Preferred orientation in drawn austenitic stainless steel

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Abstract

This paper describes the development of crystallographic preferred orientations in a type AISI 304L austenitic stainless steel as a function of increasing reductions in area achieved by drawing at room temperature.
In order to investigate these crystallographic textures X-ray diffraction measurements were made on transverse and longitudinal sections of the drawn rods and results are expressed in terms of texture coefficients and inverse pole figures.
The results have shown that the initial duplex <111> + <100> fibre texture of the rolled and solution annealed starting rod develops towards a single <111> fibre preferred orientation along the fibre axis with increasing true drawing strain.
The single <111> fibre texture is consistent with the preferred orientations observed in low stacking fault energy face centered cubic polycrystalline metals and alloys and with the lack of dynamic recrystallization due to the low homologous temperature of deformation.
Martensitic α phase was detected in the cold worked rods since the early stages of plastic deformation and was found to be strongly textured with a single <110> fibre preferred orientation, typical of polycrystalline body centered cubic metals and alloys.

Riassunto

Orientazioni preferenziali in un acciaio inossidabile austenitico trafilato

In questa memoria viene descritto lo sviluppo delle orientazioni preferenziali in un acciaio inossidabile austenitico del tipo AISI 304L in funzione di crescenti riduzioni di sezione realizzate mediante trafilatura a temperatura ambiente.
Allo scopo di studiare queste tessiture cristallografiche si sono effettuate misure di diffrazione di raggi X su sezioni trasversali e longitudinali di barre trafilate ed i risultati sono stati espressi in termini di coefficienti di tessitura e di figure polari inverse.
I risultati hanno mostrato che la doppia tessitura iniziale <111> + <100> trovata per la barra di partenza laminata a caldo e solubilizzata evolve verso una tessitura di fibra singola con la direzione <111> lungo l'asse della barra al crescere della deformazione realizzata in trafilatura.
Tale tessitura singola <111> è in accordo con la tessitura di fibra dei metalli e leghe policristallini cubici a facce centrate aventi una bassa energia dei difetti di impilamento ed anche con l'assenza di ricristallizzazione dinamica in conseguenza della bassa temperatura omologa alla quale la deformazione è stata effettuata.
Gli esami per diffrazione di raggi X hanno anche messo in evidenza la presenza di martensite α nelle barre trafilate a freddo sino dai primi stadi della deformazione ed è stato trovato che tale struttura martensitica possiede una sviluppata tessitura di fibra singola con la direzione <110> parallela all'asse longitudinale della barra, tipica dei metalli e leghe policristallini cubici a corpo centrato deformati unidirezionalmente.

Introduction

The easiest way to strengthen a single-phase alloy is by cold working it; austenitic stainless steels, having a low stacking fault energy, are among the industrially used single phase alloys with a high work hardening capacity over a wide range of temperatures.
A further contribution to the strengthening of work hardened austenitic stainless steels comes from the fact that they undergo a partial martensitic transformation when cold work is done below their M_d, strain induced, martensitic transformation temperature.
The strengthening of austenitic stainless steels connected both with work hardening and with martensitic transformation has been studied extensively from many points of view, and many properties or in service behaviours have been related to the cold-worked state of these steels.
Notwithstanding the large amount of literature on cold-worked austenitic stainless steels (for general information and comprehensive collection of data the reader is referred to references 1-3), their textured crystallographic structure, however, does not appear to have been studied to any great extent. This is probably due to the fact that strong preferred orientations do not exist in industrial products since the sequences known to generate significant texture in face centered cubic metals and alloys (very extensive deformations, intermediate to elevated temperature deformations and specific heat treatments) are not applied in commercial practice. Research studies on the subject which have appeared in the literature are mainly related to the texture of rolled products (4-8) or to crystallographic preferred orientations in weldments (9-10).
It is well known that all random polycrystalline aggregates develop preferred orientations, or textures, upon sufficient plastic deformation and that the nature of the crystallographic preferred orientations depends essentially on the crystal structure and the flow characteristics.
Crystallographic texture is often an important consideration in metals and alloys, as preferred orientation can cause striking effects on the physical properties or the mechanical behaviour.
The study of preferred orientations has been of interest for a considerable period as it provides the tool to analyze the anisotropy in the properties of cold-worked materials, and in particular it was the growing use of X-ray diffraction techniques in the period before 1950 that stimulated an interest in the textures developed during deformation and annealing.
Since then, investigations into the nature of deformed metals have been performed by optical microscopy and etched specimens, X-ray diffraction, experimental evidence of dislocations and thin-foil electron microscope techniques and a sufficient understanding of both the microstructure of deformed metals and the origin of deformation texture is nowadays established.
The aim of the present research work is to investigate one aspect of the preferred orientation of austenitic stainless steels that has not yet been examined, namely the development of fibre texture upon subsequent drawing passes.

