

High temperature mechanical properties of iron aluminides

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Abstract

Considerable attention has been given to the iron aluminide family of intermetallics over the past years since they offer considerable potential as engineering materials for intermediate to high temperature applications, particularly in cases where extreme oxidation or corrosion resistance is required. Despite efforts at alloy development, however, high temperature strength remains low and creep resistance poor. Reasons for the poor high-temperature strength of iron aluminides will be discussed, based on the ordered crystal structure, the dislocation structure found in the material, and the mechanisms of dislocation pinning operating. Alternative ways of improving high temperature strength by microstructural modification and the inclusion of second phase particles will also be considered.

Keywords

Iron aluminide intermetallics. Mechanical properties. High-temperature strength.

Propiedades mecánicas a alta temperatura de los intermetálicos Fe-Al

Resumen

Durante los últimos años se ha prestado mucha atención a la familia de intermetálicos Fe-Al, puesto que estos constituyen un considerable potencial como materiales de ingeniería en aplicaciones a temperaturas intermedias o altas, sobre todo en casos donde se necesita alta resistencia a la oxidación o corrosión. A pesar del considerable esfuerzo desarrollado para obtener aleaciones con mejores propiedades, su resistencia mecánica a alta temperatura no es muy elevada. Se discutirán los aspectos que contribuyen a la baja resistencia mecánica a temperatura elevada en función de la estructura de dislocaciones y los mecanismos de anclaje que operan en este intermetálico. Se considerarán, también, maneras alternativas para mejorar la resistencia a temperatura elevada mediante la modificación de la microestructura y la incorporación de partículas de segunda fase.

Palabras clave

Intermetálicos de tipo Fe-Al. Propiedades mecánicas. Resistencia a alta temperatura.

1. INTRODUCTION

There are two ordered intermetallics in the Fe-rich part of the Fe-Al system, namely Fe₃Al which has the DO₃ (ordered bcc) crystal structure and exists over the composition range about 25-35 atomic % Al and up to a critical temperature of about 550 °C, and FeAl which has the B2 (also ordered bcc) crystal structure and exists over the composition range about 25-50 atomic % Al and up to a temperature of about 700-1300 °C over this composition range. Both have been extensively examined with a view to understanding their

mechanical, chemical and physical properties and their development as engineering materials for various high-tech applications.

This section gives a brief overview of some of their interesting properties and possibilities for commercial applications. The present document will subsequently concentrate on their mechanical behaviour, particularly at high temperatures. Indeed, one of the limitations restricting applications of these materials is their relatively poor high temperature strength, particularly under creep conditions, and it is important to understand the reasons for such behaviour.

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Reasons for interest in the iron aluminides stem from their high Al content and the tendency to covalent bonding between the constituent Fe and Al atoms. The high Al content means that high temperature oxidation and corrosion resistance can be outstanding under particular environments, for example in oxidising atmospheres rich in oxygen or sulphur (hydrogen sulphide) and under carburising atmospheres (methane). Figure 1 illustrates such behaviour, shown in comparison with more conventional steel-base alloys. Mechanical properties at ambient temperatures are similar to those of stainless steels, in particular typically showing good ductility of the order of 5-10 %. Bonding between Fe and Al atoms takes on a partially metallic, partially covalent nature which is reflected in an increase in the values of elastic moduli such as the Young's modulus and the shear modulus. The presence of the large quantity of Al also means that the density is comparatively low, about 6.7 g/cm³ for Fe₃Al and 5.6 g/cm³ for FeAl, such that specific strength and elastic moduli become more attractive. Figure 2 illustrates^[1] some of these mechanical properties, in comparison with other engineering materials such as steels, showing particularly the good toughness with stiffness and the good strength and low density of the iron aluminides. The ductility appears to be controlled by crack nucleation and propagation, with environmental sensitivity arising through the reaction of the Al with water vapour leading to H⁺ injection near the crack tip and consequent local

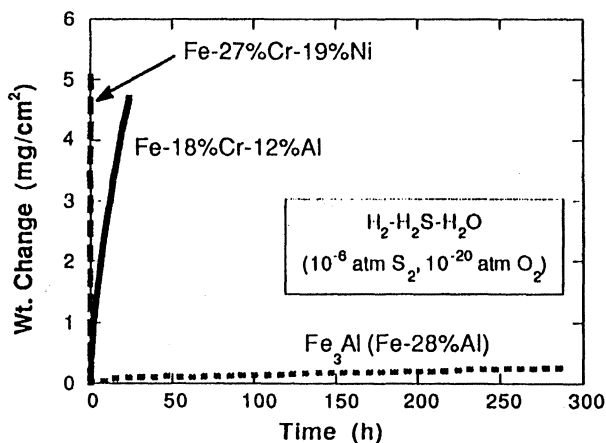
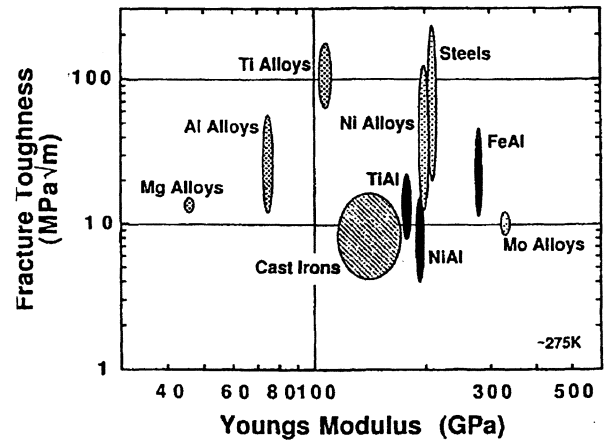


