produce a fine-grained microstructure.

Keywords: Fe-Al alloys, hot deformation, microstructure, mechanical properties

Stopy żelazo-aluminium (tzw. aluminidki żelazowe) należą do perspektywicznych zaawansowanych materiałów konstrukcyjnych do pracy w wysokich temperaturach. Oferują one kombinację takich właściwości, jak: doskonałą odporność na korozję i zużycie ścierne, małą gęstość i niższe koszty materiałowe w porównaniu do stali Fe-Cr lub Fe-Cr-Ni. W niniejszej pracy badano charakterystyki odkształcenia plastycznego na gorąco stopów Fe-Al w próbie ściskania w zakresie temperatury 600-1100°C i szybkości odkształcenia 0.001-100 s^{-1}. Stwierdzono, że naprężenie płynięcia plastycznego silnie zależy od temperatury i szybkości odkształcenia. Przy szybkości odkształcenia powyżej 10 s^{-1} i w temperaturze poniżej 800°C, na krzywej naprężenie – odkształcenie występuje spadek naprężenia płynięcia (mięknięcie). Natomiast przy niższych szybkościach odkształcenia i w temperaturach powyżej 800°C, krzywe płynięcia mają charakter krzywych typu ustalonego. Przy małych szybkościach odkształcenia (<1 s^{-1}) i wysokich temperaturach (>800°C) materiał podlega rekrystalizacji dynamicznej tworząc mikrostrukturę drobnoziarnistą.

1. Introduction

Fe-Al alloys with aluminum content from 16 to 50 at.% are usually called iron aluminides [1]. They are mainly based on Fe_3Al or FeAl intermetallic compound. These materials are commercially attractive due to their low material cost, good wear and corrosion resistance, excellent oxidation and sulfidation resistance, low density and good mechanical properties [2-4]. However, structural applications of iron aluminides have been hampered by their low room temperature ductility due to environmental hydrogen embrittlement and by poor strength above 600°C. In order to overcome these problems a lot of efforts have been made so far in the area of their mechanical behavior at the room and high temperature with putting emphasis on the relationship between their microstructure and mechanical properties [see e.g., 5, 6]. It has been found that the room temperature ductility of iron aluminides can be essentially improved by the controlling of microstructure under thermomechanical treatment.

The present study is therefore directed towards assessing the temperature and strain state dependence of the flow stress of iron aluminides based on disordered (A2) Fe(Al) alloy and ordered (B2) Fe_3Al intermetallic alloy. A microstructural examination in the compression tested samples has been carried out to correlate their mechanical behavior with their microstructural features. It should be also mentioned that this work is a part of the research program being in progress in our research

---

* DEPARTMENT OF ADVANCED MATERIALS AND TECHNOLOGIES, FACULTY OF ADVANCED TECHNOLOGY AND CHEMISTRY, MILITARY UNIVERSITY OF TECHNOLOGY, 00-908 WARSAW, KALISKIEGO 2 STR., POLAND

group on the hot working behavior of iron aluminides [7].

2. Experimental procedure

The chemical compositions of the investigated iron aluminide alloys are listed in Table 1. Both alloys were prepared by induction melting in the argon atmosphere and casting into a cylindrical steel mold. Each ingot of 24 mm in diameter and 150 mm in length was homogenized at 1100°C for 10 h. Hot-hammering steps at about 1100°C were then carried out and rods of 7.8 mm in diameter were obtained. Compression specimens with nominal dimensions of 12 mm in height and 7.6 mm in diameter were cut by electrical discharge machining (EDM) from the hot-hammered rods. Isothermal, constant-strain rate compression tests were carried out on a Gleeble 3800 testing system over a range of strain rates (0.001-100 s^{-1}) and temperatures (600-1100°C). All the tests were performed in the argon atmosphere with the heating rate of 3°C s^{-1}. The specimens were annealed for 10 min at the testing temperature prior to deformation, deformed to a true strain of ~0.65 and water cooled to the room temperature after deformation. The microstructure was observed by an optical microscopy after etching the polished specimens in $33\%CH_3COOH+33\%HNO_3+33\%H_2O+1\%HF$ reagent. Grain size measurements were carried out by using an automatic image analyzer developed by Soft Imaging Software GmbH (SIS) attached to a Philips XL30 scanning electron microscope (SEM). The following parameters of the individual grains were measured: A, grain section area and d_{eq}, equivalent circle diameter of the grain ($d_{eq} = [4A/\pi]^{1/2}$).