An austenitic stainless steel of the type AISI 304L was used in the present study and a total reduction in area limited to 75% was achieved by drawing, unlike cold work levels employed in previous studies on the development of fibre preferred orientations in other face-centered cubic metals and alloys (11-16); the above reported limit was chosen in order to investigate the range of cold work of practical applications of austenitic stainless steel, although in the literature more extreme reductions in area by drawing have been reported in order to develop particularly high tensile strengths in this type of steels (17).
The cold-work strain achieved by drawing was also limited in order to avoid too extensive martensitic transformation whose reflections in the X-ray diffraction patterns could introduce some errors in the evaluation of the preferred orientation of the austenite matrix. Furthermore, in the range of strain applied by drawing, a decrease in the elastic modulus was reported (17-18) and it has been hypothesized that such a decrease, as well as anisotropy of mechanical properties, among them Young modulus, may be related to the crystallographic texture, both in drawn (17) and rolled products (19). The results of the present research work will then be utilized as the basis to investigate the connection between crystallographic preferred orientation and elastic modulus in austenitic stainless steels.

Experimental methods

The material used was a commercial heat of type AISI 304L austenitic stainless steel, having the following chemical composition: C = 0.015%, Si = 0.39%, Mn = 1.67%, P = 0.018%, S = 0.026%, Ni = 10.37%, Cr = 19.45%, Mo = 0.10%, Cu = 0.04%. The steel was supplied in the form of a 30 mm dia. rod, in the 1050°C solution-annealed condition.
Room temperature plastic deformation was obtained by drawing; no intermediate heat treatments were introduced between the various cold-work levels.
The cold-work levels, given by the total reduction in area with respect to the initial 30 mm. dia. section, were 15.36%, 29.44%, 45.24%, 59.89 and 75.00%, corresponding to the following applied true strains ε_t = .17, .35, .60, .91 and 1.39.
A complete characterization of the mechanical properties of the steel as a function of the cold-work level is reported elsewhere (18).
The quantitative characterization of the preferred orientations was done by the X-ray diffraction technique. Diffraction data from the rod samples were obtained on the transverse section and, in some cases, on longitudinal sections passing through the rod axis; sample surfaces were prepared by conventional metallographic techniques with a final electropolishing in a sulphuric-phosphoric acid electrolyte to remove disturbed metal.
A Jeol X-ray diffractometer equipped with a filtered Mo Kα radiation was used; a scan speed of 0.5 degrees of 2Θ per minute has been utilized for all the diffraction measurements.
Diffraction intensity data were used, according to the inverse pole figure technique introduced by Harris (20) and later refined by Jetter and co-workers (21) and by Mueller and co-workers (22), to calculate the texture coefficients $T.C._{(hkl)}$. Inverse pole figures were chosen instead of direct pole figures because they yield better delineation of the texture, as they give the orientation distribution of the reference axis (the longitudinal axis of the rod) with respect to the standard crystallographic axes (21). Inverse pole figures, furthermore, provide a quantitative characterization of the preferred orientations and, in the case of fibre-textured specimens, such a characterization is complete, provided that the single grains are randomly orientated around the fibre axis (23).
In the inverse pole figure technique the diffraction intensities for the (hkl) reflections from the sample under examination are normalized with respect to the corresponding intensities from a specimen of random orientation. In order to obtain the texture coefficients $T.C._{(hkl)}$, or relative pole densities, the normalized intensities are divided by the average of all the normalized intensities, according to the following equation:

$$T.C._{(hkl)} = \frac{I_{(hkl)}/I_{o(hkl)}}{\frac{1}{n}\sum_1^n I_{(hkl)}/I_{o(hkl)}}$$

where I is the measured integrated intensity of a given (hkl) reflection in the sample under examination, I_o is the calculated theoretical intensity for the same (hkl) reflection in a randomly orientated specimen, and n the total number of reflections from the random specimen in the examined range of 2Θ. The values published in the ASTM X-ray powder data file for the Fe-austenite structure have been used for I_o.
The texture coefficients, calculated from the above reported equation, are proportional to the number of grains in the given orientation. These texture coefficients were plotted on a standard stereographic projection of the diffraction planes for the face centered cubic austenite crystals, which is called an inverse pole figure.
Isointensity contour lines were drawn around plotted points to include regions of similar texture coefficient values, in the same manner that isointensity lines are drawn on standard pole figures; in order to draw the isointensity contour lines with a continuous rate of slope change, it has been assumed that the pole

intensity variations between adjacent poles changed at a constant rate.
A value of texture coefficient greater or lower than unity indicates that the considered orientation is more or less frequent than in a random sample, while a value of unity represents a total lack of preferred orientation.

Results and discussion

Preferred crystallographic orientations resulting from large shears and rotations occurring during drawing of AISI 304L austenitic stainless steel are summarized in Fig. 1, where the inverse pole figures relating respectively to the starting rolled and solution annealed rod and to the subsequent cold-work levels are reported.
First of all it must be observed that a preferred orientation is already present in the starting material and therefore, at least in the first stages of cold-work deformation, the development of texture will be influenced by it. Many a time, in fact, it has been stressed in the literature that a common fault in many of the investigations of fibre textures is that the starting material has been assumed to be randomly orientated. The fibre texture of the rolled and solution-annealed rod is a duplex <111> + <100> one. The X-ray diffraction peaks of the double components were so strong that most of the crystallites within the rod can be assumed to have either <111> or <100> parallel to the axis of the rod itself.
The same observation can be made regarding the preferred orientation developed upon drawing in correspondence of the initial levels of strain; when increasing the strain, however, a variation in the relative amount of each component can be observed, with a progressive increase in the texture coefficient of the <111> component and a concurrent decrease in the (100) pole density. This is best illustrated in Fig. 2, where the inverse pole figure data showing the density of planes perpendicular to the rod axis are reported against the true drawing strain. The decrease in the (100) component pole density is such that for a true drawing strains greater than ≈ 1 a single <111> fibre texture is established within the rod.
The strong preferred orientation already present in the starting rod doesn't allow us to observe the development of texture predicted for the early stages of deformation (24), according to which, for preferred {111} <110> slip, crystallite orientations rapidly move away from the <110> orientation towards a broad band of orientations between <111> and <100>. The preferred orientations developed from such an initially textured rod may therefore be compared to some extent to the end texture of heavily deformed face centered cubic metals and alloys.
The preferred orientations which have been found for the drawn rod of type AISI 304L austenitic stainless steel are in agreement with the observation that the fibre texture of cold drawn face centered cubic metals and alloys consists of some combination of <111> and <100> orientations, whose relative amounts, according to English and Chin (13), are related to the stacking fault energy of the material. In fact, for the dimentionless parameter γ/Gb, where γ is the stacking fault energy, G the shear modulus and b the Burgers vector, using for γ values reported in the literature (25-26) for austenitic stainless steels very similar to the AISI 304L here examined, a value somewhat lower than $1 \cdot 10^{-3}$ can be calculated, corresponding to a low percentage of the <100> component in the duplex <111> + <100> fibre texture of cold drawn face centered cubic metals and alloys.
The <111> single fibre texture is also consistent with the fact that no dynamic recrystallization can occur during the plastic deformation as a consequence of the low homologous temperature at which the drawing passes were performed (16).
Results from X-ray diffraction measurements on longitudinal sections of some of the cold drawn rods ($\varepsilon_t = .17$, .35 and 1.39) have indicated that no cyclic or single crystal textures (23) were present; a texture coefficient higher than unity was measured only in the case of the (110) component, as could be expected since the <111> direction lies on it. The values of the texture coefficients for the (110) component on the longitudinal section have been found to increase with increasing cold work level ($T.C._{(110)} = 3.32$, 3.53 and 4.28 respectively for the above reported true drawing strains) and to be in nearly linear relationship with the texture coefficients of the <111> direction along the fibre axis.
Finally, as regards the strain-induced martensitic transformation, though a deep investigation of the matter goes beyond the aim of the present research work, first of all it must be reported that the X-ray diffraction peaks revealed the presence, from the early stages of cold work, of α martensite, while no martensitic ε phase could be detected throughout the range of applied drawing strains.
Martensitic α phase was found to be strongly textured with a single <110> fibre preferred orientation ($T.C._{(110)} = 6$, with all the other pole densities much lower than unity), typical of polycrystalline body-centered cubic metal and alloy wires (27).
No calculations of the relative amounts of austenite and α martensite phases were made since the preferred orientations of both the phases would introduce errors in the comparison of the relative integrated intensities of the X-ray diffraction peaks.