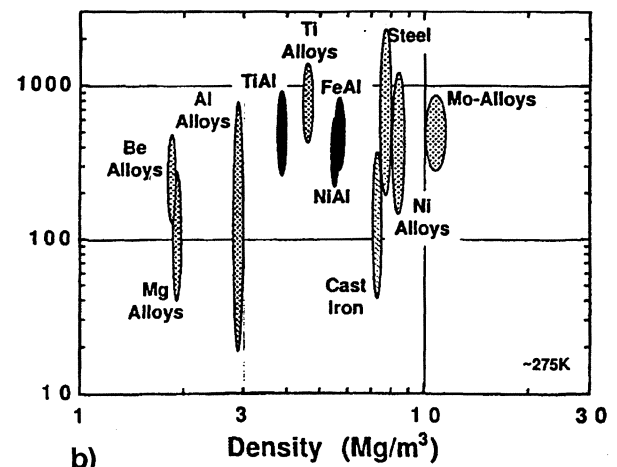
Figure 1. Weight changes of Fe₃Al, a Fe-Cr-Al alloy, and a stainless steel isothermally exposed to H₂S-H₂-H₂O at 800 °C.

Figura 1. Cambios de peso de las aleaciones Fe₃Al, Fe-Cr-Al y acero inoxidable expuestas a H₂S-H₂-H₂O a 800 °C en condiciones isoterma.

Rev. Metal. Madrid 37 (2001)



a)



b)

Figure 2. Typical properties of FeAl intermetallics, in comparison with a variety of other engineering materials^[1].

Figura 2. Propiedades típicas de intermetálicos de FeAl, en comparación con algunos materiales de ingeniería^[1].

embrittlement^[2]. Figure 3 shows results of an analysis of tensile ductility in terms of the yield stress of the particular material and the grain size - a dependence that can be explained by the influence of plastic deformation on crack nucleation and propagation.

Because of the special properties of the iron aluminides, especially those based on FeAl, a wide range of industrial applications has been proposed, and several are now at various stages of commercial evaluation and exploitation. Table I lists some of these applications, together with the specific properties that make the material interesting. As these applications develop the material is likely to transition from being a novel and expensive material to being a highly valued engineering material.

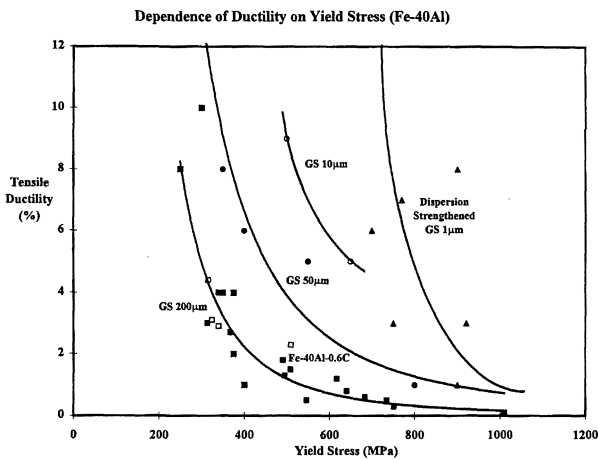


Figure 3. Influence of grain size and yield stress on the tensile ductility of Fe-40Al intermetallic^[2].

Figura 3. Influencia del tamaño de grano y del límite elástico en la ductilidad en tracción del intermetálico Fe-40Al^[2].

2. PLASTIC DEFORMATION OF FeAl - DISLOCATION TYPES

The following discussion will consider the behaviour of the B 2 intermetallic FeAl. In many senses the DO₃ ordered Fe₃Al intermetallic behaves the same, apart from showing on some occasions different types of dislocation structure, and hence much of this discussion can be taken to apply to both types of intermetallics.