TABLE

Nominal compositions (at.%) of investigated iron aluminides

Material designation	Al	Cr	Mo	Zr	B	Fe
FAP	16	5	1.0	0.10	–	balance
Fe_3Al	28	5	–	0.08	0.04	balance

3. Results and discussion

The microstructures of the starting iron aluminide alloys are shown in Fig. 1a,b. As it can be seen both initial materials, FAP and Fe_3Al have an equiaxed grain structure with the average equivalent grain diameter of about 250 and 300 μm, respectively.

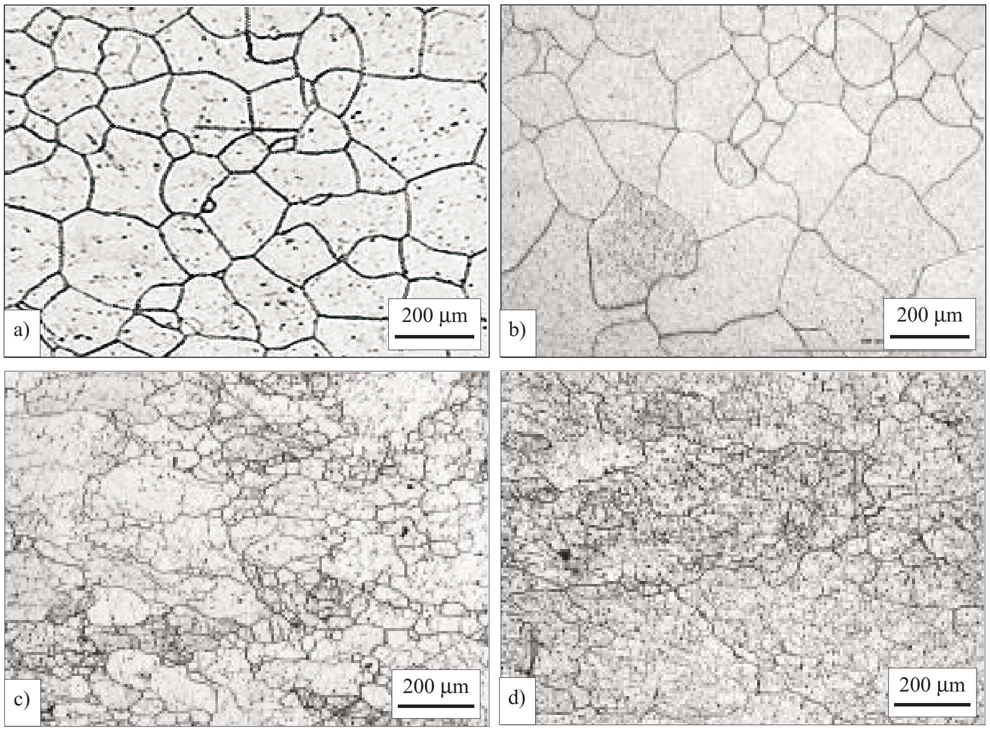

Fig. 1. Starting microstructures of investigated iron aluminides: (a) FAP and (b) Fe_3Al and microstructures showing DRX: (c) FAP deformed at 900°C/0.001 s^{-1} and (d) Fe_3Al deformed at 1100°C/0.01 s^{-1}