Fig. 1 - Axis distribution charts (inverse pole figures) for AISI 304L austenitic stainless steel rod after various drawing reductions: a) 0%, b) 15.36%, c) 29.44%, d) 45.24%, e) 59.89%, f) 75.00%.

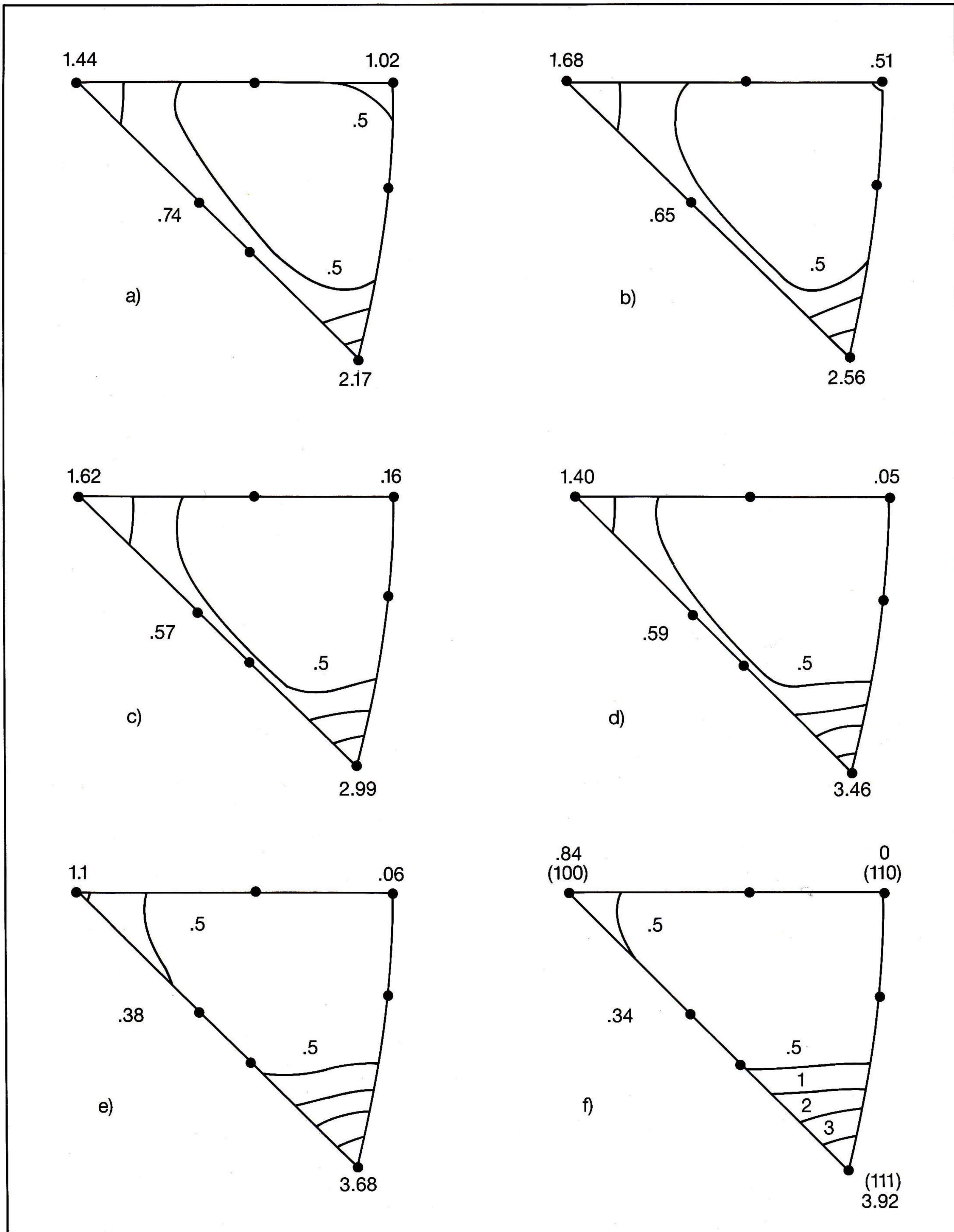
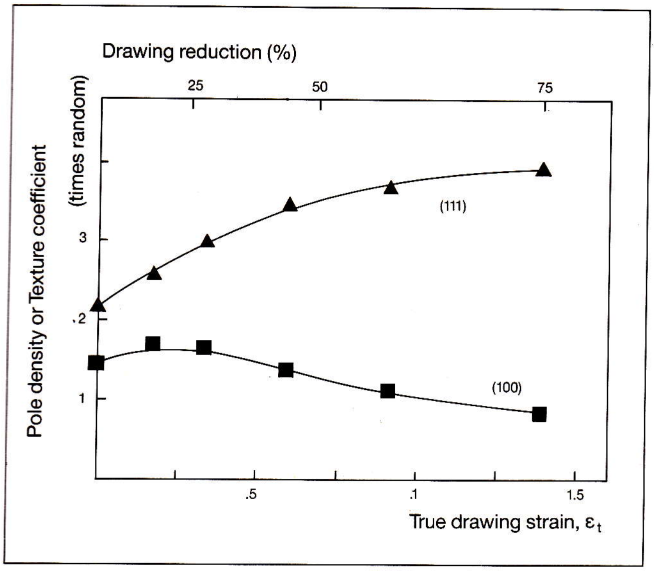


Fig. 2 - Inverse pole figure data showing the density of planes (111) and (100) perpendicular to the rod axis after various drawing reductions.



Conclusions

The preferred orientation developed in cold-drawn type AISI 304L austenitic stainless steel has been assessed by means of X-ray diffraction methods and quantitatively described in terms of inverse pole figures and pole density or texture coefficient values.
The initial duplex <111> + <100> fibre texture of the starting rod has been found to develop towards a single <111> preferred orientation along the fibre axis with increasing true drawing strain.
The single <111> fibre texture is consistent with the low stacking-fault energy of the AISI 304L stainless steel and with the lack of dynamic recrystallization due to the low homologous temperature of deformation.
Martensitic α phase has been detected in the cold worked rods from the early stages of plastic deformation and has been found to have a strong <110> single-fibre texture.

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