The crystal structure of FeAl is illustrated in figure 4, where the possible Burgers vectors of shear dislocations are also shown. Two perfect Burgers vectors are indicated, namely $\langle 100 \rangle$ and $\langle 111 \rangle$. The $\langle 111 \rangle$ vector can be broken into two partial vectors of magnitude $1/2\langle 111 \rangle$ each. The chosen dislocation type in a deformed crystal will depend on the intrinsic mobility of each dislocation type, the energy of the dislocation, and the possibilities of pinning processes operating. The energy of the dislocation is essentially given by the Burgers vector squared (b^2), such that a perfect $\langle 111 \rangle$

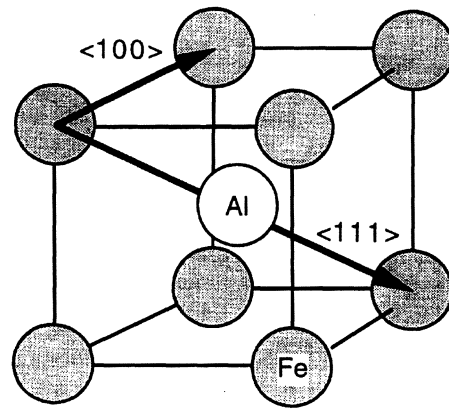


Figure 4. B2 crystal structure and possible Burgers vectors of perfect dislocations.

Figura 4. Estructura cristalina B2 y posibles vectores de Burgers de dislocaciones perfectas.

dislocation has a high (unfavourable) energy, a perfect $\langle 100 \rangle$ dislocation has a rather low energy and an imperfect $1/2\langle 111 \rangle$ dislocation has the lowest possible energy. This imperfect dislocation will trail an order fault (APB) as it moves, however, which greatly reduces its mobility. This problem is overcome as a pair of such imperfect dislocations couple together, separated by the APB fault, to constitute a superdislocation. Intrinsic mobility - due to Peierls effects - is essentially determined by the dislocation core structure, whether the core is planar or non-planar, and whether the plane of the core is a suitable glide plane. These factors mean that the $\langle 100 \rangle$ dislocation has limited mobility at ambient temperatures and the $\langle 111 \rangle$ dislocations are used by the deforming material. Pinning processes cannot operate at ambient temperatures but can become important at elevated temperatures. Figure 5 illustrates the dislocation types that may be found in such FeAl alloys, including the possible mobile dislocations and a locked dissociated superdislocation.

Table I. Applications of Iron aluminide intermetallics

Tabla I. Aplicaciones de intermetálicos de aluminios de hierro

Application	Special Properties
Heating elements (food processing/drug delivery)	Electrical resistance, absence of toxicity
Hot gas/liquid metal filters	Oxidation resistance
Dies in glass processing	Oxidation resistance, wear resistance
Heat exchanger tubing/coatings	Oxidation/erosion resistance
Helicopter/aerospace connectors	High specific moduli

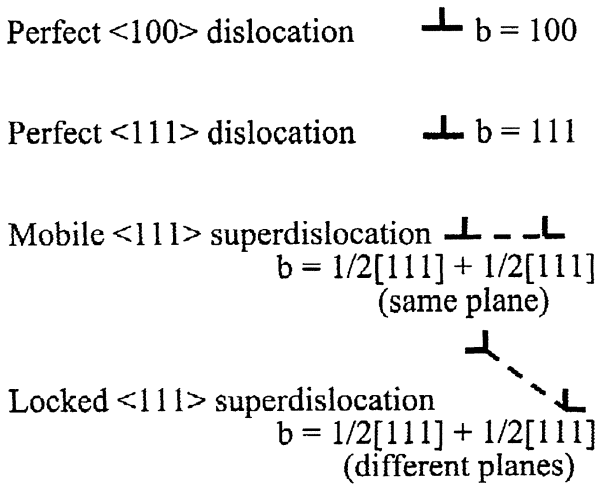


Figure 5. Possible dislocation types in a B2 crystal.

Figura 5. Posibles tipos de dislocaciones en un cristal de estructura B2.

3. INTERMEDIATE TEMPERATURE FLOW STRESS ANOMALY

As many intermetallics, the iron aluminides show the strange characteristic of becoming stronger as the temperature increases above ambient. This behaviour has been nicknamed “anomalous”, even though it is only the strange intermetallic that does not show an anomaly! The anomaly is interesting from the scientific point of view, and also from the technical since it means that the material retains good strength to higher temperatures than otherwise. Figure 6 illustrates the increase in flow stress for both Fe₃Al and FeAl alloys with

temperature over the range about 300 °C to 600 °C. The Fe₃Al example in figure 6 a) shows data on polycrystalline material, and the FeAl example in figure 6 b) shows data on single crystalline material in different orientations.

Examination of the dislocations present after deformation shows three ranges of structures. At low temperatures, from room temperature to about 300 °C, the dislocations present are mobile <111> superdislocations, that is the dislocations show no signs of locking and are on their glide planes (of type {110} or {112}). Over the temperature range of increasing flow stress, about 300-600 °C, the dislocations present are still <111> superdislocations, but now they show various signs of dislocation locking. At high temperatures, over the stress peak, the dislocations found are often of <100> type, and show no clear signs of locking. These three classes of dislocation structure are illustrated in figure 7. The important question to examine is why the initially mobile <111> superdislocations become locked at high temperatures. We can understand that when these become excessively locked the crystal will adopt alternative deformation modes, using <100> dislocations.