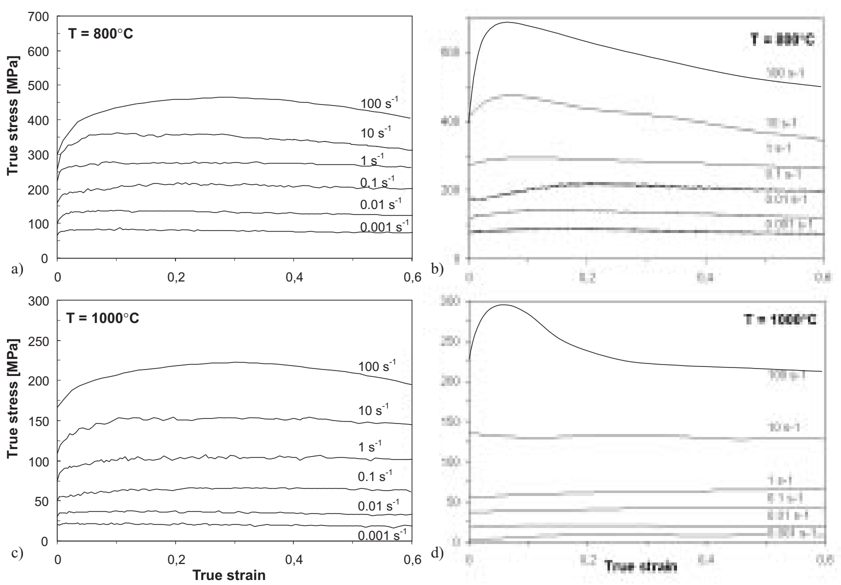

Fig. 2. True stress-true strain curves for investigated iron aluminides deformed at two selected temperatures: (a) FAP-800°C, (b) Fe_3Al – 800°C, (c) FAP – 1000°C and (d) Fe_3Al – 1000°C

The typical true stress – true strain curves recorded for both investigated alloys at the two selected temperatures, 800°C and 1000°C are shown in Fig. 2. The flow curves at strain rates lower than 10 s^{-1} are essentially of steady state type. At higher strain rates (>10 s^{-1}) the flow curves of Fe_3Al intermetallic exhibit significant flow softening after reaching a peak in the flow stress in the initial stages of deformation (Fig. 2b, d). Moreover, some oscillations on the flow curves at temperatures higher than 800°C and strain rates lower than 1 s^{-1} can be observed. Such oscillations generally suggest that dynamic recrystallization (DRX) may occur [8]. However, it is hard to attribute any specific deformation mechanism to the shapes of flow curves since different mechanisms lead to similar behavior, e.g. DRX, flow instability or globularization of lamellar structures. The investigations in this area are in progress in our research group.

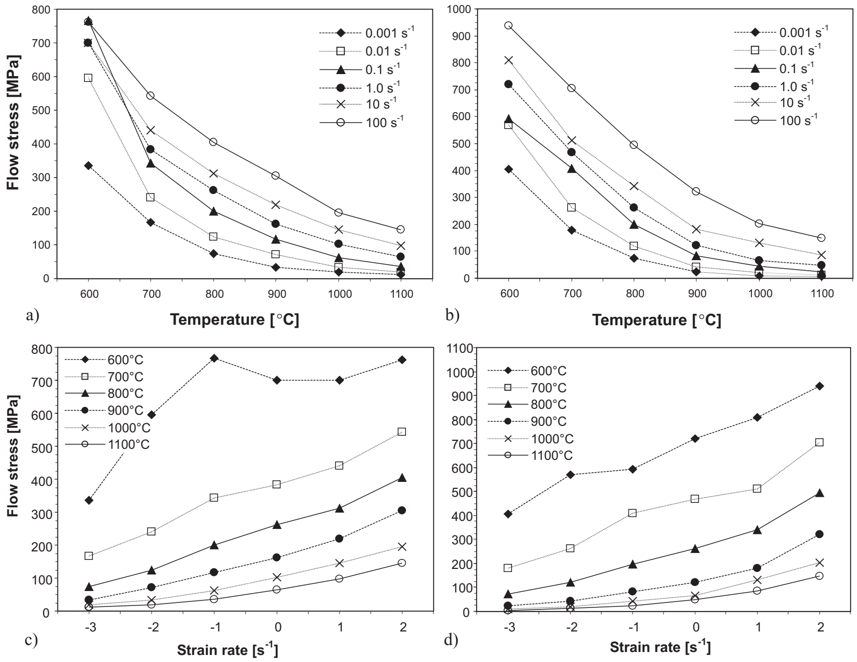

Fig. 3. Temperature and strain rate dependence of flow stress for iron aluminides at 0.6 strain: (a) FAP and (b) Fe_3Al. Variation of flow stress with strain rate at 0.6 strain: (c) FAP and (d) Fe_3Al

Fig. 3a, b shows the effect of the temperature on the flow stress of both alloys at the 0.6 strain level. It is clearly seen that the flow stress at a given strain rate decreases with increasing the temperature in a normal manner. Similar trends are observed at a given temperature, the flow stress increases with the increasing strain rate (Fig. 3c, d). However, the variation of both parameters is less at high temperatures and lower strain rates. The above mechanical characteristics clearly indicate that the deformation of both alloys is thermally activated. Our results are in good accord with the results reported by Sundar et al. [9, 10], who also observed a strong dependence of the flow stress on the temperature and the strain rate.