Table II summarises the pinning mechanisms that have been suggested to cause locking of the <111> superdislocations. Various mechanisms have been proposed or observed, some for alloys of the Fe₃Al type and some for FeAl alloys. Briefly, the mechanisms proposed are:

- (i) cross slip hardening where by the superdislocations cross slip with such high

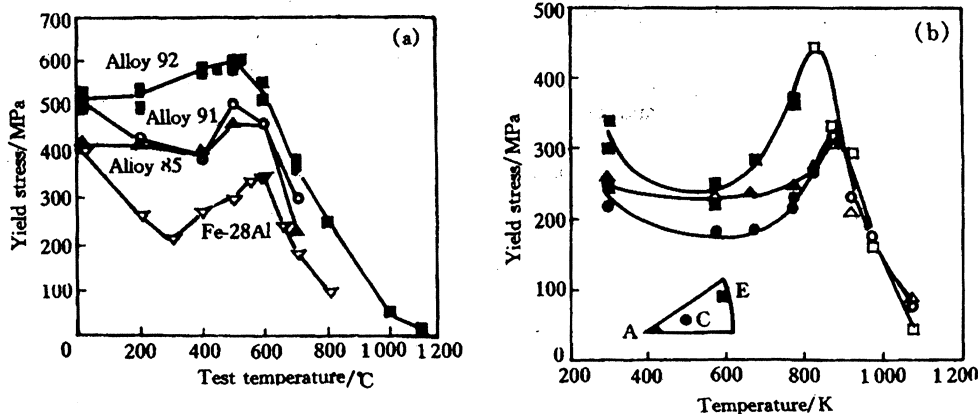


Figure 6. Examples of anomalous flow stress behaviour in Fe₃Al (a) and FeAl (b)

Figura 6. Ejemplos de comportamiento anómalo del límite elástico en Fe₃Al (a) y FeAl (b)

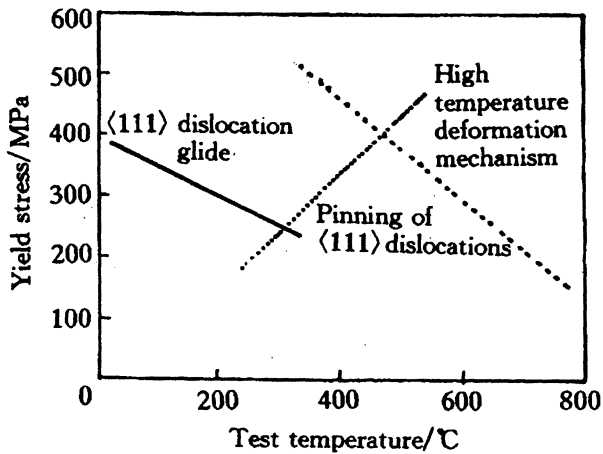


Figure 7. Schematic illustration of dislocation mechanisms operating during the stress anomaly.

Figura 7. Esquema ilustrativo de mecanismos de dislocaciones activos durante la tensión anómala.

frequency that all dislocation segments become very short and the flow stress required (a Taylor type stress) increases - this has been suggested for Fe₃Al alloys but not found for FeAl alloys.

- (ii) APB relaxation - the APB fault between the superdislocation partials is a high energy shear fault which relaxes its structure and perhaps chemistry by local atomic rearrangement or diffusion to a lower energy configuration, which makes further movement of the superdislocation difficult - this phenomenon has been observed for alloys of the Fe₃Al type but the kinetics are perhaps too slow to operate during deformation itself.
- (iii) local climb locking occurs as vacancy transfer between one of the superdislocation partials allows one partial to climb up and one to

Table II. Dislocation pinning mechanisms suggested to explain anomalous stress rise^[3]

Tabla II. Mecanismos de anclaje de dislocaciones sugeridos para explicar la anomalía de tensión^[3]

Strengthening mechanism	Aluminium content (atomic %)					
	25	30	35	40	45	50
Cross slip hardening						
Louchet-type reduction of mobile dislocations	Suggested for this range of compositions.			No evidence of the process in this compositional range.		
APB Relaxation						
Brown-type model. Observed on ageing deformed Fe-28Al and Fe ₃ Al alloys. Little evidence for Fe-35 Al.	Phenomenon observed but kinetics are very (too) slow for many alloys. Should occur for <111> dislocations on {110}, and on {112}, hence slip plane change cannot explain stress fall at high temperatures.			Phenomenon not observed at high Al content.		
Local Climb locking						
Diffusional point defect exchange between <111> partials leads to rapid climb and impossibility of further glide.	Observed on <i>post-mortem</i> samples, but studies of climb kinetics show that the process is very slow.			Phenomenon not reported in studies on these alloys.		
Glide Decomposition						
Occurs due to elastic parameters, preferentially for near-screw orientations. Requires high temperatures to separate product dislocations.	Phenomenon not observed.			Phenomenon observed or suggested.		
Vacancy Hardening						
High concentration of thermally-produced vacancies act as solute.	Concentration of thermal vacancies is very low at the stress anomaly temperature.			High hardening seen in quenched samples. Is it possible that sufficient hardening occurs at the stress anomaly temperature?		

climb down, similar to the locked configuration illustrated in figure 5 this phenomenon has been observed, but perhaps occurs too slowly to take place during deformation.

- (iv) glide decomposition involves the breakdown of the $\langle 111 \rangle$ superdislocation into a perfect $\langle 100 \rangle$ and a perfect $\langle 110 \rangle$ dislocation, and since these two will glide away on separate planes, even if they are mobile, a strong lock is produced - this process has been observed in FeAl alloys.
- (v) vacancy hardening - since FeAl especially (not Fe₃Al) has a very high vacancy concentration at high temperatures, it is suggested that a form of solution hardening occurs at the intermediate temperatures as many stationary vacancies are produced. As clear from table II, there is various evidence of several workers and on various alloys of the operation of each of these mechanisms. Figure 8 shows examples of dislocations in several of these locked configurations. These micrographs simply confirm that such pinning can occur, it does not necessarily follow that a specific process is important during deformation itself. This important distinction has made the analysis of pinning mechanisms an area of violent scientific debate in many specialist conferences!

The most likely mechanism operating is today believed to be either that of the glide decomposition model or that of vacancy hardening. Table III shows a comparison of these two mechanisms in terms of what may be termed a kinetic analysis, that is by examining the temperature and the strain rate dependence of the flow stress anomaly, that is in terms of an activation energy and an activation volume of plastic flow - and by examining the flow stress reversibility of the stress anomaly - i.e. what is the flow stress response if one suddenly jumps the temperature up or down from below or above the stress peak. While the mechanical data analyses are also inconclusive, as are the transmission electron microscope analyses, the general evidence favours the vacancy hardening mechanism as being the operating one. The important results are: that APB relaxation and climb kinetics are argued to be too slow to operate during a typical deformation experiment, cross slip locking does not seem to be active (at least for FeAl alloys), and the kinetic data is incompatible with

the glide decomposition mechanism. The most significant information is provided by the strain-rate sensitivity of the flow stress anomaly and by experiments involving temperature jumps. The stress-rising portion of the stress anomaly is shown to be totally independent of the strain rate, ranging from slow, almost creep-style strain rates up to strain rates involving high velocity impact. Jumping the temperature rapidly over the stress peak is shown to lead to a short time dependence of the hardening, confirming that this hardening is time dependent but not strain rate dependent. This implies that hardening is a state parameter, and not a rate parameter - as would be solution hardening but not a thermally-activated locking process.

While the results and analyses described here provide convincing proof of the operation of a particular strengthening mechanism, it should be emphasized that this research field is still in a state of flux, and new experiments or theoretical analyses may yet change this story. The development of our understanding over the last five years has also been a salutary one, showing how convincing scientific proof using sophisticated experimental apparatus, and widely believed by the scientific community, can suddenly be overturned by the appearance of new data from an unexpected direction.

4. LOW STRENGTH AT HIGH TEMPERATURE

At high temperatures, above the anomalous stress peak, the flow stress of the iron aluminides falls to low values, illustrated in figure 6. Such low flow stress is also reflected in a very low creep strength at potential temperatures of usage, for example 600-800 °C. The origin of this low strength is the simple dislocation structure of the dislocations responsible for deformation, namely the perfect $\langle 100 \rangle$ dislocations. These do not have a complex dissociated structure susceptible to pinning, and at high temperatures any core effects - leading to significant Peierls stresses - are readily overcome. In addition, there is at high temperature a very high vacancy concentration in the open bcc-base material but these vacancies are now mobile, in contrast to the case at the temperatures of the flow stress anomaly, such that both dislocation glide and climb processes operate easily in the materials. For these intrinsic reasons the high temperature strength properties are no better than those of a simple steel.