Fig. 4 shows the variation of average grain size as a function of the temperature. There is no any visible grain growth before DRX in both alloys. The grain size is seen to be almost constant up to about 800°C, for FAP and about 900°C, for Fe_3Al. The grain refinement at temperatures higher than 900°C can be noticed, especially for Fe_3Al intermetallic. Moreover, the grain size in both alloys is finer at higher strain rates. Typical microstructures recorded on the FAP specimen deformed at 900°C at the strain rate of 0.001 s^{-1} and Fe_3Al specimen at 1100°C and 0.01 s^{-1} are presented in Fig. 1c, d, respectively. As it can be seen, both microstructures exhibited fine newly formed grains in the matrix of original grains.

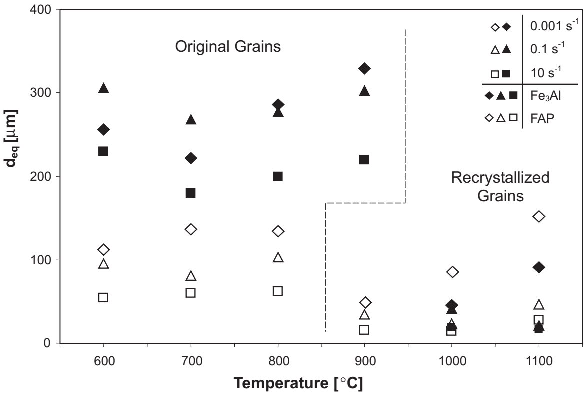

Fig. 4. Variation of grain size with temperature at different strain rates for investigated iron aluminides

Received: 10 March 2007.

4. Conclusions

In the present paper, the temperature and the strain rate dependencies of the flow stress were studied on the basis of the shapes of flow curves and microstructural analysis. Both investigated iron aluminides at temperatures lower than 800°C and higher strain rates (> 10 s^{-1}) exhibited the flow softening type of stress – strain curves, while at higher temperatures and lower strain rates ($\leqslant$ 10 s^{-1}), the flow curves were of steady – state type. Both materials recrystallized at lower strain rates (< 1 s^{-1}) and higher temperatures (> 800°C) to produce a stable fine-grained microstructure.

Acknowledgements

The authors are grateful to Prof. R. Kuziak from the Institute for Ferrous Metallurgy in Poland for his willing assistance in the experimental work as well as for helpful advice. This work has been financially supported by the Polish Ministry of Science and Higher Education.

REFERENCES

[1] V.K. Sikka, S. Viswanathan, C.G. McKamey, Structural intermetallics. in: Darolia R, Lewandowski JJ, Liu CT, Martin PL, Miracle DB, Nathal MV, editors. TMS, p. 483 (1993).

[2] S.C. Deevi, V.K. Sikka, Intermetallics **4**, 357 (1996).

[3] G. Sauthoff, Intermetallics. Wiley-VCH, 1995.

[4] R.R. Judkins, U.S. Rao, Intermetallics **8**, 1347 (2000).

[5] C.G. McKamey, D.H. Pierce, Scripta Metall Mater **28**, 1173 (1993).

[6] D.G. Morris, M. Leboeuf, Acta Metall Mater **42**, 1817 (1994).

[7] R. Łyszkowski, J. Bystrzycki, Intermetallics **14**, 1231 (2006).

[8] J. Konrad, S. Zaefferer, A. Schneider, D. Raabe, G. Frommeyer, Intermetallics **13**, 1304 (2005).

[9] R.S. Sundar, R.G. Baligidad, Y.V.R.K. Prasad, D.H. Sastry, Mater. Sci. Eng. **A258**, 219 (1998).

[10] R.S. Sundar, D.H. Sastry, Y.V.R.K. Prasad, Mater. Sci. Eng. **A347**, 86 (2003).