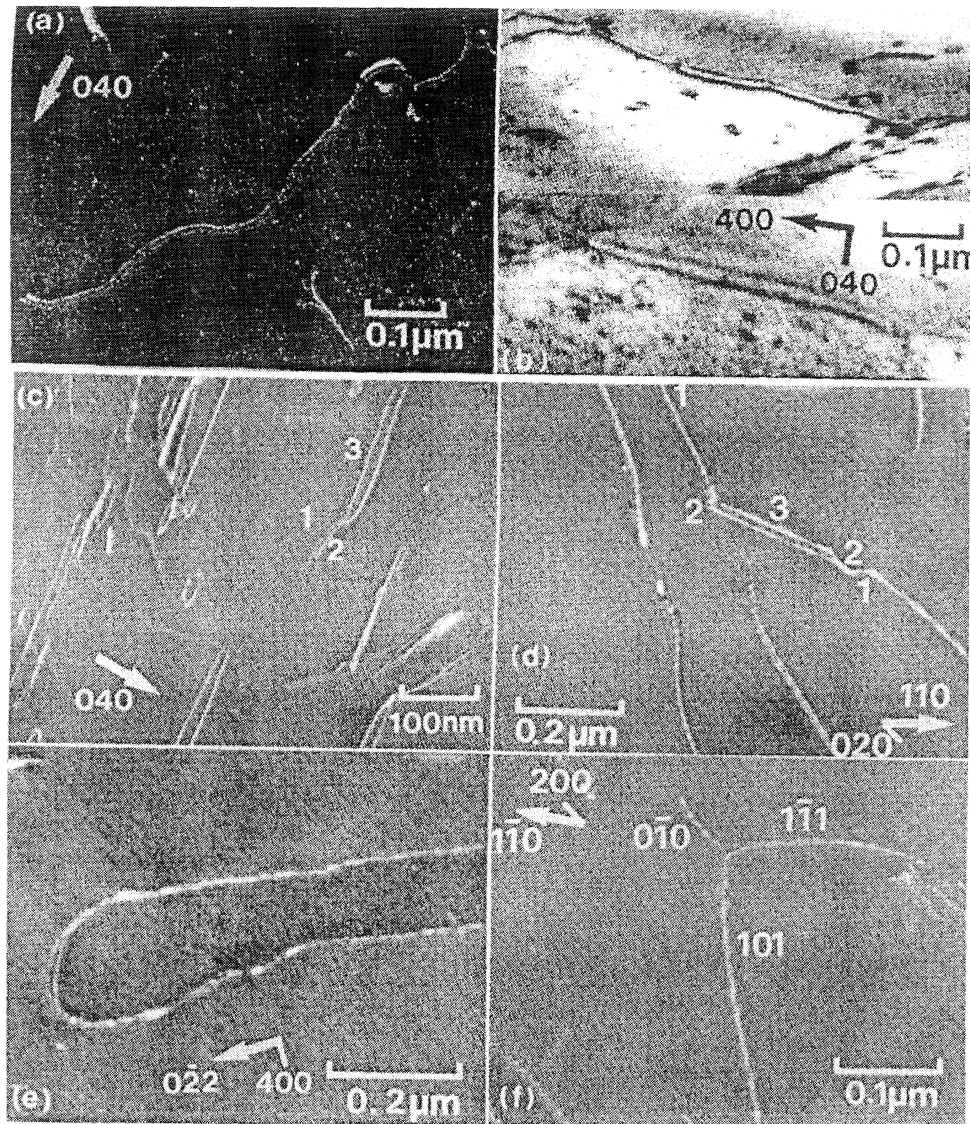


Figure 8. Examples of dislocation pinning mechanisms in FeAl: (a) glide $\langle 111 \rangle$ superdislocation, (b) the same superdislocation during APB relaxation, (c) reactions during annealing to create $\langle 100 \rangle$ dislocations, (d) reactions during deformation to create $\langle 100 \rangle$ dislocations, (e) climb-locked dislocations, (f) decomposition of $\langle 111 \rangle$ into $\langle 100 \rangle$ and $\langle 011 \rangle$. (from ref 33).

Figura 8. Ejemplos de mecanismos de anclaje de dislocaciones en FeAl: (a) deslizamiento de superdislocación $\langle 111 \rangle$, (b) la misma dislocación durante la relajación del APB, (c) reacciones para crear dislocaciones $\langle 100 \rangle$ durante el recocido, (d) reacciones para crear dislocaciones $\langle 100 \rangle$ durante la deformación, (e) dislocaciones bloqueadas por trepado, (f) descomposición de dislocaciones $\langle 111 \rangle$ a $\langle 100 \rangle$ y $\langle 011 \rangle$ (de la ref. 3).

5. HARDENING BY THE INTRODUCTION OF SECOND PHASE PARTICLES

In cases when the strength of a material is insufficient, the classical metallurgical response is to incorporate a large volume fraction of small particles to act as obstacles to dislocation motion. Such particles should be sufficiently stable that they can withstand long periods at high

temperatures without significant coarsening. Three approaches to obtaining such fine and stable particles have been examined, and are briefly reported here: the classical method of solutionising at high temperature and then precipitating out fine particles; modifying alloy composition to obtain a duplex phase mixture; and incorporating stable and insoluble dispersoid particles by a mechanical alloying approach.

Table III. Kinetic analysis of stress anomaly in Fe-Al alloys (taken from ref. 3)*Tabla III. Análisis de la cinética de la anomalía de tensión en aleaciones de Fe-Al (tomado de la ref. 3)*

Experimental observations at Fe ~ 40%Al	Glide decomposition model	Vacancy hardening model
Activation energy Q		
This analysis considers that the anomalous strengthening process is thermally activated. (Could be vacancy formation or the stress-assisted separation of <100> and <110> dislocations. Values of ~50 kJ/mol are observed, varying with orientation, 20-100 kJ/mol.	No value has been predicted. Stress direction can assist the movement of <100> or <110> dislocations away from the decomposition point, helping to stabilise the lock.	Should be $\Delta H_f / 2$ (ΔH_f is the enthalpy for vacancy formation). Thus, Q should be 40-50 kJ/mol, independent of orientation.
Activation Volume V		
High values found at low temperatures falling to the anomalous stress peak: values ~ 2,00 b ³ , falling to ~ 100 b ³ to 500 b ³ .	Thermally activated decomposition reduces the mobile segment length at higher temperatures.	Thermally activated vacancy formation reduces the mobile segment length at higher temperatures.
Flow stress reversibility		
On jumping temperature up or down the flow stress quickly finds its equilibrium, reversible value for the new temperature.	Mobile dislocations of suitable character, segments of <111> superdislocations or <100>, are recreated at each temperature.	Vacancies are quickly created on increasing temperature, but are difficult to annihilate on cooling. Is model compatible?

Possibilities of microalloying to allow solutionising at high temperatures and then precipitation at lower temperatures can be evaluated from the information^[4] given in table IV. The important features of this approach are ensuring sufficient solubility at high temperature, for example 1 %, such that a sufficient volume fraction of reinforcing phase can subsequently be precipitated, while ensuring that the solubility is low enough at low temperatures, for example at the 1000 K operating temperature. Precipitate particles of compounds such as Al₂O₃ would be tremendously stable at 1000 K, as indicated by the very low solubility of the phase, but the compound is unfortunately so stable that no more than a few parts per million can be dissolved at even the highest solutionising temperatures. On the other hand, compounds such as the perovskite Fe₃AlC phase appear to be insufficiently stable at the usage temperature, even though a reasonable quantity of this phase can be dissolved by heat treatments. Figure 9 shows an example of such carbides precipitated from a Fe-Al-C alloy by solutionising and ageing treatments^[5]. While unstable at high operating temperatures such particles may nevertheless play an important role in producing room temperature strengthening. Other precipitates, such as NbC or ZrC, have been used successfully to improve creep strength by such solutionising and

Table IV. Información on the stability of selected precipitate compounds in a FeAl matrix with an estimate of particle solubility at the likely material operating temperature (1000K). Taken from ref. 4.*Tabla IV. Información sobre la estabilidad de compuestos de precipitados seleccionados en una matriz de FeAl con la solubilidad estimada de las partículas a la temperatura probable de trabajo (1000K). Tomado de la ref. 4.*

Compound	Free energy of formation at 1000 K (kJ/mol)	Solubility at 1000 K
Al ₄ C ₃	174	5·10 ⁻²
Fe ₃ AlC	27	20%
NbC	120	7·10 ⁻⁴
SiC	55	3%
VC	95	3·10 ⁻³
WC	36	10%
AlN	200	6·10 ⁻⁶
CrN	40	9%
NbN	190	1·10 ⁻⁵
CrB ₂	85	3%
FeB/Fe ₂ B	66	2% / 7%
NbB ₂	240	6·10 ⁻⁵
TaB ₂	200	3·10 ⁻⁴
Al ₂ O ₃	1.360	1·10 ⁻¹⁵

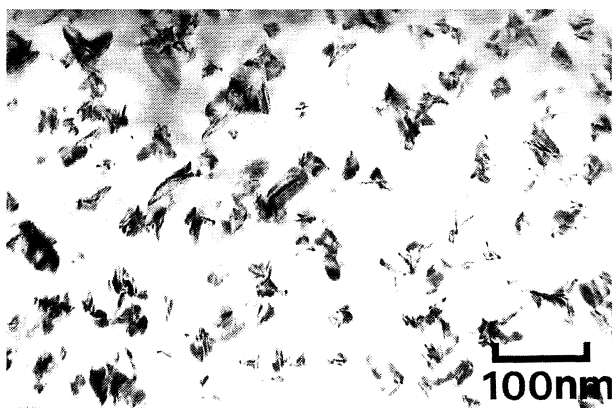


Figure 9. Fine carbide particles in a Fe-40Al-C alloy after solutionising at 1000°C and ageing for 1h at 340°C. Taken from ref.5.

Figura 9. Partículas finas de carburos en una aleación Fe-40Al-C después de solubilizar a 1.000°C y envejecer a 340°C, 1h. Tomado de la ref.5.

ageing heat treatments^[6], but the improvement is seen to be limited due to the small amount of precipitate phase which can be produced.

Another way of strengthening is to follow the approach used for the γ - γ' superalloys, namely to modify the alloy composition such that a large volume fraction of ordered phase sits in a disordered matrix. This may be done for the Fe₃Al alloys by reducing the Al content to slightly below 25 %, where a two-phase α + α'' mixture is obtained (here α is the disordered Fe(Al) matrix and α'' is the DO₃ ordered Fe₃Al phase). Figure 10 shows an example of such a microstructure, obtained by annealing an alloy containing about 23 %Al at 500 °C to give an ordered volume fraction of about 25 %. Iron aluminides do not show the stability and strength of the γ - γ' superalloys for two reasons^[7]: that the volume fraction changes sensitively with the test temperature, although this can be somewhat improved by alloying additions, for example small Si additions, and that the strength of ordering is weaker than in the γ - γ' superalloys and hence dislocations cut the ordered particles instead of being funnelled down the narrow disordered channels^[7].

The final way of strengthening the material, through the use of mechanical alloying, is illustrated in figure 11, showing a Fe-40Al alloy with 1vol % Y₂O₃ particles introduced as fine and stable dispersoids. During deformation at high temperature these fine particles lead to a considerable increase in flow strength, particularly under creep conditions^[8]. The major disadvantage

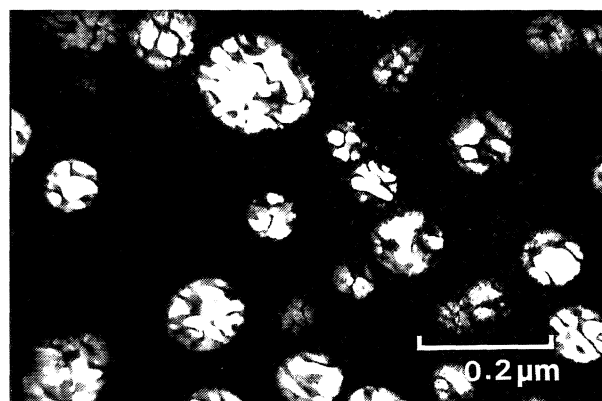


Figure 10. Microstructure of a Fe-23%Al intermetallic heat treated to show the α + α'' microstructure, that is a duplex α Fe + α'' Fe₃Al mixture. During deformation the α'' particles act to strengthen the material (ref. 7).

Figura 10. Microestructura de un intermetálico Fe-23%Al tratado térmicamente para obtener la mezcla de fases α + α'' , es decir la mezcla duplex α Fe + α'' Fe₃Al. Durante la deformación las partículas α'' contribuyen al endurecimiento del material (ref.7)

of this method of material fabrication is, of course, the considerable cost of processing which significantly reduces the opportunities for exploitation of the material. Nevertheless, considerable hopes exist for commercial application of such materials in aerospace, power generation and automotive applications.

6. SUMMARY

The present brief overview has attempted to show that the iron aluminide intermetallics, especially materials based on FeAl, offer considerable hope for their development from interesting laboratory samples to useful industrial materials.

The general plastic behaviour of such materials is dictated by the types of dislocation found, namely the Burgers vector of the overall dislocation, the core structure and its intrinsic mobility, and the decomposition structure of the dislocation and the possibilities for geometrical pinning at both room temperature and at elevated temperature. Such pinning at elevated temperature, brought on by thermally-activated modifications of core, decomposition state, or fault structure, has long been thought to be responsible for the anomalous increase in flow stress observed as the test temperature increases above room temperature. It is only recently, as the importance of immobile vacancies present in equilibrium at high temperature has become clear, that a theory

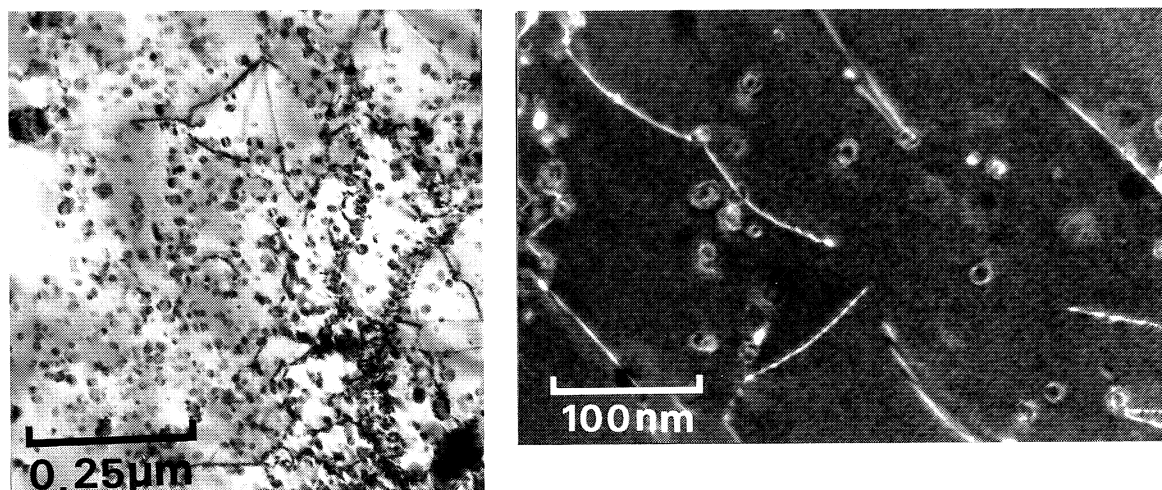


Figure 11. Y_2O_3 dispersoids in a MA Fe-40Al material, and dislocations pinned at the particles following high temperature deformation (taken from ref. 8).

Figura 11. Dispersoides de Y_2O_3 en un material MA Fe-40Al, y dislocaciones ancladas en las partículas después de la deformación a temperatura elevada (tomado de la ref.8).

based on the solid strengthening effect of such a solution of point defects has become widely accepted.

Such solid solution strengthening loses its effect under creep conditions at high temperature, as the many vacancies become sufficiently mobile that dislocation climb over obstacles is activated. For this reason, the only opportunities for high-temperature strengthening make use of extrinsic hardening methods, especially by incorporating a high density of fine and stable precipitate or dispersoid particles.

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